Laser Welding of High Strength Steels

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Abstract

S960 and S700 are two types of high strength low alloy steels (minimum yield strengths at 960 MPa and 700 MPa, respectively) developed recently by Tata Steel. These steels are typically used in heavy lifting equipment. This research examines the feasibility and characteristics of single pass autogenous laser welding (ALW), multi-pass ultra-narrow gap laser welding (NGLW) of 8 mm thick S960 and 13 mm thick S700 high strength low alloy (HSLA) steels and compared the characteristics of the welds with those of gas metal arc welding (GMAW). The work aims to understand the development of welding induced residual stresses, microstructures, microhardness, tensile properties, bending properties and Charpy impact toughness at different temperatures as produced by different welding techniques (ALW, NGLW and GMAW).

Design of experiments and statistical modelling were used to predict and optimise laser welding parameters of S960 and S700 HSLA steels. The contour method was used to measure the 2D distribution of residual stresses of the welded specimens. X-ray diffraction was carried out to measure the surface residual stresses of the welded specimens. The main novel contributions include:

(1) Development of welding procedures for ultra-NGLW of HSLA steels. The ultra-NGLW process was successfully applied to the welding of 8 mm thick S960 and 13 mm thick S700 HSLA steels with a very narrow groove (1.2-1.4 mm wide) using a moderate laser power (2-3 kW).

(2) Resolving the melt sagging problem for single pass autogenous laser welding of thick section materials. Horizontal (2G) welding position was applied to successfully resolve the melt sagging problem when single pass flat (1G) position ALW was applied to welding a 13 mm thick S700 steel plate. Computational fluid dynamic (CFD) modelling was carried out to understand the dynamic force interactions in the weld pool and the factors affecting the formation of the weld bead profile.

(3) Understanding the effects of heat input on the microstructures evolution and mechanical properties of the welded high strength steel joints. The much lower heat input for ALW of 8 mm thick S960 steel and ultra-NGLW of both 8 mm thick S960 and 13 mm thick S700 steels results in the generation of hard martensite in the narrow fusion zone (FZ) and heat affected zone (HAZ), which strengthened the welded joints but deteriorated the toughness of the welded joints. The strengthened narrow FZ and HAZ for both the ALW and ultra-NGLW of 8 mm thick S960 steels demonstrated almost the same tensile strength and elongation as the base material. A relatively high heat input for the ALW of 13 mm thick S700 steel results in the generation of bainite in the FZ, which has almost the same microstructure and hardness as the base material.

(4) Understanding the effect of solid-state phase transformation on the residual stresses of the welded specimens. It was demonstrated that the solid-state phase transformation from austenite to ferrite, bainite and martensite changes the magnitude of residual stress in the fusion zone for the welded S700 steel plates. In addition, it also changes the yield strength of the FZ, which also has a significant effect on the welding residual stress.

In summary, this work has resulted in a significantly enhanced understanding of the way in which the choice of welding process affects the properties of welded joints in high strength steels. Laser welding was found to offer strengthened welded joints. However, the laser welded joints presented low impact toughness. If the toughness of the laser welded joints can be improved, laser welding will be a promising technique for joining high strength steels.
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# Abbreviations

<table>
<thead>
<tr>
<th>Abbreviation</th>
<th>Full Form</th>
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<tbody>
<tr>
<td>AC</td>
<td>alternate current</td>
</tr>
<tr>
<td>AHSS</td>
<td>advanced high strength steel</td>
</tr>
<tr>
<td>ALW</td>
<td>autogenous laser welded</td>
</tr>
<tr>
<td>ANOVA</td>
<td>analysis of variance</td>
</tr>
<tr>
<td>AW</td>
<td>annealing + welding</td>
</tr>
<tr>
<td>AWST</td>
<td>annealing + welding + solution + tempering</td>
</tr>
<tr>
<td>BM</td>
<td>base material</td>
</tr>
<tr>
<td>BPP</td>
<td>beam parameter product</td>
</tr>
<tr>
<td>Bs</td>
<td>bainite start</td>
</tr>
<tr>
<td>CCD</td>
<td>central composite design</td>
</tr>
<tr>
<td>CCT</td>
<td>continuous cooling transformation</td>
</tr>
<tr>
<td>CE</td>
<td>carbon equivalent</td>
</tr>
<tr>
<td>CFD</td>
<td>computational fluid dynamic</td>
</tr>
<tr>
<td>CGHAZ</td>
<td>coarse grained heat affected zone</td>
</tr>
<tr>
<td>COIL</td>
<td>Chemical Oxygen-Iodine Laser</td>
</tr>
<tr>
<td>CTOD</td>
<td>crack-tip-opening displacement</td>
</tr>
<tr>
<td>CVN</td>
<td>Charpy V-Notch</td>
</tr>
<tr>
<td>CW</td>
<td>continuous wave</td>
</tr>
<tr>
<td>DC</td>
<td>direct current</td>
</tr>
<tr>
<td>DP</td>
<td>dual-phase</td>
</tr>
<tr>
<td>EC3</td>
<td>Eurocode 3</td>
</tr>
<tr>
<td>EDM</td>
<td>electric discharge machining</td>
</tr>
<tr>
<td>EDX</td>
<td>Energy Dispersive X-ray Detector</td>
</tr>
<tr>
<td>FEM</td>
<td>finite element modelling</td>
</tr>
<tr>
<td>FGHAZ</td>
<td>fine-grained heat affected zone</td>
</tr>
</tbody>
</table>
FI  factor interaction
FL  fusion line
FZ  fusion zone
GGHAZ grain growth heat affected zone
GMAW gas metal arc welding
GTAW gas tungsten arc welding
HAZ heat affected zone
HLB-SAW hybrid laser beam – submerged arc welding
HSLA high strength low alloy
HSS high strength steels
ICHAZ intercritical heat affected zone
IIW International Institute of Welding
LDH limiting dome height
M/A martensite/austenite
Ms martensite start
MS martensitic steel
NF normalization factor
NGLW narrow gap laser welding
PWHT post weld heat treatment
RPV reactor pressure vessel
RSM response surface methodology
RSW resistance spot welding
SAW submerged arc welding
SCHAZ sub-critical heat affected zone
SEM scanning electron microscope
SMAW shielded metal arc welding
TEM transmission electron microscopy
<table>
<thead>
<tr>
<th>Abbreviation</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>TMCP</td>
<td>thermo-mechanical control processing</td>
</tr>
<tr>
<td>TSAW</td>
<td>tandem submerged arc welding</td>
</tr>
<tr>
<td>UTS</td>
<td>ultimate tensile strength</td>
</tr>
<tr>
<td>XRD</td>
<td>X-ray diffraction</td>
</tr>
<tr>
<td>YS</td>
<td>yield strength</td>
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</table>
# Nomenclature

<table>
<thead>
<tr>
<th>Symbol</th>
<th>Definition</th>
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<tbody>
<tr>
<td>1G</td>
<td>flat welding position</td>
</tr>
<tr>
<td>2G</td>
<td>horizontal welding position</td>
</tr>
<tr>
<td>A</td>
<td>current (A)</td>
</tr>
<tr>
<td>$A_{c1}$</td>
<td>temperature corresponds to the start of the transformation to austenite (°C)</td>
</tr>
<tr>
<td>$A_{c3}$</td>
<td>temperature corresponds to the completion of the transformation to austenite (°C)</td>
</tr>
<tr>
<td>$b$</td>
<td>length of the remaining ligament at the notch of the Charpy impact test sample (mm)</td>
</tr>
<tr>
<td>$B$</td>
<td>specimen thickness (mm)</td>
</tr>
<tr>
<td>$C_p(T)$</td>
<td>specific heat at constant pressure as a function of temperature (J·kg$^{-1}$·K$^{-1}$)</td>
</tr>
<tr>
<td>$d_i$</td>
<td>the $i$th response desirability value</td>
</tr>
<tr>
<td>$g$</td>
<td>gravitational acceleration (9.8 m/s$^2$)</td>
</tr>
<tr>
<td>$h$</td>
<td>enthalpy (J/kg)</td>
</tr>
<tr>
<td>$h_{conv}$</td>
<td>convective heat transfer coefficient of air (W·K$^{-1}$·m$^{-2}$)</td>
</tr>
<tr>
<td>$i_{th}$</td>
<td>the $i$th response</td>
</tr>
<tr>
<td>$k$</td>
<td>thermal conductivity (W·m$^{-1}$·K$^{-1}$)</td>
</tr>
<tr>
<td>$K$</td>
<td>the drag coefficient for a porous media model in the mushy zone</td>
</tr>
<tr>
<td>$k(T)$</td>
<td>thermal conductivity as a function of temperature (W·m$^{-1}$·K$^{-1}$)</td>
</tr>
<tr>
<td>$L$</td>
<td>specimen span (mm)</td>
</tr>
<tr>
<td>$n$</td>
<td>the number of responses</td>
</tr>
<tr>
<td>$p$</td>
<td>pressure (Pa)</td>
</tr>
<tr>
<td>$P$</td>
<td>output laser power (kW)</td>
</tr>
<tr>
<td>$p_{hs}$</td>
<td>hydrostatic pressure (Pa)</td>
</tr>
<tr>
<td>$p_{\gamma}$</td>
<td>Laplace pressure (Pa)</td>
</tr>
<tr>
<td>$q$</td>
<td>power density (W/mm$^2$)</td>
</tr>
<tr>
<td>$Q_v$</td>
<td>volumetric heat flux (W/m$^3$)</td>
</tr>
</tbody>
</table>
$r$ radius of the sagging root surface (mm)

$r_e$ opening radius of the paraboloid (mm)

$r_0$ radius of any point of the paraboloid (mm)

$R^2$ a measure of the variation around the mean

$R^2_{\text{adj}}$ a measure of the variation around the mean of the adjusted model terms

$t$ time (s)

$T$ temperature ($\degree C$)

$T_0$ ambient temperature ($\degree C$)

$u(u, v, w)$ velocity components in the corresponding directions (m/s)

$V$ voltage (V)

$x_i$ variable $i$

$x_j$ variable $j$

$Y$ response of interest

$Y_i$ response $i$

$z$ the coordinate in the out-of-plane direction (mm)

$z_e$ the maximum possible value of the vertical coordinate on the paraboloid (mm)

$z_i$ the minimum possible value of vertical coordinate (mm)

$\alpha_0$ a random experimental error

$\beta$ a vector of unknown coefficients

$\beta_0$ response at the centre point

$\beta_i$ coefficient of the main linear components

$\beta_{ii}$ coefficient of the quadratic factor

$\beta_{ij}$ coefficient of the two linear factor interactions

$\gamma$ surface tension (N/m)

$\delta$ the overall desirability

$\eta$ heat source efficiency
\( \mathbf{v} \)  velocity vector

\( \rho(T) \)  density as a function of temperature \( (\text{kg/m}^3) \)

\( \nu \)  dynamic viscosity \( (\text{m}^2/\text{s}) \)
Chapter 1: Introduction

1.1 Research background and motivation

Due to their excellent strength-toughness combination, high strength/weight ratio and weldability, high strength low alloy (HSLA) steels have found wide applications as structural components, in, for example, pressure vessels, oil/gas transportation pipes, lifting equipment, vehicles and in the shipbuilding and offshore industries [1, 2]. Thin gauge HSLA steel (0.8-3 mm thickness) is mainly used in the automotive industry [3, 4]. For offshore pipelines, the current trend is towards the use of HSLA steels with a wall thickness up to 40 mm [5].

S960 and S700 are two types of low carbon, low alloy high strength steels developed recently by Tata Steel. These two high strength steels are produced by a thermo-mechanical control process. They are typically used in manufacturing heavy lifting equipment, with a minimum yield strength of 960 MPa and 700 MPa, respectively. Conventional gas metal arc welding (GMAW) is the primary technique applied to weld these high strength steels. GMAW introduces high heat input into the workpieces, which results in a large residual stress and distortion in the welded specimens. Meanwhile, these high strength steels exhibit wide soft heat affected zone (HAZ) when they are exposed to the welding thermal cycles, which leads to a deterioration of the mechanical properties of welded joints.

HSLA steels are widely employed in modern car manufacturing to reduce material cost and improve the transportation efficiency [6-9]. HSLA steels are also increasingly used in the construction of ships and offshore wind energy plants. The steel grades with higher yield strengths offer further possibilities for the design engineers in the construction of wind energy plants [10].

The application of high strength low alloy steels for structural components enables lighter and more slender products and reduces construction costs without loss of structural integrity [11, 12]. The use of HSLA steels in ships enables construction engineers to realise light-weight structures. The decrease of energy consumption of ships caused by a lower dead weight leads to an improvement regarding economical and
ecological aspects at a constant performance level [10]. Lücken et al. [10] reported that under tensile loading a thickness reduction can be realised compared to S355 of around 50% for the thermo-mechanically rolled 700 MPa high strength steel and up to 55% thickness reduction for 800 MPa high strength steel.

Welding is an indispensable manufacturing technique in the manufacture of large and complex structures [13]. However, in the welding of high strength steels, the matter of most concern is normally hot cracking which occurs only in the weld metal [14, 15], and hydrogen induced cracking in the heat affected zone (HAZ) [16, 17]. Hot cracking is also called solidification cracking, because it occurs immediately after welds are completed and sometimes while the welds are being made. Hot cracks are typically caused by excessive transverse stress, high depth to width ratio in excess of 2:1, high sulphur and phosphorus content [18, 19]. High strength steels are difficult to weld due to their high hardenability, which makes them prone to cold cracking [20]. These cracks often occur in the HAZ below the fusion boundary and are therefore sometimes termed ‘under-bead cracking’ [21]. They are attributed to the presence of hydrogen in the weld atmosphere [22] and high cooling rates, which promote martensite formation.

The widely used welding techniques for welding the HSLA steels in industry are conventional arc welding processes, such as gas tungsten arc welding (GTAW) or gas metal arc welding (GMAW), as well as submerged arc welding (SAW) and resistance spot welding (RSW). GTAW, GMAW and SAW techniques can be used to weld thick section materials, while RSW technique is mainly used to weld thin material [23-28]. Although these traditional welding techniques can satisfy the requirements for industry, the welding efficiency is very low because of their low penetration depth. In addition, the above traditional arc welding techniques can introduce a very large amount of heat input to the specimens [29], which results in large residual stresses and distortions in the welded specimens. Meanwhile, the HSLA steels exhibit heat affected zone (HAZ) softening when they are exposed to weld thermal cycles, which leads to deterioration of the mechanical properties of welded joints. It has been shown that increasing heat input leads to a wider soft heat affected zone [30]. Therefore, there is an exigent requirement to develop advanced welding techniques with high welding efficiency and narrow heat affected zones to obtain high mechanical performance welded HSLA steel joints.
Laser welding is a contact-free welding method that represents a fast and flexible technique, and the quality of the welds is normally excellent because of the accurate energy input, high power density and low heat inputs [31, 32]. Low heat input welding methods can maintain the fine grain structure that provides the strength and toughness of the welded joints, and also have the benefit of decreasing welding induced distortion [33]. Laser welds have characteristics of high aspect ratio with narrow heat affected zones and the distortion is low in contrast to other conventional arc welding methods. Therefore, laser welding is a promising welding technique to join HSLA steels to obtain high mechanical performance welded joints.

Although laser welding has the capability to weld medium and thick section materials using a single pass autogenous welding technique without filler material, there are still some disadvantages for single pass autogenous laser welding of thick section materials, such as the limits on the maximum welding penetration depth due to corresponding limits on the maximum power that is available with commercial lasers, the tight joint fit-up requirement because of the small spot size. Laser welding with filler material, in many cases, can relax somewhat the tight joint fit-up requirement for single pass autogenous laser welding and obviate the requirement of expensive high power lasers for welding medium and thick section structures.

Over the last few years, the multi-pass narrow gap laser welding (NGLW) technique has been demonstrated, which can be applied to welding medium and thick section components with a filler wire using relatively moderate laser powers [34, 35]. A filler material is generally used to fill a pre-prepared groove to join the material. Narrow gap configurations enable medium and thick section materials to be welded with a significant reduction in the required volume of material addition, and reducing the total welding time [36]. Therefore, NGLW is a potential new welding technique that could be applied to weld medium and thick section HSLA steel using a moderate laser power in the future.

Although there are many studies on laser welding of high strength steels [37-39], there are limited data reported on the S960 and S700 HSLA steels. Studies on the development of residual stress, microstructures evolution and mechanical properties of the narrow gap laser welded materials are scarce. Thus a knowledge gap exists. The lack
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of information on these two HSLA steels will affect the widely application in the structures components manufacturing industry.

1.2 Aim and objectives

1.2.1 Aim

The aim of this project is to understand the effect of heat input on microstructures evolution, mechanical properties and residual stresses of welded S960 and S700 HSLA steels by different welding techniques (GMAW, single pass autogenous laser welding and multi-pass narrow gap laser welding). Both of these two HSLA steels are newly developed high strength low alloy steels by Tata Steel. There are scarcities of reported data on these two HSLA steels. This research is intended to establish relationships between microstructures evolution, mechanical properties (such as micro-hardness, tensile properties, three-point bending properties and Charpy impact toughness) and residual stresses for S960 and S700 HSLA steels welded by different welding techniques (GMAW, single pass autogenous laser welding and multi-pass ultra-narrow gap laser welding).

1.2.2 Objectives

The objectives of this project are:

1) To develop single pass autogenous laser welding and multi-pass ultra-narrow gap laser welding (NGLW) techniques to weld 8 mm thick S960 and 13 mm thick S700 HSLA steels. These will be compared with the traditional GMAW technique.

2) To optimise multi-pass ultra-NGLW parameters to achieve sound welds in S960 and S700 HSLA steels using design of experiments and statistical modelling.
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3) To understand the microstructure evolution in the fusion zone and heat affected zone (HAZ) of the single pass autogenous laser welded, multi-pass ultra-NGLW and GMAW S960 and S700 HSLA steels.

4) To understand the mechanical properties including microhardness, tensile strength, three-point bending properties and Charpy impact toughness at different temperatures of single pass autogenous laser welded, multi-pass ultra-NGLW S960 and S700 HSLA steel joints in comparison to those of traditional GMAW joints.

5) To understand residual stress characteristics in single pass autogenous laser welded, multi-pass ultra-NGLW S700 HSLA steel joints in comparison with those of GMAW welded joints.

1.3 Scientific and technological challenges

This project has two main scientific and technological challenges: the first one is to obtain sound welds without defects in the welds for ultra-NGLW technique to weld S960 and S700 HSLA steels using a moderate laser power. The second challenge is to understand the effect of heat inputs on microstructures evolution, mechanical properties and residual stresses of welded HSLA steels by different welding techniques (single pass autogenous laser welding, multi-pass ultra-NGLW and GMAW). The phase transformation during the welding process, especially for multi-pass welding processes, and the microstructure evolution of the welded joints for HSLA steels are very complex. These phenomena determine the final mechanical properties and residual stress of the welded joints.

1.4 Thesis structure

An overview of welding of high strength steels is given in Chapter 2. The development of high strength steels and application of high strength steels are reviewed. The existing welding techniques for high strength steels and the effect of heat input on the microstructure and mechanical properties of the welded joints are discussed. In addition, the previous research on laser welding of high strength steels is also included in this
Chapter 1

Chapter. The main purpose of this literature review is to review the state of the art in laser welding of high strength steels and identify the existing knowledge gaps.

In Chapter 3, the single pass autogenous laser welding technique is developed to weld 8 mm thick S960 HSLA steel. The microstructure evolution and mechanical properties (including microhardness, tensile strength, three-point bending properties and Charpy impact toughness) were characterised and discussed.

The design of experiments and statistical modelling for the optimisation of multi-pass ultra-NGLW process for welding 6 mm/8 mm thick S960 HSLA steel are discussed in Chapter 4. Multiple parameter interactions were investigated and multi-objective optimisation was carried out.

In Chapter 5, the comparison of microstructure evolution and mechanical properties (including microhardness, tensile strength, three-point bending properties and Charpy impact toughness) were discussed for ultra-NGLW and GMAW of 8 mm thick S960 HSLA steel.

A comparative study of traditional flat (1G) and horizontal (2G) welding positions were carried out on 13 mm thick S700 HSLA steel in Chapter 6. A computational fluid dynamic (CFD) analysis was carried out to understand the dynamic forces in the weld pool and the factors affecting the formation of the melt sagging problem for the single pass autogenous laser welding of thick section materials.

In Chapter 7, the residual stress characteristics of single pass autogenous laser welded, multi-pass ultra-NGLW and GMAW 13 mm thick S700 steel specimens were discussed. X-ray diffraction and the contour method were used to measure the residual stress.

A general discussion that pulls together the varied work reported in the individual chapter is given in Chapter 8.

Finally, the conclusions and future work are given in Chapter 9.
References

Chapter 2: Literature review

2.1 Review of high strength steels

With the development of steel mechanical properties and steel production techniques, such as the thermo-mechanical control processing (TMCP), quenching and tempering processes, high strength steel structures have been widely applied, since the excellent properties of high strength steels and high performance steels have provided engineers an innovative solution to construct more efficient and economical steel structures [1-4]. According to the current European Standard Eurocode 3 (EC3) [5], the steels with a nominal yield stress equal to or above 460 MPa are called as high strength steel.

Thermo-mechanical control processing (TMCP) enables the economical production of high strength steel without the costs of quench and temper processing or expensive alloying of the steel. TMCP involves hot rolling of steel at carefully controlled temperatures (i.e. controlled rolling) and/or quenching of steel as part of the hot rolling process (i.e. direct quench). This process helps increase both the yield strength and tensile strength of the steel [6].

High strength low alloy (HSLA) steels have been developed for many years and achieved high tensile properties and good toughness [7]. HSLA alloys are classified as low carbon steels and achieve their high strength levels from low percentage alloying additions that result in precipitation hardening, or grain refinement strengthening mechanisms through thermo-mechanical processing [8-13]. Precipitation hardening is one of the most effective strengthening methods used in steels, and is achieved by producing a dispersion of particulates that serve as obstacles to dislocation movement [14, 15]. Microalloying elements, such as Nb, V and Ti, are most commonly used as the alloying elements to precipitate as carbonitrides [16-18], which can facilitate grain refinement through precipitation in austenite and acting as grain boundary pinning particles during reheating, and/or can contribute to dispersion hardening through strain induced precipitation during rolling or through fine scale precipitation after the austenite to ferrite transformation in these steels [19, 20]. Even though the strong carbide and nitride forming alloy elements (Ti, Nb, and V) added in the steel have a small amounts
(max 0.1 wt.%), they make the steel achieve a great improvement in their mechanical properties [21, 22].

The strength of HSLA steels comes from their special microstructures. In general, the microstructures of high strength steel are martensite and bainite. To achieve a combination of high strength and toughness, the microstructure of lower bainite or ferrite plus martensite has been designed for HSLA steels [23-27]. Hence, in spite of additions of alloying elements such as Mo and B to enhance lower bainite and martensite in these steels, rapid cooling after finishing rolling has been introduced. Conventional high strength steels, which obtain their strength through high alloying content and hence high carbon equivalent (CE) value, require the steels to be preheated prior to welding. In contrast, the low carbon content of the HSLA steels with low CE improves the weldability of HSLA steel and negates the requirement for weld preheating, which leads to significant cost savings through decreased labour and energy requirements and increased productivity [6, 10, 28]. The carbon equivalent (CE) can be calculated according to the following equation [29]:

\[
CE = C + \frac{Mn}{6} + \frac{(Cr + Mo + V)}{5} + \frac{(Ni + Cu)}{15}
\]  

(2.1)

2.2 Applications of high strength steels

Due to the excellent strength-toughness combination and weldability, high strength low alloy (HSLA) steels are used in a wide variety of applications as structural components, pressure vessel and linepipes, and in shipbuilding, offshore construction, automotive (pressed chassis and reinforcement parts, beams or welded tubes), seats (tubes, rails and mechanical elements), industrial vehicles, tractors, trailers and skips as chassis components, lifting and handling equipment (cranes, fork lifts, platforms, warehouse shelves, lifts), the agricultural sector for chassis and protective elements and roll bars, buildings, containers, urban lighting masts, concrete mixers [30-35]. Figure 2.1 shows the typical application of high strength low alloy steels [35]. The application of high strength low alloy steels for such components not only enables lighter and more slender structures, but also reduces construction costs without loss of structural integrity [36-38].
The weight of the structure itself has a substantial effect on the working load. A reduction of the weight of the construction itself but maintaining its loading capacity, given by the strength and safety of the construction, leads to an improvement regarding economical and ecological aspects at a constant performance level [39]. Figure 2.2 shows examples indicating that under tensile stresses the plate thickness can be reduced by approximately 60% by using the steel grade S960 instead of S355 [40]. Despite their high strength, these structural steels exhibit excellent ductility, toughness and good weldability, which have begun to be applied in many steel structures. There are strong demands world-wide on structure steel with higher strength [1-3].

Due to government regulations that require the reduction in greenhouse gas emissions and fuel consumption to protect our precious environment and mitigate the recently
realized the intensifying energy crisis, man-made global warming and smog, and consumer demands for light, and fuel-efficient vehicles, automotive industry is constantly seeking efficient methods to manufacture vehicles from lighter materials or the materials with higher strength and ductility so as to reduce the vehicle weight while guaranteeing improved safety, performance, and comfort [41-46]. All these demands continue putting pressure on the automotive manufacturers especially in an environment where there are ever-increasing global competition and continuous cost reduction demands [45].

Many materials have been involved in this competition to accomplish the requirement of automotive industries. Because of excellent strength-toughness combination and weldability of high strength and ductility, high strength steels (HSS) have been the winner, which have been increasingly utilized by automotive companies [47, 48]. High strength steel thinner sheets (0.8-3 mm thick and below) have been successfully used in automotive industry to reduce the weight of automobiles, improve the crash safety and low down the gas emissions [47, 49, 50].

There is a strong motivation in the crane manufacturing sector to increase loading capacity. Ultimate tensile strength is important to assure in weight lifting structures like cranes. High strength steels are attractive candidate materials for the cranes enables lighter and more slender structures, which effectively reduce dead weight of the cranes and lead to save fuel consumption and improve loading capacity without compromising the structural integrity [51, 52]. In addition, the use of high strength steels provides means for cost savings through design of lighter cranes, with less need for material and possible cost reductions during fabrication [53, 54].

Increasing demands for clean energy and increasing needs for transportation of higher volumes of oil and natural gas through high pressure steel pipelines, have led to an increasing demand for large amounts of steel for the installation of many pipelines around the world [55, 56]. Because pipelines are normally installed in severe environments such as permafrost or seismic regions, high performance linepipe steels with high strength, excellent low temperature toughness, good weldability and superior corrosion resistance are promising candidate materials for manufacturing pipelines [57-59]. Therefore, in recent decades the HSLA linepipe steels have been developed from
grade X60, X70 to the current X80 and X100 grades [23-27]. Such pipelines currently experience internal transmission pressures up to 15 MPa (and even higher) in low ambient temperature down to -40 °C [56]. The current trend is towards the use of grade X70 steels with a wall thickness up to 40 mm, and for X80 and X100 linepipe with wall thickness of 25 mm and below [23, 60]. The application of HSLA steels leads to increasing the operating pressure through thinner pipelines and results in an overall reduction in construction and transportation costs [56-59].

High strength steels have been employed to build some significant structures worldwide, including buildings, bridges and landmark constructions, such as New York Freedom Tower, Beijing Bird’s Nest Olympic Stadium and French cable-stayed road-bridge Millau Viaduct [61-64]. The Bird’s Nest was made using steel with the nominal yield strength of 460 MPa. A picture of the national stadium Bird’s Nest [65] is shown in Figure 2.3. In Berlin, Germany, the Sony Center utilised S460 (with the nominal yield strength of 460 MPa) and S690 (with the nominal yield strength of 690 MPa) high strength steels. In Sydney, Australia’s Star City hotel and Latitude building adopted steel structural products with the yield strength of 690 MPa [66].

![Figure 2.3 The Chinese national stadium Bird’s nest in Beijing [65].](image)

### 2.3 Existing traditional welding techniques for high strength steels

In most cases the structures are too large and complex to be produced. Generally these large structures produced fall into several components to be assembled using some connection techniques. Among the most important connection techniques, welding technology is an essential manufacturing technique in the manufacturing of large and
complex structures [67, 68]. Conventional arc welding processes, such as gas tungsten arc welding (GTAW) or gas metal arc welding (GMAW), as well as submerged arc welding (SAW) and resistance spot welding (RSW) are widely used to weld these HSLA steels [47, 52, 69-71].

SAE 4130 alloy steel is a heat treatable, high strength low alloy (HSLA) steel, which easily forms a martensitic structure after quenching. Lee et al. [72] investigated the effect of thermal refining on mechanical properties of annealed SAE 4130 by multi-pass GTAW. It was found from their research that the AW (annealing + welding) and AWST (annealing + welding + solution + tempering) occurred with a minimum hardness value at grain growth heat affected zone (GGHAZ) due to coarse grain growth; the hardness occurred with sudden drops between the multilayer welding, but was still larger than the minimum value at HAZ (heat affected zone). Mohandas and Reddy [71] compared continuous and pulse current GTAW of AISI 4340 high strength steels. The heat input was ~ 1.3 kJ/mm per pass and welding speed was 0.15 m/min. It was found from their investigation that continuous current welds exhibited columnar grain morphology, whilst pulse current welds had predominantly equiaxed grain morphology. The average fusion zone grain size was smaller in pulsed current welds and correspondingly they exhibited superior mechanical properties. Silva and Farias [73] used GTAW process to weld ASTM A106 Gr. B steel tubes with a 101.6 mm diameter and 6.6 mm thickness. The heat input was ~ 1.4 kJ/mm and welding speed was ~ 0.03 m/min.

There is a strong trend for increases in natural gas consumption worldwide, which implies continued growth of gas pipeline installation. In recent years the use of high strength steels has substantially reduced the cost of pipeline materials with X70 and X80 being commonly applied. Mechanised GMAW has now been successfully used for pipeline applications for over thirty years, and has achieved an impressive record on improving productivity over that time [74]. Yapp and Blackman [74] developed a novel narrow gap GMAW with dual torch, tandem GMA welding technique to do pipeline girth welding. This newly developed narrow gap tandem GMAW technique has been applied to pipeline girth welding with two tandem torches in a single welding head. This newly developed dual torch, narrow gap tandem GMAW system is shown in Figure 2.4 [74]. This system offered welding productivity three to four times higher than that possible with the conventional single wire GMAW technique, while still producing a
weld which was very similar to that generated by single wire welding. A cross section of narrow gap tandem GMAW system welded 14.9 mm thick X100 linepipe steel is presented in Figure 2.5 [74].

Figure 2.4 The Automated tandem GMAW pipeline girth welding system [74].

Figure 2.5 Cross-section of narrow gap tandem GMAW system welded 14.9 mm thick X100 linepipe steel [74].

Ahiale and Oh [75] carried out a comparative study of high cycle fatigue behaviours of butt weld joints in 2 mm thick advanced high strength steels with different strength levels and weld bead geometries. The welded joints were made by GMAW process on dual-phase steels (DP440 and DP590) and martensitic steel (MS) with tensile strengths of 440, 590, and 1500 MPa, respectively. The heat input was ~ 0.32 kJ/mm and welding speed was 0.6 m/min. The results show that the microstructures with the lowest hardness were found at the base metal, the sub-critical heat-affected zone (HAZ), and
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the fusion zone for DP440, DP590, and MS welded specimens, respectively. Fatigue failure of specimens without weld beads reinforcements occurred at the points of lowest hardness, and fatigue life exhibited the order MS > DP590 > DP440, similar to the order of lowest hardness values in each welded specimens.

Because submerged arc welding (SAW) offers many advantages such as ease of automation, low operator skill requirements, high deposition rate, excellent weld joint quality, ability to weld thick plates and minimum welding fumes, it has been used extensively in industry to fabricate pressure vessels, pipelines, offshore structures, marine vessels, and wind turbine towers [52, 76, 77].

Kirana and De [78] investigated the influence of leading wire current, trailing wire current pulses, and welding speed on the weld bead dimensions and mechanical properties in single pass two-wire tandem submerged welding of 12 mm thick typical HSLA steel plate. The yield strength, ultimate tensile strength and elongation of this typical HSLA steel were 497 MPa, 662 MPa and 27.9%, respectively. Two-wire tandem submerged arc welding process involved simultaneous depositions from two electrode wires with the leading wire usually connected to a direct current (DC) power source and the trailing wire connected to a pulsed alternating current (AC) power source. It was realized that the weld bead profile and mechanical properties in the tandem submerged arc welding were significantly affected by the leading and trailing wire current transients and the welding speed. The weld bead penetration was primarily influenced by the leading wire current while the weld bead width and the reinforcement height were sensitive to the trailing wire current pulses.

Moeinifar et al. [79, 80] investigated the influence of thermal cycles on the properties of the coarse grained heat affected zone in X80 HSLA steel. Four-wire tandem submerged arc welding (TSAW) system was used to weld 17.5 mm thick X80 HSLA steel plate. The wires position for the four-wire TSAW process is shown in Figure 2.6 [79]. The four-wire TSAW process, with different heat input values, was used to investigate the effect of different heat input on the welded microstructures. The heat input was ~ 3-4 kJ/mm per pass and welding speed was 1.7 m/min. The optical macrograph of the weld cross section for selected samples with different total heat input is shown in Figure 2.7. It was found that the martensite/austenite constituent appeared in the microstructure of
the heat affected zone region for all the specimens along the prior-austenite grain boundaries and between the bainitic ferrite laths. The fractional area of martensite/austenite particles due to different cooling rate was the main factor in increasing of the hardness values in the coarse grained heat affected zone.

![Figure 2.6 The wires position for the four-wire TSAW process [79].](image)

Shen et al. [52] investigated the effect of heat input on weld bead geometry of submerged arc welded ASTM A709 Grade 50 steel joints. They used single and double wires SAW to study the effect of heat input on bead reinforcement, bead width, penetration depth, contact angle, heat affected zone (HAZ) size, deposition area, penetration area and total molten area. It was found that the cooling time from 800 to 500 °C was related to various weld bead characteristics (i.e., total nugget area, heat transfer boundary lengths, bead width-to-depth ratio, and nugget parameter).

![Figure 2.7 Optical macrograph of weld cross section (a) sample with total heat input of 8.05 kJ/mm and (b) sample with total heat input of 6 kJ/mm [79].](image)
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HY-80 steel is a high strength low alloy (HSLA) steel with minimum yield strength of 550 MPa. The microstructure of this HSLA steel is combined tempered bainite and tempered martensite. Because of many attractive properties, such as good formability, weldability, and corrosion resistance, HY-80 steel has been applied in many engineering and marine constructions, including submarine pressure hulls. In order to study the mechanical performance of weldments of this HSLA steel, Yayla et al. [81] tried shielded metal arc welding (SMAW), gas metal arc welding (GMAW) and submerged arc welding (SAW) techniques on this steel with thickness of 22 mm.

Among the high strength steel, dual-phase (DP) steels have been the subject of particular attention for automotive industry owing to their good combination of high strength and ductility [47, 82]. Because of simple and cheap operation, resistance spot welding (RSW) is the primary sheet metal welding process in the manufacture of automotive assemblies [47, 83, 84]. Typically, there are about 2000–5000 spot welds in a modern vehicle [85, 86]. RSW are mainly applied to spot weld thin high strength steel sheet in automotive industry, generally around 0.8-3 mm [83, 86, 87].

2.4 Review of laser welding of high strength steels

The laser was invented in 1960 and is widely used in materials processing. Laser welding constitutes a large segment of laser materials processing market with a share of about 40% [88, 89]. The illustration of the two laser welding modes: conduction mode and keyhole mode is shown in Figure 2.8 [90]. Conduction mode welding occurs when the power density at a given welding speed is insufficient to cause boiling and therefore to generate a keyhole. Conduction mode welding is used to melt material to create a joint without significant vaporisation. While for the alternative keyhole mode welding, energy per unit length is sufficient to cause evaporation and a hole in the melt pool [91, 92]. If the laser beam is defocused and the heat input is very low, conduction mode welding is the main mechanism involved which results in high conduction losses, leading to low penetration depth and low process efficiency. Typically, conduction weld beads present a semi-spherical geometry with penetration lower than the bead width. A keyhole is formed if the laser beam irradiates the workpiece with high power density (typically >10^6 W/cm^2) [93, 94], which enables the laser to penetrate deeply and deep penetration mode mechanism becomes predominant, which can increase the welding
efficiency [95]. Low power laser is usually applied to weld thin materials in conduction mode, while high power laser can be used to weld thick materials as deep penetration keyhole mode [89].

![Conduction mode and keyhole mode laser welds](image)

Figure 2.8 Conduction mode and keyhole mode laser welds [90].

The power density in laser welding is on the order of up to $10^8$ W/cm$^2$, which is about 4 orders of magnitude higher than in conventional arc welding methods [96]. Therefore the influence of laser welding on the base metal in the vicinity of the weld is considerably lower and the microstructure degradation is minimal [96]. Laser welding is a contact-free welding technique which represents a fast and flexible technique, and the quality of the welds is normally excellent because it offers a unique combination of high welding speed, accurate energy input, high power density, low heat inputs and low heat distortion, compared with the conventional welding process [97, 98]. Besides its advantages, there are also some drawbacks with the laser welding. The capital investment for laser equipment is high. Due to the small spot size of the laser beam, the process requires stringent parts positioning, with maximum allowable gaps typically in the range of 0.1 times or less the thickness of the material [99]. The high welding speeds for laser welding lead to high solidification or cooling rates. These characteristic may in turn lead to pores or cracking in the weld [100, 101]. The rapid cooling rate could also result in the increase of hardness in the weld and HAZ, which could deteriorate the toughness of the welded joints [102]. Finally, laser welding of highly reflective materials such as aluminum, copper, and gold is more difficult than steel [103].
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Until now resistance spot welding (RSW) has been widely used in automobile fabrication, however, RSW requires many welding machines because of geometrical and structural problems [85, 86, 104, 105]. The combination of the above advantages of laser welding attracts more and more automotive manufacturers’ attentions recently [97, 104, 106, 107].

As an innovative technique, laser welding has the potential to substitute the conventional resistance spot welding (RSW) technique to save equipment cost, reduce vehicle emission, and improve fuel economy and welded products quality without compromising the safety, structural integrity and crash performance of the vehicles [102, 104, 108, 109]. With the development of new high power and high brightness laser sources, the total weld seam length of lap welded sheets in some automobiles reaches up to 100 m [110].

Xia et al. [111] investigated the effects of laser welding on strength and formability of high strength low alloy (HSLA) and DP980 (Dual Phase, 980 MPa) sheet steels. A 4 kW diode laser was used to weld these two sheet steels. Tensile, microhardness and limiting dome height (LDH) tests were used to assess the mechanical properties and formability of the welded specimens. It was found that the microstructure of the DP980 steel fusion zone was mainly comprised of martensite, the fusion zone of HSLA steel mostly consisted of ferrite side plate and bainitic structures and the soft HAZ of DP980 steel was tempered martensite, as seen in Figure 2.9(a), (b) and (c), respectively [111]. The formability of HSLA steel welded joints was insensitive to the welding process and comparable to that of parent metal. For the DP steel, welded joints formability was much lower than that of corresponding parent metal, which appeared to be due to the formation of soft zones in the outer region of the heat affected zone (HAZ) of the welds. The side views of representative dome height tested specimens are shown in Figure 2.10. In addition, it was found that an increase of welding speed resulted in a slight increase of formability of the DP steel, associated with a reduction in the microhardness difference between base metal and HAZ soft zones [111].
Figure 2.9 Microstructure of fusion zones and soft HAZ, (a) fusion zone of DP980 steel, (b) fusion zone of HSLA steel, (c) soft HAZ of DP980 steel [111].

Sharma et al. [112] investigated weldability of three grades of advanced high strength steels (TRIP780, DP980 and USIBOR) with thickness variation from 1–2 mm using a 6 kW Yb:YAG disk laser. They used penetration, weld profile, weld defects, microhardness and melting efficiency to evaluate the weldability of the steel.
Microhardness measurements indicated a substantial increase in hardness in the fusion zone. However, softening in the heat affected zone was observed in DP980 steel.

The capability for optical fiber beam delivery and the use of industrial robots makes laser welding process very flexible, easily-automated and robotized in manufacturing industry [102]. On the other hand, industrial lasers have become a very reliable and stable tool with a wide range of features and capabilities, including an excellent service that allows to use them in a very large range of applications, such as automobile, shipbuilding, aerospace, computer, medical, electronic and other industries [97, 102, 104, 110, 113]. Fiber lasers have the characteristics of small beam divergence, low maintenance costs, high efficiency, high precision and reliability, and compact size, and consequently they have attracted more attention over the last decade for cutting, welding and cladding applications [95, 107, 113, 114].

Xu et al. [114] evaluated the microstructure and mechanical properties of high-speed fiber laser welded 1.2 mm thick high strength low alloy (HSLA) and DP980 dual-phase steel joints with varying weld geometries. It was found that fusion zone consisted of martensitic structure, and HAZ contained newly-formed martensite in both steels and partially tempered martensite in DP980 steel side. While HAZ-softening was presented in DP980 steel, it was absent in HSLA steel side, as seen the representative hardness profiles across the HSLA and DP980 steel joints in Figure 2.11 [114]. Both the HSLA and DP980 steel welded joints showed a superior tensile strength, with a joint efficiency of 94–96% and 96–97%, respectively. Fatigue strength was higher in DP980 joints than in HSLA steel joints at higher stress amplitudes, but had no obvious difference at lower stress amplitudes. DP980 steel multiple linear welds exhibited a larger scatter and lower fatigue strength. Fatigue failure of HSLA steel joints occurred in the base metal at a stress amplitude above 250 MPa, and at weld concavity at a lower stress amplitude below 250 MPa. Fatigue crack in DP980 joints initiated predominantly from the weld concavity at both high and low levels of stress amplitudes.
Saha et al. [115] and Parkes et al. [116] investigated similar and dissimilar welds of 1.2 mm thick dual-phase (DP980) and high strength low alloy (HSLA) steels made by fiber laser welding. It was found that the fusion zone in the DP980 welds was consisted of fully martensitic structure, whereas HSLA and dissimilar weld fusion zone microstructure were mixture of martensite and bainite. The HAZ contained some newly formed martensite and partially tempered martensite on the DP980 steel side. A very narrow softening area was found in the DP980 steel side subcritical heat affected zone for both the similar DP980-DP980 steel and dissimilar DP980-HSLA steel welded joints, while it was absent on the HSLA side.

With the development of laser technology, new generation high power fiber lasers are available, which presents several benefits for industrial purposes, such as high power with low beam divergence, flexible beam delivery, low maintenance costs, high efficiency and compact size [95]. The new generation of high power fiber lasers can be used to weld thicker section material using single pass autogenous welding process.

The ASTM A387 Grade 91 steel is a ferritic-martensitic steel, which is extensively applied in super critical power generation plants, nuclear power systems and in the petrochemical industry [117-119]. Kumar et al. [119] used neutron diffraction to characterise the residual stresses in 9 mm thick single pass autogenous laser welded ASTM A387 Grade 91 steel plates. They compared the microstructures in the fusion zone and HAZ, and residual stresses of the welded specimens under two different heat
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inputs, namely 0.32 kJ/mm and 0.64 kJ/mm. The cross section macrostructures of the two welded joints and the lines along which neutron diffraction measurements were made are shown schematically in Figure 2.12 [119]. It can be found that the higher heat input welded joint has a much wider fusion zone and HAZ. The microstructure of the base metal was tempered martensite. The microstructures for the fusion zone and HAZ for two different heat inputs specimens are presented in Figure 2.13 [119]. It can be found that the fusion zones in both the welded joints showed an as-transformed martensitic structure within columnar prior austenite grains. There were no significant differences in the microstructures of the two specimens welded with different heat input. The HAZs in both the specimens showed an as-transformed martensitic structure within re-austenitized grains of varying size, in addition, some tempered martensitic structure were found in the HAZ. Finally, it was found that the measured longitudinal and normal components of residual stress showed a bimodal distribution across the welded joint with a low tensile or compressive trough at the weld centre flanked by high magnitude tensile peaks in parent metal adjacent to the heat affected zone boundaries. The width of the central trough and spread of the out board tensile zones were significantly greater for the high heat input weld.

![Figure 2.12 Macrograph showing welded joint cross-sections with neutron diffraction measurement lines marked, (a) heat input: 0.32 kJ/mm, (b) heat input: 0.64 kJ/mm [119].](image)
In Sumi et al.’s [120] research, high-strength steel plates with different carbon contents and carbon equivalents Ceq-WES were used to investigate the single pass autogenous laser weld metal microstructure and mechanical properties of the laser weld metal of high-strength steels. A CO$_2$ laser was used to weld the 12 mm thick high-strength steel plates with laser power of 13 kW and welding speed of 0.8 m/min. Figure 2.14 shows an example of a cross-sectional macrostructure of a full penetration laser weld [120]. The results revealed that higher carbon (C) contents caused the microstructure of the laser weld metal to change from lower bainite to martensite. Figure 2.15 shows examples of scanning electron microscope (SEM) images of the microstructure of the respective laser weld metals [120]. It was revealed from their study that laser weld metal toughness was strongly dependent on the C content of the steel plates, and excellent toughness can be obtained in weld metals with low C contents having an almost fully lower bainite structure without formation of slender martensite austenite constituent (MA). Laser welding of high-tensile-strength steel with a high C content resulted in poor toughness of the weld metal.
Figure 2.14 Example of cross-section of laser weld [120].

Quintino et al. [95] reported a brief review of the development of high power lasers, and presented some initial data on welding of 19 mm thick API 5L: X100 pipeline steel with a 8 kW fiber laser. API X100 steel is used for pipeline applications with yield strength of 735 MPa. Weld bead geometry was evaluated and transition between conduction and deep penetration welding modes was investigated. Their experimental results were analysed on the basis of the relationships between process parameters and weld bead shape and dimensions. Figure 2.16 [95] shows a plot of the results for relationship between the attained penetration depth and heat input. It can be observed that for low heat input (between 10 and 100 J/mm) the penetration depth was very low and the conduction mode prevailed. For higher heat input (from 600 to 1000 J/mm) the penetration depth increased dramatically and a deep penetration mode was observed.
The relationships between penetration depth versus laser power and welding speed represented in Figure 2.17 and Figure 2.18 [95], respectively, it can be found that penetration depth follows the usual trends observed in laser welding, namely, the weld depth has an almost direct linear relationship with the laser power and decreases exponentially with the welding speed [91].
Modern high-power laser sources allow advanced thick plate welding applications and enable a stable single pass welding process even for thick metal plates. Vollertsen et al. [121] reported an autogeneous full penetration laser beam welding of up to 20 mm thick carbon steel in a single run with 19 kW laser power. Even though high power is available to be used to weld thick section material, there is a limited maximum laser power available for commercial lasers (typically less than 20 kW). It is reported that laser penetration depth in a single pass welding is typically on the order of 1-2 mm/kW [122, 123]. Even for the most advanced fiber laser, the penetration depth for single pass autogenous laser welding that can be achieved is less than 25 mm [124, 125]. Katayama et al. [126] achieved 50-70 mm penetration depths in 304 stainless steel using 10 kW and 16 kW high power disk lasers at pressures of 0.1 kPa. To obtain low local vacuum around the workpiece is unrealistic for industrial production to weld large and thick plates.

Although single pass autogenous laser welding has been widely applied in industry, the major drawbacks of the single pass autogenous laser welding process are stringent joint tolerances and limited available laser powers at hand [105]. In addition, medium and thick section laser welding with a single pass approach often results in porosity, molten material dropout, cracking, and mis-tracking of weld seams [127-129]. The use of filler material in many cases can widen the use of laser welding by enabling good weld quality whilst retaining its many features. Firstly, the tight joint fit-up requirement for laser welding can be relaxed somewhat by using the filler material. Secondly, the use of
filler material makes it possible to tailor the weld properties by modifying the fusion zone composition and improve weld properties to satisfy the specific requirement of the product. Thirdly, multi-pass welding with filler wire technique makes it possible to weld medium and thick section materials using the available lasers and occasionally decrease the capital investment and obviate the need for expensive very high power lasers [130, 131].

Ohnishi et al. [132] investigated weld penetration characteristics on butt welding of 590 MPa high strength steel plates with a high power laser over gaps. The study was conducted on butt welding of 12 mm thick plates with a 16 kW high power disc laser together with a hot wire. It was found that penetrations were stable over a range of gaps from 0 to 0.4 mm, owing to an ejection of excess melt through a keyhole outlet at the bottom of the molten pool by a strong plume. Figure 2.19 shows the appearances of the beads and weld cross-sections with gap widths [132]. It was noted that laser welding with a hot wire can produce a fully penetrated keyhole and effectively improve the gap tolerance while suppressing both burn-through and underfilling defects [132].

<table>
<thead>
<tr>
<th>Gap width, g</th>
<th>0 mm</th>
<th>0.2 mm</th>
<th>0.4 mm</th>
</tr>
</thead>
<tbody>
<tr>
<td>Laser-irradiated appearance of weld bead</td>
<td>![Image]</td>
<td>![Image]</td>
<td>![Image]</td>
</tr>
<tr>
<td>Bottom surface appearance of weld bead</td>
<td>![Image]</td>
<td>![Image]</td>
<td>![Image]</td>
</tr>
<tr>
<td>Cross section of weld bead</td>
<td>![Image]</td>
<td>![Image]</td>
<td>![Image]</td>
</tr>
</tbody>
</table>

Figure 2.19 Typical bead appearances and cross-sections of full penetration joints produced by laser butt welding over gaps from 0 to 0.4 mm [132].

Hybrid laser/arc welding, as one of the promising welding techniques for joining medium and thick plates, retains the advantages of both the laser welding and conventional gas metal arc welding (GMAW) [10, 133]. The combination of the laser
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and the arc has a synergistic effect on increasing the penetration depth, raising the degree of gap tolerance and possibility to control the chemical composition of a weld [134-136]. Hybrid process introduces not only filler wire, but also extra heat source from the arc which will make the welding process more effective. However, excess heat input may act to soften the heat affected zone [132].

Gao et al. [137] reported that typical hybrid laser–arc welding process could be classified as two parts: the wide upper zone and the narrow nether zone, which were defined as arc zone and laser zone, respectively. This phenomenon is consistent with Cao et al.’s [10] results that the laser beam mainly melts the base metal and then the arc mainly melts the filler wire on the molten pool, forming a wide weld bead in the hybrid laser–arc welding process [10]. Cao et al. [10] successfully developed a hybrid fiber laser–GMAW process to fully penetrate 9.3 mm thick HSLA-65 steel plates using a single pass technique. A 5.2 kW continuous wave (CW) fiber laser was combined with a GMAW system in their hybrid welding system.

Ni et al. [17] investigated the microstructure and mechanical properties of a laser–GMAW hybrid welded 15 mm thick microalloyed steel plate. Their investigated results showed that the yield strength and ultimate tensile strength of weld metal were up to 713 MPa and 918 MPa, respectively. Both of them were almost 1.5 times higher than those of base metal. The Charpy impact absorbed energy at -40 ºC was also higher than that of base metal. Weld metal predominately consisted of granular bainite and carbon-free bainite. Both of them mainly contained lath morphology bainitic ferrite. The lath morphology bainitic ferrite with fine grain size played an important role in higher strength.

Double side single pass laser welding or hybrid laser–GMA welding shows a tendency to pore formation in the lower weld regions [138]. The main reason for the pore formation with laser welding processes is that it is difficult to degas completely through keyhole during laser welding of thick materials in short solidification time. In order to increase the liquidity interval of the molten pool to improve the degassing capabilities, Reisgen et al. [138] combined laser welding and submerged arc welding together to weld 38 mm thick X65 (L460MB) steel plates. Figure 2.20 shows the schematics of the hybrid laser beam – submerged arc welding (HLB-SAW) process [138]. Figure 2.21
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shows the cross and longitudinal sections of a double side, single pass HLB-SAW with smooth transitions from weld metal to the work piece surfaces [138]. The arc zones and laser zones are identified clearly.

Figure 2.20 Principle of laser beam submerged arc hybrid welding [138].

Figure 2.21 Cross- and longitudinal section of double sided, single pass HLB-SAW [138].

Grünenwald et al. [139] investigated high power fiber laser welding of thick X70 pipe steel. In their experiment, an 8 kW fiber laser was hybrid with GMAW to weld 14 mm thick X70 pipe steels. They used the hybrid laser–GMAW process for the root pass and a GMAW process for the fill or cap pass with as less passes as possible to increase the productive efficiency for pipe production. The joint preparation was 6 mm thick root face and 45° included angle welded with 0 mm gap. Figure 2.22 shows the cross-sections of a root pass and fully welded joint [139]. There are no indications of visible defects in the welded joint.
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Chen et al. [140] developed a novel double-sided hybrid fibre laser-arc welding procedure to join high strength steel, and this procedure was compared with conventional double-sided arc welding. The double-sided hybrid laser-GMAW system is shown in Figure 2.23 [140]. The double-sided hybrid laser-arc welding system consisted of two hybrid laser-arc welding systems, which were symmetrically laid on both sides of the workpiece in a horizontal position. The welding was completed with the laser leading the arc in the welding sequence with four passes, while thirteen welding passes were required for the conventional double-sided GMAW process. The weld-bead profile of the welded joint for double-sided hybrid laser-GMAW is shown in Figure 2.24 (a) [140]. The double-sided hybrid laser-GMAW process was highly efficient and saved material when utilised for joining thick-section high-strength steel when compared with the conventional double-sided GMAW process. Optical microstructures of the base metal, laser zone, and arc zone are shown in Figure 2.24(b)–(d), respectively [140]. Their results indicated that the laser zone and arc zone predominately consisted of lath martensite with a high dislocation density. The average grain sizes of the laser zone and the arc zone were smaller than that of the base metal. The results also indicated that the laser zone and the arc zone possessed higher strength when compared with the base metal because of the fine lath martensite. Meanwhile, the strength observed in the laser zone was slightly higher than that of the arc zone due to the small average effective grain size. On the contrary, the toughness of the base metal was higher than the toughness in the laser zone and the arc zone because of massive
polygonal ferrites. Meanwhile, a significant increase in the toughness of the laser zone when compared with the arc zone occurred due to an increase in the prevalence of grain boundaries with large misorientation angles.

Figure 2.23 Double-sided hybrid laser-arc welding system: (a) equipment and (b) schematic [140].

Multi-pass narrow gap laser welding is another promising advanced laser welding technique using filler material. Over the last few years, the multi-pass narrow gap laser welding technique has been demonstrated, which can be applied to weld medium and thick section components with a filler wire using relatively moderate laser powers [122, 141-144].
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Elmesalamy et al. [122, 142] successfully welded 10 and 20 mm thick 316L stainless steel using a 1 kW IPG single mode fibre laser with an ultra-narrow gap (1.5 mm gap width) and a groove angle of 6°. Double sided multiple-pass narrow gap approach was carried out to weld 20 mm thick 316L stainless steel. The cross sections of narrow gap laser welded 10 and 20 mm thick 316L stainless steel are shown in Figure 2.25 [142].

Yu et al. [145] developed a multi-pass narrow gap laser welding of 17 mm Q235 low-carbon steel plate. In order to study the effect of bevel angle and size on the weld quality, three kinds of groove with different bevel angle, opening width and root face height were investigated. A high speed camera was used to real-time monitor the welding process. Experimental results showed that using a relatively small groove size not only reduce the consumption of filler wire but also reduce the deflection of filler wire in the groove that can improve the fusion of groove. As a result, more energy was used to melt the side wall to avoid lack of fusion on the side wall of the groove. Figure 2.26 shows the images taken by the high speed camera and the cross section of the welded joint [145]. This fact implied that reduction in the groove size can eliminate the lack of fusion defect on the side wall of the groove.
2.5 The effect of heat input on the microstructure and mechanical properties of the welded joints

Welding is a local rapid heating and cooling process in a material, and a rapid melting and solidification accompanies this process. During the welding process, the heat generated by the heat source produces thermal cycles that generate the formation of heat affected zone (HAZ). The welding thermal cycle has its own unique features, such as rapid heating and cooling, high peak temperature, and negligible hold time at the peak temperature, which produces the continuous changes of microstructures and mechanical properties (hardness, strength and toughness) through the fusion zone and the HAZ [146, 147]. The properties of the welded joints are determined by the characteristics of the heat affected zone (HAZ) [96, 148]. It is generally accepted that heterogeneous microstructure in the HAZ plays a vital role in deteriorating the mechanical properties of welded joint of high strength steels [148, 149].

The cooling rate is one of the most important factors influencing microstructure evolution in either fusion zone or HAZ [150]. The higher values of the heat input, the slower the cooling rate would be, and vice versa [151]. Heat input is a major parameter affecting the microstructure and mechanical properties of the fusion zone and HAZ, which can be controlled by suitably selecting welding speed and input power [69, 152]. During the process, difficulties are experienced in controlling the heat input [153]. Heat input for arc welding can be calculated by the following equation: heat input (J/mm) =
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efficiency \times \frac{(voltage (V) \times current (A))}{welding speed (mm/s)}. The consumable electrode process, such as GMAW, exhibits an average arc efficiency of 80\% [154]. Heat input for laser welding is calculated by the equation: heat input (J/mm) = efficiency \times laser power (W)/welding speed (mm/s). The welding efficiency considering the conduction heat loss only is about 80\% for autogenous laser welding [155, 156]. In addition, some beam reflection from the filler wire takes place in the ultra-NGLW process, which could result in 30\% loss of energy applied into the workpieces [157].

Shi et al. [36] investigated the effect of welding cooling time and peak temperature on fracture toughness and microstructure feature of HAZ for an 800 MPa grade HSLA steel using thermal simulated specimens. They investigated the effect of cooling time from 800 °C to 500 °C, namely \( t_{8/5} \) on the microstructure and fracture toughness of HAZ. It is well known that lath martensite exhibits good toughness. However, the martensite/ austenite (M/A) gives a significant effect on toughness of welds. Several publications have indicated that the M/A island is the main microstructure controlling toughness [158-163]. Because M/A constituent is a hard and brittle particle, and is usually regarded as the sites of crack initiation [164]. Table 2.1 gives the results on effect of cooling time \( t_{8/5} \) on the volume fraction of M/A constituent [36]. With increasing the cooling time \( t_{8/5} \), the volume fraction of M/A constituent was increased. The maximum fraction of M/A constituent appears at \( t_{8/5} \) equal to 100 s. In fact, the higher the amount of M/A in the simulated HAZ, the lower the toughness was. Moreover, with increasing the cooling time \( t_{8/5} \), the morphology of M/A constituent would be transformed from bar to block. Typical morphology of M/A constituent is shown in Figure 2.27 with different cooling time [36]. They found that when the cooling time \( t_{8/5} \) was 18 s, the fracture toughness in the simulated HAZ was the highest. However, the toughness was deteriorated when the value of \( t_{8/5} \) was 45 s or longer. These results implied that the weld heat input should be restricted to limit the cooling time \( t_{8/5} \) equal to about 18 s [36].

36
Table 2.1 Effect of cooling time $t_{8/5}$ on volume fraction of M/A constituent [36].

<table>
<thead>
<tr>
<th>$t_{8/5}$ (s)</th>
<th>M/A constituent (%)</th>
</tr>
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<tbody>
<tr>
<td>9</td>
<td>9.6</td>
</tr>
<tr>
<td>18</td>
<td>12.8</td>
</tr>
<tr>
<td>27</td>
<td>15.2</td>
</tr>
<tr>
<td>45</td>
<td>17.8</td>
</tr>
<tr>
<td>100</td>
<td>26.2</td>
</tr>
<tr>
<td>240</td>
<td>24.2</td>
</tr>
</tbody>
</table>

Figure 2.27 M/A constituent with different cooling time: (a) $t_{8/5} = 18$ s, (b) $t_{8/5} = 45$ s, (c) $t_{8/5} = 100$ s, (d) $t_{8/5} = 240$ s [36].

Hu et al. [165] investigated the effect of heat input on the microstructures, hardness, and toughness in the simulated welding coarse grained heat affected zone (CGHAZ) of V–N high strength steel. The microstructural evolution, hardness, and toughness subjected to four different heat inputs were investigated. The results indicated that the hardness decreased with an increase in the heat input, while the toughness first increased and then decreased. Moderate heat input was optimum, and the microstructure was fine polygonal ferrite, granular bainite, and acicular ferrite with dispersive nano-scale V(C,N) precipitates. Figure 2.28 shows the micrographs of the simulated CGHAZ specimens for different $t_{8/5}$ [165]. As the $t_{8/5}$ 10 s, the microstructure was composed of lath martensite, lath bainite, and a small amount of primary Widmanstatten, as shown in Figure 2.28(a) [165]. At $t_{8/5}$ 20 s, the microstructure was predominantly fine grained granular bainite, acicular ferrite, and polygonal ferrite/allotriomorphic ferrite, as shown in Figure 2.28(b). At the $t_{8/5}$ 60 s, the microstructure was composed of acicular ferrite generated inside of the prior austenite grain and fine and discontinuous cementite distributed along the
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degenerated pearlite thin film, as shown in Figure 2.28(c) [165]. On increasing the $t_{8/5}$ to 120 s, acicular ferrite and coarse pearlite were formed during the slow cooling process, as shown in Figure 2.28(d) [165].

Figure 2.28 Micrographs of simulated CGHAZ specimens for different $t_{8/5}$: (a) $t_{8/5}$ 10 s, (b) $t_{8/5}$ 20 s, (c) $t_{8/5}$ 60 s, and (d) $t_{8/5}$ 120 s [165].

Zhang et al. [166] investigated the sensitivity of gas metal arc weld metal (WM) to welding parameters in two Nb and Ti microalloyed S690QL steels. Welding parameters were varied to obtain three cooling rates: slow, medium and fast cooling. It was found that as cooling rate increased, the predominantly acicular ferrite in Nb weld metal (WM) was substituted by bainite, with a consequence of obvious hardness increase. The microstructures of Nb WM at different cooling rates are shown in Figure 2.29 [166]. However, in Ti WM, no great variation of acicular ferrite at all cooling rates contributed to little increment of hardness. WM for Ti steel showed to be markedly less sensitive to the variations of cooling rate than that for Nb steel.
Xie et al. [167] studied the microstructural evolution in the simulated coarse grain heat affected zone (CGHAZ) of novel low carbon microalloyed steel with yield strength of 1000 MPa. The effect of size of martensite–austenite (M–A) constituent in this novel high strength steel was explored by conducting studies on simulated CGHAZ specimens with four different cooling rates. The influence of welding thermal cycles on microstructural evolution and impact toughness were also studied. The selected $t_{8/5}$ times were 10, 30, 60 and 150 s, to simulate different welding heat inputs, which were 16.2, 28.0, 39.6 and 62.7 kJ/cm, respectively. It was found that the microstructure was significantly influenced by different heat inputs, which affected the different cooling rates. Figure 2.30 shows scanning electron microscope (SEM) images of the microstructures of the simulated CGHAZ specimens at different cooling rates [167]. The figure indicates that microstructural changes dramatically occurred with increasing $t_{8/5}$ cooling time. The microstructure of the simulated CGHAZ specimen changed from lath martensite to lath bainite and then to coarse granular bainite with increase in $t_{8/5}$ time. It was indicated that toughness first increased and then decreased with increasing $t_{8/5}$ time. This was due to the reduction of the brittle martensitic phase at $t_{8/5}$ times less than and up to 60 s, whereas the reduced toughness observed on exceeding 60 s was due to the formation of blocky M–A constituents composed of martensitic proportions. It
was believed that the formation of coarse M–A constituent because of low cooling rate ($t_{85}$ of 150 s) was the main reason for the dramatic decreasing in impact toughness of the simulated CGHAZ specimens [167].

![SEM micrographs](image)

Figure 2.30 SEM micrographs of the simulated CGHAZ with different $t_{85}$, (a) 10 s, (b) 30 s, (c) 60 s, and (d) 150 s [167].

Gianetto et al. [168] investigated the influence of composition and heat input on the structure and properties of single-pass submerged arc bead-in-groove welds produced on HY80 and HSLA80 steels. For the HY80 welds, a pronounced decrease in yield strength and microhardness were observed with increasing heat input. The low-temperature notch toughness was poor for both the 1 and 4 kJ/mm welds with an improvement observed at intermediate heat inputs. The poor notch toughness at 1 kJ/mm was attributed to the formation of hard lath martensite with high yield strength. Intermediate heat input welds consisted of a fine bainite microstructure with lower yield strength, which provided improved notch toughness. At the 4 kJ/mm heat input, an increased packet size for the fully bainitic microstructure re-suited in an increase in transition temperature. The HSLA80 welds showed a small decrease in yield strength with increasing heat input and superior notch toughness independent of heat input. This
occurred as a result of the transformation to a high proportion (80%) of acicular ferrite with limited continuous/discontinuous grain boundary ferrite.

X100 is a high strength steel for pipeline applications which enables the use of thinner walled pipe, making them lighter to transport and easier to handle on site, allowing greater operating pressures and reducing overall costs [169]. In order to better understand weldability of X100 steel and microstructures transformation induced by the thermal cycle, Miranda et al. [169] made a comparison study of laser and GTA autogeneous welding processes using similar heat inputs. In their experiment, an IPG YLR-8000 fiber laser with a maximum output power of 8 kW was used to weld 19 mm thick X100 steel plates. It was found that the microstructure of the fusion zone welded by fiber laser under a low heat input (160 J/mm) was mainly constituted by martensite and bainite, while the microstructure in the fusion zone welded by GTAW with heat input of 160 J/mm was coarse grained structure mainly constituted of a side-plate ferrite and pearlite, as seen in Figure 2.31(a) and (b), respectively [169].

![Figure 2.31 Microstructure of fusion zone obtained with laser and GTA welding with heat input of 160 J/mm, (a) laser welded, (b) GTA welded [169].](image)

The heat affected zone (HAZ) for both the laser and GTA welded specimens with heat input of 160 J/mm was also mainly constituted of martensite, but the microstructure in the HAZ of GTAW specimen showed a coarse grained structure, as seen in Figure 2.32(a) and (b), respectively [169].
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However, when the heat input was increased, the martensitic transformation was suppressed, the structure became coarser and the decomposition of austenite both in the fusion zone and in the HAZ led to primary ferrite surrounding austenite grains and pearlite for laser welded specimen with heat input of 960 J/mm. In addition, austenite was transformed into ferrite and pearlite for GTAW specimen with heat input of 960 J/mm. The microstructures in the fusion zone for both the laser and GTA welded specimens are shown in Figure 2.33(a) and (b), respectively [169].

Viano et al. [170] evaluated the influence of welding speed and heat inputs on joint quality, microstructure, and mechanical properties of four wire (double tandem) submerged arc welds in HSLA 80 steel. It was found that as the heat input increased, the cooling rate decreased resulting in a larger cellular dendritic cell spacing, decreased acicular ferrite content, and coarser acicular ferrite laths. The effect of welding speed on
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delta ferrite cell spacing and prior austenite grain size was found to be co-dependent on the heat input and the thermal profile resulting from multiple electrode welding.

Mohandas et al. [70] investigated the effect of the chemical composition of the high strength low alloy steels and the welding process on the softening of the heat affected zone. Three high-strength low-alloy steels were selected for investigating the influence of alloy elements on heat affected zone softening in three welding processes, namely, shielded metal arc welding (SMAW), gas tungsten arc welding (GTAW) and gas metal arc welding (GMAW). It was observed that a steel with a high carbon equivalent exhibited maximum softening. A steel with a low carbon-equivalent with high Ms (martensite start) and Bs (bainite start) temperatures coupled with minimum critical cooling time for nil martensite and full martensite exhibited the least softening in low heat input welding (shielded metal arc welding (SMAW)), whilst a steel with longer critical cooling time for full martensite exhibited more resistance to softening in high heat input welding (GMAW). In general, for the high heat input welding processes (GTAW and GMAW), a maximum degree of softening was observed. This soft region therefore exhibited low hardness and therefore low strength and hence was a weak area in any mechanical testing. However, external cooling methods, such as copper backing and argon purging, was found to be useful in reducing the tendency for softening [70].

2.6 Residual stresses measurements

Residual stress is that which remains in a body that is stationary and at equilibrium with its surroundings. Residual stresses can be introduced into the material by virtually any manufacturing processes that cause thermal or compositional gradients or that involve plastic deformation. Thus, they can be produced by forming, joining, machining, heat treatment, abrasion and many other relatively simple processes [171-173]. In many cases where unexpected failure has occurred, this has been due to the presence of residual stresses which have combined with the service stresses to seriously shorten component life. On the other hand, compressive stresses are sometimes introduced deliberately, as in shot peening, which is used to improve fatigue resistance [173].

Welding processes introduce localised rapid heating and cooling cycles, which result in regions in the vicinity of the weld undergoing severe thermal cycles, as well as localised
fusion in the weld. The severe thermal cycles cause non-uniform heating and cooling in the material, thus generating inhomogeneous expansion and contraction in the weldment [174]. When the fused region solidifies, the accompanying contraction exerts a pull on the surrounding material which may be prevented from complying by constraint, and then misfit strains will be introduced into the parts being welded. If these misfit strains are modest, they can be accommodated by elastic strains. However, when these misfit strains increase to a level which cannot be sustained elastically, localised plastic deformation is induced, which leads to the development of residual stresses [175, 176]. The associated misfit strains often lead to near yield tensile stresses at room temperature, especially in the welding direction [177]. All this means that in the absence of transformation effects, significant tensile stresses reside in the vicinity of the weld after the component has reached thermal equilibrium [175, 176, 178].

Although tensile residual stresses in welded joints are the usual residual stress status in the fusion zone and heat-affected zone, compressive stresses can be generated under some circumstances when solid state phase transformations take place during the welding process. The volumetric expansion resulting from the solid state phase transformation translates into compression stress [175].

Tensile residual stresses are of particular concern because they can contribute to fatigue crack development in a structure even under compressive cyclic loading [179]. They are also considered to be the main reason for the initiation and propagation of stress corrosion cracking (SCC) and fracture [174, 175, 180].

Due to their important contribution to failure and their almost universal presence, the knowledge of residual stress is crucial for prediction of the strength of any engineering structure. However, the prediction of residual stresses is a very complex problem. For this reason, it is important to have reliable methods for the measurement of these stresses and to understand the level of information they can provide [171, 172, 181]. The measurement methods can be classified into two categories. Destructive techniques are based on the principle that the strain measured in a material as a result of sectioning, or removing, part of the material is directly related to the residual stresses in the material before the cut is made. Examples include hole drilling, layer removal, compliance and contour methods. Non-destructive methods rely on the measurement of
a change in a certain physical property of the material, such as lattice spacings or magnetic properties, owing to the residual stresses. Diffraction methods, which use X-rays or neutrons, are currently among the most popular measurement techniques, though they do have some limitations [182, 183].

### 2.6.1 Residual stress measurements by the X-ray diffraction method

The internal residual stress measured using X-ray diffraction was determined by the sin²ψ method. There have been numerous papers on how this sin²ψ method can be used to calculate residual stress [183-186]. Here only a brief introduction is given.

X-ray diffraction method employs Bragg’s law to estimate the residual strains present in the atomic plans [187]. In this method, a monochromatic X-ray beam of sufficient intensity is made to be incident on the atomic planes. The reflected beam from successive planes of atoms is observed. Bragg’s law defines the condition for diffraction through the following equation:

\[ n\lambda = 2dsin\theta \]  

where \( \lambda \) is the wave length of incident X-rays, \( \theta \) is the angle between the incident or reflected beam and the reflecting planes, \( d \) is the interplanar spacing, and \( n \) is the order of reflection (\( n=1, 2, 3, \ldots \)). Eq. (2.2) shows that, if the wavelength of X-rays is known, \( d \) can be determined by measuring the angle \( \theta \). In the presence of residual stresses, \( d \) changes, leading to a shift in X-ray diffraction peaks, which this is a measure of the residual stress.

Figure 2.34 shows the configuration generally followed for residual stress measurements [188]. \( P_1, P_2 \) and \( P_3 \) refer to three orthogonal directions relative to the sample under investigation and \( L_1, L_2 \) and \( L_3 \) describe the laboratory or measurement frame of reference. The angles \( \psi \) and \( \phi \) define the relationship between \( P_1 \) and \( L_i \) axes; \( \psi \) describes the angle between the surface normal (\( P_j \)) and the direction of strain being measured (\( L_i \)); \( \phi \) denotes the angle between one of the principle stress axes (\( P_i \)) and the projection of the measured strain direction (\( L_j \)) onto the specimen surface. In the widely used sin²ψ method, diffraction measurements are made at several tilt angles \( \psi \). If \( d_{\phi\psi} \) is the interplanar spacing in the direction described by the angles \( \phi \) and \( \psi \) obtained from
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the position of diffraction peak for a given ‘h k l’ plane, the strain along \( L_3 \) may be obtained as

\[
(\varepsilon_{33})_{\phi \psi} = \frac{d_{\phi \psi} - d_{\phi \psi = 0}}{d_{\phi = 0}}
\]  

(2.3)

where \( d_{\phi \psi} = 0 \) refers to the stress-free interplanar spacing value. The strain may be expressed in terms of strains \( \varepsilon_{ij} \) in the sample coordinate system, by tensor transformation. For the general case of two tilt measurements at angles \( 0^\circ \) and \( \psi \), the surface residual stress \( \sigma_\phi \) can be expressed as

\[
\sigma_\phi = \frac{E}{1 + \nu} \frac{1}{\sin^2 \psi} \frac{d_{\phi \psi} - d_{\phi \psi = 0}}{d_{\phi = 0}}
\]  

(2.4)

where \( E \) and \( \nu \) are Young’s Modulus and Poisson’s ratio, respectively. The term \( E/(1 + \nu) \) is a constant. Using the linear relationship given in Eq. (15) for surface residual stress, the lattice strain \( \Delta d/d \) plotted against \( \sin^2 \psi \) will produce a straight line whose gradient is a function of \( \sigma_\phi \), \( \nu \) and \( E \). From the gradient of a least squares straight line fit through the data points for various tilt angles, \( \sigma_\phi \) can be calculated [176, 188].

Figure 2.34 Axial system for residual stress measurements [188].

2.6.2 Residual stress measurements by the neutron diffraction method

Like other diffraction techniques neutron diffraction relies on elastic deformations within a polycrystalline material that cause changes in the spacing of the lattice planes
from their stress free value. Measurements are carried out in much the same way as with X-ray diffraction, with a detector moving around the sample, locating the positions of high intensity diffracted beams. The greatest advantage that neutrons have over X-rays is the very large penetration depths that neutrons can obtain, which makes them capable of measuring at near surface depths of around 0.2 mm down to bulk measurements of up to 100 mm in aluminium or 30 mm in steel. With high spatial resolution, neutron diffraction can provide complete three-dimensional strain maps of engineered components. This is achieved through translational and rotational movements of the component [189, 190].

There are essentially two neutron diffraction techniques, namely, conventional $\theta/2\theta$ scanning and time of flight approaches. These two methods have developed largely because of the two forms in which neutron beams are available, i.e. either as a continuous beam from a reactor source, or as a pulsed beam from a spallation source. The former is well suited to conventional $\theta/2\theta$ scanning, whereby shifts $\Delta \theta$ in a single ‘$h k l$’ diffraction peak are monitored according to $\varepsilon = \frac{\Delta d}{d} = -\cot \theta \Delta \theta$ ($\theta$ is Bragg scattering angle, $\Delta \theta$ is change in the Bragg scattering angle), while the latter is well suited to the time of flight method. In this case, the diffraction profile is not collected as a function of the Bragg angle $\theta$, rather the Bragg angle is held constant (usually $2\theta = 90^\circ$) and the incident wavelength $\lambda$ varied. This is because within each pulse of neutrons leaving the moderated target there is a large range of neutron energies. Naturally, the most energetic neutrons arrive at the specimen first, the least energetic last. Consequently, the energy and hence wavelength of each detected neutron can be deduced from the time that has elapsed since the pulse of neutrons was produced at the target, i.e. from the time of flight. In this case, the strain is given by $\varepsilon = \Delta t/t$, where $t$ is the time of flight. As the strain resolution is dependent upon the accuracy of the measurement of the time of flight, high resolution instruments tend to have large flight paths (> 100 m) [171].

2.6.3 Residual stress measurements by the deep hole drilling method

The deep-hole drilling (DHD) technique is a semi-destructive residual stress measurement method [191]. It is a mechanical strain relaxation technique used to determine through-thickness residual stresses by measuring strains during stress relief.
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from the removal of a small amount of material [192, 193]. The measurement procedure for residual stress measurement consists of four steps, as shown in Figure 2.35. First, two reference bushes are attached on the front and back surfaces of the component to avoid drill bell-mouthing and provide stress-free references. Then, a small reference hole is drilled through the sample where stresses are to be determined (Figure 2.35A). This is followed by accurate measurement of the hole diameter (Figure 2.35B). The diameter measurement is taken at different angular positions around the reference hole and equal intervals along the hole axis. In step 3 (Figure 2.35C), a core of material coaxial to the reference hole is trepanned. Trepanning the column containing the reference hole releases residual stresses. Finally, hole diameter is re-measured at the same angular positions and depths (Figure 2.35D) [194, 195].

![Figure 2.35 Schematic of the deep hole drilling technique [195].](image)

The change in diameter of the reference hole before and after trepanning is used to determine the radial distortion of the hole. The trepanned core can be considered as a series of independent plates containing a small hole subjected to uniform uniaxial stresses. When an elastic plate containing a (reference) hole is subjected to a remote and uniform stress fields \(\sigma_{xx}, \sigma_{yy}\), and \(\sigma_{xy}\), there is a change in diameter \(\Delta d(\theta, z)\). This is a function of angle \(\theta\) around the hole and through the thickness \(z\) of the plate.
The normalised radial distortion, $\overline{U}_{rr}$, around the hole can be related to the in-plane residual stresses $\sigma_{xx}$, $\sigma_{yy}$ and $\sigma_{xy}$ by [194].

\[
\overline{U}_{rr}(\theta) = \frac{\Delta d(\theta)}{d(\theta)} = -\frac{1}{E} \left[ \sigma_{xx}(1 + 2\cos 2\theta) + \sigma_{yy}(1 - 2\cos 2\theta) + \sigma_{xy}(4\sin 2\theta) \right]
\]  

(2.5)

The method determines the local stresses in the absence of the reference hole. The normalised radial distortion is obtained from the difference in the reference hole diameter before and after trepanning.

\[
\Delta d(\theta) = d'(\theta) - d(\theta)
\]  

(2.6)

Here, $d(\theta)$ and $d'(\theta)$ correspond to the diameter of the reference hole before and after trepanning. This is repeated at $m$ different angular measurements around the reference hole and for every depth, $z$, along the hole axis. The distortion is linear with respect to the unknown in-plane stresses (perpendicular to the axis of the reference hole) according to Equation (1), so that

\[
\overline{U}_{rr}(\theta) = \frac{\Delta d(\theta)}{d(\theta)} = -\frac{f(\theta)\sigma_{xx} + g(\theta)\sigma_{yy} + h(\theta)\sigma_{xy}}{E}
\]  

(2.7)

The analysis assumes that the out-of-plane stress, $\sigma_{zz}$, is zero. At a specified location along the axis of the reference hole, Equation (3) can be rewritten as

\[
\overline{U}_{rr}(\theta) = -\frac{1}{E} M_{2D} \sigma
\]  

(2.8)

where $M_{2D}$ corresponds to the matrix of angular functions and $E$ is Young’s modulus. Measurements are obtained at $m$ angles, and a least-squares fit to the diametral strains can be used to determine the stresses, and a pseudo-inverse matrix is used [196, 197].

\[
\hat{\sigma} = -EM_{2D}^* \overline{U}_{rr}
\]  

(2.9)

where

\[
M_{2D}^* = (M_{2D}^T M_{2D})^{-1} M_{2D}^T
\]  

(2.10)
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\( M_{2D}^* \) is the pseudo-inverse of matrix \( M_{2D} \), \( M_{2D}^T \) is the transpose of matrix \( M_{2D} \) and \( \hat{\sigma} \) is the optimum stress vector that best fits the measured diametral distortions [194, 195].

2.6.4 Residual stress measurements by the contour method

The contour method is a newly developed relaxation technique for residual stress evaluation by Prime in 2000 [198]. The concept of the contour method is derived from Bueckner’s elastic superposition principle [199], which allows measurement of residual stresses normal to a plane of interest [200, 201]. Figure 2.36 illustrates Bueckner’s superposition principle [202] as applied to sectioning parts for residual stress measurement. Step A shows the undisturbed part with the residual stresses that are to be determined. The part is cut in two on the plane \( x = 0 \) and deforms as residual stresses are released by the cut. B shows half of the part in the post-cut state with partially relaxed stresses. The surface contour is measured at this point. C is an analytical step that starts with a stress-free body and then the surface created by the cut is displaced back to its original flat shape. Assuming elasticity, superimposing the partially relaxed stress state in B with the change in stress from C gives the original residual stress throughout the part:

\[
\sigma^A(x, y, z) = \sigma^B(x, y, z) + \sigma^C(x, y, z)
\]

where \( \sigma \) without subscripts refers to the entire stress tensor. \( \sigma^A \) can be referred to as the original stress, \( \sigma^B \) as the remaining stress, and \( \sigma^C \) as the change stress since it quantifies how much the stress has relaxed from sectioning.
Figure 2.36 Bueckner’s superposition principle applied to residual stresses in a sectioned part: \( A = B + C \) [203].

The contour method involves four steps: specimen cutting, measurement of profile, data smoothing and finite element analysis [204].

Specimen cutting is the first and the most critical step in implementing the contour method, as the subsequent procedures of contour measurement, data reduction and stress analysis are all reliant on the quality of the cutting [205]. As with other relaxation methods [206], it is assumed that cutting process relaxes the residual stresses elastically and does not induce any stresses into the material. Proper constraint of a specimen to avoid its movement during cutting is essential. Additionally, the ideal cutting process would generate a perfectly straight cut, without any further removal of material from the cut surfaces. Wire electrodischarge machining (EDM) is thought to come closest to these conditions. “Skim cut” settings, which are normally used for better precision and a finer surface finish, were used because they also minimize any recast layer and cutting-induced stresses [207]. An example of wire EDM cutting was performed through a 12 mm thickness along the transverse direction of the weld, across the middle width (240 mm) of the welded plate, as indicated in Figure 2.37 [205].
Figure 2.37 Schematic illustration of welded plate and cut arrangement for the contour method [205].

After the wire EDM cut, a contoured surface is formed owing to the release of residual stresses, which needs to be measured on both cut surfaces. Previous experiments have shown that the use of a coordinate measurement machine (CMM) or a laser scanner provides adequately accurate measurement for a variety of specimens [205, 207, 208].

The measured surface displacements contain high frequency noise, which is caused by measurement error and surface roughness from EDM cutting. Data smoothing is performed to extract the form of the surfaces (filtering out the noise) and to average the displacements from the opposite cut faces. Data smoothing was performed based on the valid data using a running average, in which the middle value is replaced by the average of three adjacent data. Extrapolation of the removed data was conducted using a spline function. Smoothing of the measurements to minimize the errors in the data is, therefore, crucial to achieve accurate stress evaluation with the contour method [205]. Figure 2.38 shows an example of scanned surface before and after data reduction and smoothing of a cross-sectional deformation on a cut plane [205].
Figure 2.38 Scanned surface before and after data reduction and smoothing. (a) detail of the scanned surface, (b) the averaged, smoothed and extrapolated data [205].

The residual stresses are finally inferred from the measured and averaged displacement by use of the FE method according to Bueckner’s superposition principle [199]. The averaged and smoothed surface data are input, as an initial displacement boundary condition, into a one-half finite element (FE) model simulating the geometry of the specimen, with opposite sign to that measured [173]. The whole analysis is assumed to behave elastically [209]. Figure 2.39 shows an example of a longitudinal residual stress through the thickness of a variable-polarity plasma-arc welded 2024-T351 aluminium alloy plate obtained using the contour method [205].

Figure 2.39 A cross-sectional longitudinal residual stress map (x = 278 mm, y = 12 mm): the thickness of the plate is magnified by 2 for clarity. All stress values in MPa. [205]
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2.7 Summary

This literature review mainly introduces the development of high strength steels, the application of high strength steels, the conventional welding techniques to join the high strength steels, the laser welding of high strength steels, effect of heat input on the microstructure and mechanical properties of the welded joints, and residual stress measurement methods. A summary is carried out based on the literature review.

Outstanding strength-toughness combination and weldability have encouraged the use of high strength steels in various applications including pressure vessel, linepipes, shipbuilding, offshore construction, automotive industry, lifting and handling equipment [30-35, 47]. The application of high strength steels in the structural components not only decreases component size and structure weight, but also reduces consumption of materials without compromising the structural integrity [3, 51, 52, 210].

Conventional arc welding processes, such as gas tungsten arc welding (GTAW), gas metal arc welding (GMAW) and submerged arc welding (SAW), as well as resistance spot welding (RSW) are widely used to weld the high strength steels [47, 52, 69, 70, 148]. However, these conventional welding processes have relatively high heat inputs which result in slow cooling rates of the weldment, and this can produce grain growth and subsequent softening in the heat affected zone (HAZ), which leads to deteriorate the mechanical properties of welded joint of high strength steels [211-214].

Laser welds have characteristics of high aspect ratio with narrow heat affected zone and the distortion is low in contrast to the conventional arc welding methods. Laser welding has been widely applied in industry because of the excellent advantages mentioned above plus optical fiber beam delivery and the use of industrial robots. Single pass autogenous laser welding (weld without filler material) is one of the typical features of laser welding [124]. However, there are still some disadvantages for single pass autogenous laser welding of medium and thick section materials, such as the tight joint fit-up requirement and the limited maximum laser power available for commercial lasers (typically less than 20 kW) [102, 122, 123]. In addition, single pass autogenous laser welding medium and thick section materials often results in porosity, molten material dropout, cracking, and mis-tracking of weld seams [127-129]. The single pass
autogenous laser welding of medium and thick section materials requires high capital investment on the high power lasers.

Hybrid laser–arc welding combines laser welding and arc welding into one process, which can compensate the disadvantage of individual process [10, 133, 215]. The combination of the laser and the arc has a synergistic effect on increasing the penetration depth, filling a large gap (bridge-ability) and possibility to tailor the weld properties by modifying the chemical composition of the fusion zone [10, 134-136]. However, an extra heat source from the arc is introduced into the hybrid laser–arc welding process makes the heat input of hybrid laser–arc welding greater than that of autogenous laser welding process [102]. Another disadvantage of hybrid laser–arc welding process is the large number of parameters that have to be carefully adjusted and controlled to get a stable condition [103, 215].

Multi-pass narrow gap laser welding technique with filler material can be used to weld medium and thick section materials using relatively moderate laser powers [142, 145]. The application of the multi-pass narrow gap laser welding technique can not only obviate the need for expensive high-power lasers but also reduce the consumption of filler material. Multi-pass narrow gap laser welding technique can eliminate the limits for single pass autogenous laser welding, such as improving the gap bridging ability, relaxing the tight joint fit-up requirement and tailoring the weld properties by modifying the fusion zone composition [111, 136]. In addition, this welding technique provides an approach to solve molten material dropout and cracking in the weld these occur in single pass autogenous laser welding medium and thick section materials.

According to the metallurgical characteristics of the HSLA steel, heat input of welding process significantly affects the mechanical properties of heat affected zone (HAZ) [36]. The characterisation of HAZ softening is generally happened in arc welding of HSLA steels [70, 216, 217]. The softening behaviour in the HAZ is attributed to tempering in this region [218]. The degree of softening in the HAZ is a function of the weld thermal cycle, which is a characteristic of the welding process [70, 217, 219, 220]. The softening characteristics depend also on the kinetics of the phase transformations of the steel and are a function of the chemical composition of the steel [70, 219]. The properties of the welded joints are determined by the characteristics of the HAZ [109,
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The softening in the HAZ will lead to degradation in mechanic properties, such as loss of strength of the welded joints and increasing the probability of fatigue failures at the weakest zones [70, 216]. Since the mechanical properties of steel are highly dependent on the microstructure, and distribution of each phase present [221]. In this sense, the heat input generally has a strong influence on the welding metallurgy [148].

The detrimental effect of HAZ softening can be reduced by increasing the welding speed and reducing the input power (i.e. lowering the heat input), which decreases the width of the softened zone, hence improving the mechanical properties [222-225]. The cooling rate is one of the most important factors influencing microstructure evolution in either fusion zone or HAZ [150]. The higher values of the heat input, the slower the cooling rate would be, and vice versa [151].

It has therefore advantageous to develop advanced welding procedures that minimize heat input and reduce the width of the soft HAZ. Compared the heat input per unit length between typical arc and laser welds, it shows that the heat input in laser welds is typically almost an order of magnitude lower than in arc welds [123]. Autogenous laser welding has lower line energy than laser arc hybrid welding or arc welding processes and higher cooling rates (2000–3000 °C/s) [102].

X-ray and neutron beams are two main types of radiation for residual measurement. Neutron diffraction and X-ray diffraction are two popular non-destructive techniques that have been used to residual stress measurement in engineering components [226, 227]. These two beams provide very different levels of penetration into most polycrystalline metals and ceramics, and so their applications vary [228]. The penetration depth of X-rays for most of metallic materials within tens of microns, however neutrons have the advantage over X-rays that for wavelengths comparable to the atomic spacing, their penetration into engineering materials is typically many centimetres [183, 229]. X-ray diffraction is mainly used to measure the surface residual stress [227]. Neutron diffraction methods have revolutionized the capability to non-destructively measure the residual stresses deep within welded components and structures [230]. However, some components are too thick for neutron measurement and welds are often problematic because of spatial variations in the reference lattice constant caused by chemistry changes or microstructural gradients in and around the weld [231-
or the presence of microstresses [233]. Another challenge for diffraction-based techniques is that it is necessary to obtain the stress-free lattice parameter $d_0$. For welded specimens the stress-free lattice parameter is especially challenging to obtain due to the compositional and microstructural variations encountered [228, 234]. The deep hole drilling method is a semi-destructive residual stress measurement method [191], which has had the most success on very thick components but measures only a 1D stress profile [194].

The contour method is a destructive method used to measure residual stresses, which allows the 2D residual stress maps to be obtained directly from displacement maps with minimum time and computational cost. An important inherent advantage of using the contour method instead of neutron diffraction, especially in the context of welded samples, is that it is not necessary to obtain the stress-free lattice parameter $d_0$ [234]. Furthermore, this method is not sensitive to microstructure and does not require complex equipment [173, 235]. However, there are two main limitations for the method: the full 2D residual stress map can be obtained only for one stress direction [204]. Also, like other mechanical strain relief techniques [193, 236], it is prone to plasticity induced-error when measuring stresses that are close to the yield strength of the material [237, 238], and the residual stress in the near-surface region is very challenging to obtain reliably [237].

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Chapter 3: Microstructure and mechanical properties of autogenous laser welded S960 high strength steel

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Abstract

S960 steel is an advanced low carbon and low alloy ultra-high strength steel (with a minimum yield strength of 960 MPa) developed by Tata Steel. At present, there is a scarcity of data for laser welding of such a material. In this study, 8 mm thick hot rolled and quenched S960 high strength low alloy (HSLA) steel plates were welded using a 16 kW fiber laser system. The microstructure, microhardness, and tensile properties were characterised, Charpy impact testing and three-point bending testing were carried out, and fracture surfaces were investigated. Preliminary results suggest that the laser welding process can produce single-pass welds which are free of macroscopic defects. The microstructures in the fusion zone and heat affected zone were predominately martensite and some self-tempered martensite, with grain size variation in different sub-zones. The tensile properties of the laser welded joint matched those obtained for the base material, with failure occurring in the base material away from the weld. While the welded joint performed well when subjected to bending, the impact toughness was reduced when compared with that of the base material.

Keywords: self-tempering, fiber laser, martensite, S960 high strength steel, toughness
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3.1 Introduction

Due to their excellent combination of strength and toughness, high strength/weight ratio and weldability, high strength low alloy (HSLA) steels have found widespread applications as structural components, in pressure vessels, oil/gas transportation pipes, and in the shipbuilding and offshore industries [1, 2]. HSLA steels can be used in lightweight, high-performance structural applications to increase their loading capacity [3]. The approach to strengthen HSLA steels is through small additions of alloying elements that promote precipitation hardening, grain refinement through thermo-mechanical processing, or transformation hardening through the formation of bainite and/or martensite.

Welding is an essential manufacturing technique for joining high strength steels, as it is required to mitigate the complexity of manufacturing large structures [4]. The low carbon content and low alloy content of HSLA steels results in a low carbon equivalent (CE) value, which can reduce the preheating temperature or negate the requirement for weld preheating [5, 6]. When compared to laser welding, conventional arc welding processes lead to relatively high heat inputs, resulting in lower cooling rates across the weldments, and this can cause softening in the heat affected zone (HAZ) [7, 8], and reduced strength for the weldment as a whole [9, 10]. This is a particular issue for higher strength HSLA steels, but one which could be addressed by specific processing involving rapid water cooling. Alternatively, laser welding is a contact-free welding method which offers flexibility and high welding speeds, and the quality of the welds is normally excellent because of the high power density and low heat inputs [11, 12]. Low heat input welding methods can produce the fine grain structures that provide the strength and toughness of the joint, and they also have the benefit of decreasing welding induced distortion [6]. Laser welds generally have a high aspect ratio (weld depth to width ratio) with a narrow heat affected zone, and the distortion is low in contrast to conventional arc welding methods. Laser welding has been widely applied in industry because of the advantages mentioned above, plus the capability for optical fiber beam delivery and the use of industrial robots. Among commercial laser systems, fiber lasers have the characteristics of excellent beam quality and high brightness, and consequently they have attracted more attention over the last decade for cutting, welding and cladding applications [13, 14].
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Tata Steel are currently developing an advanced low carbon, low alloy (S960) steel with a minimum yield strength of 960 MPa. Because of its high specific strength, good impact toughness and formability, it is a promising material for applications in the heavy crane sector. Ruukki Metals [15] are also developing Optim 960 QC steel with a strength of 960 MPa. The Optim 960 QC strip steels are made by modern hot strip rolling and direct-quenching processes, which are similar to the method employed by Tata Steel.

Welding is an indispensable processing technique for using such materials in engineering structures. Arc welding processes are typically employed to join HSLA steels. Němeček et al. [16] carried out a comparative study on laser and gas metal arc welding (GMAW) of two high strength steels, namely TRIP 900 steel and martensitic DOCOL 1200 steel. It was found that the strength of GMA welds was significantly reduced in comparison to the parent materials, while laser welding resulted in the strength of the base material being nearly unaffected. In addition, the hardness in the laser weld and heat affected zone (HAZ) were approximately the same as that of the base material, whereas the GMA weld exhibited signs of softening in the weld and HAZ. Lee et al. [17] investigated laser welding, gas tungsten arc welding (GTAW) and gas metal arc welding (GMAW) of dual-phase steel (DP780). It was found that the hardness of the fusion zone (FZ) and HAZ increased with increasing cooling rate (laser > TIG > MAG). Laser and GTA welded specimens exhibited equal strength to that of the base metal, while the strength of the GMAW specimens was reduced by a broad and soft FZ and HAZ. Due to the potential for HAZ softening of S960 steels using conventional arc welding processes, laser welding is potentially an improved fabrication process for this steel.

In the present work, an autogenous single pass laser welding process (without filler wire) was employed in order to examine the weldability of this steel. The microstructure, microhardness, tensile properties, Charpy impact toughness and bending performance were examined in order to evaluate the joint performance and the characteristics of autogenous laser welded S960 HSLA steel joints.
3.2 Material and experimental procedures

The as-received base material (BM) used in this study, which was supplied by Tata Steel, was in the form of 8 mm thick S960 HSLA steel hot rolled strips, which had been rapidly water cooled to room temperature. The chemical composition and a scanning electron micrograph of the BM are shown in Table 3.1 and Figure 3.1, respectively. The carbon equivalent (CE) of the investigated steel was calculated according to the following equation [18]:

\[
CE = C + \frac{Mn}{6} + \frac{(Cr + Mo + V)}{5} + \frac{(Ni + Cu)}{15}
\]  

(3.1)

which is recommended by the International Institute of Welding (IIW). All concentrations are defined in weight percent. The CE of S960 steel is 0.496, as seen in Table 3.1. The carbon content of the S960 HSLA steel is low at 0.088%. The Ms (martensite start) temperature is high (~ 450 °C), and so the martensite will self-temper immediately after transformation. Bainite and tempered martensite are distinguished based on the orientation of the carbides. Lower bainite has small carbides that are parallel to each other, upper bainite has larger carbides which are between the laths of bainitic ferrite, and tempered martensite has carbides which are orientated along certain crystallographic planes (so not parallel). The microstructure of the BM consists of a mixture of tempered martensite and bainite, as shown in Figure 3.1.

| Table 3.1 Chemical composition of S960 steel (wt. %). |
|--------------|-----|-----|-----|-----|-----|-----|-----|-----|-----|
| C            | Mn  | P   | S   | N   | B   | Ca  | Si  | Al  | Cu  |
| 0.088        | 1.51| 0.008| 0.002| 0.0089| 0.0022| 0.012| 0.055| 0.033| 0.014|
| Sn           | Cr  | Ni  | Mo  | Ti  | Nb  | V   | Fe  | CE  |     |
| 0.001        | 0.472| 0.023| 0.248| 0.025| 0.040| 0.050| Bal.| 0.496|     |
The as received materials were sliced by wire electric discharge machining (EDM) in order to achieve good joint fit-up for the autogenous laser welding experiments. The plates used for autogenous laser welding had dimensions of approximately 200 mm × 100 mm × 8 mm. After welding the samples were measured as 200 mm × 200 mm × 8 mm.

A continuous wave fiber laser (IPG YLS-16000) with a maximum power of 16 kW was used in the laser welding experiments. The beam parameter product was 10 mm mrad. The laser beam was delivered to the robotic welding cell through an optical fiber with a 300 µm core diameter. The laser beam emitted from the end of the optical fiber was collimated by a lens with a 150 mm focal length and then focused onto the specimen surface using a lens with a 400 mm focal length. The measured focus spot size and Rayleigh length were 0.8 mm and 15 mm, respectively. The laser welding head was mounted on a 6-axis KUKA robot. A schematic representation of the laser welding setup is shown in Figure 3.2.
Before welding, the parent material in the vicinity of the weld track was cleaned using sand blasting to remove the surface oxide. Acetone was used to clean the surface after sand blasting, and then the materials were clamped on the work table to ensure adequate restraint. Autogenous laser welding was carried out perpendicular to the rolling direction of the base material. The top and back surfaces of the specimens were shielded using argon gas to protect the molten weld pool during the welding process. Argon gas was blown onto the top surface with a flow rate of 12 l/min, and the backing gas was blown through a side tube to an internal chamber under the specimens with a flow rate of 8 l/min. The laser head was tilted by 8° to protect the laser head from laser reflection.

After the welding experiments, the specimens were cut on a plane that was transverse to the welding direction and macrograph sections were mounted in hot-setting epoxy resin. The specimens for metallographic examination were subsequently ground and polished using an automatic polishing machine, followed by etching in a solution of 2% Nital for about 2 s. The macrostructure of the joint and the microstructure of the weld were examined using an optical metallurgical microscope (KEYENCE VHX-500F) and a Philips XL 30 scanning electron microscope (SEM).

Micro-indentation hardness mapping profiles across the welded joint were measured using a load of 300 g and a dwell period of 15 s with a Vickers micro-hardness machine (Wilson Hardness Tukon 2500). In the fusion zone and the HAZ, the spacing between hardness indentations was 0.2 mm and for areas further away from the weld it was 0.4 mm.
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Tensile test specimens for the as-received base material and the welded samples were produced in accordance with ASTM E8M-04. Three-point bending test samples were cut according to BS EN ISO 15614-1. Both face and root bending tests were carried out with three test pieces for each type of test. Bend testing was executed until an angle of 180° was reached according to BS EN ISO 5173: 2010+A1:2011. Sub-size Charpy impact test samples (8 mm thick) were prepared following BS EN 10045-1:1990. The notches were located in the fusion zone (FZ), on the fusion line (FL) and in the HAZ (FL + 0.5 mm), respectively to examine the impact toughness of the fusion zone, fusion line and HAZ for laser welded samples. Three replicates for tensile, bending and Charpy impact testing were prepared to reduce experimental uncertainties. Samples prepared for tensile, three-point bending and Charpy impact tests were extracted from the steady state region of the welds using wire EDM, with the long axis of each specimen type being normal to the welding direction. Tensile test coupons and three-point bending specimens were sliced to coincide with the rolling direction of the material, and the configurations for each are shown in Figure 3.3. The weld reinforcements in the face and root regions were removed by manual grinding before the tensile, three-point bending and Charpy impact tests were conducted. Tensile tests were carried out on an Instron model 8500 electronic universal test machine at room temperature. Three-point bending tests were carried out on a Dartec bending test machine at room temperature. The diameter of the former and the gap distance for the three-point bending test are shown in Figure 3.4. The Charpy impact tests were carried out on a Zwick Roell Charpy impact test machine at -40 °C, -20 °C, 0 °C and at room temperature, respectively. Each test sample was kept at the related test temperature for half an hour to ensure that the temperature in each specimen was uniform. Following the tensile and Charpy impact tests, all of the fracture surfaces of the tested specimens were examined using a Zeiss EVO 50 SEM installed with an Energy Dispersive X-ray Detector (EDX) to reveal the fracture morphology and to gain understanding of the fracture mechanisms.
Based on previous experience with single pass autogenous laser welding of high strength steels, autogenous laser butt welding of 8 mm thick S960 steel was carried out at a welding speed of 1.08 m/min and with the laser power varying between 6 and 6.7 kW. The laser focal position was set 4 mm below the top surface of the plates.

After welding, top and back weld appearances of the specimens at different welding speeds are shown in Figure 3.5. The specimens were cut and prepared for metallography in order to examine the fusion zone and HAZ. Weld cross sections corresponding to different combinations of welding parameters are presented in Figure 3.6.
Figure 3.5 Top and back weld appearances of the specimens at a welding speed of 1.08 m/s for 8 mm thick S960 plates using different laser powers, (a) 6 kW, (b) 6.2 kW, (c) 6.7 kW and (d) 6.5 kW.

Figure 3.6 Weld bead cross sections at a welding speed of 1.08 m/s for 8 mm thick S960 plates using different laser powers, (a) 6 kW, (b) 6.2 kW, (c) 6.7 kW and (d) 6.5 kW.
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In Figure 3.5 and Figure 3.6 it can be seen that there is lack of penetration for the welds made with laser powers of 6 and 6.2 kW, and that there is undercut in the weld made with a laser power of 6.7 kW. However, a sound weld without any defects was achieved with a laser power of 6.5 kW. On this basis, the optimised welding parameters for the 8 mm thick plates are summarised in Table 3.2.

Table 3.2 Optimised autogenous laser welding parameters for 8 mm thick S960 steel.

<table>
<thead>
<tr>
<th>Power (kW)</th>
<th>Welding speed (m/min)</th>
<th>Top shielding gas flow (l/min)</th>
<th>Back shielding gas flow (l/min)</th>
<th>Focal position (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>6.5</td>
<td>1.08</td>
<td>12</td>
<td>8</td>
<td>-4</td>
</tr>
</tbody>
</table>

3.3 Results

3.3.1 Macrostructure characteristics

The macrostructure of the autogenous laser welded S960 HSLA steel joint is shown in Figure 3.6(d). It can be seen that the fusion zone exhibits an hour-glass shape, which is a characteristic of keyhole welding. There is no evidence of defects such as porosity or undercut. The macrostructure of the joint can be split into several distinct regions, such as the fusion zone (FZ) in the centre, the heat affected zone (HAZ) and the base material (BM). The fusion zone consists of coarse columnar dendritic grains, which are oriented perpendicular to the fusion boundary. Since the rate of heat extraction is maximum in the direction perpendicular to the fusion boundary, the grains tend to grow in that direction, resulting in a columnar grain structure in the fusion zone [19, 20]. Different sub-zones within the HAZ experience different peak temperatures during the welding thermal cycle, owing to differences in distance from the weld centerline. The grain size within the HAZ varied accordingly. The HAZ can be divided into four different sub-zones: the coarse-grained HAZ (CGHAZ), the fine-grained HAZ (FGHAZ), the intercritical HAZ (ICHAZ) and the sub-critical HAZ (SCHAZ) adjacent to the BM. Among these, the CGHAZ experienced the highest peak temperature, which caused the increase in the grain size in this zone.
3.3.2 Microstructural characteristics

The peak temperature and the cooling rate are the most important factors in determining the microstructural evolution within each sub-zone of the welded joint [21, 22]. The characteristics of the microstructures within each of the different sub-zones (SCHAZ, ICHAZ, FGHAZ, CGHAZ and FZ) of the laser welded joint were examined using an SEM and micrographs are presented in Figure 3.7. The microstructure in the SCHAZ is similar to the base material, comprising lower bainite and auto-tempered martensite. The ICHAZ consists of an overtempered base material microstructure and regions with martensite and auto-tempered martensite. The microstructure within the FGHAZ is a mixture of equiaxed martensite and auto-tempered martensite, with a small prior-austenite grain size. The microstructure of the CGHAZ is a mixture of equiaxed martensite and auto-tempered martensite with a large prior-austenite grain size. The microstructure of the FZ is a mixture of elongated martensite and auto-tempered martensite, within a large columnar grain structure.
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Figure 3.7 SEM micrographs showing the microstructures in the different sub-zones of the welded joint (a) SCHAZ, (b) ICHAZ, (c) FGHAZ, (d) CGHAZ, (e) FZ, (prior austenite grain boundaries are signified with arrows).

3.3.3 Microhardness

The hardness across the welded joint varies significantly because of the phase transformations that have taken place in the FZ and HAZ during the welding thermal cycle. Figure 3.8(a) shows a microhardness map of the entire welded joint. It clearly shows the variation in hardness across the fusion zone, HAZ and base material through differences in colour. Figure 3.8(b) shows a hardness profile extracted across the mid-section of the sample. The array of hardness mapping indents is presented in Figure 3.8(c), and each sub-zone is marked. From these results, it can be seen that the average hardness in the FZ is approximately 375 HV$_{0.3}$. The hardness increases from the CGHAZ to the FGHAZ with a peak hardness value of 385 HV$_{0.3}$ in the FGHAZ, and then the hardness drops rapidly toward the SCHAZ. The hardness of the BM is approximately 310 HV$_{0.3}$. Narrow hardness “valleys” are seen in the SCHAZ close to the BM, in which the hardness locally drops to lower than the BM hardness. These hardness results are similar to Siltanen et al.’s results [15] for laser welded Optim 960 QC steel. The hardness of the lower region of the BM is noticeably higher than for other regions, and this may have been caused by different cooling rates in the rolling process when this plate was manufactured.
3.3.4 Tensile and bending properties

Typical engineering stress-strain plots for the BM and laser welded S960 steel joint are shown in Figure 3.9. It can be seen that, while the apparent elongation was slightly lower for the laser weld, the yield strength (YS) and ultimate tensile strength (UTS) remained very close to that of S960 steel. The tensile test results are summarized in Table 3.3, which lists average values and standard deviations. It must be remembered that the welded specimens were not homogenous. This means the recorded values for the yield stress and, in particular, the elongation are not truly representative of any microstructural region. In addition, they will also vary with the choice of gauge length (50 mm in this case). The YS, UTS and apparent elongation were found to be 1018 MPa, 1030 MPa and 8%, respectively, for the laser welded joint. All tensile failures for the welded specimens were in the BM well away from the weld, as is shown in Figure 3.10. The YS, UTS and apparent elongation of the laser welded samples were observed to be the same as those for the BM, which had a YS of 1026 MPa, a UTS of 1035 MPa.
and an apparent elongation of 8.5%. The hardness profiles in Figure 3.8 showed that, in the as-welded condition, the majority of the welded region had been strengthened by the welding process, so it is likely that the weld region did not yield during the tensile test. This is likely to explain the slightly lower elongation that was recorded for the laser welded joint. The reductions of area for the BM and laser welded specimens were 59.6% and 55.6%, respectively.

![Figure 3.9 Representative plots of engineering stress versus engineering strain for S960 steel and the laser welded S960 steel joint.](image)

<table>
<thead>
<tr>
<th>Test specimens</th>
<th>Yield strength (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
<th>Reduction of area (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Base material (8 mm)</td>
<td>1026 ± 7.4</td>
<td>1035 ± 11.0</td>
<td>8.5 ± 0.3</td>
<td>59.6 ± 0.4</td>
</tr>
<tr>
<td>Laser weld (8 mm)</td>
<td>1018 ± 10.6</td>
<td>1030 ± 13.5</td>
<td>8.0 ± 1.0</td>
<td>55.6 ± 0.7</td>
</tr>
</tbody>
</table>

![Figure 3.10 Fracture locations for the tensile test specimens, (a) base material, (b) welded samples.](image)
Typical fracture surfaces from the tensile tests carried out on both the base material and welded specimens were examined using an optical microscope and an SEM, and the results are shown in Figure 3.11. The welded tensile test specimens broke in the base material. The fracture surface of the welded specimen was the same as for the base material. The macro fracture surfaces for the BM and the welded sample are presented in Figure 3.11(a) and (c). All of the specimens split into two segments, with the boundary between the segments coinciding with the mid-thickness position within the plate. It is speculated that this splitting is due to the specific rolling and fabrication process for the plate, which can lead to variation in the chemical composition in the through-thickness direction [23]. As plastic deformation accumulates, a crack may first develop parallel to the rolling direction and perpendicular to the fracture surface, before ultimately deviating to produce the final fracture. High magnification SEM fractographs for both the base material and the welded specimen are shown in Figure 3.11(b) and (d). These reveal ductile failures in each case, as the fractographs are composed predominately of large voids and deep equiaxed dimples. Interestingly, there were some obvious spherically-shaped inclusions located within the large voids. A qualitative indication of the chemical composition for these inclusions was obtained by energy dispersive X-ray spectroscopy (EDX). The EDX spectrum indicated that the inclusion particles were rich in Ca, O, Al, Mn and S, as can be seen in Figure 3.11(e) and (f).
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Figure 3.11 SEM micrographs showing the fracture surface morphology for the tensile test specimens, and the results of EDX analysis: (a) macro fracture surface from base material, (b) high magnification fracture surface from base material, (c) macro fracture surface from welded specimen, (d) high magnification fracture surface from welded specimen, (e) EDX spectrum for the inclusion particles in the base material, (f) EDX spectrum for the inclusion particles in the welded specimen.

Root and face bend tests were carried out on the as-welded joints, and the resulting specimens are shown immediately after testing (Figure 3.12(a)) and after liquid dye penetrant testing (Figure 3.12(b)). The welds were subjected to significant plastic deformation and no surface cracks were visible after either root or face bend testing,
which indicates that the laser welded joints exhibited good ductility and adequate bending strength.

![Image](image1.png)

Figure 3.12 Three-point bending test results, (a) original samples after bending test, (b) bent samples after dye penetrant testing.

### 3.3.5 Charpy impact properties

The Charpy impact test pieces used in this work were sub-sized (55 mm × 10 mm × 8 mm), owing to the limitation imposed by the thickness of the base material (8mm). In order to compare the results with those of standard 55 mm × 10 mm × 10 mm samples, corrections were applied to account for the reduced thickness. A number of studies on corrections for sub-sized Charpy V-Notch (CVN) specimens have been carried out [23-27]. Schubert et al. [26] described a normalization factor \( NF = \frac{Bb^2}{KL} \) to correct for sub-sized test results, where \( B \) is the specimen thickness, \( b \) is the length of the remaining ligament at the notch, \( K \) is the modified stress concentration factor at the notch and \( L \) is the specimen span. Test data obtained with sub-sized specimens can be extrapolated to corresponding results for a standard specimen using \( NF \). Chao et al. [27] tested the toughness of DP590 steel using sub-sized CVN specimens. In their case the only non-standard dimension for the test specimens was the thickness (5.5 mm). It was postulated that the only correction that was required was for the thickness. Therefore, they estimated CVN impact energies for standard specimens by multiplying the test data they obtained on their sub-sized specimens by a factor of 1.82 (i.e. 10/5.5). In this work we have a similar situation in that the only non-standard dimension is the specimen thickness. Accordingly, we have followed a similar approach in applying corrections to
the energies we measured on 8 mm thick specimens, *i.e.* by multiplying the absorbed energies by a factor of 1.25 (*i.e.* 10/8).

The original absorbed energy results for the sub-sized specimens, and the corresponding estimates for the absorbed energies in standard specimens, are shown at different test temperatures in Figure 3.13(a) and (b) for the BM and specimens with notches placed at the weld centre line, at the FL and in the HAZ region. In order to assess the fracture mechanism, the proportion of the fracture surface that was subject to brittle fracture is plotted for each type of specimen in Figure 3.13(c). There were some specimens with notches placed at the weld centre line (tested at room temperature) and the FL (tested at 0 °C and room temperature) for which the crack path deviated into the HAZ and base material. Many researchers have reported that the laser and electron beam welding processes may present difficulties for toughness testing because of the narrow fusion zones, together with a higher degree of strength overmatching of the joint [28-31]. Elliott reported that the tendency for the fracture to deviate into the base metal rather than propagate through the fusion zone can lead to misleading results [32]. The deviation of the crack during fracture may result in the test results for the specimens with notches in the weld metal and at the FL exhibiting significant scatter. Selected sub-sized fractured samples both with and without crack deviation are shown in Figure 3.14. Table 3.4 summarises each of the three test results for the three cases that were prone to crack deviation, so that the implications for experimental scatter can be assessed.
Figure 3.13 Absorbed energy results and proportion of brittle fracture on the fracture surface for the base material, weld metal, fusion line and HAZ impact specimens in S960 steel at different temperatures, (a) original absorbed energy results for sub-size specimens (55 mm × 10 mm × 8 mm), (b) corresponding estimates for absorbed energies in standard-sized specimens, (c) proportion of brittle fracture on the fracture surface.

Figure 3.14 Selected fractured samples with and without crack deviation (notch at FL, tested at room temperature).

Table 3.4 Summary of test results for the cases those were prone to crack deviation.

<table>
<thead>
<tr>
<th>Test specimens</th>
<th>Test No.</th>
<th>Deviation</th>
<th>Absorbed energy (J)</th>
<th>Proportion of brittle fracture (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Notch in the FZ</td>
<td>1</td>
<td>No</td>
<td>50</td>
<td>43</td>
</tr>
<tr>
<td>(room temperature)</td>
<td>2</td>
<td>To HAZ and BM</td>
<td>132</td>
<td>22</td>
</tr>
<tr>
<td></td>
<td>3</td>
<td>To HAZ and BM</td>
<td>122</td>
<td>25</td>
</tr>
<tr>
<td>Notch at the FL</td>
<td>1</td>
<td>To BM</td>
<td>149</td>
<td>0</td>
</tr>
<tr>
<td>(0 °C)</td>
<td>2</td>
<td>To BM</td>
<td>136</td>
<td>0</td>
</tr>
<tr>
<td></td>
<td>3</td>
<td>To BM</td>
<td>144</td>
<td>0</td>
</tr>
<tr>
<td>Notch at the FL</td>
<td>1</td>
<td>To BM</td>
<td>141</td>
<td>0</td>
</tr>
<tr>
<td>(room temperature)</td>
<td>2</td>
<td>To BM</td>
<td>141</td>
<td>0</td>
</tr>
<tr>
<td></td>
<td>3</td>
<td>No</td>
<td>91</td>
<td>34</td>
</tr>
</tbody>
</table>

An overall trend can be seen in the original test results and the converted results in Figure 3.13(a) and (b): the absorbed energies increase with an increase in the test
temperature. The proportion of brittle fracture for the BM and specimens with a notch placed at the weld centre, at the FL and in the HAZ region at different test temperatures in Figure 3.13(c) shows results consistent with the absorbed energies in Figure 3.13(a) and (b). The proportion of the brittle fracture region generally drops with an increase in the test temperature. A greater proportion of brittle fracture on the fracture surface leads to a lower absorbed energy.

The base material achieves good toughness, with converted average absorbed energy values of approximately 125 J, 144 J, 156 J and 158 J at test temperatures of -40, -20, 0 and 23 °C, respectively. It can be seen that all of the absorbed energy values for the specimens with notches placed at the weld centre, at the FL and in the HAZ region are lower than the corresponding values for the BM from -40 °C to room temperature. The converted absorbed energy values for the specimens with notches at the FL and in the HAZ show similar values at test temperatures of -40 °C and -20 °C, varying from 35 to 60 J. The converted absorbed energies for the specimens with notches at the weld centre are lower than for those specimens with notches at the FL and in the HAZ at test temperatures of -40 °C and -20 °C, varying gradually from 20 J to 25 J. The specimens with a notch at the FL tested at 0 °C and room temperature achieved higher absorbed energies than the specimens with a notch in the HAZ at the corresponding test temperatures, because the cracks deviated in to the BM. The converted average absorbed energy values varied from 143 J to 124 J. The converted average absorbed energy values for the specimens with a notch at the weld centre and in the HAZ were similar, varying from 65 to 90 J at 0 °C and at room temperature.

The macro-fracture surfaces of the Charpy impact test specimens were examined using an optical microscope. The proportion of brittle fracture on the fracture surfaces was measured and the results are summarized in Figure 3.13(c). All of the BM specimens fractured in a ductile manner, i.e. there was no brittle fracture region, regardless of test temperature. The proportion of brittle fracture for the specimens with a notch either in the weld, at the FL or in the HAZ, dropped dramatically as the test temperature increased from -40 °C to room temperature. The proportion of brittle fracture for the specimens with a notch at the weld centre dropped markedly from 84 to 31% as the test temperature increased from -40 °C to room temperature. There were two specimens with notches at the weld centre, which were tested at room temperature, for which
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fracture deviated in to the HAZ and BM, resulting in a proportion of brittle fracture of 23%. Another fractured without deviation with a brittle fracture region comprising 43% of the fracture surface, as is shown in Table 3.4. The average brittle fracture region fraction for the specimens with notches in the HAZ dropped markedly from 72 to 36% as the test temperatures increased from -40 °C to room temperature. The overall trend for the average brittle fracture region fraction, for the specimens with notches at the FL, was a sharp drop from 79 to 34% as the test temperatures increased from -40 °C to room temperature. However, the cracks for the three specimens with notches at the FL deviated in to the BM at a test temperature of 0 °C, as is mentioned in Table 3.4. All three replicate specimens fractured in an entirely ductile manner. There were two specimens with notches at the FL, which were tested at room temperature, for which the fracture path deviated in to the BM, giving a completely ductile fracture, while another fractured with no crack path deviation and this sample exhibited a brittle fracture region fraction of 34%, as is shown in Table 3.4.

The macroscopic fracture surfaces of selected specimens with fracture path deviation during Charpy impact testing are presented in Figure 3.15. Among the specimens with notches at the weld centre tested at room temperature, two exhibited fracture path deviation into the HAZ and BM, while another fractured without deviation. The undeviated specimen achieved a lower absorbed energy value (50 J), and exhibited a larger brittle fracture region, covering approximately 43% of the fracture surface (specimen on left in Figure 3.15(a)), while the deviated specimens achieved a higher absorbed energy value (122 J) and had a brittle fracture region covering approximately 25% of the fracture surface (see specimen on right). The boundary between the region of ductile fracture and the region of subsequent brittle fracture is clearly evident in Figure 3.15(a). All of the specimens with notches at the FL which were tested at 0 °C showed fracture path deviation in to the BM. The specimens exhibited a completely ductile fracture surface and achieved a high absorbed energy (144 J), as can be seen in Figure 3.15(b). Two of the specimens with notches at the FL which were tested at room temperature fractured with crack path deviation to BM, while another fractured with no deviation. The fractured specimens tested at room temperature (Figure 3.15(c)) present two very different fracture surfaces: the un-deviated sample on the left reveals that the crack initially propagated in a ductile manner before continuing to propagate in a brittle manner over part (~ 34 %) of the fracture surface, and the absorbed energy was slightly
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lower in this specimen (91 J). In contrast, the specimen with fracture deviation to the BM presents a completely ductile fracture surface (141 J).

Figure 3.15 Selected macro-fractured surfaces with fracture deviation, (a) notch at the weld centre tested at room temperature, (b) notch at the FL tested at 0 °C, (c) notch at the FL tested at room temperature.

Figure 3.16(a)-(d) present the fracture surface morphologies and EDX results for the base material after impact testing at temperatures ranging from -40 °C to room temperature. The fracture surfaces of the BM reveal a typical ductile fracture, with fracture surfaces showing dimples and microvoids, with some of the dimples containing spherical inclusions. Selected inclusion particles were examined using EDX, and all of the EDX results indicated similar inclusion compositions. One of the selected EDX results, corresponding to the inclusion particle shown in Figure 3.16(a), is presented in Figure 3.16(e). The results are similar to those for the tensile fracture samples: the inclusion particles are rich in Ca, O, Al, Mn and S.
Figure 3.16 Fractured surfaces for the base material after Charpy impact testing. (a) BM (-40 °C), (b) BM (-20 °C), (c) BM (0 °C), (d) BM (23 °C), (e) EDX spectrum for an inclusion particle in the BM (-40 °C).

Cleavage fracture was confirmed in those specimens with the notch at the weld centre, which were tested at -40 and -20 °C, and these specimens fractured with low absorbed energies. The brittle regions on the fracture surfaces revealed a cleavage dominated fracture, as is shown in Figure 3.17(a) and (b). At 0 °C and at room temperature, the specimens with notches at the weld centre fractured with noticeably higher absorbed energies, and these specimens are shown in Figure 3.17(c) and (d). The brittle regions
on these fracture surfaces revealed a cleavage dominated fracture with a small proportion of dimples.

Figure 3.17 Fractured surfaces for the specimens with a notch in the weld after Charpy impact testing, (a) weld (-40 °C), (b) weld (-20 °C), (c) weld (0 °C), (d) weld (23 °C).

The specimens with a notch at the FL which were tested at -40 and -20 °C fractured with brittle fracture region fractions of approximately 80% and 70%, respectively. The brittle regions on the fracture surfaces revealed a cleavage dominated fracture, as is shown in Figure 3.18(a) and (b). In contrast the specimens tested at 0 °C, which showed fracture path deviation to the BM and achieved a higher absorbed energy, revealed a ductile fracture surface with dimples and microvoids, some containing spherical shaped inclusions, as is shown in Figure 3.18(c). Two of the specimens with notches at the FL which were tested at room temperature showed fracture path deviation to BM, and these achieved higher absorbed energy values. These fracture surfaces also showed dimples and microvoids and spherical inclusions, as shown in Figure 3.18(d). According to the EDX results, all of the inclusion particles contain similar elements. EDX results for the inclusion particle in Figure 3.18(c) are shown in Figure 3.18(f). The inclusion particles are rich in Ca, O, Al, Mn and S. However, another specimen tested at room temperature,
which showed no fracture path deviation, exhibited approximately 34% brittle fracture. The brittle regions on the fracture surfaces revealed a cleavage dominated fracture, as is evident in Figure 3.18(e).
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Figure 3.18 Fractured surfaces for the specimens with a notch at the FL after Charpy impact testing, (a) FL (−40 °C), (b) FL (−20 °C), (c) FL (0 °C), (d) FL with fracture deviated (23 °C), (e) FL without fracture deviated, (f) EDX spectrum for the inclusion particle in the specimen with a notch at the FL (0 °C).

The fracture surfaces for the specimens with notches located in the HAZ are shown in Figure 3.19, and it can be observed that cleavage fracture was confirmed in the brittle regions for all of the specimens, regardless of the test temperature. The specimens tested at −40 and −20 °C revealed a cleavage dominated fracture in the brittle regions on the fracture surfaces, as is shown in Figure 3.19(a) and (b). The specimens tested at 0 °C and at room temperature also exhibited a cleavage dominated fracture, but with a small proportion of dimples (Figure 3.19(c) and (d)). The spherical shaped inclusion particles located in the microvoids in Figure 3.19(c) are also rich in Ca, O, Al, Mn and S, as can be seen in the corresponding EDX results (Figure 3.19(e)). In addition, it can be seen from Figure 3.17–Figure 3.19 that the cleavage facets on the fracture surfaces of the weld and FL specimens are larger than they are for the HAZ specimens.
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Figure 3.19 Fractured surfaces for the specimens with a notch in the HAZ after Charpy impact testing, (a) HAZ (-40 °C), (b) HAZ (-20 °C), (c) HAZ (0 °C), (d) HAZ (23 °C), (e) EDX spectrum for the inclusion particle in the HAZ (0 °C).

3.4 Discussion

3.4.1 Microstructure characteristics in different sub-zones

The development of microstructure within the different zones of a welded joint will be dependent on the local thermal cycles, and the peak temperature that is reached during
welding is particularly important in controlling weld microstructures. For steels, the $\text{Ac}_1$ and $\text{Ac}_3$ transformation temperatures refer to transformation temperatures during heating; the $\text{Ac}_1$ temperature corresponds to the temperature at which the BM begins to transform to austenite, and the $\text{Ac}_3$ temperature corresponds to the completion of the transformation to austenite. The transformation temperatures for this steel were measured using dilatometry. During heating the $\text{Ac}_1$ temperature was found to be 717 °C and this value was found not to be particularly sensitive to heating rate. The $\text{Ac}_3$ temperature varied from 890 to 930 °C as the heating rate increased from 150 to 450 °C/s. The $\text{Ac}_3$ temperature for this steel during the welding process would be expected to be higher than 930 °C due to the greater heating rates that are likely to be experienced during the laser welding process.

In the SCHAZ, during the laser welding process, the peak temperature does not quite reach the $\text{Ac}_1$ temperature, and so no material is re-austenized in this region [33]. The original microstructure of the BM is retained, but some tempering of the BM is likely to occur. The microstructure in this SCHAZ is tempered martensite and tempered lower bainite, as is shown in Figure 3.7(a), which can be seen to be similar to the BM in Figure 3.1. However, recovery may happen in the SCHAZ because of the relatively high peak temperature (although lower than the $\text{Ac}_1$ temperature) during the welding process, which may result in a reduction in the dislocation density and softening of the microstructure in this region.

During the laser welding process, if the peak temperature increases to a value between the $\text{Ac}_1$ and $\text{Ac}_3$ temperatures, some austenite grains will nucleate on the tempered martensite and lower bainite packets or grain boundaries [22, 34, 35]. During this partial transformation, owing to the fast cooling rates associated with the laser welding process, this newly generated austenite transforms to martensite, which can increase the hardness and lower the toughness of this region [33, 36]. The un-transformed base material is tempered by the welding thermal cycle. The martensite within the ICHAZ reflects the original banded morphology of the BM, as is shown in Figure 3.7(b). The banded microstructure in the BM reflects micro-segregation in the original BM, and it is in these regions that transformation to austenite first begins.
Moving closer to the fusion line, the peak temperatures in the FGHAZ are slightly higher than the $\text{Ac}_3$ temperature. This results in complete re-austenitisation of the BM, but there is limited austenite grain growth due to the relatively low peak temperatures and the short periods of time at these temperatures in the FGHAZ \cite{22, 37}. During rapid cooling following welding, this fine-grained austenite transforms completely to martensite, and the martensite can self-temper during the cooling process, as is seen in Figure 3.7(c).

Adjacent to the fusion line, the peak temperature is raised to 1300 °C or higher (but lower than the melting point) for a sufficient period to ensure complete re-austenitisation, and rapid grain growth \cite{22, 38}. During rapid cooling following welding, martensite forms in the CGHAZ, and some martensite will self-temper during the cooling process, as seen in Figure 3.7(d). The grain size within the CGHAZ is larger than that of the FGHAZ. The prior austenite grain boundaries are indicated by arrows.

The FZ was melted during the heating process, and it subsequently solidified during the cooling process. In this fast solidification process, a large number of very large austenite grains formed, some of which were columnar and orientated parallel to the direction of heat flow. Under the very fast cooling conditions following welding, these very large austenite grains again transformed to martensite. After the martensite transformation was finished, the martensite was partly self-tempered during subsequent cooling. The large prior austenite grain boundaries and martensitic structure of the FZ can be seen clearly in Figure 3.7(e).

### 3.4.2 Mechanical properties

The fast heating and cooling rates associated with laser welding result in the generation of martensite in the FZ, CGHAZ and FGHAZ, and the partial transformation to austenite in the ICHAZ results in regions of martensite. This results in the hardness in the FZ, CGHAZ and FGHAZ increasing by 65 to 75 HV$_{0.3}$, and the hardness in the ICHAZ increasing by 30 to 40 HV$_{0.3}$ when compared with the hardness in the BM (~310 HV$_{0.3}$). No phase transformation occurred in the SCHAZ. However, there may be a reduction in dislocation density due to the recovery in this region, which led to development of a narrow soft region in the SCHAZ (~285 HV$_{0.3}$).
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There is an approximately proportional relationship between strength and hardness in steels, with the harder material having higher strength, although this is not always the case [39]. The strengths of the fusion zone and each sub-zone within the HAZ were increased mainly by the generation of martensite. Some researchers [40, 41] have investigated the correlation between the microstructure and the fracture toughness within the heat affected zone in SA508 steel. Lee et al. [41] reported that the yield strengths were over 1100 MPa within the CGHAZ and FGHAZ, and that these values were approximately double the yield strength of the base material (~ 500 MPa). In the same study, the microhardness values in the CGHAZ and FGHAZ were approximately double that of the base material for SA508 steel in the as-welded condition. Furthermore, the yield strength for the ICHAZ was approximately 600 MPa, and, the hardness in the ICHAZ was a little higher than that in the BM because of the partially transformed martensite in this region. The hardness of the FZ, CGHAZ, FGHAZ and ICHAZ in the as-welded S960 steel is approximately 1.2 times that of the BM. According to Lee et al.’s [41] results, this suggests that the yield strength of the FZ, CGHAZ and FGHAZ could be ~ 1.2 times that of the base S960 steel. Since the microstructure in the ICHAZ was partially transformed to martensite, the yield strength could equal or even exceed that of the BM. In addition, there is a narrow soft SCHAZ, which could result in a lower yield strength in this narrow region when compared with the BM. Laser welding resulted in the hardness across the majority of the joint being greater than that of the base material. The laser welded tensile test specimens failed in the BM without losing any strength, so it would appear that the very narrow soft sub-zones were not detrimental to the overall strength of the welded joint.

In the case of the laser welded specimen, the fusion zone and HAZ (with the exception of the SCHAZ) had a much higher hardness than the BM, while the SCHAZ was a little softer than the BM, but this region is very narrow (~ 0.6 mm), as can be seen in Figure 3.8. The hard fusion zone and the hard regions of the HAZ can therefore act to constrain plastic deformation within the SCHAZ, thereby resulting in the majority of the tensile plastic deformation accumulating in the BM, leading to subsequent fracture in the BM [17, 42, 43], as seen in Figure 3.10. Lee et al. [17] reported that this concentration of deformation in the BM was confirmed by optical observation of the local strain distribution in their test on laser welded DP780 steel.
The generation of hard martensite in the FZ and HAZ strengthened the laser welded joint but also increased the brittleness of the joint. This is demonstrated by the results in Figure 3.13: the absorbed energy values for the specimens with a notch placed at the weld centre, at the FL and in the HAZ are lower than the corresponding values for the BM from -40 °C to room temperature. However, post weld heat treatment could be applied to improve the toughness of the laser welded joint. It was found from the Charpy impact testing that some of the test specimens with notches at the weld centre and at the FL fractured with cracks deviating in to the HAZ and BM (Figure 3.14 and Table 3.4), which gave misleading impact toughness results. This deviation can be attributed to the mismatch in properties across the laser welded joint, as well as to the narrowness of the fusion zone and HAZ.

It must be mentioned here that the FL is not a perfectly straight line. A notch aligned with the FL would be expected to sample portions of the FZ and/or CGHAZ. The HAZ specimens were produced with the notch at a 0.5 mm offset from the FL, toward the HAZ side, and so the notch may “sample” the CGHAZ and/or FGHAZ. It could be found from the hardness results in Figure 3.8, that the hardness value at the FL is close to and also a little lower than that within the FZ; while the hardness value in the HAZ (FL + 0.5 mm) is a little higher than that within the FZ. In addition, the prior austenite grain size in the FZ is larger than that at the FL, in the CGHAZ and in the FGHAZ. A slightly softer FZ (compared with the hardness in the HAZ) could improve the toughness of the FZ to some extent. However, a larger austenite grain size would reduce toughness, as austenite grain boundaries act as barriers to crack propagation. The absorbed energy values for the FL and HAZ specimens are close to and a little higher than those for the specimens with a notch at the weld centre at -40 °C and -20 °C. This means that the coarse grains in the FZ are having a more significant and detrimental influence on Charpy toughness than any benefits which may be attributed to the slightly reduced hardness. The high magnification SEM fracture images in Figure 3.17–Figure 3.19 reveal that the cleavage facet size in the weld and FL fracture specimens is larger than that in the HAZ specimens, which can also be attributed to the different prior austenite grain sizes in each sub-zone. The FL specimens tested at 0 °C and at room temperature, and the weld centre specimens tested at room temperature, fractured with cracks deviating in to the HAZ and BM. As such, the absorbed energy values for the FL specimens tested at 0 °C and at room temperature, and the weld centre specimens, gave
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higher and misleading impact toughness results while exhibiting significant scatter. If the scatter is taken in to account, it can be found that the absorbed energy values for the specimens with a notch in the HAZ and at the weld centre, tested at 0 °C and room temperature, are very close. The toughness of the specimens with a notch at the weld centre and at the FL is not very sensitive to the two factors described above at higher test temperatures (0 °C and room temperature).

3.5 Conclusions

From this investigation the following conclusions were derived:

(1) It is possible to weld 8 mm thick S960 HSLA steel plates in a single weld pass, without any macroscopic defects, using an autogenous laser welding process with a laser power of 6.5 kW and welding speed of 1.08 m/min.

(2) The tensile strength of the laser welded specimens matched the strength of the BM, with all welded specimens failing in the BM. The laser welded joint also performed well in bending tests.

(3) Autogenous laser welding produced a fusion zone microstructure comprising very coarse lath martensite, with some self-tempered martensite. In the heat affected zone, the microstructures were also predominantly martensitic with some self-tempered martensite, with the prior austenite grain size reducing gradually moving from the CGHAZ to ICHAZ. The SCHAZ retained the microstructure of the BM.

(4) The maximum hardness coincided with the FGHAZ, where the average value was approximately 385 HV$_{0.3}$, for the welding parameters used in this study. The average hardness in the FZ was approximately 375 HV$_{0.3}$. The hardness in the HAZ dropped rapidly when moving from the FGHAZ to the SCHAZ. There were narrow hardness “valleys” in the SCHAZ, in which the hardness locally dropped below that for the base material.

(5) The generation of martensite in the FZ and HAZ led to increased hardness in the weld zone, but also a reduction in the impact toughness in comparison with the base
material. Low Charpy impact toughness for the autogenous laser welded joints is unfavourable

References

Chapter 3


Chapter 4: Process-parameter interactions in ultra-narrow gap laser welding of high strength steels

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Abstract

S960 and S700 are two new high strength steels recently developed by Tata Steel. Laser welding of thick section steels has been widely used in offshore construction and shipbuilding. Melt sagging is a typical defect for single pass autogenous laser welding of thick section materials. Multi-pass narrow gap laser welding techniques become more attractive because they can weld thicker sections of material with a moderate laser power and suppress the melt sagging problem. In addition, this approach requires less filler material, and the cumulative heat input to the material is reduced when compared with traditional arc welding techniques. However, there are many variables involved in this narrow gap laser welding technique, which makes this process more complicated than single pass autogenous laser welding. In this study, the effects of multi-pass ultra-narrow gap laser welding parameter interactions (\textit{i.e.} laser power, welding speed and wire feed rate) on laser weld quality and the welding efficiency for S960 high strength steel plates were investigated. Moderate laser powers of 1 to 2 kW were used to weld S960 high strength steel plates with a very narrow parallel groove (1.2 mm). Statistical design of experiments was carried out to assess the process parameter interactions and to optimise the ultra-narrow gap laser welding parameters. Validation experiments indicate that the proposed models predict the responses adequately within the range of welding parameters that were used. Defect-free welds in 6 mm thick S960 steel were obtained with only two passes, using the optimised welding parameters, and these optimised parameters were successfully transferred to the welding of 8 mm thick S960
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steel. In addition, they were also successfully transferred to the welding of 13 mm thick S700 steel with a small modification. The optimised narrow gap laser welded joints show almost the same tensile properties as the base material, with failures occurring in the base material away from the weld.

Keywords: high strength steel, ultra-narrow parallel groove, solid filler wire, statistical modelling

4.1 Introduction

High strength steels are used in a wide variety of applications such as for structural components, pressure vessels and oil/gas transportation pipes, in shipbuilding, offshore construction (wind power and oil exploration) and in the automotive industry [1]. The application of high strength steels leads to not only to reductions in weight but also to compact structures [2]. Thin gauge high strength steels (0.8-3 mm thick and below) are widely used in the automotive industry [3-5]. For offshore pipelines, the current trends are towards the use of grade X70 steels with a wall thickness up to 40 mm, and for X80 and X100 pipeline steels with a wall thicknesses of 25 mm and below [6].

Compared to traditional arc welding techniques, laser welding presents many advantages for the welding of high strength steels, such as rapid welding speeds, deeper penetration depths, narrower heat affected zones and lower thermal distortions [7, 8]. However, there are limits to the maximum laser power that is available in commercial lasers (typically less than 20 kW), which restricts the maximum welding depth using the single pass autogenous laser welding technique. It was reported that the laser penetration depth in single pass welding is typically in the order of 1-2 mm/kW [9, 10]. This will restrict the application of single pass laser welding to thick section components, such as thick offshore pipelines and pressure vessels. Autogenous laser welding (without a filler material) requires precise fit-up, which is often considered as a barrier to the wider applications of laser welding in the industry. In addition, thicker section laser welding with a single pass approach often results in porosity, molten material dropout, cracking, and mis-tracking of weld seams [7, 11, 12].
Over the last few years, multi-pass narrow gap laser welding techniques have been demonstrated, which can be applied to the welding of very thick section components, using a filler wire and moderate laser powers. Zhang et al. [13] demonstrated that a 50 mm thick 316L stainless steel plate could be welded using a narrow gap laser welding approach with a laser power of 8 kW and a groove width of 2-4 mm with a taper of 5°. In addition, the introduction of a filler wire into the welding groove can improve the metallurgy and the properties of the weld. These researchers effectively suppressed solidification cracks through controlling the chemical composition of the filler wire [13]. Elmesalmy et al. [14] successfully welded 316L stainless steel plates that were 20 mm in thickness using a 1 kW IPG single mode fiber laser with an ultra-narrow gap (1.5 mm gap width) and a taper of 3° from both sides using a multi-pass narrow gap laser welding technique. Li et al. [15] reported a combined approach which employed autogenous laser welding, laser welding with a filler wire and the hybrid laser-gas metal arc welding (GMAW) welding technique to weld a 30 mm thick Q235 steel plate using a multi-layer, multi-pass process and a “Y” shaped groove.

Narrow gap laser welding is an emerging welding technique that can be applied to welding thick section materials with a narrow groove and a filler wire. However, this technique involves more complex process parameters when compared with autogenous laser welding as it has more interrelated variables. The relative distance between the laser beam and the filler wire, the laser power, the focal point, the angle between the wire and the laser beam, the welding speed, the filler wire diameter and the wire feed rate all have influences on the quality of the weld. All these variables make the narrow gap laser welding process more complicated than the single pass autogenous laser welding technique [16, 17]. The interactions between these welding parameters directly affect the quality of a welded joint. Traditionally, many welding trials with different welding parameters are carried out to optimise the welding parameters in order to obtain defect-free welded joints. This approach is inefficient, which leads to long process development times, consuming a large amount of test materials and energy.

Various statistical optimisation methods, such as response surface methodology (RSM), the Taguchi method and computational networks have been applied to defining the desired output variables through developing mathematical models that specify a relationship between the input parameters and output variables [18, 19]. In Benyounis
and Olabi’s [20] review of the application of these different statistical and numerical approaches, it was noted that RSM performs better than other techniques when a large number of experiments is not affordable. RSM is a statistical method that can be used to establish relationships between input variables and responses based on linear and quadratic polynomial equations. After the most important input variables have been identified, these equations can be optimised to span a reasonable range of process parameters [17, 20].

Elmesalamy et al. [9] used a statistical modelling approach to optimise the welding parameters for ultra-narrow gap laser welding of 316L stainless steel using a maximum laser power of 1 kW. In their work, the influences of laser power, welding speed and the wire feed rate on the integrity of the weld beads, the gap bridgability, the weld bead overlap factor, surface oxidation and the welding efficiency were investigated and optimised. Shi et al. [17] studied the interactions between welding parameters and the geometry of single beads using statistical methods. In their study, RSM based on a central composite design (CCD) was used and the validated models were applied to the multi-pass narrow gap laser welding of 20 mm thick AH32 high-strength ship steel. In order to obtain a reasonable range of geometric parameters for the single bead, the range in laser power was selected to be from 3 to 5 kW, while the range in welding speed was selected to be from 0.4 to 0.6 m/min, and the wire feed rate was selected to vary from 3 to 5 m/min. These workers successfully welded the 20 mm thick high-strength ship steel plates after applying the models they created. However, some porosity and lack of fusion defects still existed in the weld beads.

As can be seen from the above review, statistical modelling is an effective approach for investigating multiple parameter interactions in narrow gap laser welding. However, no attempt has been made so far to assess the laser welding of S960 or S700 high strength steel plates using the narrow gap method. These two steels are relatively new, having only recently been developed by Tata Steel. This study, therefore, aims to investigate process parameter interactions and study the influences of welding input parameters (laser power, welding speed and wire feed rate) on the three responses (i.e. integrity of the weld bead, the weld bead width and the number of filling passes) for the ultra-narrow gap laser welding of S960 high strength steel plates with thicknesses of 6 and 8 mm. The optimised welding parameters were also transferred to the welding of 13 mm
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thick S700 high strength steel plates. The tensile properties were examined in order to evaluate the quality of the optimised ultra-narrow gap laser welded high strength steel joints.

4.2 Materials and experimental procedures

The materials used in this study were 6 and 8 mm thick S960 and 13 mm thick S700 high strength steel plates. S960 steel and S700 steel are both advanced high strength low alloy steels with minimum yield strengths of 960 and 700 MPa, respectively. Both steels were recently developed by Tata Steel. These two steels are produced by thermomechanically controlled processing (TMCP) technology, which involves hot rolling of the steel at carefully controlled temperatures \(i.e.\) controlled rolling and/or quenching of the steel as part of the hot rolling process \(i.e.\) direct quench. TMCP technology achieves a high strength with low levels of alloying through the exploitation of strengthening mechanisms such as precipitation hardening and/or grain refinement. The solid filler wires used for the welding of the S960 and S700 steels in this study were referred to as Union X96 (ER120S-G) with a diameter of 0.8 mm and Carbofil 2NiMoCr (ER120S-G) with a diameter of 1 mm, respectively. The chemical compositions for the steels and filler wires are given in Table 4.1, while the mechanical properties are shown in Table 4.2. All of these data were provided by Tata Steel. The carbon equivalent (CE) values for the investigated steels and filler wires were calculated according to the IIW equation [21]:

\[
CE = C + \frac{Mn}{6} + \frac{(Cr + Mo + V)}{5} + \frac{(Ni + Cu)}{15}
\]  

(4.1)

Table 4.1 Chemical compositions of base S960 and S700 steels, and Union X96 and Carbofil 2NiMoCr filler wires (wt.%).

<table>
<thead>
<tr>
<th>Material</th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>N</th>
<th>B</th>
<th>Ca</th>
<th>Si</th>
<th>Al</th>
<th>Cu</th>
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<tbody>
<tr>
<td>S960 steel</td>
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<td></td>
<td></td>
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<tr>
<td>C</td>
<td>0.088</td>
<td>1.51</td>
<td>0.008</td>
<td>0.002</td>
<td>0.0089</td>
<td>0.0022</td>
<td>0.012</td>
<td>0.055</td>
<td>0.033</td>
<td>0.014</td>
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<td>0.023</td>
<td>0.248</td>
<td>0.025</td>
<td>0.040</td>
<td>0.050</td>
<td>Bal.</td>
<td>0.496</td>
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<td>Union X96</td>
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<td>C</td>
<td>0.12</td>
<td>0.80</td>
<td>1.90</td>
<td>0.45</td>
<td>0.55</td>
<td>2.35</td>
<td>Bal.</td>
<td>0.793</td>
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<td>S700 steel</td>
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<td>C</td>
<td>0.068</td>
<td>1.476</td>
<td>0.009</td>
<td>0.001</td>
<td>0.05</td>
<td>0.073</td>
<td>0.495</td>
<td>0.19</td>
<td>0.03</td>
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<td>B</td>
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<td>0.0045</td>
<td>Bal.</td>
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<td>Carbofil</td>
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</table>
Table 4.2 Mechanical properties of base S960 and S700 steels, and Union X96 and Carbofil 2NiMoCr filler wires.

<table>
<thead>
<tr>
<th>Material</th>
<th>Yield strength 0.2 % (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
<th>Impact value (J)</th>
</tr>
</thead>
<tbody>
<tr>
<td>S960 steel</td>
<td>≥ 960</td>
<td>1000</td>
<td>7</td>
<td>120 (-40 °C)</td>
</tr>
<tr>
<td>Union X96</td>
<td>930</td>
<td>980</td>
<td>14</td>
<td>50 (-50 °C)</td>
</tr>
<tr>
<td>S700 steel</td>
<td>≥ 700</td>
<td>760</td>
<td>12</td>
<td>≥ 42 (-40 °C)</td>
</tr>
<tr>
<td>Carbofil 2NiMoCr</td>
<td>≥ 960</td>
<td>≥ 940</td>
<td>≥ 15</td>
<td>≥ 47 (-40 °C)</td>
</tr>
</tbody>
</table>

A schematic diagram of the setup for the ultra-narrow gap laser welding process is given in Figure 4.1. A continuous wave fiber laser (IPG YLS-16000) with a maximum achievable power of 16 kW was used in these experiments. The beam parameter product of the laser was 10 mm mrad and the beam was optically delivered (with a fiber core diameter of 300 μm) to the output lenses. The laser beam emitted from the end of the optical fiber was collimated with a lens with a 150 mm focal length and then focused onto the specimen surface using a lens with a 400 mm focal length. The measured focal spot size and Rayleigh length were 0.8 mm and 15 mm, respectively. The laser welding head was mounted on a 6-axis KUKA robot. The filler wire was fed into the leading edge of the molten pool at an angle of 30° with respect to the specimen surface. Argon gas was blown onto the weld pool surface through a copper tube at a flow rate of 12 l/min, and the backing gas was delivered through a side tube to a shielding chamber under the specimen at a flow rate of 8 l/min to protect the back surface from oxidation.

Figure 4.1 Schematic representation of setup for ultra-narrow gap laser welding.
Usually a “Vee” groove is prepared for narrow gap laser welding, and a filler wire is required in order to fill the groove using a multi-pass approach [13-15, 17]. However, the use of a Vee groove increases the volume of filler material that is consumed and the number of filling passes that is required. In Yu et al.’s [22] investigation on multi-pass narrow gap laser welding of 17 mm Q235 low-carbon steel plates, it was found that using a relatively small groove not only reduced the consumption of filler wire but it also reduced the deflection of the filler wire in the groove, which can improve fusion within the groove. So, in the present work, a parallel groove with a gap width of 1.2 mm was prepared for the S960 steel plates and a gap width of 1.4 mm was employed for the S700 steel plates. The groove geometries are shown in Figure 4.2.

![Groove geometries for ultra-narrow gap laser welded specimens, (a) 6 mm thick S960 steel, (b) 8 mm thick S960 steel, and (c) 13 mm thick S700 steel.](image)

Prior to the welding experiments, the parent material in the vicinity of the machined groove was cleaned by sand blasting in order to remove the surface oxide, and it was then cleaned with ethanol. The welding trials for the design of experiments and statistical modeling stage, as well as for the process optimisation, were carried out on 6 mm thick S960 steel plates. The optimised welding parameters were then transferred to the welding of 8 mm thick S960 and 13 mm thick S700 steel plates. After welding, cross-sectional samples were extracted from the plates. The samples were ground and polished to a 1 µm finish, and then etched in a solution of 2% Nital for about 2 s. The assessment of the macrostructures for the welded joints was carried out using a KEYENCE VHX-500F optical microscope. Image analysis software (ImageJ) was
applied to measure the weld bead widths and the cross sectional areas of the welded samples.

Autogenous laser welding was used first to weld the 2 mm thick root face. For root pass welding, a laser power of 2 kW was selected. The laser focal point was set 4 mm below the top surface of the specimen, i.e. the focal position coincided with the top surface of the landing at the weld root. The welding speed was varied from 20 mm/s (1.2 m/min) to 35 mm/s (2.1 m/min) in increments of 5 mm/s (0.3 m/min). Figure 4.3 shows the cross sections of the root pass welds at different welding speeds. It can be seen that all the root passes resulted in sound weld beads without defects. However, from the underside appearances of the root welds (see Figure 4.4), it can be seen that there is discontinuous weld penetration with a welding speed of 35 mm/s (2.1 m/min). Full penetration was consistently achieved for welding speeds ranging from 20 mm/s (1.2 m/min) to 30 mm/s (1.8 m/min). However, the underside weld appearance was much smoother with a welding speed of 20 mm/s (1.2 m/min) when compared with 25 mm/s (1.5 m/min) and 30 mm/s (1.8 m/min). So welding parameters of 2 kW and 20 mm/s (1.2 m/min) were selected to weld the root face in the ultra-narrow gap laser welding experiments.
Figure 4.3 Cross sections of the root pass welds for 6 mm thick S960 steel plates at different welding speeds, (a) 20 mm/s (1.2 m/min), (b) 25 mm/s (1.5 m/min), (c) 30 mm/s (1.8 m/min) and (d) 35 mm/s (2.1 m/min).

Figure 4.4 Underside appearance of root welds at different welding speeds, (a) 20 mm/s (1.2 m/min), (b) 25 mm/s (1.5 m/min), (c) 30 mm/s (1.8 m/min) and (d) 35 mm/s (2.1 m/min).

After root pass welding, the laser beam was focused on the top surface of the specimens to perform the subsequent filling passes. A larger laser beam diameter at the beam/material interaction point in the groove would increase the coverage of the laser beam within the groove, which would be beneficial from the point of view of melting the filler wire and the groove walls in order to avoid lack of fusion during the filling passes.

Tensile test specimens for the as-received 8 mm thick S960 and 13 mm thick S700 base materials and the optimised welded samples were produced in accordance with ASTM E8M-04. Three replicates for tensile testing were prepared in each case to reduce experimental uncertainties. Samples prepared for tensile testing were extracted from the steady state region of the welds using wire electric discharge machining (EDM), with
the long axis of each specimen type being normal to the welding direction. Tensile test coupons were sliced to coincide with the rolling direction of the material, and the configuration is shown in Figure 4.5. The weld reinforcements in the face and root regions were removed by manual grinding before the tensile tests were conducted. Tensile tests were carried out on an Instron model 8500 electronic universal testing machine at room temperature.

![Figure 4.5 Dimension and configuration of tensile test samples.](image)

### 4.3 Design of experiments and statistical modelling approach

Response surface methodology (RSM) is one of the most widespread statistical modelling approaches to identifying the significant processing parameters and quantifying their relationships with particular measured outputs [17, 23, 24]. The concept of RSM was introduced by Box and Wilson in 1951 [25]. RSM is a set of mathematical and statistical approaches that are useful for modelling and predicting one or more responses that are influenced by a number of input variables with the particular aim of optimising the responses [26, 27]. The functional relationship between a response of interest ($Y$) and input variables ($x_i$ and $x_j$) can be illustrated by Eq. (2) [9, 24]:

$$Y = \alpha_0 + \sum_{i=1}^{k} \beta_i x_i + \sum_{i<j}^{k} \beta_{ij} x_i x_j + \sum_{i=1}^{k} \beta_{ii} x_i^2 + \beta_0$$

(4.2)

where, $\alpha_0$ is the random experimental error, $\beta$ is a vector of $p$ unknown coefficients, $\beta_0$ is the response at the centre point; $\beta_i$ is the coefficient of the main linear components; $\beta_{ij}$ is the coefficient of the two linear factor interactions, and $\beta_{ii}$ is the coefficient of the quadratic factor.

The ultra-narrow gap laser welding technique involves several variable parameters, which affect the responses of interest and the optimum welding parameters. Central composite design (CCD) is one of the most popular experiment designs, and it has been
found to be most efficient in terms of the number of runs required [17, 24]. A CCD design consists of three types of design points (central, axial, and factorial points) [9, 24].

Analysis of variance (ANOVA) can be used to determine the significant parameters and set the optimal level for each parameter, which is tested against regression. An $F$ test was applied to each term in the model to check if the regression model was significant and to establish the significant model terms and their significance levels. The model terms are indicated as “significant” when values of “prob>F” are less than 0.05. In addition, the step-wise regression method was also applied to eliminate the insignificant model terms automatically. The “lack of fit value” represents the variation of actual data around the fitted model. If the “lack of fit” is not significant according to the $F$ test results, this indicates that the model fits the data and this is desirable. The $R^2$ and the adjusted $R^2$ values are the other two criteria for assessing models. The $R^2$ parameter is used to indicate the adequacy of a fitted regression model [9, 28]. $R^2$ is a measure of the variation around the mean, and the adjusted $R^2$ value is a measure of the variation around the mean of the adjusted model terms. “$R^2_{adj}$” is calculated according to Eq. (3) [9]:

$$R^2_{adj} = \frac{n-1}{n-p} + (1 - R^2)$$  \hspace{1cm} (4.3)

where $p$ is the number of model parameters and $n$ is the number of experiments. Scatter diagrams were also used for regression. Perturbation curves and response surface graphs were derived for each model. “Adeq Precision” measures the signal-to noise ratio of each model. If the ratio is greater than 4, it indicates that the model is desirable [9, 17, 28, 29].

Based on the fit summaries, suitable response models for the response factors should be selected. In Khan et al.’s work on the optimisation of process parameters for the laser welding of martensitic stainless steels, two factor interaction (2FI) models were statistically recommended [30].

In this investigation, a CCD experimental plan with three factors and five levels was used to implement the design matrix. The welding input variables were laser power,
welding speed and wire feed rate. In order to find out the working range for each variable, preliminary bead-on-plate trials were carried out with a filler wire, by varying one of the process parameters at a time and checking the working ranges for acceptable weld quality. The integrity of the weld bead, the weld bead width and the number of filling passes were selected as the responses. The statistical modeling software package Design-Expert 7.0 was used to code the variables and to establish the design matrix. The independent process variables and factor levels, and the design matrix are shown in Table 4.3 and Table 4.4, respectively. A set of 17 multi-pass ultra-narrow gap laser welding experiments was carried out with different welding parameters according to the design matrix. One sample was manufactured in the case of axial and factorial points, but the experiment for the central point was repeated three times.

Table 4.3 Independent variables and experimental design levels.

<table>
<thead>
<tr>
<th>Variable</th>
<th>Level 1 (-1)</th>
<th>Level 2 (-0.5)</th>
<th>Level 3 (0)</th>
<th>Level 4 (0.5)</th>
<th>Level 5 (1)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Laser power (kW)</td>
<td>1.0</td>
<td>1.2</td>
<td>1.5</td>
<td>1.8</td>
<td>2.0</td>
</tr>
<tr>
<td>Welding speed (mm/s)</td>
<td>10</td>
<td>13.6</td>
<td>19</td>
<td>24.4</td>
<td>28</td>
</tr>
<tr>
<td>Wire feed rate (mm/s)</td>
<td>10</td>
<td>19</td>
<td>33</td>
<td>46</td>
<td>55</td>
</tr>
</tbody>
</table>

Table 4.4 Design matrix with code independent process variables.

<table>
<thead>
<tr>
<th>Experiment No.</th>
<th>Point type</th>
<th>Laser power (kW)</th>
<th>Welding speed (mm/s)</th>
<th>Wire feed rate (mm/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Fact</td>
<td>1.0</td>
<td>28</td>
<td>10</td>
</tr>
<tr>
<td>2</td>
<td>Fact</td>
<td>2.0</td>
<td>28</td>
<td>10</td>
</tr>
<tr>
<td>3</td>
<td>Axial</td>
<td>1.5</td>
<td>19</td>
<td>46</td>
</tr>
<tr>
<td>4</td>
<td>Fact</td>
<td>2.0</td>
<td>10</td>
<td>10</td>
</tr>
<tr>
<td>5</td>
<td>Axial</td>
<td>1.2</td>
<td>19</td>
<td>33</td>
</tr>
<tr>
<td>6</td>
<td>Axial</td>
<td>1.5</td>
<td>13.6</td>
<td>33</td>
</tr>
<tr>
<td>7</td>
<td>Center</td>
<td>1.5</td>
<td>19</td>
<td>33</td>
</tr>
<tr>
<td>8</td>
<td>Fact</td>
<td>2.0</td>
<td>28</td>
<td>55</td>
</tr>
<tr>
<td>9</td>
<td>Axial</td>
<td>1.5</td>
<td>19</td>
<td>19</td>
</tr>
<tr>
<td>10</td>
<td>Axial</td>
<td>1.8</td>
<td>19</td>
<td>33</td>
</tr>
<tr>
<td>11</td>
<td>Fact</td>
<td>2.0</td>
<td>10</td>
<td>55</td>
</tr>
<tr>
<td>12</td>
<td>Center</td>
<td>1.5</td>
<td>19</td>
<td>33</td>
</tr>
<tr>
<td>13</td>
<td>Axial</td>
<td>1.5</td>
<td>24.4</td>
<td>33</td>
</tr>
<tr>
<td>14</td>
<td>Center</td>
<td>1.5</td>
<td>19</td>
<td>33</td>
</tr>
<tr>
<td>15</td>
<td>Fact</td>
<td>1.0</td>
<td>10</td>
<td>10</td>
</tr>
<tr>
<td>16</td>
<td>Fact</td>
<td>1.0</td>
<td>10</td>
<td>55</td>
</tr>
<tr>
<td>17</td>
<td>Fact</td>
<td>1.0</td>
<td>28</td>
<td>55</td>
</tr>
</tbody>
</table>
4.4 Results and discussion

The experimental results for the weld bead integrity, the weld bead width and the number of filling passes are listed in Table 4.5. Two factor interaction (2FI) models were statistically selected for the responses: weld bead integrity, weld bead width and the number of filling passes. The measured responses were analysed using the Design-Expert 7.0 software. The weld bead integrity can be calculated according to [9]:

\[
\text{Weld bead integrity} = \left(1 - \frac{\text{area of voids and lack of fusion}}{\text{total weld bead area}}\right) \times 100\%
\]  

(4.4)

The weld bead width was calculated according to the average width of each filling pass. The number of filling passes is the total number of filling passes required to fill the groove.

<table>
<thead>
<tr>
<th>Experiment No.</th>
<th>Point type</th>
<th>Responses</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>Weld bead integrity (%)</td>
</tr>
<tr>
<td>1</td>
<td>Fact</td>
<td>52</td>
</tr>
<tr>
<td>2</td>
<td>Fact</td>
<td>40</td>
</tr>
<tr>
<td>3</td>
<td>Axial</td>
<td>99.7</td>
</tr>
<tr>
<td>4</td>
<td>Fact</td>
<td>99.1</td>
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<tr>
<td>5</td>
<td>Axial</td>
<td>97.9</td>
</tr>
<tr>
<td>6</td>
<td>Axial</td>
<td>99.3</td>
</tr>
<tr>
<td>7</td>
<td>Center</td>
<td>99.5</td>
</tr>
<tr>
<td>8</td>
<td>Fact</td>
<td>98.1</td>
</tr>
<tr>
<td>9</td>
<td>Axial</td>
<td>99.9</td>
</tr>
<tr>
<td>10</td>
<td>Axial</td>
<td>98</td>
</tr>
<tr>
<td>11</td>
<td>Fact</td>
<td>100</td>
</tr>
<tr>
<td>12</td>
<td>Center</td>
<td>98.3</td>
</tr>
<tr>
<td>13</td>
<td>Axial</td>
<td>99.7</td>
</tr>
<tr>
<td>14</td>
<td>Center</td>
<td>98.9</td>
</tr>
<tr>
<td>15</td>
<td>Fact</td>
<td>99.6</td>
</tr>
<tr>
<td>16</td>
<td>Fact</td>
<td>94.1</td>
</tr>
<tr>
<td>17</td>
<td>Fact</td>
<td>99.2</td>
</tr>
</tbody>
</table>

Typical macrographs from welded specimens made with different welding parameters are shown in Figure 4.6. Examples of some typical weld defects are shown, for example, the wire failing to fill the groove completely (Figure 4.6(a)). This can arise due to the wire getting caught in the weld groove, and the interruption to the filling
Chapter 4

process can lead to a large cavity in the groove. In sample 1 (Figure 4.6(a)) this occurred in one of the early filling passes, and subsequent filling passes could not rectify the problem and fill the groove completely, leading to a low weld bead integrity. This also occurred in sample 2, as is evident in Table 4.5. Porosity and lack of side-wall fusion were also evident in some welds, as is highlighted in Figure 4.6(b).

Figure 4.6 Selected macrographs showing the bead shape, bead width, the number of passes and some typical defects. The number on each macrograph indicates the sample number.

4.4.1 Statistical models and analysis

4.4.1.1 Weld bead integrity model

The weld bead integrity is one of the most important factors to consider in assessing the quality of a weld. Good weld bead integrity often ensures that the mechanical properties (i.e. tensile strength, toughness, and fatigue properties) of an ultra-narrow gap laser welded joint are acceptable. In this investigation, the objective of this model was to maximise the weld bead integrity and thereby obtain a sound weld without defects (porosity and lack of fusion).

The ANOVA results for the weld bead integrity 2FI model are given in Table 4.6. An $F$ value of 6.42 from the ANOVA results indicates that the model is significant. Values of prob>F less than 0.05 indicate that the model terms are significant, and vice versa. In this case, B (welding speed), C (wire feed rate) and the interaction of B and C (i.e. BC) are significant process parameters for the weld bead integrity model. The parameter A (laser power) as well as the interactions AB and AC are insignificant process parameters
for the weld bead integrity model. These values represent adequate signals from the model, and they can be used to navigate the whole design space. Model validation measures are given as \( R^2 = 0.7938 \) and adjusted \( R^2 = 0.6701 \). The adequate precision measures the signal-to-noise ratio. The value of 8.773 is greater than 4 and it indicates an adequate signal.

\[
\text{Table 4.6 ANOVA table for weld bead integrity 2FI model.}
\]

<table>
<thead>
<tr>
<th>Source</th>
<th>Sum of squares</th>
<th>df</th>
<th>Mean square</th>
<th>F value</th>
<th>p-value</th>
<th>Prob&gt;F</th>
</tr>
</thead>
<tbody>
<tr>
<td>Model</td>
<td>3979.55</td>
<td>6</td>
<td>663.26</td>
<td>6.42</td>
<td>0.0054</td>
<td>significant</td>
</tr>
<tr>
<td>A-power</td>
<td>6.69</td>
<td>1</td>
<td>6.69</td>
<td>0.065</td>
<td>0.8043</td>
<td></td>
</tr>
<tr>
<td>B-welding speed</td>
<td>1222.78</td>
<td>1</td>
<td>1222.78</td>
<td>11.83</td>
<td>0.0063</td>
<td></td>
</tr>
<tr>
<td>C-wire feed rate</td>
<td>1160.13</td>
<td>1</td>
<td>1160.13</td>
<td>11.22</td>
<td>0.0074</td>
<td></td>
</tr>
<tr>
<td>AB</td>
<td>42.78</td>
<td>1</td>
<td>42.78</td>
<td>0.41</td>
<td>0.5345</td>
<td></td>
</tr>
<tr>
<td>AC</td>
<td>37.41</td>
<td>1</td>
<td>37.41</td>
<td>0.36</td>
<td>0.5608</td>
<td></td>
</tr>
<tr>
<td>BC</td>
<td>1509.75</td>
<td>1</td>
<td>1509.75</td>
<td>14.61</td>
<td>0.0034</td>
<td></td>
</tr>
<tr>
<td>Residual</td>
<td>1033.64</td>
<td>10</td>
<td>103.36</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Lack of fit</td>
<td>1032.92</td>
<td>8</td>
<td>129.11</td>
<td>358.65</td>
<td>0.0028</td>
<td>significant</td>
</tr>
<tr>
<td>Pure error</td>
<td>0.72</td>
<td>2</td>
<td>0.36</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Cor total</td>
<td>5013.18</td>
<td>16</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

From the above analysis, the final mathematical model for the weld bead integrity in terms of coded factors as derived by the design expert software package is as shown below:

\[
\begin{align*}
\text{Weld bead integrity} &= 92.55 - 0.88 \times A - 11.84 \times B + 11.53 \times C - 2.31 \times A \times B + 2.16 \\
& \quad \times A \times C + 13.74 \times B \times C \quad (4.5)
\end{align*}
\]

### 4.4.1.2 Weld bead width model

The width of the weld bead is another important factor in assessing the quality of the weld. It demonstrates the ability of the filler wire to fill the prepared groove without lack of fusion. The wider the weld bead is, the better the mixing of the filler wire and the base material will be, but if the weld bead is too wide then more energy will be required per unit length of weld and the productivity will drop. A suitable width for the weld bead will ensure that the laser energy is used effectively and it will assist in ensuring that a sound welded joint with satisfactory mechanical properties will be obtained. In this investigation, the objective of this model was to achieve a weld bead
width that exceeded the width of the weld groove by a specified amount, thereby producing a sound weld without lack of fusion.

The results of ANOVA are given in Table 4.7. The $F$ value of 4.09 implies that the model is significant. Values of prob>F less than 0.05 indicate that the model terms are significant. In this case $B$ (welding speed) is a significant model term. $A$ (laser power), $C$ (wire feed rate), and the interactions $AB$, $AC$ and $BC$ are insignificant model terms for the weld bead width model. The “lack of fit F value” of 3.41 implies that the lack of fit is not significant relative to the pure error. These values would suggest that we have adequate signals from the model, and that they can be used to navigate the whole design space. Model validation measures are given as $R^2 = 0.7104$ and adjusted $R^2 = 0.5366$. The adequate precision measures the signal-to-noise ratio. The value of 7.185 is greater than 4 and it indicates an adequate signal.

Table 4.7 ANOVA table for the weld bead width 2FI model.

<table>
<thead>
<tr>
<th>Source</th>
<th>Sum of squares</th>
<th>df</th>
<th>Mean square</th>
<th>F value</th>
<th>p-value Prob&gt;F</th>
</tr>
</thead>
<tbody>
<tr>
<td>Model</td>
<td>0.43</td>
<td>6</td>
<td>0.072</td>
<td>4.09</td>
<td>0.0247</td>
</tr>
<tr>
<td>A-power</td>
<td>0.015</td>
<td>1</td>
<td>0.015</td>
<td>0.87</td>
<td>0.3727</td>
</tr>
<tr>
<td>B-welding speed</td>
<td>0.34</td>
<td>1</td>
<td>0.34</td>
<td>19.01</td>
<td>0.0014</td>
</tr>
<tr>
<td>C-wire feed rate</td>
<td>0.040</td>
<td>1</td>
<td>0.040</td>
<td>2.29</td>
<td>0.1609</td>
</tr>
<tr>
<td>AB</td>
<td>2.812E-003</td>
<td>1</td>
<td>2.812E-003</td>
<td>0.16</td>
<td>0.6981</td>
</tr>
<tr>
<td>AC</td>
<td>1.012E-003</td>
<td>1</td>
<td>1.012E-003</td>
<td>0.057</td>
<td>0.8155</td>
</tr>
<tr>
<td>BC</td>
<td>0.038</td>
<td>1</td>
<td>0.038</td>
<td>2.14</td>
<td>0.1739</td>
</tr>
<tr>
<td>Residual</td>
<td>0.18</td>
<td>10</td>
<td>0.018</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Lack of fit</td>
<td>0.16</td>
<td>8</td>
<td>0.021</td>
<td>3.41</td>
<td>0.2468</td>
</tr>
<tr>
<td>Pure error</td>
<td>0.012</td>
<td>2</td>
<td>6.033E-003</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Cor total</td>
<td>0.61</td>
<td>16</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

From the above analysis, the final mathematical model in terms of coded factors as derived by the design expert software package is as given below:

\[
\text{Bead width} = 1.47 + 0.042 \times A - 0.20 \times B + 0.068 \times C + 0.019 \times A \times B - 0.011 \times A \times C - 0.069 \times B \times C
\] (4.6)

4.4.1.3 Number of filling passes model

The number of filling passes embodies the efficiency with which the filler wire fills the groove in ultra-narrow gap laser welding. Experimental results indicate that different
numbers of filling passes are required to fill the ultra-narrow groove completely with different combinations of welding parameters. It is known amongst welding practitioners that the preparation work prior to welding is much more time-consuming than the welding process itself, and also that the interactions among the welding parameters are complex. However, an increase in the number of filling passes that is needed to fill the groove will increase the probability of generating welding defects such as porosity and lack of fusion in the weld. In addition, a higher number of filling passes implies a low filling efficiency for an individual filling pass, which will increase the cumulative heat input to the specimen. More welding passes could also result in a larger butterfly distortion of the weld, which will increase the possibility of the filler wire failing to fill in the groove completely due to problems with access. In this investigation, the objective of this model was to minimise the number of filling passes while achieving a high welding efficiency.

ANOVA of the number of filling passes reduced 2FI model (Table 4.8) gives the results of analysis of variance. The $F$ value of 6.05 implies that the model is significant. Values of $\text{prob}>F$ less than 0.05 indicate that the model terms are significant. In this case, B (welding speed) is a significant model term. A (laser power), C (wire feed rate) and the interaction BC are insignificant model terms for the number of filling passes model. These values suggest that we have adequate signals from the model, and they can be used to navigate the whole design space. Model validation measures are given as $R^2 = 0.7288$ and adjusted $R^2 = 0.6083$. The adequate precision measures the signal-to-noise ratio. The value of 8.567 is greater than 4 and it indicates an adequate signal.

Table 4.8 ANOVA table for the number of filling passes reduced 2FI model.

<table>
<thead>
<tr>
<th>Source</th>
<th>Sum of squares</th>
<th>df</th>
<th>Mean square</th>
<th>F value</th>
<th>p-value Prob&gt;F</th>
</tr>
</thead>
<tbody>
<tr>
<td>Model</td>
<td>3.8</td>
<td>4</td>
<td>0.95</td>
<td>6.05</td>
<td>0.0121 significant</td>
</tr>
<tr>
<td>A-power</td>
<td>0.38</td>
<td>1</td>
<td>0.38</td>
<td>2.42</td>
<td>0.1539</td>
</tr>
<tr>
<td>B-welding speed</td>
<td>0.95</td>
<td>1</td>
<td>0.95</td>
<td>6.03</td>
<td>0.0364</td>
</tr>
<tr>
<td>C-wire feed rate</td>
<td>0.65</td>
<td>1</td>
<td>0.65</td>
<td>4.14</td>
<td>0.0725</td>
</tr>
<tr>
<td>BC</td>
<td>0.16</td>
<td>1</td>
<td>0.16</td>
<td>1.02</td>
<td>0.3382</td>
</tr>
<tr>
<td>Residual</td>
<td>1.41</td>
<td>9</td>
<td>0.16</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Lack of fit</td>
<td>1.41</td>
<td>7</td>
<td>0.20</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Pure error</td>
<td>0</td>
<td>2</td>
<td>0</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Cor total</td>
<td>5.21</td>
<td>13</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
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From the above analysis, the final mathematical model in terms of coded factors as derived by the design expert software package is as given below:

Number of passes = 2.77 − 0.24 × A + 0.54 × B − 0.47 × C + 0.24 × B × C              (4.7)

4.4.2 Diagnostics of the models

The normality of the residual data is normally checked in order to validate the statistical models. The residuals (the deviations between the predicted and actual values of responses) should follow a normal distribution with a mean value of 0. The method of least squares is usually employed in regression analysis. In an ideal scenario, a straight line is produced, which indicates no abnormalities [17, 31]. Figure 4.7 shows the normal plots of residuals for the weld bead integrity, the weld bead width and the number of filling passes. It can be seen that most of the points are located on the straight line or close to the straight line, except for a limited number of individual points, which indicates that the models are effective.

Figure 4.7 Residual plots for each model, (a) weld bead integrity model, (b) weld bead width model and (c) number of filling passes model.
4.4.3 Effects of welding parameters on the responses

4.4.3.1 Effects of welding parameters on the weld bead integrity

The perturbation plot for the effects of welding parameters on the weld bead integrity is shown in Figure 4.8. The welding speed and the laser power have an inverse correlation with the integrity of the weld bead. At higher welding speeds, the weld bead integrity reduces significantly. This may be because a fast welding speed results in a fast cooling rate. Pores may not have sufficient time to escape from the molten pool. In addition, the molten filler wire does not have sufficient time to fill the groove well before it solidifies, which may lead to some lack of fusion in the weld. The laser power has only a slight influence on the weld bead integrity. However, the wire feed rate has a direct correlation and a stronger effect, with an increase in the wire feed rate leading to a marked increase in the weld bead integrity.

![Perturbation plot](image)

Figure 4.8 Perturbation plot showing the effects of all factors on the weld bead integrity, x axis: variables vary at different levels. A is laser power, B is welding speed and C is wire feed rate.

The response surfaces for the weld bead integrity model are shown in Figure 4.9. The relationship between the laser power and welding speed is shown in Figure 4.9(a). It indicates that the weld bead integrity will be maximised with a high laser power and a low welding speed. However, the weld bead integrity drops dramatically with a high laser power and a high welding speed. This may be because the high welding speed results in less filler wire being deposited in the groove during each weld pass, and the deposited metal may shrink significantly because there is less weld metal available to
resist the shrinkage. Shrinkage in the fusion zone makes it progressively more difficult for the filler wire to fill the groove properly in the following welding passes, and this increases the possibility of generating of lack of fusion or porosity. Furthermore, owing to the high welding speed, the molten filler wire will solidify quickly and this can make it difficult for pores to escape from the molten pool, leading to formation of porosity. Figure 4.9(b) presents the relationship between the laser power and the wire feed rate. The integrity of the weld bead can be improved by increasing the laser power and the wire feed rate simultaneously. The higher wire feed rate will enable the molten filler wire to fill the groove completely when employing a higher laser power, which will help in avoiding lack of fusion. In addition, the higher deposition rate for each pass will assist in resisting shrinkage of the weld bead, which will lead to less distortion and make it easier to deposit filler material in the following weld passes, thereby enabling the groove to be filled successfully. The relationship between the welding speed and the wire feed rate is shown in Figure 4.9(c). The weld bead integrity is not strongly dependent on the wire feed rate and the welding speed when high values of these parameters are employed in combination. However, the weld bead integrity drops dramatically when higher welding speeds are used in combination with a low wire feed rate. This may because that the combination of a high welding speed and low wire feed rate reduces the cross sectional area of material that is deposited in each welding pass, which means less material is available to resist the shrinkage in the vicinity of the groove. It then becomes more difficult to add the filler material successfully in later weld passes.
4.4.3.2 Effects of welding parameters on the weld bead width

Figure 4.10 is a perturbation plot that illustrates the effect of welding parameters on the weld bead width. It is evident that both the laser power and the wire feed rate demonstrate a direct correction with the weld bead width. An increase in either the laser power or the wire feed rate at a certain welding speed will lead to corresponding increases in the heat input or the quantity of material that is deposited per unit length of weld, respectively, and both of these will translate to an increase in the bead width. The welding speed, however, has a strong inverse correlation with the bead width. An increase in the welding speed reduces the heat input of the welding process, which generally results in a reduction in the size of the molten pool, and hence a reduction in the bead width.
Figure 4.10 Perturbation plot showing the effects of all factors on the weld bead width, x axis: variables vary at different levels. A is the laser power, B is the welding speed and C is the wire feed rate.

The response surfaces for the weld bead width model are shown in Figure 4.11. From the relationship between the laser power and the welding speed in Figure 4.11(a), it can be deduced that the combination of a high laser power and a low welding speed will increase the weld bead width. Conversely, the choice of a lower laser power and a high welding speed will lead to a dramatic reduction in the weld bead width, because this combination of parameters will lead to a low heat input. The relationship between the laser power and the wire feed rate is shown in Figure 4.11(b). It can be seen that the weld bead width increases with increases in both the laser power and the wire feed rate. As was mentioned earlier, increasing either the heat input or the cross-sectional area of the weld bead will lead to the production of larger weld beads and higher bead widths. Figure 4.11(c) presents the relationship between the welding speed and the wire feed rate. It can be seen that reductions in the welding speed and increases in the wire feed rate also increase the weld bead width. Here the effect of the welding speed can also be explained in terms of its influence on the weld heat input. Conversely, increases in the welding speed and reductions in the wire feed rate will lead to dramatic reductions in the weld bead width.
Figure 4.11 Response surface graphs for the width of the weld bead, (a) laser power-welding speed, (b) laser power-wire feed rate, (c) welding speed-wire feed rate. A is the laser power, B is the welding speed and C is the wire feed rate.

**4.4.3.3 Effects of welding parameters on the number of filling passes**

The perturbation plot for the effects of welding parameters on the number of filling passes is shown in Figure 4.12. It can be seen that the laser power and the wire feed rate are inversely correlated with the number of filling passes, whilst the welding speed has a direct correlation with the number of filling passes. The correlation between laser power and the number of filling passes is weaker than that for the wire feed rate. Consequently, increasing the wire feed rate and decreasing the welding speed would constitute an effective strategy for reducing the number of filling passes that is required to complete the welded joint.
Figure 4.12 Perturbation plot showing the effects of all factors on the number of filling passes, x axis: variables vary at different levels. A is the laser power, B is the welding speed and C is the wire feed rate.

The response surface graph in Figure 4.13 indicates that the number of filling passes drops with increases in the wire feed rate and also with reductions in the welding speed. Increases in the wire feed rate and reductions in the welding speed lead to an increase in the volume of filler material that is deposited per unit length and, consequently, the average height of each filling pass. This will mean that a weld of a given thickness will be completed in fewer passes. Increasing the wire deposition rate will also lead to reduced joint completion times.

Figure 4.13 Response surface graphs for the number of filling passes.
4.4.4 Welding parameters optimisation and validation

4.4.4.1 Multiple-objective optimisation

In order to obtain high quality narrow gap laser welded joints, the welding parameters need to be optimised. There are many welding parameters involved in the ultra-narrow gap laser welding process, and there are interactions between the effects that these welding parameters have on weld quality. That means that taking measures to improve one response may deteriorate another. In such cases, a multiple variable optimisation may be employed. Such an approach can be advantageous because it can identify the working range for each of the variables in order to satisfy the specified optimisation criteria for all of the responses simultaneously. Statistical optimisation is achieved through a dimensionless objective function called “desirability”. The desirability function approach involves transforming each response, \( Y_i \), into a dimensionless value, \( d_i \), between 0 and 1 [9, 24]. A higher \( d_i \) value implies that the corresponding response value is more desirable. The individual desirability of each response can be calculated using Eqs. (4.8) and (4.9) [9, 32]. The shape of the desirability function can be changed for each goal through the weight field \( w_{ti} \). Weights are used to give more emphasis to the upper/lower bounds or to emphasise the target value. Importance varies from the least important value of 1, indicated by (+), to the most important value of 5, indicated by (+++++) [32].

For the goal of maximum, the desirability can be defined by

\[
d_i = \begin{cases} 
0, & Y_i \leq \text{Low}_i \\
\left( \frac{Y_i - \text{Low}_i}{\text{High}_i - \text{Low}_i} \right)^{w_{ti}}, & \text{Low}_i < Y_i < \text{High}_i \\
1, & Y_i \geq \text{High}_i
\end{cases} \quad (4.8)
\]

For the goal of minimum, the desirability can be defined by

\[
d_i = \begin{cases} 
1, & Y_i \leq \text{Low}_i \\
\left( \frac{\text{High}_i - Y_i}{\text{High}_i - \text{Low}_i} \right)^{w_{ti}}, & \text{Low}_i < Y_i < \text{High}_i \\
0, & Y_i \geq \text{High}_i
\end{cases} \quad (4.9)
\]
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The final goal of the optimisation is to find a good set of parameters that satisfy all of the goals [9, 32]. Each criterion (maximisation or minimisation) has a specific importance factor according to its influence on the final weld bead quality [9, 29]. This value is specified for each optimisation criterion. In order to satisfy multi-objective optimisation, the desirability function is defined by the geometric mean of all individual desirabilities that range from 0 for the least desirable settings to 1 for the most desirable process settings [9, 24]. The function is defined as [9, 24, 32]

\[
\delta = \left( \prod_{i=1}^{n} d_i \right)^{\frac{1}{n}}
\]

(10)

This equation represents the overall desirability function, where \( \delta \) is the overall desirability, \( n \) is the number of responses and \( d_i \) is the \( i \)th response desirability value. In this work, both numerical and graphical optimisation approaches were used by selecting the desired goals for each factor and response.

4.4.4.2 Numerical and graphical optimisation

The numerical optimisation approach involved combining the final goals into an overall desirability function. However, each optimisation criterion has a different effect on the final weld bead quality. In such cases, each optimisation criterion can be assigned a different weight according to the degree of influence it has in the process optimisation [9, 29]. In this investigation, the numerical multiple-response optimisation criterion is to obtain maximum weld bead integrity, maximum weld bead width in order to improve the weld quality, and the minimum the number of filling passes in order to improve the welding efficiency. The weld bead integrity model was given the highest weight value of 5 (++++) due to its importance on the final quality of the weld and the mechanical properties of the welded joint. The weld bead width is also important to the mechanical properties of the welded joint, so it was given the second highest weight value of 4
The number of filling passes was given the weight value of 3 (+++). This criterion was deemed to be of lower importance than obtaining a sound weld.

According to the above optimisation procedure, the details of the most desirable solution and the response values are presented in Table 4.9. In order to facilitate comparison with arc welding processes in the future, the units for the welding speed and the wire feed rate were converted from mm/s to m/min. The total desirability value is 90.5%. The weld bead integrity is 100%. The weld bead width is 1.82 mm and the number of filling passes is 1. In addition, the desirability shape function (wire feed rate: 55 mm/s (3.3 m/min)) for the ultra-narrow gap laser welding process is shown in Figure 4.14. The desirability only has one peak region, and it is obtained by increasing the laser power and decreasing the welding speed.

Table 4.9 Multiple-objective optimisation results.

<table>
<thead>
<tr>
<th>Laser power (kW)</th>
<th>Welding speed (mm/s)</th>
<th>Wire feed rate (mm/s)</th>
<th>Weld bead integrity (%)</th>
<th>Weld bead width (mm)</th>
<th>No. of filling passes</th>
<th>Desirability (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>2</td>
<td>10 (0.6 m/min)*</td>
<td>55 (3.3 m/min)*</td>
<td>100</td>
<td>1.82</td>
<td>1</td>
<td>90.5</td>
</tr>
</tbody>
</table>

Note: *denotes converted unit of welding speed and wire feed rate to m/min.

Figure 4.14 Desirability shape function for ultra-narrow gap laser welding (wire feed rate 55 mm/s (3.3 m/min)).
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Only limited sets of optimised welding parameters can be obtained in the numerical optimisation process, so it is not convenient for application in industry. However, in the graphical optimisation process, it is possible to define regions in which the process parameters simultaneously satisfy the different optimisation criteria. The critical responses can be overlayed on a single contour plot, and such a plot is referred to as an overlay plot. The graphical optimisation displays the region of feasible response values in the factor space. The results of the graphical optimisation are the overlay plots, which are quite convenient for rapid application in industry when choosing the values of the welding parameters [9, 29]. The maximum and minimum limits for the optimisation constraints for the graphical optimisation were specified according to the numerical optimisation results: i.e. integrity of the weld bead $\geq 95\%$, width of the weld bead $\geq 1.5$ and number of filling passes $\leq 2$. The overlay plot for all the models according to the optimisation constraints is shown in Figure 4.15. The yellow area highlights the region comprising optimum conditions, for which all criteria are satisfied simultaneously. According to Figure 4.14, a higher laser power and a lower welding speed provide the optimum conditions for each individual constraint. The intersection area for all constraints is consistent with the numerical optimisation results in Figure 4.14.

![Overlay plot]

Figure 4.15 Optimum welding parameters overlay plot considering all constraints for all variables.
4.4.4.3 Validation of the optimised welding parameters

New sets of welding experiments based on the optimised welding parameters, as obtained from the maximum desirability solution in Table 4.9, were carried out to evaluate the reliability of the modelling work. The 8 mm thick S960 steel was machined with the same width of groove (1.2 mm) for ultra-narrow gap laser welding. Fig. 18 shows the weld cross sections for the 6 mm thick and 8 mm thick S960 steel welds. One filling pass was required for ultra-narrow gap laser welding of 6 mm thick S960 steel and two filling passes were required for ultra-narrow gap laser welding of 8 mm thick S960 steel. There was no evidence of lack of fusion or porosity.

![Figure 4.16](image_url) Cross-sections of ultra-narrow gap laser welds in S960 steel, (a) 6 mm thick cross section, (b) 8 mm thick cross section (laser power: 2 kW, welding speed: 0.6 m/min, wire feed rate: 3.3 m/min).

It was found that melt pool sagging was one of the characteristics of single pass autogenous laser welding of 13 mm thick S700 steel in the flat (1G) position, as shown in Figure 4.17. To solve this melt sagging problem, an attempt was made to transfer the above mentioned optimised welding parameters (laser power: 2 kW, welding speed: 0.6 m/min, wire feed rate: 3.3 m/min) to the welding of 13 mm thick S700 steel plates using the narrow gap laser welding (NGLW) technique.
Figure 4.17 Typical melt sagging defects for single pass autogenous laser welding of 13 mm thick S700 steel in the flat (1G) position, (a) top weld appearance, (b) weld cross section, (c) underside appearance.

Some lack of fusion was found between the two filling passes when using the above optimised welding parameters, as can be seen in Figure 4.18(a). This may because the input energy was absorbed primarily by the side walls of the groove. The laser power was then increased to 3 kW, with the welding speed and wire feed rate unchanged, to weld the 13 mm thick S700 steel plates. Figure 4.18(b) shows the weld cross sections for the 13 mm thick S700 steel weld made with a laser power of 3 kW. Three filling passes were required for ultra-narrow gap laser welding of 13 mm thick S700 steel plates. There was no evidence of lack of fusion or porosity. In addition, the melt sagging problem was resolved by using this multi-pass ultra-narrow gap laser welding technique to weld this 13 mm thick S700 steel.
4.4.4.4 Tensile properties

Tensile testing was carried out on the as-received 8 mm thick S960 and 13 mm thick S700 base materials and the optimised narrow gap laser welded samples. Typical engineering stress–strain plots for the base materials and the narrow gap laser welded (NGLW) steel joints are shown in Figure 4.19. It can be seen that both of the NGLW steel joints present almost the same tensile properties as the base material. The tensile test results are summarised in Table 4.10, which lists average values and standard deviations. The yield strength (YS), ultimate tensile strength (UTS) and apparent elongation for the NGLW S960 steel samples were found to be 1019 MPa, 1040 MPa and 8.3%, respectively. The YS, UTS and apparent elongation of the NGLW S960 steel samples were observed to be the same as those for the base S960 steel, which had a YS of 1026 MPa, a UTS of 1035 MPa and an apparent elongation of 8.5%. The YS, UTS and apparent elongation for the NGLW S700 steel samples were found to be 714 MPa, 748 MPa and 12.8%, respectively. The YS, UTS and apparent elongation of the base S700 steel were 724 MPa, 763 MPa and 12.9%, respectively. All tensile failures for the welded specimens were in the base material well away from the weld, as is shown in
Figure 4.20. This means that the optimised NGLW specimens retained good tensile properties. It should be noted, however, that the elongations measured in this work do not capture the behaviour of the weld region itself very effectively, since the elongations were averaged across the weld zone and the parent material. While these preliminary results are encouraging, further work would be required in order to obtain a more detailed assessment of the mechanical properties across the welded joints.

Figure 4.19 Representative engineering stress versus engineering strain curves for base material and the ultra-NGLW joints, (a) S960 steel, (b) S700 steel.

Table 4.10 Tensile properties for the base material and the ultra-NGLW specimens.

<table>
<thead>
<tr>
<th>Test specimen</th>
<th>Yield strength (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Base S960 (8 mm)</td>
<td>1026 ± 7</td>
<td>1035 ± 11</td>
<td>8.5 ± 0.3</td>
</tr>
<tr>
<td>NGLW S960 (8mm)</td>
<td>1019 ± 8</td>
<td>1040 ± 11</td>
<td>8.3 ± 0.3</td>
</tr>
<tr>
<td>Base S700 (13 mm)</td>
<td>724 ± 7</td>
<td>763 ± 5</td>
<td>12.9 ± 0.8</td>
</tr>
<tr>
<td>NGLW S700 (13 mm)</td>
<td>714 ± 5</td>
<td>748 ± 5</td>
<td>12.8 ± 0.4</td>
</tr>
</tbody>
</table>
The following conclusions can be drawn from this investigation:

(1) Ultra-narrow gap laser welding was successfully carried out on 6 and 8 mm thick S960 and 13 mm thick S700 high strength steel plates using a relatively moderate laser power (2-3 kW) and a narrow (1.2-1.4 mm) parallel groove.

(2) The multi-pass ultra-narrow gap laser welding technique provides an alternative solution to solving the melt sagging problem when welding medium and thick section materials.

(3) From the predictions of the statistical model, it was found that the weld quality and the welding efficiency can be improved through increasing the laser power and the wire feed rate, and reducing the welding speed.

(4) The optimised welding parameters were successfully transferred to the welding of 8 mm thick S960 steel with two filling passes, and to the welding of 13 mm thick S700 steel with three filling passes.

(5) The tensile properties of the optimised narrow gap laser welded specimens matched those of the base material, with all welded specimens failing in the BM.
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Chapter 5: A comparative study of microstructure and mechanical properties of ultra-narrow gap laser and GMA welded S960 high strength steel

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Abstract

The microstructural characteristics, mechanical properties, including micro-hardness, tensile properties, three-point bend test and Charpy impact at different test temperatures of 8 mm thick S960 high strength steel plates were investigated following their joining by multi-pass ultra-narrow gap laser welding (NGLW) and gas metal arc welding (GMAW) techniques. It was found that the microstructure in the fusion zone (FZ) for the ultra-NGLW joint was predominantly martensite mixed with some tempered martensite, while the FZ for the GMAW joint was mainly consisted of very fine acicular ferrite with small amount of martensite. The strength of the ultra-NGLW specimens was comparable to that of the base material (BM), with all welded specimens failed in the BM in the tensile tests. The tensile strength of the GMAW specimens was reduced approximately by 100 MPa compared with the base material by a broad and soft HAZ with failure located in the soft HAZ. Both the ultra-NGLW and GMAW specimens performed well in three-point bending tests. The GMAW joints exhibited better Charpy impact toughness than the ultra-NGLW joints.

\textbf{Keywords:} as welded condition, heat input, cooling rate, toughness, hardness, martensite
Chapter 5

5.1 Introduction

With the development of the thermo-mechanical rolling, quenching and tempering process technique, high strength low alloy (HSLA) steels have been developed for many years and achieve high tensile properties and good toughness [1]. HSLA steels are classified as low carbon steels and achieve their high strength levels from low percentage alloying additions that result in precipitation hardening, or grain refinement strengthening mechanisms through thermo-mechanical processing [2]. In general, the microstructures of high-strength steel are martensite and/or bainite. Microalloying elements in high strength low alloy (HSLA) steels are used mainly for precipitation hardening and grain size refinement [3]. To achieve a combination of high strength and toughness, the microstructure of lower bainite or ferrite plus martensite has been designed for HSLA steels [4, 5]. HSLA steels are used in a wide variety of applications as structural components, pressure vessel and linepipes, and in shipbuilding, offshore construction, automotive, industrial vehicles, lifting and cranes. The application of HSLA steels for such components enables lighter and more slender products and reduces construction costs without loss of structural integrity [6, 7]. Hulka et al. [8] reported that under tensile stresses the plate thickness can be reduced by around 60% by using the steel grade S960 instead of S355.

In the manufacturing of structural components, welding is an indispensable manufacturing technique for joining the small components into the final complex products and the application of welding technique can mitigate the complexity of manufacturing large structures [9]. Conventional arc welding are the main welding technique applied in industry to join the high strength steels. However, these traditional arc welding procedures have relatively high heat inputs which result in low cooling rates of the weldment, and this can produce softening in the heat affected zone (HAZ) [10], giving low strength of the weldment [11, 12].

Compared to traditional arc welding techniques, laser welding has many advantages, such as high power density, deep penetration depth, which generates a narrower heat-affected zone. In addition, thermal distortions and residual stress can be reduced since the lower heat input applied per unit length [13, 14]. Furthermore, the capability for optical fiber beam delivery and the use of industrial robots makes laser welding process easily-automated in manufacturing industry [15]. Fiber lasers have the characteristics of
small beam divergence, low maintenance costs, high efficiency, high precision and reliability, and compact size, and consequently they have attracted more attention over the last decade for cutting, welding and cladding applications [16-18].

Single pass autogenous laser welding without filling any other material has been widely applied in industry [19]. However, single pass autogenous laser welding technique has many disadvantages, including the precise fit-up requirements prior to welding and the limited weld range due to limited maximum laser power available for commercial lasers (typically less than 20 kW). In addition, thick section laser welding with a single pass approach often results in porosity, molten material dropout, cracking, and mis-tracking of weld seams [20-22].

Multi-pass laser welding with filler wire is a feasible technique to weld medium and thick section material using the limited laser power at hand. In addition, the introduction of a filler wire into the welding groove can improve the metallurgy and the properties of the weld. Joining of thick plates is accomplished by using filler materials to fill the prior prepared welding groove [23]. Usually a “Vee” groove is prepared. However, the thicker the plate, the wider the groove is, which results in lower productivity and more consumption materials. Over the last few years, the multi-pass narrow gap laser welding technique has been demonstrated. Multi-pass narrow gap laser welding technique can be applied to weld medium and thick section materials with a filler wire using relatively moderate laser powers. This technique has many advantages, including increase productivity, saving consumption filler materials, releasing precise fit-up requirements, improving the metallurgical properties of the welds, improving the melt sagging problem, and reducing residual stress and distortion [24-26].

Elmesalamy et al. [25] successfully welded 20 mm thick 316L stainless steel plate using a 1 kW IPG single mode fibre laser with an ultra-narrow gap (1.5 mm gap width) and a taper of 6° from double sides using multiple-pass narrow gap laser welding approach. Shi et al. [27] carried out multi-pass narrow gap laser welding technique to weld a 20 mm thick AH32 high strength ship steel with a gap width of 3 mm and taper of 5° and 10°. However, some porosity and lack of fusion defects still existed in the weld beads.
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It can be found from the above reviews on the narrow gap laser welding technique, most of these investigations are just focused on the welding technique and welding parameters optimisations. Studies on the microstructures evolution and mechanical properties of the narrow gap laser welded materials are scarce.

Tata Steel has developed an advanced low carbon, low alloy S960 steel with a minimum yield strength of 960 MPa. This steel is a promising high strength steel for application in the heavy crane sectors due to its high specific strength, good impact toughness and formability. Ruukki Metals [28] are also developing Optim 960 QC steel with strength of 960 MPa. The Optim 960 QC strip steels are made by modern hot strip rolling and direct-quenching processes, which is similar to the method employed by Tata Steel.

So far, there is a knowledge gap on the microstructure and mechanical properties of narrow gap laser welded S960 HSLA steel. In the present work, a comparative study on ultra-narrow gap laser welding (NGLW) and traditional gas metal arc welding (GMAW) was carried out on the S960 HSLA steel. A very narrow parallel groove (width was 1.2 mm) was prepared to do the ultra-narrow gap laser welding experiment using a moderate laser power (2 kW). The microstructures and mechanical properties, such as tensile strength, microhardness, Charpy impact toughness and bend performance of the welded components were investigated and compared for the above two welding techniques in the as-welded condition.

5.2 Material and experimental procedures

The base material used in this study is a newly developed S960 high strength low alloy high strength (HSLA) steel by Tata Steel. The base material is hot rolled strips, which had been rapidly water cooled to low temperature. The dimensions (length × width × thickness) of the base material were 200 mm × 100 mm × 8 mm. After the welding the samples were measured as approximately 200 mm × 200 mm × 8 mm. The chemical composition and the scanning electron microscope (SEM) micrograph of the base material (BM) are shown in Table 5.1 and Figure 5.1, respectively. The carbon equivalent (CE) of the investigated steel is calculated according to the following equation [17]:

\[
CE = C + \frac{Mn}{6} + \frac{Cr + Mo + V}{5} + \frac{Cu}{15}
\]
which is recommended by International Institute of Welding (IIW). All concentrations are defined in weight percent. The CE of S960 steel is 0.496, as seen in Table 5.1. The Ms (martensite start) temperature is high (∼ 450 °C), and so the martensite will self-temper immediately after transformation. Bainite and tempered martensite are distinguished based on the orientation of the carbides. Lower bainite has small carbides that are parallel to each other, upper bainite has larger carbides which are between the laths of bainitic ferrite, and tempered martensite has carbides which are orientated along certain crystallographic planes (so not parallel) [17]. The microstructure of the BM is a mixture of tempered martensite and bainite, as shown in Figure 5.1. The filler wire used in this study is matching Union X96 (ER120S-G). The chemical composition and mechanical properties of the solid filler material Union X96 are depicted in Table 5.2 and Table 5.3, respectively. The CE of the Union X96 filler wire is 0.79. The filler wire diameter is 0.8 mm.

Table 5.1 Chemical composition of S960 steel (wt. %).

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>N</th>
<th>B</th>
<th>Ca</th>
<th>Si</th>
<th>Al</th>
<th>Cu</th>
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</thead>
<tbody>
<tr>
<td></td>
<td>0.088</td>
<td>1.51</td>
<td>0.008</td>
<td>0.002</td>
<td>0.0089</td>
<td>0.012</td>
<td>0.055</td>
<td>0.033</td>
<td>0.014</td>
<td></td>
</tr>
<tr>
<td>Sn</td>
<td>0.001</td>
<td>0.472</td>
<td>0.023</td>
<td>0.248</td>
<td>0.025</td>
<td>0.040</td>
<td>0.050</td>
<td>Bal.</td>
<td>0.496</td>
<td></td>
</tr>
</tbody>
</table>

Figure 5.1 SEM micrograph of the base material.
Table 5.2 Chemical composition of the Union X96 filler wire (wt.%).

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
<th>Mo</th>
<th>Ni</th>
<th>Fe</th>
<th>CE</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.12</td>
<td>0.80</td>
<td>1.90</td>
<td>0.45</td>
<td>0.55</td>
<td>2.35</td>
<td>Bal.</td>
<td>0.79</td>
</tr>
</tbody>
</table>

Table 5.3 The mechanical properties of the Union X96 filler wire.

<table>
<thead>
<tr>
<th>Material</th>
<th>Yield strength 0.2% (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
<th>Impact value (-50 °C) (J)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Union X96</td>
<td>930</td>
<td>980</td>
<td>14</td>
<td>50</td>
</tr>
</tbody>
</table>

To weld the 8 mm thick S960 steel, a butt joint with a parallel groove configuration (groove width was 1.2 mm) was prepared for multi-pass ultra-narrow gap laser welding (NGLW), as shown in Figure 5.2(a). To obtain full penetration of the GMA weld in the 8 mm thick S960 steel, a butt joint with a V-groove configuration with a 1 mm thick root face and 1 mm gap and 60° groove angle was prepared, as shown in Figure 5.2(b). Because the conventional V-groove is generally applied in practice, the V-groove geometry was chosen for the GMAW in this study.

Prior to the welding experiment, the vicinity of the weld track on the base material was sand blasted to clean the surface oxide. The removal of the oxide can also make the groove surfaces rougher, which can have an effect on improving the absorption and reflections of the laser beam especially when using a 1 µm wavelength laser [28]. Acetone was used to clean the surface after sand blasting, and then the materials were clamped on the work table to ensure adequate restraint. Both the ultra-NGLW and GMAW were carried out perpendicular to the rolling direction of the base material.
A Miller Axcess 450 gas metal arc welding (GMAW) power source was used for the manual GMA welding experiments. To protect the molten weld pool and obtain good penetration depth during welding process, the specimens were shielded using a mixture of 80% argon and 20% CO\textsubscript{2} with gas flow rate of 22 l/min. A continuous wave fibre laser (IPG YLS-16000) equipment with a maximum output power of 16 kW was used for the ultra-NGLW experiments. The laser beam has an emission wavelength of 1070 nm. The beam parameter product of the laser was 10 mm mrad and the beam was optically delivered (with a fibre core diameter of 300 μm) to the output lenses. The laser beam emitted from the end of the optical fibre was collimated by a lens with a 150 mm focal length and then focused onto the specimen surface using a lens with a 400 mm focal length. The measured focus beam size and Rayleigh length were 0.8 mm and 15 mm, respectively. The laser head was mounted on a 6-axis KUKA robot. The filler wire was fed into the leading edge of the molten pool at an angle of 30° with respect to the specimen surface. Pure argon gas was blown onto the top weld pool surface through a copper tube with a flow rate of 12 l/min, and the back shielding gas was blown through a side tube to a shielding chamber under the specimens to protect the back surface of the specimens at a flow rate of 8 l/min. The schematic of the ultra-NGLW process setup is shown in Figure 5.3. The robot end-effector holds the laser head perpendicular to the workpiece and has the capability of moving in 3D work space.

![Figure 5.3 Schematic representation of ultra-narrow gap laser welding setup.](image)

After the welding, the welded specimens were sectioned transverse to the welding direction, then mounted ground, polished, and etched with a 2% Nital solution. The macrostructure and microstructure of the welded joints investigation were performed.
using a KEYENCE VHX-500F optical microscope and Philips XL 30 scanning electron microscope (SEM).

Microindentation hardness mapping profiles across the welded joints were measured using a 300 g test load and a dwell period of 15 s with a Vickers microhardness machine (Struers DuraScan 50) having a fully automated testing cycle. For the NGLW specimen, because of the ultra-narrow fusion zone and heat affected zone (HAZ), an indent interval of 0.2 mm was set in the area close to the fusion zone and HAZ and the indent interval for other areas further away from the fusion zone and HAZ was set as 0.4 mm. Whilst, because of the wide fusion zone and HAZ for the GMA welded specimen, the indent interval was set as 0.4 mm across the fusion zone.

Tensile test specimens for the as-received base material and the welded samples were prepared according to the ASTM E8M-04. The gauge length for the tensile test was 50 mm. Three-point bending test samples were sliced in accordance with the BS EN ISO 15614-1. Both face and root bending tests were carried out with two test pieces for each type of test. Bend testing was executed until an angle of 180° was reached according to BS EN ISO 5173: 2010+A1:2011. Sub-size Charpy impact test samples (8 mm thick) were prepared following BS EN 10045-1:1990. Three replicates for tensile and Charpy impact testing were prepared to reduce experimental uncertainties. The mechanical properties test samples were sliced along the rolling direction of the material using wire electric discharge machining (EDM), and the configurations for each are shown in Figure 5.4. The weld reinforcements in the face and root regions were removed by manual grinding before the mechanical properties tests were carried out. The notches for the Charpy impact test coupons were located in the fusion zone, on the fusion line (FL) and in the HAZ, respectively to examine the impact toughness of the fusion zone (FZ), fusion line (FL) and HAZ for the welded samples. In order to fix the right position of Charpy V notch (CVN), the impact specimens were macro etched with 2% Nital solution to exhibit the outline of fusion line and HAZ. For the ultra-NGLW samples with notches located in the HAZ were sampled as FL + 0.5 mm. Because the fusion zone for GMA welded specimens is a ‘V’ shaped area, there is an angle (~ 60º) between the fusion line and the bottom surface of the specimens. The usual procedure for the FL notch is to pass through the HAZ (real FL) at the mid-thickness point of the material (sample 50% weld metal and 50% HAZ) to delegate the nominated FL, while the HAZ
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notch is to sample the notch at FL + 2 mm. The CVN position was located at the FL and HAZ for GMA welded samples schematically shown in Figure 5.5. Tensile tests were performed at room temperature using an Instron model 8500 electronic universal test machine. Three-point bending tests were carried out on a Dartec bending test machine at room temperature. The diameter of the former and the gap distance for the three-point bending test are shown in Figure 5.6. The Charpy impact tests were carried out on a Zwick Roell Charpy impact test machine at -40 °C, -20 °C, 0 °C and at room temperature, respectively. Each test sample was kept at the related test temperature for half an hour to ensure that the temperature in each specimen was uniform. Following the tensile and Charpy impact tests, all of the fracture surfaces of the tested specimens were examined using Zeiss EVO 50 SEM installed with an Energy Dispersive X-ray Detector (EDX) to identify the fracture morphology and the fracture mechanisms.

Figure 5.4 Dimensions and configurations of specimens for mechanical property testing, (a) tensile test sample, (b) Charpy impact test sample, (c) three-point bending test sample.

Figure 5.5 Schematic representation of the sample with notch on the FL and HAZ for GMA welded samples.
5.3 Results

5.3.1 Welding parameters

There are three main variable welding parameters (*i.e.* laser power, welding speed and wire feed rate) for ultra-NGLW process, statistical modeling was applied to optimise the welding parameters. The detailed optimising procedure was demonstrated in [29]. The welding parameters for both manual GMAW and ultra-NGLW of S960 steel are shown in Table 5.4 and Table 5.5, respectively. The focal position was set on the root surface to obtain full penetration for the root pass with a focused beam. The focal position was set on the top surface of the specimens for the filling passes to melt the filler wire and cover the groove in order to avoid lack of fusion on the side wall. From previous experience with single pass autogenous laser welding of 8 mm thick S960 steel, the optimum laser power was 6.5 kW [17]. In this investigation, moderate power ultra-NGLW was carried out using laser power of 2 kW, which will have a benefit to save capital investment. Heat input for arc welding can be calculated by the following equation: heat input (J/mm) = efficiency × (voltage (V) × current (A))/welding speed (mm/s). The consumable electrode process, such as GMAW, exhibits an average arc efficiency of 80% [30]. Heat input for laser welding is calculated by the equation: heat input (J/mm) = efficiency × laser power (W)/welding speed (mm/s). The welding efficiency considering the conduction heat loss only is about 80% for autogenous laser welding [31, 32]. In addition, some beam reflection from the filler wire takes place in
the ultra-NGLW process, which could result in 30% loss of energy applied into the workpieces [33].

The comparison of energy and consumed filler material between ultra-NGLW and GMAW techniques are summarised in Table 5.6. It can be found that the maximum power used for GMAW is 4.7 kW, while ultra-NGLW is 2 kW. The heat input for each welding pass for GMAW is approximately 5-8 times that of ultra-NGLW. The cumulative heat input (the heat input summation of all welding passes) for GMAW is ~6 times that of ultra-NGLW. The welding speed for ultra-NGLW is ~1.5 times that of GMAW. The consumed filler material for GMAW is ~6 times that of ultra-NGLW.

Table 5.4 Optimised GMAW parameters for 8 mm thick S960 steel.

<table>
<thead>
<tr>
<th>Welding pass</th>
<th>Voltage (V)</th>
<th>Current (A)</th>
<th>Welding speed (m/min)</th>
<th>Wire feeding rate (m/min)</th>
<th>Shielding gas flow (l/min)</th>
<th>Heat input (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>First</td>
<td>27</td>
<td>175</td>
<td>0.4</td>
<td>4.0</td>
<td>22</td>
<td>0.57</td>
</tr>
<tr>
<td>Second</td>
<td>27</td>
<td>165</td>
<td>0.46</td>
<td>4.0</td>
<td>22</td>
<td>0.46</td>
</tr>
<tr>
<td>Third (torch weave)</td>
<td>27</td>
<td>168</td>
<td>0.26</td>
<td>4.0</td>
<td>22</td>
<td>0.84</td>
</tr>
</tbody>
</table>

Table 5.5 Optimised ultra-NGLW parameters for 8 mm thick S960 steel.

<table>
<thead>
<tr>
<th>Welding pass</th>
<th>Power (kW)</th>
<th>Welding speed (m/min)</th>
<th>Wire feeding rate (m/min)</th>
<th>Top shielding gas flow (l/min)</th>
<th>Back shielding gas flow (l/min)</th>
<th>Focal position (mm)</th>
<th>Heat input (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Root pass</td>
<td>2</td>
<td>1.2</td>
<td>-</td>
<td>12</td>
<td>8</td>
<td>-6</td>
<td>0.08</td>
</tr>
<tr>
<td>Filling pass</td>
<td>2</td>
<td>0.6</td>
<td>3.3</td>
<td>12</td>
<td>8</td>
<td>0</td>
<td>0.11</td>
</tr>
<tr>
<td>Filling pass</td>
<td>2</td>
<td>0.6</td>
<td>3.3</td>
<td>12</td>
<td>8</td>
<td>0</td>
<td>0.11</td>
</tr>
</tbody>
</table>

Table 5.6 Comparison of energy and consumed filler material between ultra-NGLW and GMAW.

<table>
<thead>
<tr>
<th>Welding technique</th>
<th>Maximum power used (kW)</th>
<th>No. of welding passes</th>
<th>Welding speed (m/min)</th>
<th>Consumed filler material (g)</th>
<th>Cumulative heat input (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>GMAW</td>
<td>4.7</td>
<td>3</td>
<td>0.4</td>
<td>89</td>
<td>1.87</td>
</tr>
<tr>
<td>Ultra-NGLW</td>
<td>2</td>
<td>3</td>
<td>0.6</td>
<td>16</td>
<td>0.3</td>
</tr>
</tbody>
</table>

5.3.2 Macrostructure

It is seen that the cross-section of the welds (Figure 5.7(a) and (b)) exhibited a significant microstructural change in the fusion zone (FZ) and heat affected zone (HAZ) for both the GMAW and ultra-NGLW joints. Both the welded joints are full penetration,
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and there are no macroscopic defects such as cracks, incomplete fusion, and porosity. The GMAW specimen presents a ‘V’ shaped fusion zone and ultra-NGLW specimen shows a nearly rectangle fusion zone. Three welding passes can be distinguished clearly for both the GMAW and ultra-NGLW joints. The various regions, i.e., base material (BM), fusion zone (FZ) and heat affected zone (HAZ) are visible. The top width of fusion zone and the width of HAZ for the GMAW specimen were approximate 12 mm and 3.5 mm, respectively. Whilst, the widths of the FZ and HAZ for the ultra-NGLW specimen were approximate 1.4 mm and 1 mm, respectively. Even at a relatively low magnification, evidence of coarse columnar grains which nucleated epitaxially from the fusion line and grew toward the weld centreline can be observed. During the fusion zone solidification, grains tend to grow in the direction of the maximum heat extraction.

Clearly, the width of the fusion zone for the GMAW specimen is largely determined by the choice of the weld groove geometry. While it is possible to argue that a more appropriate comparison would have involved a narrow-groove GMAW weld groove geometry, in place of a conventional Vee-groove, such a geometry would generally not be applied in practice, so the Vee-groove geometry was chosen on the basis that it is likely to be most representative of industrial practice.

![Figure 5.7 Cross-section profiles the welded joints, (a) GMAW, (b) ultra-NGLW.](image)

5.3.3 Microstructure

Because different locations of the HAZ suffered from different thermal cycles, especially different peak temperatures, during the welding process, the grain size within the HAZ varied accordingly. In order of increasing peak temperature, the HAZ can be
subdivided into the sub-critical HAZ (SCHAZ), the intercritical or partially austenitised HAZ (ICHAZ), the fine grained HAZ (FGHAZ), and finally the coarse grained HAZ (CGHAZ) which is beside the fusion zone itself.

Both the GMAW and ultra-NGLW processes involve multi-welding passes to fill up the joints. The weld beads covered by the subsequent passes introduce new thermal cycle to the initial generated FZ and HAZ, which will have a significant effect on the final microstructures.

The characteristics of the microstructures within each of the different sub-zones for the ultra-NGLW and GMAW joints were examined using an SEM. The microstructure of the upper part and lower part of the welded joints were examined individually for both the ultra-NGLW and GMAW joints. The weld bead in the upper part was formed in the last filling pass, so there was no more additional heat effect added in this part. However, the lower part endured some more additional thermal cycles from the subsequent filling passes. In addition, a reheated CGHAZ was generated in the GMAW joint. This sub-zone is located in the middle of the wide HAZ and surrounds the ‘V’ shaped FZ, as indicated in Figure 5.7(a). The microstructures of each different sub-zone for the upper and lower part of the ultra-NGLW sample are presented in Figure 5.8 and Figure 5.9, respectively. In the upper part, the SCHAZ retains the microstructure of the BM, consisting bainite and auto-tempered martensite. The ICHAZ consists of an over tempered BM microstructure and regions with partially transformed high carbon martensite and auto-tempered martensite. In the FGHAZ, the microstructure is mixed with equiaxed martensite and auto-tempered martensite, with a small prior-austenite grain size. The microstructure in the CGHAZ is a mixture of equiaxed martensite and auto-tempered martensite with a large prior-austenite grain size. Elongated martensite is predominated in the FZ with a small amount of auto-tempered martensite, and the prior-austenite grain shows a large columnar grain structure.
The microstructure in the lower part of SCHAZ, ICHAZ, FGHAZ, CGHAZ and FZ shows a similar microstructure to the corresponding sub-zones in the upper part of the ultra-NGLW joint, as shown in Figure 5.9. However, the microstructure of the sub-zones in the lower part presents more tempered martensite. In order to avoid duplicating the information in Figure 5.8(a), the microstructure in the lower part of the SCHAZ is not presented in Figure 5.9.
Both the upper and lower part of the GMAW sample were examined using SEM, the microstructures in the corresponding sub-zones for the upper and lower part did not show a big difference. Representative SEM micrographs illustrating microstructures of each sub-zone in the GMAW specimen are shown in Figure 5.10. The microstructure in the SCHAZ is similar to the base material, comprising bainite and auto-tempered martensite. In the ICHAZ, some partially transformed martensite and auto-tempered martensite were found, and the other regions retained the BM microstructure. The microstructure in the FGHAZ is predominated of equiaxed bainite with a mixture of small amount of martensite and tempered martensite, and with a small prior-austenite grain size. The microstructure in the CGHAZ is the same as that in the FGHAZ, but with a large prior-austenite grain size. In the reheated CGHAZ, some partially transformed martensite was observed encircled around the prior austenite grain boundary and the generated bainite in the CGHAZ during the prior welding pass was retained inside of the partially transformed martensite circle, as shown in Figure 5.10(e).
A lot of tempered martensite were found in the circular partially transformed sub-region (Figure 5.10(f)) under higher magnification. In the FZ, the microstructure is predominated of very fine acicular ferrite with some martensite. Some micro-voids were found in the FZ because of the drop off of the inclusions during the polishing process.
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Figure 5.10 SEM micrographs showing the microstructures in different sub-zones of GMAW joint, (a) SCHAZ, (b) ICHAZ, (c) FGHAZ, (d) CGHAZ, (e) low magnification of reheated CGHAZ, (f) high magnification of reheated CGHAZ, (g) FZ, (prior austenite grain boundaries are signified with arrows).

5.3.4 Microhardness

The hardness maps of entire weld for both the ultra-NGLW and GMAW joints are shown in Figure 5.11(a) and (b), respectively. The FZ, HAZ and BM can be distinguished clearly from the hardness contour map because of the hardness variation in different zones. Figure 5.11(c) and (d) show hardness profiles extracted across the welded joints at 3 mm below the top surface for ultra-NGLW and GMAW samples, respectively. It is seen that the hardness of the base material for these two welded specimens shows some difference, approximately 320-345 HV$_{0.3}$. This may because the two plates for the welding experiments were extracted from different strips, which may not cool consistently during the steel strips manufacturing process. The hardness values of the FZ are higher in ultra-NGLW sample compared to GMAW sample. The average hardness in the FZ for the ultra-NGLW sample is approximately 380 HV$_{0.3}$. The peak hardness is located in the un-tempered FZ and CGHAZ with hardness value of ~ 390 HV$_{0.3}$. The hardness rapidly drops from the CGHAZ toward the SCHAZ. The hardness value in the tempered FZ region was reduced to ~ 330 HV$_{0.3}$, which is located between the two welding passes, as shown in Figure 5.11(a). The lowest Vickers hardness is located in the very narrow soft SCHAZ with hardness value of ~ 310 HV$_{0.3}$. However, there is a wide soft HAZ for the GMAW joint, as seen in Figure 5.11(b). The lowest Vickers hardness is ~ 280 HV$_{0.3}$. In addition, a ‘V’ shaped relatively higher hardness contour can be observed in the wide soft HAZ, which surrounds the ‘V’ shaped FZ. This region is corresponding to the reheated CHGAZ. The peak hardness for the
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GMAW sample is located in the FZ with hardness value fluctuated around 350 HV$_{0.3}$. The hardness in the HAZ has a large variation from 285 to 325 HV$_{0.3}$.

Figure 5.11 Microhardness measurements on the welded joint, (a) hardness map of the entire ultra-NGLW joint, (b) hardness map of the entire GMAW joint, (c) hardness profile across the ultra-NGLW joint, (d) hardness profile across the GMAW joint.
5.3.5 Tensile and bending properties

The representative engineering stress versus engineering strain curves for base S960 steel, the ultra-NGLW and GMAW S960 steel joints are illustrated in Figure 5.12. All of these curves are smooth and continuous. The details of the transverse tensile test results are given in Table 5.7. It must be mentioned here that the welded specimens are not homogenous, especially in the FZ and HAZ. This means the values recorded for the yield stress and the elongation are not truly representative of any particular sub-zones. In addition, they will also vary with the choice of gauge length (50 mm in this case). All fractured samples after tensile test are shown in Figure 5.13. It is apparent that the ultra-NGLW specimens present almost the same tensile properties as the BM even though laser welding results in a narrow soft SCHAZ. It suggested that the ultra-NGLW process did not deteriorate the tensile properties of the S960 HSLA steel. However, the GMAW specimens show inferior tensile properties when compared to the BM. The yield strength (YS), ultimate tensile strength (UTS) and apparent elongation for the ultra-NGLW joints were obtained to be 1019 MPa, 1040 MPa, and 8.3%, respectively, which were very close to those of the BM which had a YS of 1026 MPa, UTS of 1035 MPa and an apparent elongation of 8.5%, respectively. All ultra-NGLW specimens failed in the BM well away from the weld during the tensile test, as is shown in Figure 5.13(b). Whilst GMAW specimens demonstrate a YS of 903 MPa, a UTS of 936 MPa and an apparent elongation of 7.1%, respectively, which present an approximately 100 MPa lower strength and 1.4% less elongation when compared to the BM. All tensile failures for the GMAW specimens were in the soft HAZ, as is shown in Figure 5.13(c). The reductions of area for the BM, ultra-NGLW and GMAW specimens were 59.6%, 54.6% and 56.1%, respectively.
Figure 5.12 Representative engineering stress versus engineering strain curves for base S960 steel, the ultra-NGLW and GMAW S960 steel joints.

Table 5.7 Tensile properties for the base material, the ultra-NGLW and the GMAW specimens.

<table>
<thead>
<tr>
<th>Test specimens</th>
<th>Yield strength (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
<th>Reduction of area (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Base material (8 mm)</td>
<td>1026 ± 7</td>
<td>1035 ± 11</td>
<td>8.5 ± 0.3</td>
<td>59.6 ± 0.4</td>
</tr>
<tr>
<td>NGLW (8mm)</td>
<td>1019 ± 8</td>
<td>1040 ± 11</td>
<td>8.3 ± 0.3</td>
<td>54.6 ± 0.9</td>
</tr>
<tr>
<td>GMAW (8 mm)</td>
<td>903 ± 7</td>
<td>936 ± 9</td>
<td>7.1 ± 0.8</td>
<td>56.1 ± 0.5</td>
</tr>
</tbody>
</table>
Fracture surfaces of the tensile specimens were investigated to examine the fracture mechanisms of the steels. The fractured samples were examined under optical microscope and SEM. All ultra-NGLW specimens failed in the BM well away from the weld, whilst all GMAW specimens fractured in the soft HAZ, as is shown in Figure 5.13(b) and (c), respectively. All fractured surfaces show similar macro and micro morphology. Representative fracture surfaces of the failed tensile test samples are shown in Figure 5.14. From the overall morphology of the fractured surface, it can be seen that the specimen splits into two segments (Figure 5.14(a)), with the boundary between the segments coinciding with the mid-thickness position within the plate. It is speculated that this splitting is due to the specific rolling and fabrication process for the plate, which can lead to variation in the chemical composition in the through-thickness direction [34]. As plastic deformation accumulates, a crack may first develop parallel to the rolling direction and perpendicular to the fracture surface, before ultimately deviating to produce the final fracture. High magnification observations reveal that the fracture surface is composed of predominately of large voids and deep equiaxed dimples, as seen in Figure 5.14(b). The dimples in the SEM photomicrograph of the fractured surfaces validate the ductile fracture behaviour of the tensile test samples. The macro and micro morphology of the fracture surfaces reflect the BM, ultra-NGLW and GMAW specimens have good ductility. There are many spherically-shaped inclusion particles located within the large voids. The chemical elements of these inclusion particles were identified by energy dispersive X-ray spectroscopy (EDX). The EDX spectrum (Figure 5.14(c)) indicates that the inclusion particles are rich in Ca, O, Al, Mn, S and Fe.
Both the root and face bend tests for ultra-NGLW and GMAW joints were carried out on the as-welded joints, and the resulting specimens are shown immediately after testing (Figure 5.15). The welds were subjected to significant plastic deformation and no surface cracks were visible after either root or face bend testing, which indicates that both the ultra-NGLW and GMAW joints exhibited good ductility and adequate bending strength.

Figure 5.14 SEM micrographs showing the fracture surface morphology for the tensile test specimens, and the results of EDX analysis: (a) macro fracture surface, (b) high magnification fracture surface, (c) EDX spectrum for the inclusion particles.
5.3.6 Charpy impact properties

Because the thickness of the base material used in this study was 8 mm, the Charpy impact test in this work was carried out using sub-size specimens (55 mm × 10 mm × 8 mm). Corrections were applied to account for the sub-size specimens to compare the results with those of standard samples (55 mm × 10 mm × 10 mm). There are many previous studies on corrections for sub-sized Charpy V-Notch (CVN) specimens [34-38]. Schubert et al. [37] investigated a normalization factor $NF = Bb^2/KL$ to correct for sub-sized test results, where $B$ is the specimen thickness, $b$ is the length of the remaining ligament at the notch, $K$ is the modified stress concentration factor at the notch and $L$ is the specimen span. Test data obtained with sub-sized specimens can be extrapolated to corresponding results for a standard specimen using $NF$. Chao et al. [38] tested the toughness of DP590 steel using sub-sized CVN specimens. In their case the only non-standard dimension for the test specimens was the thickness (5.5 mm). It was postulated that the only correction that was required was for the thickness. Therefore, they estimated CVN impact energies for standard specimens by multiplying their test data they obtained on their sub-sized specimens by a factor of 1.82 (i.e. 10/5.5). In this work, a similar approach was followed in applying corrections to the energies on 8 mm thick specimens, i.e. by multiplying the absorbed energies by a factor of 1.25 (i.e. 10/8).

The original absorbed Charpy impact energy results for the sub-sized specimens, and the corresponding estimates for the absorbed energies in standard specimens, are shown at different test temperatures in Figure 5.16(a) and (b) for the BM and GMAW and
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ultra-NGLW specimens with notches sampled at the weld centre line, at the FL and in the HAZ region. The proportion of the fracture surface that was subject to brittle fracture is plotted for each type of sample in Figure 5.16(c) to exam the fracture mechanism.

Figure 5.16 Absorbed energy results and proportion of brittle fracture on the fracture surface for the BM, FZ, FL and HAZ impact specimens in GMAW and ultra-NGLW S960 steel at different temperatures, (a) original absorbed energy results for sub-size specimens (55 mm × 10 mm × 8 mm), (b) corresponding estimates for absorbed energies in standard-sized specimens, (c) proportion of brittle fracture on the fracture surface.

Both the original absorbed energy results and the converted absorbed energy results for both the GMAW and ultra-NGLW specimens obey an overall trend: the absorbed energies increase with an increase in the test temperature, as shown in Figure 5.16(a) and (b). The proportion of brittle fracture region for the tested specimens are summarised in Figure 5.16(c), which presents consistent results with the absorbed energies in Figure 5.16(a) and (b). A greater proportion of brittle fracture region on the fracture surface leads to a lower absorbed energy. There was no brittle fracture region for all the BM specimens, regardless of test temperature, which demonstrates all the BM specimens fractured in a ductile manner with good toughness.
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All of the absorbed energy values for the GMAW and ultra-NGLW specimens with notches sampled at the FZ, at the FL and in the HAZ region are lower than the corresponding values for the BM from -40 °C to room temperature. It can be observed from Figure 5.16 that the GMAW joints present better impact toughness than that for the ultra-NGLW joints.

The macroscopic fracture surfaces of selected specimens tested at -20 °C are presented in Figure 5.17. BM and ultra-NGLW specimen with notch in the HAZ fracture surfaces present completely ductile regions, as seen in Figure 5.17(a) and (d). Brittle fracture region takes predominately part of the fracture surface for the ultra-NGLW specimen with notch at the FZ, as seen in Figure 5.17(b). It can be observed that ultra-NGLW specimen with notch at the FL fractured with a brittle fracture region comprising 40% of the fracture surface, as shown in Figure 5.17(c). A very small proportion of brittle fracture region (~ 8-10%) can be observed on the fractured surfaces of GMAW specimens with notch at the FZ and in the HAZ, as shown in Figure 5.17(e) and (g). Brittle fracture region takes approximately half of the fracture surface for the GMAW specimens with notch at the FL, as shown in Figure 5.17(f).
Selected Charpy impact fracture surfaces were examined using SEM. All the ductile fracture regions present dimples and microvoids, with some of the dimples containing spherical inclusions. Selected ductile fracture region SEM image is shown in Figure 5.18(a). The inclusions particle located in the microvoid were examined using EDX. The EDX results corresponding to the inclusion particle shown in Figure 5.18(a), is presented in Figure 5.18(d). The inclusion particles are rich in Al, S, Ca, Mn and Fe. Cleavage fracture was confirmed in the ultra-NGLW specimens with the notch at the FZ and GMAW specimen with notch at the FL, which were tested at -20 °C, and these specimens fractured with low absorbed energies. The brittle regions on the fracture surfaces revealed a cleavage dominated fracture, as is shown in Figure 5.18(b) and (c).
Figure 5.18 Fractured surfaces for the selected specimens after Charpy impact testing at -20 °C, (a) BM, (b) ultra-NGLW specimen with notch at the FZ, (c) GMAW specimen with notch at the FL, (d) EDX spectrum for an inclusion particle in (a).
5.4 Discussion

5.4.1 Microstructure in different sub zones

The microstructures evolution within the different sub-zones of a welded joint is dependent on the local thermal cycles, and the peak temperature that is reached during the welding process. Grain size of HAZ are coarser near the fusion line, it becomes finer as it goes far. However, in the SCHAZ, during the welding process, the peak temperature does not quite reach the Ac\(_1\) temperature, and so no material is re-austenized in this region [39]. There is no metallographic colour change in this SCHAZ after etched using Nital solution. So this sub-zone cannot be distinguished under optical microscope. However, SEM examination of this region for both the ultra-NGLW and GMAW joints indicated that the phases apparent in this sub-zone retained the microstructure of the BM, mixed with tempered martensite and tempered bainite, as is shown in Figure 5.8(a) and Figure 5.10(a).

The peak temperature experienced in the HAZ reduces gradually from the CGHAZ (adjacent to the fusion line) to the ICHAZ (close to SCHAZ and BM). In the CGHAZ, the peak temperatures are raised to 1300 °C or higher (but lower than the melting point) for a sufficient period to ensure complete re-austenization, and rapid grain growth [40, 41]. The FGHAZ encountered temperatures slightly higher than the Ac\(_3\) temperature. This results in complete re-austenization of the material, but there is limited austenite grain growth due to the relatively low peak temperatures and for short periods of time at these temperatures in the FGHAZ [42, 43]. Whilst the ICHAZ experienced temperature between the Ac\(_1\) and Ac\(_3\) temperatures, the material in this sub-zone has a partial transformation into austenite. The FZ was melted during the heating process, and it subsequently solidified during the cooling process. In addition, very large grain size austenite was generated in the FZ.

The cooling rate is one of the most important factors influencing microstructure evolution in either fusion zone or HAZ regions [43]. The higher values of the heat input, the slower the cooling rate would be, and vice versa [44]. Because the FZ is located in the area directly worked by the heat source and the cooling rate in the FZ is relatively slower than the other sub-zones. Comparing the heat input per unit length between
typical arc and laser welds, it is shown that the heat input in laser welds is typically almost an order of magnitude lower than in arc welds [45]. In this research, the heat input of each welding pass for GMAW is \( \sim 0.5\text{–}0.8 \text{ kJ/mm} \), while the heat input for ultra-NGLW is \( \sim 0.1 \text{ kJ/mm} \). Autogenous laser welding has lower line energy than arc welding processes and higher cooling rates (2000–3000 °C/s) [15].

Owing to the fast cooling rate in the ultra-NGLW process, the transformed austenite in the ICHAZ, FGHAZ, CGHAZ and FZ was quenched. Martensite was formed in these sub-zones. Because of the high Ms temperature (\( \sim 450 \) °C) for S960 steel, some formed martensite has more chance to be auto-tempered in the following cooling process. These sub-zones included predominantly martensite in conjunction with some tempered martensite. The grain size varied gradually from the FZ to the FGHAZ because of the gradually reduced peak temperature, as seen in Figure 5.8. In the ICHAZ, the partially transformed austenite quenched into martensite, while the un-transformed material was tempered by the welding thermal cycle. The martensite within the ICHAZ reflects the original banded morphology of the BM, which is parallel to the rolling direction, as is shown in Figure 5.8(b). In addition, more tempered martensite was observed in the lower part of the welded joint, especially the area between the two welding passes. This phenomenon can be attributed to the additional heat effect introduced from the subsequent welding pass to the prior welding pass. This newly generated thermal cycle has a temper effect on the prior welding pass.

For the GMAW joints, the arc has a relatively broad heating area and the higher heat input introduced into the joint makes it cooling rate much slower than the ultra-NGLW joint. In the slow cooling rate the newly formed austenite in the HAZ transformed into bainite, some small amount of martensite and tempered martensite were also formed in the HAZ. The grain size also reduced gradually with the increase of the distance to the FZ because of the gradually reduced peak temperature, as seen in Figure 5.10. In the ICHAZ, the partially transformed austenite transformed into martensite and some tempered martensite. This sub-zone is close to the SCHAZ and BM, the cooling rate in the sub-zone may be relatively faster than in the FGHAZ and CGHAZ. Because the FZ cools from melt, the cooling rate in the FZ is relatively slower than in the HAZ. Abundant fine acicular ferrite with some martensite were formed in the cooling process. In addition, there is a reheated CGHAZ formed because of the multi-heat effect from
Chapter 5

the multi-welding passes. In the reheat process, some austenite nucleated on the prior austenite grain boundaries, but the growth of these newly generated austenite were suppressed because of the relatively low peak temperature and fast cooling rate from the reheat process. Martensite and tempered martensite were transformed from the austenite with very small grain size (~ 2 µm), which are encircled around the prior austenite grain boundaries. While the generated bainite in the CGHAZ was retained inside of the partially transformed martensite circle, as seen in Figure 5.10(e) and (f).

5.4.2 Understanding the mechanical properties

The ultra-NGLW introduced less heat input to the welded joint, which results in a rapid quench rate. The FZ and HAZ, which consist of martensite as the principal phase with some tempered martensite, would be of high cross-section microhardness and with a lack of toughness. In addition, the low heat input and rapid cooling rate caused the formation of narrow HAZ in the ultra-NGLW joint. The different peak temperatures and thermal cycles in this narrow transition region in the HAZ caused a sharp microhardness dropping from CGHAZ to SCHAZ, seen in Figure 5.11. The SCHAZ endured a high temperature tempering (although lower than the Ac₁ temperature) during the welding process, which may result in a reduction in the dislocation density and softening of the microstructure in this very narrow region. This may result in a lower hardness than the BM, as seen in Figure 5.11(c). However, GMAW introduced more heat input to the welded joint, which can lead to a slower cooling rate. Abundant acicular ferrite with some martensite were formed in the FZ of the GMAW joint, which result in the hardness in the FZ being comparable or a little higher than the BM. Due to the relatively higher cooling rate in the HAZ and the subsequent tempering effect from the multi-pass welding process, the HAZ was predominated of bainite with a mixture of small amount of martensite and tempered martensite, which results in a broader HAZ than ultra-NGLW joint and a much wider and softer HAZ than the ultra-NGLW joint.

It is well known that hardness is an indication of the strength. In general, an increase in hardness results in an increase in strength but decreasing formability or ductility [46]. Surface hardness mapping can be employed in general to depict the strength map. It can be deduced from the hardness map of entire weld for the ultra-NGLW joint in Figure 5.11(a) that the FZ and HAZ (except the SCHAZ) were strengthened by the ultra-
NGLW process. Some thin tempered layers between the two welding passes are observed, which are attributed the tempering effect from the tempering effect by the subsequent welding pass on the prior welding pass. These tempered thin layers may have the same or comparable strength as the BM. However, a very narrow SCHAZ can also be observed from the hardness mapping results, which would be the weakest region for the ultra-NGLW joint.

In addition, it can also be deduced from the hardness map of entire weld for the GMAW joint in Figure 5.11(b) that the FZ has the same or a slightly higher strength when compared to the BM. The wide soft HAZ would make the welded joint lose strength. However, the very thin ‘V’ shaped reheated CHGAZ may strengthen the wide soft HAZ a little.

In the transverse tensile test, all the ultra-NGLW specimens failed in the BM well away from the FZ with almost the same tensile properties as the BM. However, the GMAW specimens failed in the soft HAZ show inferior tensile properties when compared to the BM. The high hardness in the FZ and HAZ for the ultra-NGLW joint results in greater yield strength than the BM. Conversely, both the yield and ultimate tensile strength of the GMAW joint are considerably reduced by the softness of its HAZ. Compared to the base metal, the elongation to failure is reduced in all of the welded samples, albeit to a much greater extent by GMAW (1.4%) than with ultra-NGLW (0.2%). This loss of ductility is mainly attributed to a localization of strain during deformation; i.e., the non-welded base metal has a homogenous microstructure, whereas the welded joint exhibits variation in microstructure and hardness across the welded joint. The local strain generated during tension is therefore concentrated within a specific region, resulting in premature fracture compared to the uniform deformation that occurs in the BM [47]. Nevertheless, these results do demonstrate that the ultra-NGLW joint can actually retain its strength with only a relatively small loss of ductility which is attributed to the very narrow strengthened FZ and HAZ. However, due to the much wider FZ with a little strengthened region for GMAW specimens, the GMAW specimens show much reduction on the elongation than the ultra-NGLW specimens. However, the much softer HAZ of the GMAW specimens are likely to be highly ductile, which present higher reduction of area than of the ultra-NGLW specimens.
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It is apparent that the ultra-NGLW specimens present almost the same tensile properties as the BM with failure in the BM not in the narrow soft SCHAZ, whereas the GMAW results in a substantial deterioration of all tensile properties with failure in the soft HAZ. In the case of the ultra-NGLW specimens, the FZ and HAZ (with the exception of the SCHAZ) had a much higher hardness than the BM, while the SCHAZ was a little softer than the BM, but this region is very narrow (~ 0.8 mm). The hard FZ and the hard regions of the HAZ can therefore act as a strong constraint on plastic deformation in the adjacent SCHAZ, thereby leading to the majority of the tensile plastic deformation accumulating in the BM and subsequent fracture in the BM [47-49], as seen in Figure 5.13. Lee et al. [47] reported that this concentration of deformation in the BM was confirmed by optical observation of the local strain distribution in their test on laser welded DP780 steel. However, it should be mentioned that this plastic deformation constrain by the surrounding strengthened material happens only in the case when the softened zone is narrow and the degree of softening is relative low. Otherwise, the wide softened zone as well as low hardness will result in the decreasing of the tensile strength of the welded joint [48]. Thereby, failure therefore occurred in the soft HAZ for the GMAW samples.

The rapid cooling process characteristic of ultra-NGLW process makes hard martensite generated in the FZ and HAZ, which strengthen the welded joint, but also increased the brittleness of the joint. This is demonstrated by the results in Figure 5.16: the absorbed energy values for the ultra-NGLW specimens with a notch sampled at the FZ, at the FL and in the HAZ are lower than the corresponding values for the BM from -40 °C to room temperature, especially for the FZ, which has a much lower absorbed energy values. It implies that the FZ of the ultra-NGLW specimens present lower toughness. However, the ultra-NGLW specimens with a notch placed in the HAZ presents high absorbed energy values, which are just a slightly lower than those for the BM. For the specimens with notch in the HAZ, the notch centre was sampled at FL + 0.5 mm. It was easy to offset the notch centre in the ICHAZ or even SCHAZ if there was a machining error because of the narrow HAZ. In addition, the cracks were easy to deviate in to the SCHAZ and BM. The specimens with notch at the FL presents higher absorbed energy values than those with notch at the FZ at corresponding test temperatures changed from -40 °C to room temperature. Because prior austenite grain size in the FZ is very large, the grain boundaries have a weaker barriers prohibiting for the propagation of cracks.
The GMAW specimens with notch at the FZ and FL present higher absorbed energy values than those ultra-NGLW samples with notch at the FZ and FL at corresponding test temperatures changed from -40 °C to room temperature. This can be attributed to the relatively softer microstructure in the GMAW joint. The higher absorbed energy values for the GMAW specimens with notch at the FZ, FL and in the HAZ at corresponding test temperatures follow an order: HAZ > FZ > FL. In addition, the absorbed energy values for the GMAW specimens with notch at the FZ are very close or even a little higher than those for the FL. For the specimens with notch in the FZ, the microstructure in the FZ was predominated of fine acicular ferrite, which had almost the same or a little higher hardness as the BM. However, there were a lot of inclusions in the FZ, which may cause the toughness in the FZ to be inferior to the BM. The FL notch was sampled to pass through the HAZ (real FL) at the mid-thickness point of the material (sample 50% weld metal and 50% HAZ) to delegate the nominated FL. So the machined FL notch centre would cover half CGHAZ and some reheated CGHAZ, which had very large prior austenite grain size. The grain boundaries would also have a weaker barriers prohibiting for the propagation of cracks for the FL samples and the relatively harder reheated CGHAZ also had some deteriorated effect on the toughness of the FL. While the HAZ notch was sampled at FL + 2 mm, which would also cover some CGHAZ on the upper part of the welded joint and some reheated CGHAZ on the lower part of the welded joint. Even though the HAZ was much softer than the BM, the toughness of the HAZ was also inferior to the BM. This means that the coarse grains in the CGHAZ are having a more significant and detrimental influence on Charpy toughness than any benefits which may be attributed to the slightly reduced hardness. In addition, the relatively harder reheat CGHAZ may be also detrimental to the toughness of the HAZ.

5.5 Conclusions

From this investigation the following conclusions were derived:

(1) The heat input for each welding pass, the cumulative heat input (the heat input summation of all welding passes) and the consumed filler material for the GMAW process are much higher than those for the ultra-NGLW process.
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(2) The tensile strength of the ultra-NGLW specimens were comparable to the strength of the BM, with all welded specimens failing in the BM. While the GMAW specimens failed in the soft HAZ with an approximately 100 MPa lower strength than the BM. The elongation of ultra-NGLW specimens is similar to that of the BM, while the GMAW specimens show much reduced elongation compared with that of the BM. Both the ultra-NGLW and GMAW specimens performed well in three-point bending tests.

(3) Microstructure in the FZ of the ultra-NGLW was predominantly martensite mixed with some tempered martensite. The HAZ was also predominantly martensitic with some self-tempered martensite, with the prior austenite grain size reducing gradually moving from the CGHAZ to ICHAZ. However, in the FZ of the GMAW joint, the microstructure was predominated of very fine acicular ferrite with some martensite. While microstructure in the HAZ mainly consisted of bainite with a mixture of small amount of martensite and tempered martensite, with the prior austenite grain size reducing gradually moving from the CGHAZ to ICHAZ.

(4) The peak hardness measured for the ultra-NGLW specimen was located in the untempered FZ and CGHAZ, which was ~ 40 HV$_{0.3}$ than that of the FZ for the GMAW specimen. The hardness in the tempered FZ was ~ 60 HV$_{0.3}$ lower than the untempered FZ for the ultra-NGLW specimen, which was located between the two welding passes. There was a very wider softening HAZ for the GMAW specimen, while the ultra-NGLW specimen had a very narrow softening region located in the SCHAZ.

(5) The GMAW joint demonstrated better impact toughness than the ultra-NGLW joint. The generation of martensite in the FZ of the ultra-NGLW joint makes the FZ brittle.

References


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Chapter 6: Comparison of laser welds in thick section S700 high-strength steel manufactured in flat (1G) and horizontal (2G) positions

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Abstract

Lack of penetration, undercut and melt sagging are common welding defects for single-pass laser welds in thick plates, particularly when using a traditional 1G welding position (laser directed towards ground). This investigation shows, for the first time, that welding 13 mm thick high-strength S700 steel plates in the 2G position (laser beam perpendicular to the direction of gravity) can mitigate some of the common welding defects including undercut and sagging. A computational fluid dynamic analysis indicates that the 2G welding position can assist in achieving an appropriate balance between surface tension, hydrostatic pressure (gravity) and recoil-pressure from the metal vapour.

\textbf{Keywords:} laser; welding; high strength steel
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6.1 Introduction

High strength low alloy (HSLA) steels have been widely used for many years, due to their high strength and toughness. HSLA steels are used in a variety of applications such as in structural components, pressure vessels and fluid transportation pipes, in shipbuilding, offshore construction, automotive applications, and lifting and handling equipment [1, 2]. The application of HSLA steels enables lighter and more slender products to be employed and reduces construction costs without loss of structural integrity [3]. Recently, fiber laser welding has been receiving attention due to the advantages of high power, high beam quality, flexible optical fibre beam delivery, and high energy efficiency, which enable high penetration welds to be produced at fast welding speeds [4, 5]. During solidification of the weld pool, the longitudinal and transverse shrinkage stress variations along the keyhole axis are much lower than for most other welding technologies. This results in low buckling and bending of the workpieces [6].

There are several common defects in single pass laser welding of thick section materials: lack of penetration, undercut on the top weld surface and melt sagging at the root section of the weld, which occur in single-pass welding of thick metallic plates in the flat welding position (i.e. 1G, with the laser beam facing towards the ground) due to imbalances of the surface tension and the hydrostatic pressure in the melt pool [7].

A backing plate is often employed to prevent the molten pool sagging excessively [8]. Jones et al. [9] took advantage of the electromagnetic (Lorentz) force in an electrically conducting liquid metal to compensate for the force of gravity and support the weld pool during overhead electron beam welding. The excitation devices were placed below the weld to generate the upward Lorentz force in order to support the weight of the molten pool. The idea to use this Lorentz force in high power laser beam welding of 20 mm thick stainless steel plates, in the 1G position, was developed in Bachmann et al.’s [7] work. However, the equipment and application of the Lorentz force technique can be complex, and it can be difficult to remove a backing plate after welding.

Welding in the horizontal (2G) position plays an important role in the manufacture of large and heavy structures, and in some cases it is the only viable welding position.
However, there is a lack of deep understanding of autogenous laser welding in the 2G position, particularly for the welding of thick section materials.

There have been very few investigations of 2G position laser welding. In order to investigate the relationships between penetration depth and welding parameters in the welding of thick plates, Okado et al. [10] and Wani et al. [11] developed a combined laser system using a 6 kW YAG laser and a 10 kW Chemical Oxygen-Iodine Laser (COIL). This system was used for bead-on-plate welding tests on thick 304 stainless steel and aluminum alloy plates in the 2G welding position. However, they did not explore the physics of thick section autogenous laser welding in the 2G position.

This paper compares the characteristics of high power fibre laser welding of thick section S700 high strength steel in the 1G and 2G positions, for the first time, filling some of the knowledge gaps in the understanding of weld defect formation. In particular, computational fluid dynamic (CFD) modelling was carried out to understand the dynamic forces on the weld pool and the factors affecting the formation of the weld bead profile.

6.2 Experimental procedure

The as-received base material (BM) provided by Tata Steel, was 13 mm thick S700 HSLA steel in the form of hot rolled strips (chemical composition: 0.068% C, 1.476% Mn, 0.009% P, 0.001% S, 0.05% Si, 0.073% Al, 0.495% Cr, 0.19% Mo, 0.03% Nb, 0.044% V, 0.0018% B, 0.0045% N and balance Fe), which had been rapidly water cooled to low temperature producing a bainitic microstructure.

A continuous wave (CW) fiber laser (IPG YLR-16000) was used in these laser welding experiments with a maximum available laser power of 16 kW and a beam parameter product (BPP) of 10 mm.mrad delivered with an optical fiber, 300 µm in diameter. The laser beam emitted from the optical fiber was collimated by a lens with a 150 mm focal length and then focused onto the specimen surface using a lens with a 400 mm focal length. The measured focus size and Rayleigh length were 0.8 mm and 15 mm, respectively. The laser head was mounted on a 6-axis KUKA robot. A schematic
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representation of the laser welding setup for 1G and 2G welding positions is shown in Figure 6.1.

Figure 6.1 Schematic representation of the laser welding setup for (a) 1G welding position and (b) 2G welding position.

The top and the back surfaces of the specimen were shielded using argon gas to protect the molten weld pool during welding, using flow rates between 8 and 12 l/min.

A significant number of preliminary welding trials on the 13 mm thick S700 HSLA steel plates helped to establish that it was beneficial to set the focal position below the top surface of the specimens, in order to minimise the melt sagging problem, which was in agreement with earlier published results [7]. A focal position of 8 mm below the top surface of the specimens was found to work well, and thus was employed in the welding experiments for both 1G and 2G welds. The Design Expert 7.0 software package was used for designing the experiments. The experiments were designed based on two welding variables (laser power and welding speed), employing five levels for each. A response surface method (RSM) was used to identify the most significant factors influencing the weld bead characteristics, including penetration depth and the extent of undercut.

After laser welding, specimens were cut from the welded plates on a transverse section with respect to the welding direction. The specimens were subsequently ground and polished using an automatic polishing machine, followed by etching in a solution of 2% Nital for about 2 s. The macrostructure of each joint was examined using an optical
microscope (KEYENCE VHX-500F) and microstructures were examined using a Philips XL 30 scanning electron microscope (SEM).

6.3 Results

A set of 11 laser welding experiments in both the 1G and 2G positions was performed using various welding parameters, according to the design matrix shown in Table 6.1. Using the 1G welding position one can achieve a deeper penetration depth compared with that for the 2G welding position when using the same welding parameters. However, the undercut depth was also larger for the 1G welds, as seen from the results in Table 6.1.

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<th>Power (kW)</th>
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It was found from the welding trials that weld pool sagging is one of the characteristics of 13 mm thick welds in S700 steel in the 1G position. A periodic sagging defect was generated on the weld root side. The transition from a partially penetrated weld to a fully penetrated weld with weld pool sagging was dramatic and resulted from small changes in welding parameters. Indeed, it was very difficult to obtain a good full penetration weld without melt sagging using the 1G welding position in 13 mm thick S700 steel. Weld pool sagging leads to deep undercut on the top surface of the weld. Kawahito and Katayama [12] reported that the laser welding process window for the production of sound welds was very narrow using the 1G welding position when welding thick section 304 stainless steel plates. The minimisation of weld pool sagging

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is one of the most important challenges when one is seeking to avoid undercut defects in single pass laser welding of thick section materials.

Figure 6.2 shows the interactions between laser power and welding speed in affecting the penetration depth for the 1G and 2G welding positions. There is a large set of welding parameters (high power and low welding speed) that can be used to achieve full penetration for the 1G welding position, as seen in the red region in Figure 6.2(a). However, melt sagging was unavoidable for full penetration welds when using the 1G welding position. In contrast, a smaller range of welding parameters can be used to obtain full penetration using the 2G position (Figure 6.2(b)), without sagging, as will be seen later.

![Figure 6.2 Contour map showing the effects of laser power and welding speed on the penetration depth. (a) 1G position, (b) 2G position.](image)

The interactions between laser power and welding speed in affecting the undercut depth for the 1G and 2G welding positions are shown in Figure 6.3. Changes in welding parameters have a big effect on the undercut depth in the 1G welding position, as seen in Figure 6.3(a), while changes in welding parameters have a smaller effect on the undercut depth when using the 2G position, as seen in Figure 6.3(b). Much deeper undercut was formed in the 1G welding position than in the 2G welding position when using the same welding parameters.
In the parameter optimisation investigation, a numerical multiple response optimisation criterion was used to reach the maximum penetration depth (13 mm) and a minimum in the undercut depth in order to improve the weld quality. In order to satisfy multi-objective optimisation, the desirability function is defined by the geometric mean of all individual desirabilities that range from 0 for the least desirable settings to 1 for the most desirable process settings [13]. The function is defined as [13]:

$$\delta = (\prod_{i=1}^{n} d_i)^{\frac{1}{n}}$$  \hspace{1cm} (6.1)

This equation represents the overall desirability function, where $\delta$ is the overall desirability, $n$ is the number of responses and $d_i$ is the $i_{th}$ response desirability value. In this research, both numerical and graphical optimisation approaches were used by selecting the desired goals for each factor and response.

The optimised welding parameters for both the 1G and 2G welding positions were 13 kW and 0.72 m/min, and these were used on the specimens in each case. A comparison of the top and back weld appearances, and weld cross sections using these two groups of optimised welding parameters are shown for each position in Figure 6.4. It can be seen that sound welds were obtained using the 2G welding position with the above optimised welding parameters, while periodic sagging was observed on the back of the weld for the 1G welding position. It was found that the microstructure in the weld for both the 1G and 2G positions was bainite, as shown in Figure 6.5(a). The microstructure in the heat affected zone (HAZ) for both the 1G and 2G positions was bainite mixed with...
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martensite, as shown in Figure 6.5(b). Several welding trials were carried out in the hope of obtaining sound welds using the 1G welding position. However, we were unable to obtain sound welds (due to either lack of penetration or melt sagging).

![Figure 6.4: Top weld appearance, weld cross section and weld underside appearance using optimised welding parameters in the 1G and 2G positions (13 kW, 0.72 m/min). (a) 1G position, (b) 2G position.](image)

![Figure 6.5: The microstructures of the weld and HAZ, (a) weld, (b) HAZ.](image)

6.4 Discussion

The results in Table 6.1 demonstrate that using the 1G welding position can lead to a deeper penetration depth when compared with the 2G position, when using the same welding parameters. This may be because, in the 1G position, gravity is acting in the same direction as the laser beam, which aids the flow of metal towards the lower part of
the weld pool. In contrast, gravity was acting in a direction that was perpendicular to the penetration direction for the 2G welding position. Under the action of gravity, the heat transfer from the high temperature metal could be biased in a direction that is perpendicular to the beam axis. A schematic representation of the pressure balance in the weld pool for both the 1G and 2G welding positions is shown in Figure 6.6.

![Figure 6.6 Schematic representation of the pressure balance in the weld pool. (a) 1G position, (b) 2G position.](image)

The undercut depth for the 1G welding position was deeper than that for the 2G welding position with the same welding parameters. This may be attributable to a higher degree of fluid flow, driven by gravity, towards the root of the weld. For the full penetration 1G welds, the drop out of metal on the underside of the weld led to deeper undercut on the top side of the weld. Deep undercut (sample 5) also occurred for 2G welding position when high enough energy was introduced to weld the material.

Three-dimensional computational fluid dynamic (CFD) modelling was performed in order to understand the dynamic forces within the weld pool, and the weld geometry that is formed with the optimised welding parameters. This was performed using FLUENT software. In the CFD model, the Navier–Stokes mass, energy and momentum balance equations were used:

Mass conservation equation:
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\[ \nabla \cdot \mathbf{v} = 0 \quad (6.2) \]

where \( \mathbf{v} \) is the velocity vector.

Navier–Stokes equation:

\[ \frac{\partial \mathbf{v}}{\partial t} + \mathbf{v} \cdot \nabla \mathbf{v} = -\frac{1}{\rho} \nabla p + \nu \nabla^2 \mathbf{v} - K \mathbf{v} + \mathbf{g} \quad (6.3) \]

where \( \rho \) is the fluid density, \( p \) is the pressure, \( \nu \) is the dynamic viscosity, \( K \) is the drag coefficient for a porous media model in the mushy zone and \( g \) is the gravitational acceleration.

Energy conservation equation:

\[ \frac{\partial h}{\partial t} + \mathbf{v} \cdot \nabla h = \frac{1}{\rho} \nabla \cdot (k \nabla T) \quad (6.4) \]

where \( h \) is the enthalpy, \( k \) is the thermal conductivity, and \( T \) is the temperature.

The fluid flow in the weld pool is primarily driven by the combination of surface tension and the buoyancy force. The material properties used for the simulation were obtained from the literature [14, 15].

The boundary condition used in this study was as follows: Marangoni shear stresses at the metal-air surface due to surface tension variations with temperature:

\[ \eta \frac{\partial u}{\partial z} = \frac{\partial y}{\partial T} \frac{\partial T}{\partial x}, \eta \frac{\partial v}{\partial z} = \frac{\partial y}{\partial T} \frac{\partial T}{\partial y} \quad (6.5) \]

where \( \gamma \) is the surface tension and \( u = (u, v, w) \) are the velocity components in the corresponding directions. The surfaces were also subject to a slip condition. The inlet temperature was set to room temperature. The velocity in the normal direction at the inlet boundary was set to be constant.

Figure 6.7 shows the calculated weld pool profiles for the 1G and 2G positions. When the surface tension is not high enough to balance the hydrostatic pressure and the recoil
pressure, melt sagging occurs in the 1G position, and the molten metal drops out, so that an undercut can form after solidification (Figure 6.7(a)). However, in the 2G position, gravity is acting in a direction perpendicular to the penetration direction. This enables the surface tension to balance the pressure in the molten pool. The molten metal remains stable in the weld pool with full penetration, but without melt sagging (Figure 6.7(b)). A full penetration weld with a very small undercut (0.2 mm) can form. A comparison of the measured weld appearances and cross sections for the 1G and 2G welding positions in Figure 6.4(a) and (b), with the simulated weld pools in Figure 6.7(a) and (b), reveals very good agreement.

Figure 6.7 Simulated fluid flow and the weld pool dynamic profiles, (a) 1G position, (b) 2G position.

The simulated longitudinal section and the cross-section of the dynamic weld pool for the 1G and 2G positions, respectively, are presented in Figure 6.8. There is a high pressure at the bottom of the molten pool for the 1G position, with a value of approximately 1000 Pa, as seen in Figure 6.8(a). This result is close to the hydrostatic pressure of the melt column, \( p_{hs} \), according to:

\[
p_{hs} = \rho gh
\]  

(6.6)

where \( \rho \), \( g \) and \( h \) are the density, the gravitational acceleration and the height of the column of molten metal, respectively. Using the density of molten steel (approximately \( 7.6 \times 10^3 \text{ kg/m}^3 \)) \([14, 15]\), the calculated hydrostatic pressure of the melt column is approximately 970 Pa. The simulated pressure at the bottom of the molten pool (~ 1000 Pa) for the 1G position in Figure 6.8(a) is very close to the calculated hydrostatic pressure, which validates the simulated results from another perspective.
Figure 6.8 Simulated fluid flow and pressure distribution in the weld pool, (a) longitudinal section of the 1G weld pool, (b) cross section of 2G weld pool.

From the perspective of the Laplace pressure, $p_\gamma$, at the curved and sagging root surface of the weld pool,

$$p_\gamma = \frac{\gamma}{r} \quad (6.7)$$

where $\gamma$ is the surface tension coefficient of the molten metal and $r$ is the radius of the sagging root surface. For carbon steel, the surface tension can reach a value of approximately 1.65 N/m [14, 15] and the subsequent value of the Laplace pressure for the experimentally observed sagging root with a 2.5 mm radius can be up to 660 Pa, which is lower than the hydrostatic pressure of approximately 1000 Pa for the 13 mm thick S700 steel (without considering the recoil pressure) which drives the sagging of the molten pool.

The molten metal in the sample welded in the 2G position has a lower pressure (Figure 6.8(b)). There is a relatively high pressure (~ 300 Pa) in the lower region of the molten pool due to the influence of gravity acting on the pool, and the resulting hydrostatic pressure. However, this relatively high pressure is located at the boundary between the molten pool and solid material (the fusion line). As such, this pressure can be sustained by solid material. Moreover, due to the movement of the molten metal near the top of the pool under the influence of gravity, there is a small cavity generated at this location, which results in a negative pressure (~ -100 Pa). The central region of the molten pool has a low pressure of ~ 30 Pa. This low pressure in the 2G molten pool makes it feasible
for surface tension to balance this pressure during welding, thereby stabilising the weld pool.

6.5 Conclusions

High quality single-pass autogenous laser welds in 13 mm thick S700 steel have been demonstrated using the 2G welding position. Indeed, this work has demonstrated that gravity-driven weld pool drop-out, which is associated with the welding of thick plates, can be prevented by taking advantage of the 2G welding position. It was found to be very difficult to obtain high quality welds using the 1G position, for autogenous single pass laser welding, without employing any other support for the weld pool. The CFD modelling has shown that the pressure in the molten pool is lower when using the 2G welding position.

References

Chapter 6


Chapter 7: Residual stress distributions in S700 high strength steel welds made with ultra-narrow-gap-laser, single-pass-autogenous-laser and gas-metal-arc welding processes

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Abstract

S700 is a new high strength steel recently developed by Tata Steel. This paper describes the residual stress distributions in a gas metal arc weld (GMAW), a single pass autogenous laser weld (ALW) and a multi-pass ultra-narrow gap laser weld (NGLW) in 13 mm thick S700 high strength steel plates, as measured by X-ray diffraction and the contour method. It was found that solid-sate phase transformations from austenite to ferrite, bainite and martensite not only change the magnitude of residual stresses for these three different welded specimens, but they also change the yield strength in the fusion zone, which can have a significant effect on the welding residual stresses. The width of the region sustaining tensile stress for these three different welded specimens decreased in the following order: GMAW $>>$ ALW $>$ ultra-NGLW. A characteristic ‘M’ shaped residual stress profile across the weld was observed for all the above three different welded specimens, with high tensile peaks situated roughly at the heat affected zone/base material boundary. Measured stresses close to the weld centreline were noticeably less tensile than the adjacent peaks in the heat affected zones.

Keywords: as welded condition, bainite, hardness, martensite, solid state phase transformation, 2G welding position
7.1 Introduction

Owing to their excellent strength-toughness combination and weldability, high strength low alloy (HSLA) steels have been widely used as structural components, pressure vessels and oil/gas transportation line-pipes, and in the shipbuilding, offshore construction, automotive, and lifting/crane industries [1, 2]. In some civil construction and heavy industry applications, such as bridges, power plants, and ship hull structures, an emphasis on the safety factor for the structural design is the primary driving force for the application of extra thick (up to ~ 50 mm) steel plates and pipes [3, 4]. Conventional arc welding processes, such as gas tungsten arc welding (GTAW) or gas metal arc welding (GMAW), as well as submerged arc welding (SAW) are widely used to weld these HSLA steels [5-7].

The power density in laser welding is on the order of $10^6-10^8 \text{ W/cm}^2$, which is about 4 orders of magnitude higher than in conventional arc welding methods [8]. Laser welding presents many advantages over traditional arc welding, such as higher welding speeds, deeper penetration, narrower fusion zones (FZ) and heat affected zones (HAZ), as well as lower heat input and distortion [9, 10]. The capability for optical fiber beam delivery and the use of industrial robots also makes the laser welding process easily automated in manufacturing industries [11].

Laser autogenous welding (without filler material) is one of the most common set-ups for laser welding [12]. Presently, the application of laser welding typically involves single pass autogenous welding without any filling material [13]. However, precise fit-up requirements, and the maximum power that is available with commercial lasers (typically less than 20 kW) hinder the wider application of the single pass autogenous laser welding technique, particularly in joining thick section materials. However, multi-pass laser welding with a filler wire is a feasible technique to weld thick section materials using the limited laser power at hand. Over the last few years, the multi-pass narrow gap laser welding (NGLW) technique has been demonstrated, which can be applied to the welding of thick section components with a filler wire using relatively moderate laser powers [14, 15].
Welding processes introduce localised rapid heating and cooling cycles, which result in regions in the vicinity of the weld undergoing severe thermal cycles, as well as localised fusion. The severe thermal cycles cause non-uniform heating and cooling in the material, thus generating inhomogeneous expansion and contraction in the weldment [16]. When the fused region solidifies, the accompanying contraction exerts a pull on the surrounding material which may be prevented from complying by constraint, and then misfit strains will be introduced into the parts being welded. If these misfit strains are modest, they can be accommodated by elastic strains. However, when these misfit strains increase to a level which cannot be sustained elastically, localised plastic deformation is induced, which leads to the development of residual stresses and component distortions [15, 17]. The associated misfit strains often lead to near yield tensile stresses at room temperature, especially in the welding direction [18]. In the absence of transformation effects, significant tensile stresses reside in the vicinity of the weld after the component has reached thermal equilibrium [15, 17, 19]. In the case of phase transformable steels, such as ferritic steels, it is also known that solid state phase transformations can have a significant effect on residual stress distributions [17, 18, 20, 21]. Although tensile residual stresses in welded joints are the norm in the fusion zone and heat-affected zone, compressive stresses can be generated under some circumstances when solid-state phase transformations take place during the welding process. The strains resulting from the solid state phase transformation translate into compressive stress [17].

The presence of residual stresses in the weldment can be detrimental to the performance of the welded product. Such stresses can be detrimental when they reduce the tolerance of the material to an externally applied force, as is the case with welded joints [21]. In addition, they can be detrimental to the performance of a material or the life of a component [20]. Tensile residual stresses are generally detrimental, as they increase the susceptibility of a weld to fatigue damage (even under compressive loading), stress corrosion cracking and fracture [16, 17].

Narrow gap laser welding is a very promising technique for joining thick section materials; however, most of the reported narrow gap laser welding work has been focused on the optimisation of welding parameters and procedures [14, 22]. Elmesalamy et al. [15] reported a study on residual stresses in multi-pass narrow gap
laser welding of 316L stainless steel using the contour method. However, to date there is no published work on residual stress investigations for narrow gap laser welding of high strength transformable steels.

The contour method is a newly developed stress-relaxation-based technique for residual stress evaluation which was developed by Prime in 2000 [23], and the supporting theory is based on a variant of Bueckner’s elastic superposition principle [24, 25]. This principle allows for the measurement of residual stresses normal to a plane of interest, which is typically a plane on which a cut is made through the material. A characteristic of the contour method is that it is possible to obtain a full-field, 2D map of residual stresses normal to the cutting plane. The application of the contour method primarily involves four steps: specimen cutting, contour measurement, data reduction and stress analysis [26]. Zhang et al. [27] validated contour method results by using non-destructive diffraction data obtained from the same weld before cutting. There was excellent agreement between the results that were obtained from the two methods. An explicit description of the contour method and its application procedure can be found elsewhere [27].

In this paper, the contour method was used to measure full longitudinal residual stress maps (i.e. maps for stresses acting parallel to the welding direction) for a gas metal arc weld (GMAW), a single pass autogenous laser weld (ALW) and a multi-pass ultranarrow gap laser weld (NGLW) in S700 high strength low alloy steel. S700 steel plates 13 mm in thickness were investigated, and the residual stress results were compared for the three different welding techniques.

7.2 Material and experimental procedures

The as-received base material (BM), which was supplied by Tata Steel, was 13 mm thick S700 HSLA steel. It had a minimum yield strength of 700 MPa. The chemical composition and a scanning electron micrograph (SEM) of the BM are given in Table 7.1 and Figure 7.1, respectively, while the mechanical properties appear in Table 7.2. The microstructure of the BM was bainite. The chemical composition and mechanical properties of the solid Carbofil 2NiMoCr (ER120S-G) filler material, which was used for the manual GMAW and the ultra-NGLW, are also given in Table 7.1 and Table 7.2,
respectively. The diameter of the filler wire was 1.0 mm. It can be seen that this filler wire overmatched the base material.

Table 7.1 Chemical compositions for base S700 steel and Carbofil 2NiMoCr filler wire (wt.%).

<table>
<thead>
<tr>
<th>Material</th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Si</th>
<th>Al</th>
<th>Cr</th>
<th>Mo</th>
<th>Nb</th>
<th>V</th>
</tr>
</thead>
<tbody>
<tr>
<td>S700 steel</td>
<td>0.068</td>
<td>1.476</td>
<td>0.009</td>
<td>0.001</td>
<td>0.05</td>
<td>0.073</td>
<td>0.19</td>
<td>0.03</td>
<td>0.044</td>
<td></td>
</tr>
<tr>
<td></td>
<td>0.0018</td>
<td>0.0045</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.46</td>
</tr>
<tr>
<td>Carbofil 2NiMoCr</td>
<td>0.08</td>
<td>1.5</td>
<td>0.6</td>
<td>≤ 0.015</td>
<td>≤ 0.018</td>
<td>0.4</td>
<td>2.2</td>
<td>0.6</td>
<td>Bal.</td>
<td>0.68</td>
</tr>
</tbody>
</table>

Figure 7.1 SEM micrograph of the base material showing bainitic microstructure.

Table 7.2 Mechanical properties of base S700 steel and Carbofil 2NiMoCr filler wire.

<table>
<thead>
<tr>
<th>Material</th>
<th>Yield strength 0.2% (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
<th>Impact value (J)</th>
</tr>
</thead>
<tbody>
<tr>
<td>S700 steel</td>
<td>≥ 700</td>
<td>760</td>
<td>12</td>
<td>≥ 42 (-40 °C)</td>
</tr>
<tr>
<td>Carbofil 2NiMoCr</td>
<td>≥ 960</td>
<td>≥ 940</td>
<td>≥ 15</td>
<td>≥ 47 (-40 °C)</td>
</tr>
</tbody>
</table>

Plates with dimensions of 200 (length) × 100 (width) × 13 mm (thickness) were machined from the received strip and, after butt welding, the samples measured 200 × 200 × 13 mm. A butt joint without a gap was prepared for single pass autogenous laser welding. A parallel groove configuration with a 2 mm thick root face was machined for the multi-pass ultra-narrow gap laser welding (NGLW), as shown in Figure 7.2(a). The groove width was 1.4 mm. In order to obtain full penetration in the GMA weld, a butt joint with a V-groove configuration was prepared, with a 1 mm thick root face, using a root gap of 1.6 mm, and a groove angle of 60°, as shown in Figure 7.2(b). The prepared
surfaces were sand blasted in the vicinity of the weld track in order to remove rust. Acetone was used to clean the surfaces after sand blasting. Before welding, all of the specimens were clamped on the work table using two clamping jaws at each transverse side of the plates to ensure adequate restraint, as shown in Figure 7.3. All of the welding operations were carried out using the same clamping system and the same restraint arrangements, in order to ensure a useful comparison. While some relaxation of the residual stresses will have taken place upon releasing the clamps, the clamping configuration was such that this will have had the largest effect on stresses acting transverse to the welding direction, and in this study we are focusing on stresses acting in the longitudinal direction. In all cases, the welding direction was perpendicular to the rolling direction of the base material.

![Figure 7.2 Schematic representation of weld groove: (a) groove shape for ultra-NGLW, (b) groove shape for GMAW.](image)

A Miller Axcess 450 gas metal arc welding (GMAW) power source was used for the manual GMA welding experiments using the flat (1G) welding position. A gas mixture comprising 80% argon and 20% CO₂ with a gas flow rate of 22 l/min was used, to protect the molten weld pool from oxidation, and to obtain a good penetration depth during the welding process. A continuous wave fibre laser (IPG YLS-16000) with a maximum output power of 16 kW was used for both the ALW in the horizontal (2G) welding position and the ultra-NGLW experiments in the 1G position. Melt sagging or drop-out is a common welding defect for single pass laser welding of thick section materials in the 1G position [28]. The surface tension of the molten metal cannot always compensate for the hydrostatic pressure in the melt for full penetration welding of thick
section materials in the 1G position. This can result in sagging on the root side and at the top surface of the weld when the specimen thickness is above a threshold [29, 30]. In this study, single pass ALW in the 2G position was successfully carried out on the 13 mm thick S700 steel plates, while avoiding the melt sagging problem. Ultra-NGLW of 13 mm thick S700 steel was carried out in the 1G welding position. The laser beam had an emission wavelength of 1070 nm. The beam parameter product for the laser was 10 mm mrad and the beam was optically delivered (with a fibre core diameter of 300 µm) to the output lenses. The laser beam emitted from the end of the optical fibre was collimated by a lens with a 150 mm focal length and then focused onto the specimen surface using a lens with a 400 mm focal length. The measured focal spot size and Rayleigh length were 0.8 mm and 15 mm, respectively. The laser head was mounted on a 6-axis KUKA robot. The filler wire was fed into the leading edge of the molten pool at an angle of 30° with respect to the specimen surface. Pure argon gas was blown on to the top surface of the weld pool through a copper tube at a flow rate of 12 l/min, and the backing gas was blown through a side tube into a blind square chamber beneath the specimens, at a flow rate of 8 l/min, to protect the back surface of the specimens. Schematic representations of the setup for the ALW and for the ultra-NGLW process are shown in Figure 7.3. The robot end-effector holds the laser head perpendicular to the workpiece and has the capability of moving in 3D space.

Cross-sectional samples were removed from the welded plates by wire electro-discharge machining (EDM). The samples were first ground and then polished to a 1 µm Ra surface finish. The macrostructure of the weld region was revealed by etching with 2%
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Nital solution. Assessments of the macrostructure and the microstructure for the welded joints were performed using a KEYENCE VHX-500F optical microscope and a Philips XL 30 scanning electron microscope (SEM).

Micro-indentation hardness profiles across the mid-thickness region of the welded joints were obtained using a 300 g test load and a dwell period of 15 s using a Vickers microhardness machine (Struers DuraScan 50). For the ALW and ultra-NGLW specimens, the fusion zone and heat affected zone (HAZ) were narrow, so an indent spacing of 0.2 mm was set in the area close to the fusion zone and HAZ, and the indent spacing for other areas further away from the fusion zone (FZ) and HAZ was set to be 0.4 mm. Owing to the relatively wide fusion zone and HAZ for the GMA welded specimen, the indent spacing was set to be 0.4 mm across the entire weld region.

The residual stress measurements on the top surface of the specimens were carried out using a Phillips portable X-ray diffractometer, based on chromium (Cr K-α) radiation and employing collimator with a diameter of 1 mm. The measurements were carried out according to the NPL Good Practice Guide [31]. The instrument is a computer controlled system and it has functions for data storage, data processing and the display of results. The position of the peaks arising from \{211\} diffracting planes was at a diffraction angle (2θ) of approximately 156.1° [31]. The detectors’ ψ tilt angles varied from −27 deg to +27 deg, using 11 scans. Peaks from different tilt angles were recorded and fitted by the cross-correlation method: from the change in peak positions the residual stresses were calculated using the sin²ψ method [31]. The error bars for the X-ray diffraction measurements were calculated based on the diffraction peaks, and the deviation of the diffraction peak fits from linearity, in d vs. sin²ψ plots. Electrolytic polishing was carried out to remove the oxide layer before the measurements took place. The X-ray diffraction (XRD) measurements were carried out on the top surface of the welded specimens across the mid-length of the weld bead (see Figure 7.4). A measurement interval of 1 mm was set in the vicinity of the fusion zone and HAZ and the measurement interval for other areas, further away from the fusion zone and HAZ, was set to be 10 mm.
The contour method measurements were carried out on the same welded specimens after completion of the surface X-ray diffraction measurements. Specimen cutting is the first and the most critical step in the contour method, as subsequent procedures such as the contour measurement, data reduction and stress analysis are critically dependent on the cutting quality [27, 32]. In particular, a precisely straight cut must be performed, additional material should not be removed from the cut surfaces and, as far as possible, the cut should not cause plastic deformation [33]. In this study, all of the welded plates were cut into two halves for the contour method measurements. The specimens were cut along a transverse plane at the mid-length position, as shown in Figure 7.4, to obtain a map of longitudinal stress. The cuts were made using an Agie Charmilles FI 440 ccS wire EDM, using a brass wire with a diameter of 250 μm. The part was submerged in temperature-controlled deionized water throughout the cutting process. A ‘skim cut’ or low-heat-input setting was used to minimize cutting-induced stresses. For each contour cut, the test specimen was symmetrically clamped to the bed of the EDM using fixtures placed close to the cut line. To prevent any thermal stresses, the weld specimen and all clamps were allowed to come to thermal equilibrium in the water tank prior to clamping. It has been proposed that rigid clamping close to the cut line will reduce plasticity errors. However, rigid clamping is often not feasible in practice [34].

After the wire EDM cutting, the cut specimens were removed from the clamping fixture. The surface deformations on both of the cut surfaces, owing to the relaxation of stresses, were measured using a NanoFocus laser scanner with a nominal accuracy of ±
0.1 \mu m. The deformed surfaces on each side of the measurement plane were measured with a measurement spacing of 200 \mu m \times 200 \mu m, so that there were roughly 65,000 data points for each surface.

Following the measurement of surface deformation, the measured displacement data for the two cut surfaces were averaged. This is essential in order to eliminate or at least minimize any effects that shear stresses may have had on the relaxation of stresses at the cut surfaces [23]. The resulting averaged surface contour is then numerically smoothed before inputting the data into a finite element (FE) simulation. Data smoothing is vital because any variation in the averaged profile due to surface roughness produces an amplified effect during the calculation of residual stresses. A level of smoothing is determined by choosing fitting orders during data reduction. One of the most popular data smoothing techniques is a 3D cubic-spline-based algorithm [35]. The algorithm is based on joining polynomials at constant intervals (knots) to form a smooth spline. The knot spacing is adjusted to determine the best fit to the data by means of a least-squares analysis [36].

Finally, the smoothed surface contours were input to a finite element (FE) model to calculate the residual stresses according to Bueckner’s superposition principle [23]. Using the averaged surface contour, half the original specimen was modelled in a linear elastic finite element analysis that applied the negative of the smoothed surface profile as a set of boundary conditions on the plane of the cut. The FE analysis was developed using ABAQUS 6.12. The sample geometry was meshed with approximately 124,384 quadratic (20 node) reduced integration hexahedral elements (C3D20R). The FE model assumed a homogeneous, isotropic, linearly elastic material with a Young’s modulus of 210 GPa and a value for Poisson’s ratio of 0.3.

7.3 Results

7.3.1 Welding parameters

The welding parameters for the manual GMAW, the single pass ALW and the multi-pass ultra-NGLW of S700 steel are shown in Table 7.3, Table 7.4 and Table 7.5, respectively. The optimised ALW welding parameters were developed in previous work
The ultra-NGLW parameters were also based on previous work by the authors, with the detailed optimisation procedure described in Ref. [37]. A comparison of the cumulative energy input and the consumption of filler material for the ALW, ultra-NGLW and GMAW techniques is given in Table 7.6.

Table 7.3 GMAW parameters for 13 mm thick S70 steel.

<table>
<thead>
<tr>
<th>Welding pass</th>
<th>Voltage (V)</th>
<th>Current (A)</th>
<th>Welding speed (m/min)</th>
<th>Wire feed rate (m/min)</th>
<th>Shielding gas flow rate (l/min)</th>
<th>Heat input (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1st</td>
<td>30</td>
<td>225</td>
<td>0.44</td>
<td>4.3</td>
<td>22</td>
<td>0.74</td>
</tr>
<tr>
<td>2nd</td>
<td>30</td>
<td>244</td>
<td>0.44</td>
<td>4.3</td>
<td>22</td>
<td>0.76</td>
</tr>
<tr>
<td>3rd (torch weave)</td>
<td>30</td>
<td>244</td>
<td>0.29</td>
<td>4.3</td>
<td>22</td>
<td>1.21</td>
</tr>
<tr>
<td>4th (offset to left)</td>
<td>30</td>
<td>242</td>
<td>0.47</td>
<td>4.3</td>
<td>22</td>
<td>0.74</td>
</tr>
<tr>
<td>5th (offset to right)</td>
<td>30</td>
<td>257</td>
<td>0.47</td>
<td>4.3</td>
<td>22</td>
<td>0.79</td>
</tr>
</tbody>
</table>

Table 7.4 ALW parameters for 13 mm thick S70 steel.

<table>
<thead>
<tr>
<th>Welding pass</th>
<th>Power (kW)</th>
<th>Welding speed (m/min)</th>
<th>Shielding gas flow rate (l/min)</th>
<th>Backing gas flow rate (l/min)</th>
<th>Focal position (mm)</th>
<th>Heat input (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Root pass</td>
<td>2</td>
<td>1.2</td>
<td>-</td>
<td>12</td>
<td>-11</td>
<td>0.08</td>
</tr>
<tr>
<td>Filling pass</td>
<td>3</td>
<td>0.6</td>
<td>3.3</td>
<td>12</td>
<td>8</td>
<td>0.17</td>
</tr>
<tr>
<td>Filling pass</td>
<td>3</td>
<td>0.6</td>
<td>3.3</td>
<td>12</td>
<td>8</td>
<td>0.17</td>
</tr>
<tr>
<td>Filling pass</td>
<td>3</td>
<td>0.6</td>
<td>3.3</td>
<td>12</td>
<td>8</td>
<td>0.17</td>
</tr>
</tbody>
</table>

Table 7.5 Ultra-NGLW parameters for 13 mm thick S70 steel.

<table>
<thead>
<tr>
<th>Welding pass</th>
<th>Power (kW)</th>
<th>Welding speed (m/min)</th>
<th>Wire feed rate (m/min)</th>
<th>Shielding gas flow rate (l/min)</th>
<th>Backing gas flow rate (l/min)</th>
<th>Focal position (mm)</th>
<th>Heat input (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Root pass</td>
<td>2</td>
<td>1.2</td>
<td>-</td>
<td>12</td>
<td>8</td>
<td>-11</td>
<td>0.08</td>
</tr>
<tr>
<td>Filling pass</td>
<td>3</td>
<td>0.6</td>
<td>3.3</td>
<td>12</td>
<td>8</td>
<td>0</td>
<td>0.17</td>
</tr>
<tr>
<td>Filling pass</td>
<td>3</td>
<td>0.6</td>
<td>3.3</td>
<td>12</td>
<td>8</td>
<td>+4</td>
<td>0.17</td>
</tr>
<tr>
<td>Filling pass</td>
<td>3</td>
<td>0.6</td>
<td>3.3</td>
<td>12</td>
<td>8</td>
<td>+8</td>
<td>0.17</td>
</tr>
</tbody>
</table>

Table 7.6 Comparison of energy input and consumed filler material for GMAW, ALW and NGLW techniques.

<table>
<thead>
<tr>
<th>Welding technique</th>
<th>Maximum power (kW)</th>
<th>No. of welding passes</th>
<th>welding speed (m/min)</th>
<th>Consumed filler material (g)</th>
<th>Cumulative heat input (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>GMAW</td>
<td>7.7</td>
<td>5</td>
<td>0.44</td>
<td>216</td>
<td>4.24</td>
</tr>
<tr>
<td>ALW</td>
<td>13</td>
<td>1</td>
<td>0.72</td>
<td>0</td>
<td>0.87</td>
</tr>
<tr>
<td>Ultra-NGLW</td>
<td>3</td>
<td>4</td>
<td>0.6</td>
<td>31</td>
<td>0.59</td>
</tr>
</tbody>
</table>

It can be seen from Table 7.4 that for the single pass ALW of 13 mm thick S700 steel, the laser power was 13 kW. However, a moderate laser power (3 kW) was applied for the ultra-NGLW of 13 mm thick S700 steel (Table 7.5), which means that this technique
can be applied with lower capital investment than is required for ALW. Heat input for arc welding can be calculated by the following equation: heat input = arc efficiency × (voltage × current) / welding speed. A consumable electrode process, such as GMAW, typically exhibits an average arc efficiency in the vicinity of 80% [38]. The heat input for laser welding is calculated by the equation: heat input = efficiency × laser power / welding speed. The welding efficiency considering only conduction heat losses is about 80% for autogenous laser welding [39, 40]. In addition, some beam reflection from the filler wire takes place in the ultra-NGLW process, which could result in 30% of the energy applied to the workpieces being lost [41].

It can be seen from Table 7.6 that the maximum power used for GMAW is 7.7 kW, and that the power used for ALW is 13 kW, while the power used for ultra-NGLW is 3 kW. The heat input for each welding pass for both GMAW and ALW is ~4-5 times greater than that of ultra-NGLW. The cumulative heat input for GMAW is ~5-7 times greater than for both ALW and ultra-NGLW. The welding speed for ALW and ultra-NGLW is approximately 1.4-1.6 times greater than that for GMAW, while the consumed filler material for GMAW is ~7 times greater than that for ultra-NGLW.

7.3.2 Macrostructure

Figure 7.5(a), (b) and (c) show the macrostructures of the single pass ALW, the multi-pass ultra-NGLW and the GMAW joints, respectively. The fusion zone of the ALW joint exhibits a ‘nail’-shaped cross section with a ‘nail head’ width of 5.4 mm, while the width between the parallel portions of the fusion boundaries on either side of the weld seam, toward the lower regions of the welded joint, was 2.2 mm. The width of the HAZ appears to extend ~2 mm on either side of the fusion zone. The ultra-NGLW joint exhibits a nearly rectangular fusion zone profile with a width of ~1.8 mm, and the width of the HAZ is ~1.2 mm. The multiple filling passes can be distinguished clearly in Figure 7.5(b). The GMAW specimen exhibits a ‘Vee’ shaped fusion zone. The top width of the fusion zone and the width of the HAZ for the GMAW specimen were approximately 22 mm and 4.4 mm, respectively. Owing to the higher cumulative heat input associated with the GMAW technique, when compared with the ALW and ultra-NGLW techniques, the width of the HAZ in the GMAW specimen is ~2 times that for the ALW specimen and ~4 times that for the ultra-NGLW specimen.
Clearly, the width of the fusion zone for the GMAW specimen is largely determined by the choice of the weld groove geometry. At the root of the weld, the fusion zone width is similar to the width of the fusion zone for the ALW specimen, but at the top surface, the width of the GMAW specimen is approximately 4-5 times greater than that of the ALW specimen. While it is possible to argue that a more appropriate comparison would have involved a narrow-groove GMAW weld groove geometry, in place of a conventional Vee-groove, such a geometry would generally not be applied in practice for a plate thickness of 13 mm, so the Vee-groove geometry was chosen on the basis that it is likely to be most representative of industrial practice at this thickness.

7.3.3 Microstructure

The microstructures in the FZ and HAZ for the GMAW, ALW and ultra-NGLW specimens are shown in Figure 7.6. The microstructure in the FZ for the GMAW joint is dominated by fine acicular ferrite with a small amount of martensite, as seen in Figure 7.6(a). Bainite and small amounts of martensite were observed in the HAZ for the GMAW specimen, as shown in Figure 7.6(b). Bainite and small quantities of martensite were formed in the FZ for the ALW specimen, as seen in Figure 7.6(c), while martensite
was formed in the HAZ for the ALW specimen (Figure 7.6(d)). Martensite can be observed in both the FZ and HAZ in the ultra-NGLW specimen, mixed with some auto-tempered martensite, as can be seen in Figure 7.6(e) and (f), respectively.

Figure 7.6 SEM micrographs showing the microstructures in the FZ and HAZ for the different welded specimens, (a) FZ of GMAW, (b) HAZ of GMAW, (c) FZ of ALW, (d) HAZ of ALW, (e) FZ of ultra-NGLW, (f) HAZ of ultra-NGLW.
7.3.4 Microhardness

The hardness profiles across the welded joints at the mid-thickness position are shown in Figure 7.7. The FZ in the GMAW specimen exhibits a slightly higher hardness (~ 280 HV$_{0.3}$) than the base material (~ 250 HV$_{0.3}$). However, a very wide (~ 4 mm) soft HAZ can be observed in the hardness profile for the GMAW joint, with a lowest Vickers hardness value of ~ 200 HV$_{0.3}$. In the ALW joint, the hardness increased dramatically when moving from the base material (BM) into the HAZ. The peak hardness occurred in the HAZ with a hardness value of ~ 335 HV$_{0.3}$. Interestingly, the FZ retained a similar hardness to that for the BM in the ALW joint. This is consistent with a predominantly bainitic microstructure being observed in the fusion zone (Figure 7.6(c)), whereas martensite was observed in the HAZ (Figure 7.6(d)). The hardness profile for the ultra-NGLW joint matched the visible extent of the HAZ boundaries, with the hardness rising rapidly when moving from the base material (~ 250 HV$_{0.3}$) to a plateau in the FZ of ~ 375 HV$_{0.3}$. The hardness level on this plateau is indicative of martensitic and bainitic microstructures, which were seen in Figure 7.6(e) and 6(f). In addition, both the ALW and ultra-NGLW joints exhibited very narrow (0.4 mm) softened regions within the HAZ, where minimum harness values were ~ 240 HV$_{0.3}$.

![Figure 7.7 Measured hardness profiles across the different welded joints at the mid-thickness position.](image-url)
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7.3.5  Residual stresses in the different welded joints

Figure 7.8 shows the distributions of residual stress acting in the longitudinal direction as measured by XRD, as a function of distance from the weld centreline, on the top surface of each sample. The residual stress results show an area of significant compressive stress (~ −150 MPa) toward the transverse edges of the plates, which was not expected, and this may have been introduced during plate manufacture, prior to welding. The measured residual stress profiles for the three welded specimens show a high degree of symmetry with respect to the weld centre-line. This gives some level of confidence in the results. In addition, the three residual stress profiles present similar distributions: the longitudinal residual stress shows a bimodal distribution (‘M’ shape) with a trough coincident with the weld centreline and the two peaks are roughly situated immediately beyond the HAZ-BM boundary on either side. The peak residual stresses for the GMAW, ALW and ultra-NGLW specimens were approximately 400, 600 and 650 MPa, respectively. The residual stress distribution for the GMAW specimen shows a much wider plateau region than is the case for the ALW and ultra-NGLW specimens. However, the residual stress in the plateau region for the GMAW specimen was generally between 100-200 MPa in tension, whereas for the ALW and ultra-NGLW specimens the corresponding stresses were around 300-500 MPa in tension.

![Residual stress profiles](image)

Figure 7.8 Residual stress profiles measured by XRD at the mid-length position and on the top surfaces of the welded S700 steel plates, (a) residual stresses as a function of distance from the weld centreline, (b) magnified residual stress profiles in the vicinity of the fusion zone and HAZ.
Cross-sectional maps of the residual stresses acting in the longitudinal direction, as measured by the contour method, are shown in Figure 7.9 and Figure 7.10 for the GMAW, ALW and ultra-NGLW specimens. The stress distributions exhibit a certain level of asymmetry with respect to the weld centreline, and this is most pronounced in the case of the GMAW specimen where, of course, the deposition sequence for the weld beads was asymmetric. However, the contour maps for the ALW and ultra-NGLW specimens also exhibit asymmetry to a lesser extent, and this is likely to be a consequence of plasticity during the cutting process, which may also affect the measured residual stresses [42]. In all cases, tensile stresses are seen in the vicinity of the FZ and HAZ, whereas compressive stresses are seen further away from the weld. The residual stress fields vary significantly across the weld and HAZ. The thin regions of compression within ~ 1 mm of the top surface are likely to be unreliable. These unrealistically high compressive stresses near to the top surfaces were most likely obtained due to plasticity and bulging errors in the cutting step [43] or, particularly in the case of the capping beads, due to poor fitting of the varying profile in these regions [32]. It was suggested by Prime et al. [43] that the near-surface stresses are very challenging to obtain reliably using the contour method. This is likely to apply to stresses measured within ~ 1 mm of the specimen surfaces.

Figure 7.9 Longitudinal residual stress maps as determined by the contour method for (a) GMAW, (b) ALW, (c) ultra-NGLW specimens (units of stress: MPa). In all cases, the cutting direction was from right to left.
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The GMAW specimen reveals a wider region of tensile residual stress than the ALW and ultra-NGLW specimens. There is a large concentrated region of tensile residual stress at the root of the FZ, with the measured peak stress being ~ 600 MPa in magnitude. The ALW specimen exhibits a narrower region of tensile residual stress than the GMAW specimen, with two small concentrated tensile residual stress regions which occur at ~ 2 mm below the top surface and another two such regions occurring within ~ 2 mm from the back surface, where peak stresses of ~ 700 MPa were measured. The ultra-NGLW specimen exhibits a narrower region of tensile residual stress than the ALW specimen. The stress distribution appears to be asymmetric, with peak stresses occurring to one side of the last pass, and these are ~ 800 MPa in magnitude. There is also a small but distinct region of tensile residual stress which appears to coincide with the second of the three filling passes in the ultra-NGLW joint, and this region has a peak value of ~ 600 MPa, which is close to the yield strength of the base material at room temperature.

Line plots for the longitudinal residual stresses in the GMAW, ALW and ultra-NGLW specimens are depicted in Figure 7.11. The line plots display the residual stresses that were measured at a distance of 3 mm below the top surface (Figure 7.11(a)) and at a
distance of 3 mm above the bottom surface (Figure 7.11(b)). Near to the top of the weld, the longitudinal residual stresses for the GMAW specimen are almost symmetric about the weld centreline. The longitudinal stress profile for the GMAW specimen exhibits a bimodal distribution (‘M’ shape) across the welded joint with a low stress trough (~ 0 MPa) at the weld centre flanked by tensile peaks (~ 600–700 MPa) roughly coinciding with the maximum extent of the HAZ. These observations are quite similar with those of Kundu et al. [44], who measured residual stress distributions in electron beam welded P91 steel plates. However, the peaks and troughs in the ‘M’ profiles for the ALW and the ultra-NGLW specimens are narrower and less pronounced, and they are also noticeably asymmetric.

Figure 7.11 Line plots showing longitudinal residual stress profiles for the GMAW, ALW and ultra-NGLW specimens at (a) a distance of 3 mm below the top surface of each sample and (b) a distance of 3 mm above the bottom surface of each sample.

7.4 Discussion

7.4.1 FZ and HAZ microstructures

The cooling rate is one of the most important influences on microstructure evolution in both the fusion zone and HAZ regions [45]. The higher the heat input, the slower the cooling rate [46], and vice versa. The highest proportion of martensite in the fusion zone was observed in the ultra-NGLW, and it can be seen from Table 7.3, Table 7.4 and Table 7.5 that the heat input per pass was significantly lower for ultra-NGLW that it was for the other processes. The fusion zone microstructure for the ALW sample was
predominantly bainite, while the fusion zone microstructure for the GMAW sample was predominantly acicular ferrite. The low fractions of martensite in the fusion zones for these samples can be attributed largely to the higher heat input for a given weld pass. While laser welding is generally associated with lower heat inputs per pass than GMAW, the ALW specimen required a high laser power and a low welding speed in order to penetrate the plate in a single pass, and this resulted in a relatively high heat input in this case.

Perhaps the most interesting feature in the weld microstructures is the contrast between the fusion zone microstructure in the ALW (bainite) and the HAZ microstructure (martensite) in the same sample. This is particularly interesting because this is an autogenous weld, so the nominal composition of the fusion zone should match the composition of the parent material. In a search for an explanation, the authors checked for obvious differences in the prior-austenite grain sizes in each microstructural zone but there did not appear to be a significant difference. Another important feature to note, however, is that the ALW was made with a high laser power (13 kW) – much higher than for ultra-NGLW (3 kW). It is possible that this high laser power led to the preferential evaporation of some alloying elements from the molten weld pool (e.g. Mn), thereby bringing about a change in the composition of the fusion zone, and in turn its hardenability. Such evaporation is known to occur [47], so this would appear to be a plausible explanation, but further work would be required to confirm this. Owing to the very low heat input and fast cooling rate in the ultra-NGLW process, martensite was formed in both the FZ and HAZ in the ultra-NGLW specimen, mixed with some tempered martensite, as seen in Figure 7.6(e) and (f), respectively.

### 7.4.2 Microhardness

The microhardness plots in Figure 7.7 reveal that the extent of the softened zone in the GMAW sample is significantly larger than it is in either the ALW or the ultra-NGLW samples. Although the heat inputs per pass for the GMAW and ALW samples were broadly similar, the higher power and high power density in ALW enabled a significant increase in the welding speed when compared to GMAW. This in turn results in a higher melting efficiency [19], which means that a greater proportion of the heat that is transferred to the material results in melting, rather than simply heating the surrounding
material. Both the extent and the degree of softening in the GMAW sample appear to be much greater than for either of the laser-based processes, as is evident through the minimum hardness in the GMAW joint (~ 200 HV$_{0.3}$). Furthermore, the softened zone in both laser welds appears to be only ~ 0.4 mm wide. The FZ in the GMAW sample has a slightly higher hardness value (~ 280 HV$_{0.3}$) than for the base material (~ 250 HV$_{0.3}$). This may be attributed to the fine acicular ferrite microstructure in the FZ and the overmatched filler wire used to fill the groove.

The hardness traverse for the ALW supports the observation that martensite is present in the HAZ while the fusion zone comprises a predominantly bainitic microstructure. The hardness traverse for the ultra-NGLW also seems consistent with the observation of predominantly martensitic microstructures in both the FZ and HAZ. Owing to the very low heat input for the ultra-NGLW process, the cooling rate in the FZ and HAZ for the ultra-NGLW joint would be expected to be very fast, which would favour the formation of hard martensite. The hardness rises rapidly from the base material level of ~ 250 HV$_{0.3}$ through the HAZ to a plateau of ~ 375 HV$_{0.3}$ in magnitude in the FZ for the ultra-NGLW joint.

It is well known that hardness can provide a good indication of strength. In general, an increase in hardness is associated with a corresponding increase in strength [48]. Through an investigation of the hardness and the yield strength in the sub-zones of the HAZ in SA508 steel, Lee et al. [49] found that the yield strength of the HAZ is proportional to the hardness. In accordance with Lee et al.’s [49] results, it can be surmised that the yield strength of the FZ for the ultra-NGLW joint might be ~ 1.5 times that of the base material, i.e. ~ 1050 MPa. The yield strength of the FZ for the GMAW specimen may be ~1.1 times of the yield strength of the base material at room temperature, i.e. ~ 770 MPa. On the other hand, the yield strength of the FZ for the ALW specimen may be the same as that of the base material at room temperature. These changes in properties are significant because the yield stress often limits the magnitude of residual stresses.
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7.4.3 Residual stress distributions in the welded joints

Many features of the residual stress distributions seem to be credible and consistent with findings that have been published previously. For example, the longitudinal stress profiles measured by XRD for the GMAW and ultra-NGLW samples exhibit an ‘M’ shape that is symmetric about the weld centreline, having tensile peaks roughly coinciding with the HAZ–BM interface. While the ‘M’ shaped stress distribution is less obvious in the case of the X-ray measurements on the ALW, which appear more like a top hat distribution, the ‘M’ shaped nature is more evident in the contour method measurements for this sample (Figure 7.10). The other contour method measurements also show evidence for ‘M’ shaped residual stress profiles.

The trough in the longitudinal residual stress profiles at the weld centre can be attributed to strains associated with the ferritic, bainitic and martensitic transformations, which include a volumetric expansion, during cooling of the regions of the weldment that were heated to above the Ac₁ temperature; that is the entire fusion zone and the majority of the HAZ. These observations and their explanation are consistent with those summarised in a review by Francis et al. [17], where results were presented in order to demonstrate that the strains associated with the bainitic and martensitic transformations can compensate for thermal contraction in the fusion zone and HAZ for ferritic steels, which usually can reverse or at least reduce the tensile stresses within the FZ. The M-shaped profiles contrast with those for a non-transforming material where a single tensile peak would be expected at or near to the weld centre, dropping to low-level compressive stresses in the far field [50].

It is worth pointing out that the extent to which the stresses are reduced in the trough region of the ‘M’ profile, due to the effects of solid-state phase transformations, is dependent on the transformation temperature for the steel on cooling, with lower transformation temperatures leading to less tensile or more compressive stresses [17]. S700 steel has a martensite-start temperature of ~ 460°C, and a bainite-start temperature of ~ 540°C when a cooling rate of 60°C s⁻¹ is applied. These temperatures are relatively high, so the stress-compensating effects offered by solid-state phase transformations will not be as great as they are in other steels which have lower transformation temperatures for the formation of the same phases.
The tensile peaks in stress are in the regions which roughly coincide with the maximum extent of the HAZ, \textit{i.e.} those regions heated to a temperature just below the \(\text{Ac}_1\) temperature, as these regions experience only thermal contraction during cooling of the welded joint \([51]\). Indeed, the extents of the tensile regions of residual stress tend to correlate with the extents of the metallurgical zones. Thus, for the GMAW specimen, which has a wide FZ and a wide HAZ, the tensile regions are significantly wider than they are for the ALW and ultra-NGLW specimens (Figure 7.10).

The highly tensile region located at the root of the GMAW joint is consistent with results published by Hosseinzadeh and Bouchard \([52]\), who measured the longitudinal residual stresses in a multi-pass girth welded pipe in P91 steel using the contour method. Interestingly, there is also a large region within the fusion zone, toward the centre and top of the GMAW specimen, that is sustaining very little stress. This region is likely to correspond to the last weld pass that was deposited, since this weld pass will have experienced the effects of the transformation strains during cooling without being subject to further thermal cycles. Similar observations were reported by Paddea \textit{et al.} \([53]\) for a submerged arc weld in P91 (ferritic/martensitic) steel.

The high power density of the laser heat source tends to produce much narrower tensile residual stress regions than those in the GMAW specimen. In the case of the ALW specimen, the peak tensile stresses are in the order of 800 MPa, which exceeds the yield strength of the base material at room temperature. A relatively high heat input was required to penetrate the 13 mm thick S700 steel plate using single pass ALW. This resulted in the width of the tensile region for the ALW being larger than that for the ultra-NGLW. Since the ALW was a single pass keyhole weld, one would expect the residual stress distribution would remain relatively consistent through the entire plate thickness. To a first order approximation this is the case. In contrast, the variation in the residual stresses in the through-thickness direction seems to be more pronounced in the case of the ultra-NGLW, which required three weld passes in order to complete the joint.

The much lower heat input for the ultra-NGLW process and very narrow FZ (due to the very narrow groove) result in a narrower tensile residual stress region than for the ALW specimen. There is evidence of an M-shaped residual stress distribution in the contour
map presented in Figure 7.10, although the stress distribution is noticeably asymmetric. There is no underlying reason for the stress distribution to exhibit such asymmetry, and the authors believe that this is an artefact resulting from plasticity during the cutting process. In all of the contour maps presented in Figure 7.10, the cutting direction was from left to right. Asymmetry of this type is also evident in the contour maps for residual stresses in the GMAW and ALW samples, albeit to a lesser extent. There is also no underlying reason for the stress distribution to be asymmetric for the ALW sample. Based on these observations, the authors believe that the contour maps presented show a tendency to underestimate the stresses on the left hand side of the weld centreline and over estimate the stresses on the right hand side. This assertion is supported by the X-ray measurements (Figure 7.8), which do not reveal any asymmetry in the stresses at the top surface.

There are reasons for the stress distribution to vary in the through-thickness direction for the ultra-NGLW sample, as three weld passes were required in order to complete the joint. Since the through-thickness direction was perpendicular to the cutting direction, the authors believe the variations in stress in the through-thickness direction are likely to be more reliable. An autogenous laser weld pass was applied to fuse the root, using a very low heat input (0.08 kJ/mm), and the extent of the tensile stresses appears to be reduced towards the bottom of the ultra-NGLW.

Some of the tensile stresses that were measured using the contour method exceed the yield stress of the S700 parent steel. Deng and Murakawa [54] demonstrated that the yield strength of the weld metal has a marked effect on the final longitudinal residual stress in the weld zone and its vicinity. They found that higher yield strengths due to martensitic transformations in the weld can result in increased residual stresses in GMAW of modified 9Cr–1Mo steel pipes. James et al. [55] reported that the application of an overmatched filler material produced a higher tensile residual stress in the weld when compared to using an undermatched filler material in their investigation on residual stresses in GMAW of RQT701 high strength steel plates. With these findings in mind, the magnitudes of the measured stresses seem plausible.

When using the contour method it is possible for the peak tensile stresses to be underestimated owing to errors associated with plasticity during the cutting operation.
There are some signs that the results presented in this work have been affected by plasticity to some extent. For example, it is very likely that for the ALW and ultra-NGLW, the true stress distributions are likely to be symmetric or very nearly symmetric about the weld centreline. It is also very likely that the stress distribution presented for the ALW sample in Figure 7.10(b) is closer to being symmetric about the weld centreline, as opposed to being slightly offset from the weld centreline. In contour method measurements, it is also possible to ‘over smooth’ the surface contour when undertaking the spline fitting operation. To date, there is no established best practice for this step and this must therefore remain as a source of uncertainty in the measurements. Nevertheless, the authors took great care with the processing of the contour data to ensure that the spline fitting procedure achieved adequate resolution.

7.5 Conclusions

The following conclusions can be drawn from this work:

(1) In GMAW and laser welding of S700 steel, solid-state phase transformations have noticeable effects on the longitudinal residual stress distributions in the vicinity of the weld, through promoting an ‘M’ shaped residual stress distribution. This contrasts with a typical profile for a non-transforming material, where a single tensile peak would be expected at or near to the weld centreline.

(2) In the case where a single pass keyhole weld is made, such as with autogenous laser welding, the ‘M’ shaped residual stress distribution is likely to appear through the thickness of the plate being welded. If, however, a multipass welding technique is applied the ‘M’ shaped residual stress distribution is less likely to be present through the entire plate thickness.

(3) The peak residual stresses in GMAW and laser welds in S700 steel tend to arise to either side of the weld centreline, coinciding approximately with the HAZ–BM interface. The magnitude of the peak stresses appears to be in the order of the yield stress of the parent steel (~700 MPa), but can be somewhat higher due to the formation of strong, hard phases within the weld fusion zone and heat-affected zone.
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(4) The extent of the tensile regions appears to correlate with the extent of the metallurgical zones in the welds produced with each process. As such, the GMAW specimen exhibited tensile stresses over a wider region than both the ALW and ultra-NGLW specimens. The ultra-NGLW specimen, owing to its low heat input, produced the narrowest regions of tensile stress.

(5) There are significant benefits associated with choosing a laser based welding process over GMAW for the welding of S700 steel. These include the fact that the extent of softening in the heat-affected zone is significantly reduced through using a laser-based process. In addition, laser-based welding processes offer significant productivity gains, since the travel speeds are lower with GMAW and a greater number of weld passes is required for a joint of the same thickness. The consumption of filler material is also greatly reduced in the case of ultra-NGLW, or eliminated altogether in the case of single-pass autogenous laser welding.

References

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[28] Bachmann M, Avilov V, Gumenyuk A, Rethmeier M. Experimental and numerical investigation of an electromagnetic weld pool support system for high power laser beam


Chapter 8: General discussion

8.1 Microstructure characteristics

The microstructure variation within the different zones of a welded joint is dependent on the local thermal cycles. The peak temperature that is reached during welding and cooling rate are particularly important factors influencing microstructure evolution in either fusion zone or HAZ regions [1]. The higher the heat input, the slower the cooling rate [2], and vice versa. Heat input is a major parameter affecting the microstructure and mechanical properties of the fusion zone and HAZ, which can be controlled by suitably selecting welding speed and input power [3, 4]. In this research, for 8 mm thick S960 steel, the heat input of each welding pass for GMAW was ~ 0.5–0.8 kJ/mm, the heat input for single pass autogenous laser welding (ALW) was ~ 0.33 kJ/mm, and the heat input of each welding pass for ultra-narrow gap laser welding (NGLW) was ~ 0.1 kJ/mm. The heat input for each welding pass for GMAW was approximately 1.5–2.4 and 5–8 times that of single pass ALW and ultra-NGLW, respectively. For 13 mm thick S700 steel, the heat input of each welding pass for GMAW was ~ 0.7–1.2 kJ/mm, the heat input for single pass ALW was ~ 0.87 kJ/mm, and the heat input of each welding pass for ultra-NGLW is ~ 0.17 kJ/mm. The GMAW presents comparative heat input for each welding pass with that of the ALW. The heat input for each welding pass for the GMAW is ~ 4–7 times that of the ultra-NGLW. The heat input for each welding pass for the ALW is ~ 5 times that of the ultra-NGLW. Low heat input was introduced into the specimens for the single pass ALW of 8 mm thick S960 steel and multi-pass ultra-NGLW of 8 mm thick S960 steel and 13 mm thick S700 steel. It was reported that autogenous laser welding has lower line energy than arc welding processes and higher cooling rates (2000–3000 °C/s) [5].

Owing to the fast cooling rate in the single pass ALW and multi-pass ultra-NGLW of 8 mm thick S960 steel, the transformed austenite in the HAZ and FZ was quenched. Martensite was formed in these sub-zones. Because of the high Ms temperature (~ 450 °C) for S960 steel, some formed martensite has more chance to be auto-tempered in the following cooling process. These sub-zones included predominantly martensite in conjunction with some tempered martensite. The grain size varied gradually from the FZ to the FGHAZ because of the gradually reduced peak temperature. For the multi-pass ultra-NGLW of the 8 mm thick S960 steel, more tempered martensite was observed in the lower
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part of the welded joint, especially the region between the two welding passes. This phenomenon can be attributed to the additional heat effect introduced from the subsequent welding pass to the prior welding pass. This newly generated thermal cycle has a temper effect on the prior welding pass.

For the GMAW of 8 mm thick S960 steel, the arc has a relatively broad heating area and the higher heat input introduced into the joint makes it cooling rate much slower than the ALW and ultra-NGLW joints. In the slow cooling rate the newly formed austenite in the HAZ transformed into bainite, some small amount of martensite and tempered martensite were also formed in the HAZ. The grain size also reduced gradually with the increase of the distance to the FZ because of the gradually reduced peak temperature. Because the FZ cools from the melt, the cooling rate in the FZ is relatively slower than in the HAZ. Abundant fine acicular ferrite with some martensite were formed in the cooling process.

Due to the fast cooling rate for the multi-pass ultra-NGLW of 13 mm thick S700 steel, martensite was formed in both the FZ and HAZ in the ultra-NGLW specimen, mixed with some tempered martensite. While laser welding is generally associated with lower heat inputs per pass than GMAW, but the ALW specimen required a high laser power and a low welding speed in order to penetrate the 13 mm thick S700 steel plate in a single pass, and this resulted in a relatively high heat input in this case. The weld microstructures are the contrast between the fusion zone microstructure in the ALW (bainite) and the HAZ microstructure (martensite) in the same sample. This is particularly interesting because this is an autogenous weld, so the nominal composition of the fusion zone should match the composition of the parent material. This may because the ALW was made with a high laser power (13 kW) – much higher than for ultra-NGLW (3kW). It is possible that this high laser power led to the preferential evaporation of some alloying elements from the molten weld pool (e.g. Mn), thereby bringing about a change in the composition of the fusion zone, and in turn its hardenability. Such evaporation is known to occur [6]. The evaporation of manganese in the weld would have effect on promotion of bainitic transformation and suppression of martensitic generation [7, 8].

The slower cooling rate for GMAW of 13 mm thick S700 steel resulted in almost the same microstructure in the FZ and HAZ as those for the GMAW of 8 mm thick S960 steel. The microstructure in the FZ for the GMAW joint was predominated of fine acicular ferrite
with a mixture of small amount of martensite. Bainite with a small amount of martensite were generated in the HAZ for the GMAW specimen.

8.2 Effect of microstructure evolution on the mechanical properties of the welded joints

The fast cooling rates associated with laser welding result in the generation of martensite in the FZ and HAZ. This results in a marked increase of hardness in the FZ and HAZ for the single pass ALW of 8 mm thick S960 steel and multi-pass ultra-NGLW of 8 mm thick S960 and 13 mm thick S700 steel. The high power single pass ALW of 13 mm thick S700 steel presented the microstructure predominated by bainite in the FZ and martensite in the HAZ. The FZ retained almost the same hardness as the base material for the single pass ALW 13 mm thick S700 steel joint. There was a dramatically hardness increase in the HAZ for the single pass ALW 13 mm thick S700 steel joint. The over tempering in the SCHAZ for the single pass ALW and multi-pass ultra-NGLW process led to development of a narrow soft region in the SCHAZ, which showed a slightly lower hardness when compared with the base materials.

The arc for GMAW has a broad heating source. This in turn results in a low melting efficiency, which means that a small proportion of the heat that is transferred to the material results in melting, rather than simply heating the surrounding material. In addition, GMAW introduced more heat input to the welded joint, which can lead to a slower cooling rate. The microstructure in the FZ for both of the GMAW of S960 and S700 steel were predominated by acicular ferrite, which resulted in the hardness in the FZ being comparable or a little higher than that of the base material. Due to the slower cooling rate and the subsequent tempering effect from the multi-pass welding process, the HAZ presented a much wider and softer HAZ than that of the ultra-NGLW joint. In addition, the soft HAZ for both the GMAW of S960 and S700 steel showed much lower hardness than that of the base material.

There is an approximately proportional relationship between strength and hardness in steels [9]. In general, increase in hardness results in increase in strength whereas decreasing formability or ductility [10]. It can be deduced from the hardness results for the single pass ALW of 8 mm thick S960 steel and multi-pass ultra-NGLW of 8 mm thick S960 steel and 13 mm thick S700 steel that the FZ and HAZ (except the SCHAZ) were
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strengthened mainly by the generation of martensite. It can be deduced from the hardness results for the single pass ALW of 13 mm thick S700 steel that the FZ could have a comparable strength with that of the base material, while the HAZ was markedly strengthened by the generation of martensite when compared with the base material. However, a very narrow SCHAZ would be the weakest region for the single pass ALW of S960 and S700 steel, as well as the multi-pass ultra-NGLW S960 and S700 steel.

It can also be deduced from the hardness results for the GMAW of S960 and S700 steel that the FZ could have a comparable or slightly higher strength when compared to the base material. The wide soft HAZ would make the GMAW joints lose strength.

In the transverse tensile test, all the single pass ALW and multi-pass ultra-NGLW S960 steel failed in the base material well away from the FZ with almost the same tensile properties as the base material. However, the GMAW S960 steel failed in the soft HAZ showed inferior tensile properties when compared to the base material. The high hardness in the FZ and HAZ for the single pass ALW and multi-pass ultra-NGLW S960 steel resulted in greater yield strength than the BM. Conversely, both the yield and ultimate tensile strength of the GMAW joint are considerably reduced by the softness of its HAZ.

In the transverse tensile test, it is apparent that the single pass ALW and multi-pass ultra-NGLW S960 steel presented almost the same tensile properties as the base material with failure in the base material not in the narrow soft SCHAZ, whereas the GMAW S960 steel resulted in a substantial deterioration of all tensile properties with failure in the soft HAZ. The higher hardness in the FZ and HAZ (with the exception of the SCHAZ) for the single pass ALW and multi-pass ultra-NGLW S960 steel resulted in greater yield strength than the base material, while the SCHAZ was a slightly softer than the base material, but this region was very narrow (~ 0.6–0.8 mm). The hard FZ and the hard regions of the HAZ can therefore act as a strong constraint on plastic deformation in the adjacent SCHAZ, thereby leading to the majority of the tensile plastic deformation accumulating in the base material and subsequent fracture in the BM [11-13]. However, it should be mentioned that this plastic deformation constrain by the surrounding strengthened material happens only in the case when the softened zone is narrow and the degree of softening is relative low. Otherwise, the wide softened zone as well as low hardness will result in the decreasing of the tensile strength of the welded joint [12]. Thereby, failure therefore occurred in the soft HAZ for the GMAW specimens.

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The generation of hard martensite in the FZ and HAZ strengthened the single pass ALW and multi-pass ultra-NGLW S960 steel joints but also increased the brittleness of the joints. The slower cooling rate in the GMAW of S960 steel deteriorated the strength of the welded joint, but showed better impact toughness when compared with the single pass ALW and multi-pass ultra-NGLW S960 steel. However, post weld heat treatment could be applied to improve the toughness of the laser welded joints.

### 8.3 Effects of welding parameters on the weld quality of the NGLW process

The weld bead integrity and weld bead width are two of the most important factors to consider in assessing the quality of a weld. The number of filling passes embodies the efficiency with which the filler wire fills the groove in ultra-narrow gap laser welding. From the predictions of the statistical model, it was found that the weld quality and the welding efficiency can be improved through increasing the laser power and the wire feed rate, and reducing the welding speed.

The integrity of the weld bead can be improved by increasing the laser power and the wire feed rate simultaneously. The higher feed rate will enable the molten filler wire to fill the groove completely when employing a relatively higher laser power, which will help in avoiding lack of fusion. The slower welding combined with higher wire feed rate could increase the wire deposition rate for each pass. The higher deposition rate for each pass will assist in resisting shrinkage of the weld bead, which will lead to less distortion and make it easier to deposit filler material in the following weld passes, thereby enabling the groove to be filled successfully. Furthermore, owing to the slower welding speed, the molten filler wire will solidify slowly and this can make it easy for pores to escape from the molten pool, leading to less porosity in the weld.

The weld bead width increases with increases in both the laser power and the wire feed rate. Increasing either the heat input or the cross-sectional area of the weld bead will lead to the production of larger weld beads and higher bead widths. Reductions in the welding speed and increases in the wire feed rate also increase the weld bead width. Here the effect of the welding speed can also be explained in terms of its influence on the weld heat input.
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Increases in the wire feed rate and reductions in the welding speed lead to an increase in the volume of filler material that is deposited per unit length and, consequently, the average height of each filling pass. This will mean that a weld of a given thickness will be completed in fewer passes. Increasing the wire deposition rate will also lead to reduced joint completion times.

8.4 Comparison of flat (1G) and horizontal (2G) positions single pass autogenous laser welding process

Melt sagging or drop-out is a common welding defect for single pass laser welding of thick section materials in the 1G position [14]. In the 1G position, gravity was acting in the same direction as the laser beam, which aided the flow of metal towards the lower part of the weld pool. The surface tension of the molten metal cannot always compensate for the hydrostatic pressure in the melt for full penetration welding of thick section materials in the 1G position. This can result in sagging on the root side of the weld when the specimen thickness is above a threshold. In contrast, gravity was acting in a direction that was perpendicular to the penetration direction for the 2G welding position. This enabled the surface tension to balance the pressure in the molten pool. The molten metal remained stable in the weld pool with full penetration. The pressure in the molten pool was much lower for the 2G welding position when compared with that for the 1G position, which was beneficial to obtain sound welds without melt sagging defect.

8.5 Effect of solid-state phase transformation on residual stress

The measured longitudinal residual stress distributions for the GMAW, ALW and ultra-NGLW samples exhibit an ‘M’ shaped residual stress profiles, having tensile peaks roughly coinciding with the HAZ–BM interface. The trough in the longitudinal residual stress profiles at the weld centre can be attributed to strains associated with the ferritic, bainitic and martensitic transformations, which include a volumetric expansion, during cooling of the regions of the weldment that were heated to above the $A_{c1}$ temperature; that is the entire fusion zone and the majority of the HAZ. These observations and their explanation are consistent with those summarised in a review by Francis et al. [15], where results were presented in order to demonstrate that the strains associated with the bainitic and martensitic transformations can compensate for thermal contraction in the fusion zone and HAZ for ferritic steels, which usually can reverse or at least reduce the tensile stresses.
within the FZ. The tensile peaks in stress are in the regions which roughly coincide with
the maximum extent of the HAZ, i.e. those regions heated to a temperature just below the
Ac$_1$ temperature, as these regions experience only thermal contraction during cooling of
the welded joint [16].

The extent to which the stresses are reduced in the trough region of the ‘M’ profile, due to
the effects of solid-state phase transformations, is dependent on the transformation
temperature for the steel on cooling, with lower transformation temperatures leading to less
tensile or more compressive stresses [15]. S700 steel has a martensite-start temperature of
~ 460°C, and a bainite-start temperature of ~ 540°C when a cooling rate of 60 ºC s$^{-1}$ is
applied. These temperatures are relatively high, so the stress-compensating effects offered
by solid-state phase transformations will not be as great as they are in other steels which
have lower transformation temperatures for the formation of the same phases.

Some of the tensile stresses that were measured using the contour method exceed the yield
stress of the S700 base material. Deng and Murakawa [17] demonstrated that the yield
strength of the weld metal has a marked effect on the final longitudinal residual stress in
the weld zone and its vicinity. They found that higher yield strengths due to martensitic
transformations in the weld can result in increased residual stresses in GMAW of modified
9Cr–1Mo steel pipes. James et al. [18] reported that the application of an overmatched
filler material produced a higher tensile residual stress in the weld when compared to using
an undermatched filler material in their investigation on residual stresses in GMAW of
RQT701 high strength steel plates. With these findings in mind, the magnitudes of the
measured stresses seem plausible.

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Chapter 9: Conclusions and future work

9.1 Conclusions

An investigation was conducted on the development and comparison of single pass autogenous laser welding (ALW), multi-pass ultra-narrow gap laser welding (NGLW) and gas metal arc welding (GMAW) of newly developed high strength low alloy steels – S960 and S700 in collaboration with Tata Steel. The materials were 8 mm in thickness for S960 and 13 mm in thickness for S700 high strength low alloy (HSLA) steels. These steels have minimum yield strengths of 960 MPa and 700 MPa, respectively. Both of this two HSLA steels are typically used in heavy lifting equipment. Design of experiments and statistical modelling were applied to optimise the welding parameters to obtain sound welds. Microstructure evolution in the fusion zone and heat affected zone (HAZ) for these two HSLA steels welded by these three different welding techniques were investigated and compared. Mechanical properties including micro-hardness, tensile strengths, three-point bending properties and Charpy impact toughness at different temperatures were evaluated and compared between these three different welded joints. The contour method and X-ray diffraction were applied to measure the residual stresses of the welded specimens. The research provides useful scientific data to support the industrial applications of these two HSLA steels. The main findings and conclusions presented in the thesis are summarised as follows:

9.1.1 Welding techniques

(1) It is possible to weld 8 mm thick S960 HSLA steel plates without any macroscopic defects, using a single pass autogenous laser welding (ALW) process at flat position (1G).

(2) It is difficult to obtain sound single pass autogenous laser welds in 13 mm thick S700 steel without deep undercut defects at the 1G position due to weld pool drop-outs driven by gravity. However, it was demonstrated that high quality single-pass autogenous laser welds in 13 mm thick S700 steel can be obtained using the horizontal position (2G).
(3) The computational fluid dynamic (CFD) modelling results demonstrate that the pressure in the molten pool at the 2G welding position is lower when compared with the 1G welding position. It is beneficial to balancing the pressure in the molten pool and obtaining sound welds without melt sagging defect when welding the 13 mm thick S700 steel plates at 2G welding position.

(4) Design of experiments and statistical modelling were successfully carried out to optimise multi-pass ultra-NGLW of 8 mm thick S960 and 13 mm thick S700 steel plates. A relatively moderate laser power (2-3 kW) and a narrow (1.2-1.4 mm) parallel groove were used in this welding technique, which could obviate the need for expensive very high power laser for welding medium and thick section materials using single pass autogenous laser welding technique.

(5) The multi-pass ultra-NGLW technique provides an alternative solution to solving the melt sagging problem when welding medium and thick section materials.

9.1.2 Microstructures

(1) The microstructure of the fusion zone (FZ) and heat affected zone (HAZ) for the single pass ALW of 8 mm thick S960 steel and multi-pass ultra-NGLW of 8 mm thick S960 and 13 mm thick S700 steel were predominantly martensite with some auto-tempered martensite.

(2) The microstructure of the FZ for the single pass ALW of 13 mm thick S700 steel was predominantly bainite mixed with a small amount of martensite, while martensite was the dominated microstructure with some bainite in the HAZ.

(3) The microstructure of the FZs for both the GMAW of 8 mm thick S960 and 13 mm thick S700 steels were predominantly fine acicular ferrite mixed with a small amount of martensite, while the HAZs were predominated by bainite with a mixture of small amount of martensite.
9.1.3 Mechanical properties

(1) The generation of hard martensite in the FZs and HAZs for the single pass ALW of 8 mm thick S960 steel and multi-pass ultra-NGLW of 8 mm thick S960 and 13 mm thick S700 steel led to a great increase of hardness in the FZs and HAZs.

(2) The dominated microstructure in the FZ for the single pass ALW of 13 mm thick S700 steel was bainite, which retained almost the same hardness as the base material. While the hard martensite in the HAZ caused a markedly increase of hardness in the HAZ.

(3) A very narrow soft region (~0.4-0.8 mm) occurred in HAZ for both the single pass ALW and multi-pass ultra-NGLW S960 and S700 steel. The hardness in these narrow soft HAZ was a slightly lower than the base material and these soft HAZ was adjacent to the base material.

(4) The hardness in the FZ for both the GMAW of 8 mm thick S960 and 13 mm thick S700 steels demonstrated a comparable or a slightly higher hardness than the base material. However, both the GMAW of 8 mm thick S960 steel and 13 mm thick S700 steel presented a much wider (~ 4 mm) soft HAZ than laser welded specimens, with hardness much lower than the base material.

(5) The tensile strength for both the single pass ALW and multi-pass ultra-NGLW of 8 mm thick S960 steel were comparable to the strength of the BM, with all welded specimens failed in the base material even there was a very narrow soft HAZ. While the GMAW specimens failed in the soft HAZ with an approximately 100 MPa lower strength than the base material.

(6) All the single pass ALW, multi-pass ultra-NGLW and GMAW of 8 mm thick S960 steel performed well in three-point bending tests.

(7) The GMAW of 8 mm thick S960 steel joint demonstrated better impact toughness than both the single pass ALW and multi-pass ultra-NGLW joints. The generation of
hard martensite in the FZs and HAZs for the ALW and ultra-NGLW joints results in increase in strength whereas decreasing the toughness of the welded joints.

9.1.4 Residual stresses

(1) In the GMAW and laser welded S700 steel, solid-state phase transformations have noticeable effects on the longitudinal residual stress distributions in the vicinity of the weld, through promoting an ‘M’ shaped residual stress distribution.

(2) The peak residual stresses in the GMAW and laser welds in S700 steel tend to arise to either side of the weld centreline, coinciding approximately with the HAZ–BM interface. The magnitude of the peak stresses appears to be in the order of the yield stress of the parent steel (~ 700 MPa), but can be somewhat higher due to the formation of strong, hard phases within the weld fusion zone and heat-affected zone.

(3) The extent of the tensile regions appears to correlate with the extent of the metallurgical zones in the welds produced with each process. As such, the GMAW specimen exhibited tensile stresses over a wider region than both the ALW and ultra-NGLW specimens. The ultra-NGLW specimen, owing to its low heat input, produced the narrowest regions of tensile stress.

9.2 Future work

This research focused on examining the feasibility and characteristics of single pass ALW, multi-pass ultra-NGLW of 8 mm thick S960 and 13 mm thick S700 HSLA steels and compared with those of traditional GMAW. The residual stresses, microstructures, microhardness, tensile properties, bending properties and Charpy impact toughness at different temperatures of these three different welded specimens were evaluated to provide data to support the industrial application of these two HSLA steel. However, there are still several areas in which further research is required. Some of the most important are included as follows.
9.2.1  **Toughness of the laser welded HSLA steel joints**

The toughness of the single pass ALW, multi-pass ultra-NGLW and GMAW was evaluated by Charpy impact tests at different temperatures. In order to fully demonstrate the toughness of the joint, the notches were sampled at the weld centre, fusion line and HAZ. However, it was found from the test results that both the single pass ALW and multi-pass ultra-NGLW of 8 mm thick S960 steel demonstrated worse toughness than the GMAW joints, especially at lower temperature, *i.e.* -40 °C and -20 °C. The low toughness of the laser welded joints is attributed to the hard martensite microstructure in the FZ and HAZ, which strengthen the laser welded joints, but deteriorated the toughness of the welded joints. It is necessary to develop a reasonable local post weld heat treatment (PWHT) sequences to do local tempering in the laser welded joints to improve the toughness of the laser welded HSLA steel joints without loss the strength of the laser welded joints.

9.2.2  **Fatigue properties**

The heavy lifting equipment will involve variable load under the real service environment. Fatigue property is a very important mechanical properties index to evaluate the lifespan of the heavy lifting equipment. The fatigue property of the single pass ALW and multi-pass ultra-NGLW HSLA steels should be evaluated and compare with GMAW specimens.

9.2.3  **Multi-physical numerical modelling and simulation of NGLW process**

To deeply understand the NGLW process, the future work should include numerical modelling and simulation. The hydrodynamic of molten pool, vaporization of the metal in the molten pool, the thermal effect of laser irradiating, and their interactions should be investigated by CFD (for fluid simulation). The residual stress measurement procedures are complex and the measurement processes are time consuming. A coupled thermal-stress model should be developed to investigate the residual stress distribution in the future work. NGLW of high strength steels involve very complex solid-state phase transformation and multi-pass interactions, which has a dramatic effect on the
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final residual stress results. In order to accurately simulate the residual stress induced by
NGLW of high strength steels, the solid-state phase transformation and multi-pass
interactions should be taken into account in the finite element model.

9.2.4 Understanding the precipitates generated in the FZ and HAZ

There were many precipitates generated in the FZ and HAZ, especially in the lasered
welded specimens. These precipitates were generated by self-tempering and tempering
by the subsequent welding passes for the ultra-NGLW process, which have a significant
effect on the mechanical properties of the welded joints. Transmission electron
microscopy (TEM) can be used to investigate the precipitates in the future.
Appendix A:

The work shown in the Appendix was carried out during my PhD on laser welding of SA508 nuclear pressure vessel steel and compared with traditional gas tungsten arc welding (GTAW). This work is part of the New Nuclear Manufacturing (NNUMAN) programme. SA508 steel is applied for the nuclear reactor pressure vessels, which is different from the S960 and S700 high strength low alloy steels. This work also investigated the microstructure and mechanical characteristics of the laser welded SA508 steel joint, as is shown in the Appendix.
Appendix A

Microstructure and mechanical characteristics of a laser welded joint in SA508 nuclear pressure vessel steel

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Abstract

SA508 steels are typically used in civil nuclear reactors for critical components such as the reactor pressure vessel. Nuclear components are commonly joined using arc welding processes, but with design lives for prospective new build projects exceeding 60 years, new welding technologies are being sought. In this exploratory study, for the first time, autogenous laser welding was carried out on 6 mm thick SA508 Cl.3 steel sheets using a 16 kW fiber laser system operating at a power of 4 kW. The microstructure and mechanical properties (including microhardness, tensile strength, elongation, and Charpy impact toughness) were characterized and the microstructures were compared with those produced through arc welding. A three-dimensional transient model based on a moving volumetric heat source model was also developed to simulate the laser welding thermal cycles in order to estimate the cooling rates included by the process. Preliminary results suggest that the laser welding process can produce welds that are free of macroscopic defects, while the strength and toughness of the laser welded joint in this study matched the values that were obtained for the parent material in the as-welded condition.

Keywords: nuclear steels, pressuriser, pressurised water reactor, reactor pressure vessel, structural integrity, weld toughness

A.1 Introduction

The life span and safe operation of the reactor pressure vessel (RPV), which is one of most critical components in nuclear power plant, depends on durability of the RPV
Appendix A

materials in high temperature, high pressure and radioactive environments. The need for materials with higher strength, toughness and resistance to irradiation embrittlement is rising due to increases in the power generation capacity and the design lives of nuclear power plants [1-8]. SA508 steel grades have been used in the manufacture of many RPV’s for pressurized water reactors because they offer the combination of strength, good ductility, fracture toughness, homogeneity with respect to mechanical properties, and they are economic [9-12]. RPVs are fabricated by welding thick ring-forgings or plates of SA508 steel together. These are typically realized using arc welding, which is followed by post-weld heat treatment in order to restore toughness in the heat affected zone (HAZ). While arc welding technologies are well established for these components, an increase in the availability of high power lasers, which are capable of welding moderately thick sections of steel with much higher welding speeds and reduced distortion, provides an incentive to consider whether laser welding would offer any advantages in welding components manufactured from SA508 steels.

The conventional welding methods applied in manufacturing the nuclear pressure vessels are gas-tungsten arc welding (GTAW) and sub-merged arc welding (SAW) [13-15]. In Verón et al.’s [14] research to assess the susceptibility to stress-relief cracking in the HAZ, multi-pass SAW was employed to weld 140 mm thick SA508 Class 2 steel with a heat input of 1.8 kJ/mm for each pass. Kim et al. [16] reported conventional SAW welding of SA 508 class 3 steel with a heat input of 3 kJ/mm per pass. Murty et al. [13] found that for a multi-pass submerged arc weld in SA533B steel, the widths of the weld metal and the HAZ were 26 and 12 mm, respectively. Logsdon [17] welded 64 mm thick SA533 Gr. A Cl. 2 steel plates using multi-pass narrow gap GTAW with a 10 mm wide groove and a heat input of 1.6 kJ/mm for each pass. It can be seen that these traditional welding techniques generally employ higher heat inputs than laser welding, which will increase the width of the HAZ and may lead to larger distortions and higher residual stresses after welding. This will be compounded if more weld passes and the addition of more filler material are required, due to the employment of a wider weld groove, and these factors may also contribute to increased production costs.

Compared with traditional arc welding technique, laser welding has its own advantages, such as a high power density, and the associated ability to make a narrow weld with a narrow HAZ, using a low heat input with a high welding speed, and achieving lower
Appendix A

levels of residual stress and distortion, while consuming less filler material [18, 19]. In addition, laser welding can be realised using remote control, because a laser beam can be delivered using an optical fiber and the welding head can be mounted on an industrial robot. This characteristic makes laser welding amenable to producing the high quality welds that are required in nuclear environments. Indeed, the application of laser welding to moderately thick sections of austenitic stainless steel has been explored previously. Zhang et al. [20] first reported multi-passes narrow gap welding of 50 mm thick type 316L plates with an 8 kW disk laser. Elmesalamy et al. [21] successfully welded 20 mm thick 316L stainless steel using a 1 kW IPG single mode fibre laser with an ultra-narrow gap (1.5 mm gap width), welding from both sides using multiple-pass narrow gap approach. In spite of this, no work has been reported on the characteristics of laser welds in SA508 steel.

The solid-state phase transformations that take place during the welding of low alloy steels can be very complex, and in some steels it can be very difficult to predict the resulting microstructures in the different sub zones of welded joints. The cooling rates in the different sub zones will determine the phase transformations that take place during the welding process in combination with the continuous cooling transformation (CCT) diagram for the steel under investigation. The temperature histories during the welding process can be recorded using thermocouples. However, thermocouples can only measure the temperature histories at discrete points. It can also be difficult to guarantee that the temperatures at all of the measurement locations can be recorded correctly. Finite element modelling (FEM) is an alternative method for investigating the thermal cycles during the welding process.

In this investigation, single pass autogenous laser welding was developed to join SA508 Cl. 3 steel plates. Autogenous GTA welding was carried out to provide a benchmark for the laser welding of this steel. The microstructures and mechanical properties, such as tensile strength, microhardness, and Charpy impact toughness of the welded components were investigated in the as-welded condition. Simulations based on a moving volumetric heat source model were also carried out to quantify the effects of the welding thermal cycles on the microstructural variation across the autogenous laser welds in SA508 steel. The numerical solutions were generated using the commercial software package ANSYS, and they were compared with experimental results to
validate the numerical model. The validated model was then used to predict the thermal history of the laser welding. This article describes the experiments and modelling, and reports on the preliminary findings arising from this work.

A.2 Material and experimental procedures

The as-received base material (BM) used in this study was quenched and tempered SA508 Cl.3 steel. The chemical composition of SA508 Cl. 3 steel is presented in Table A.1. The carbon equivalent (CE) is a parameter that is often used to evaluate the weldability of steels, and it is defined to convert the percentage of alloying elements other than carbon to the equivalent carbon concentration, from the standpoint of the hardenability of the steel. The CE of the investigated steel was calculated according to Ref. [22], and is given by:

\[
CE = C + \frac{Mn}{6} + \frac{(Cr + Mo + V)}{5} + \frac{(Ni + Cu)}{15}
\]  

(A.1)

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Ni</th>
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<td>0.04</td>
<td>0.02</td>
<td>Bal.</td>
<td>0.60</td>
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</table>

Table A.1 Chemical composition of SA508 Cl.3 steel (wt. %).

As seen from Table A.1, the CE of SA508 is 0.60. The Ms (martensite start) temperature is around 420 °C according to Suzuki’s continuous cooling transformation (CCT) diagram for SA508 Grade 3. Cl. 1 steel [23]. The AC\textsubscript{1} and AC\textsubscript{3} temperatures are approximately 700 °C and 800 °C, respectively. An optical micrograph and a scanning electron microscope (SEM) image of the microstructure of the base SA508 steel are shown in Figure A.1(a) and (b), respectively. The specimens were mechanically polished and etched in 2 % Nital solution. The microstructure of the base material (BM) is a tempered upper bainitic structure. The fine precipitates have been identified by various researchers, and they are known to be $M_7C_3$ or $M_{23}C_6$ [6, 12, 24].
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Figure A.1 Microstructure of the base material: (a) optical micrograph and (b) SEM micrograph.

The as received SA508 Cl.3 block was sliced into several 6 mm and 2 mm thick plates by wire electric discharge machining (EDM) for welding experiments. The material dimensions for autogenous laser welding were approximately 6 mm × 50 mm × 100 mm and for manual autogenous GTA welding were approximately 2 mm × 50 mm × 100 mm.

The experiments were carried out using a continuous wave fiber laser (IPG YLS-16000) with a maximum power of 16 kW. The beam parameter product was 10 mm mrad for a processing fiber 300 µm in diameter. The laser beam emitted from the end of the optical fiber was collimated by a lens with a 150 mm focal length and then focused onto the specimen surface using a lens with a 400 mm focal length. The measured focus size and Rayleigh length were 0.8 mm and 15 mm, respectively. The laser head was mounted on a 6-axis KUKA robot. A schematic illustration of the laser welding set up is shown in Figure A.2.

Figure A.2 Schematic of the laser welding configurations.
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A Miller Dynasty 350 GTA welding power source was used for the manual autogenous GTA welding experiments. Prior to welding, the samples were sand blasted to remove the oxide layer. After the sand blasting, acetone was used to clean the surface and then the base materials were clamped to ensure adequate restraint was applied. Autogenous laser welding and GTA welding was then carried out. To protect the molten weld pool during the welding process, the top and the back surface of the specimens were shielded using argon gas.

The macrostructure of the welded joints and the microstructure of welds were examined using an optical metallurgical microscope (KEYENCE VHX-500F) and Philips XL 30 scanning electron microscope (SEM). Surface hardness measurements were conducted using a Struers Duramin-2 Vickers microhardness tester.

Microindentation hardness profiles were measured across the weld and they were located on the top, middle and bottom of a macrograph section across the laser welded joint, and across the weld at the mid thickness position of the plate for the manual GTA welded joint. The hardness was measured using a load of 300 g and a dwell period of 15 s with a Vickers microhardness machine (Struers Duramin-2). Three measurements were carried out for each indent to minimize the error. The hardness traverses were carried out across the weld with indentations at intervals of 0.2 mm in the fusion zone and HAZ, and at intervals of 0.4 mm in the BM.

The specimens for the evaluation of the static tensile strength of the as-received base material and the welded sample were produced according to ASTM E8M-04. Sub-size Charpy impact test samples were prepared following the intent of BS EN 10045-1:1990. The notch was located in the fusion zone to test the toughness of the weld metal for the laser welded samples. The width of these Charpy specimens was limited by the thickness of the plates being welded, i.e. to 6 mm. Each test was repeated on three separate and nominally identical coupons to minimize uncertainties. Charpy and cross-weld tensile specimens were extracted from the steady state region of the welds using the EDM process. The sizes and shapes of the samples extracted from the base material and welded samples are shown in Figure A.3. The weld reinforcements in the face and root regions of the weld specimens were removed by manual grinding with abrasive paper before the tensile and Charpy impact tests were carried out. Tensile tests were
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carried out on an Instron model No. 4507 electronic universal test machine at room
temperature. Charpy impact tests for both the base material and welded samples were
carried out on a Zwick Roell Charpy impact test machine at -40 °C, -20 °C, 0 °C and at
room temperature. Each coupon was held at the relevant test temperature for half an
hour prior to the test to ensure that the temperature was uniform throughout the
specimen. Following the tensile and Charpy impact tests, all the fracture surfaces of the
tested specimens were examined using a Zeiss EVO 50 SEM installed with an Energy
Dispersive X-ray Detector (EDX), to study the fracture morphology and to determine
the fracture mode.

Figure A.3 Dimensions of tensile and Charpy impact test samples, (a) tensile test sample, (b) Charpy
impact test sample.

Initial trails were carried out using a bead-on-plate configuration, as opposed to joining
two distinct plates, in order to optimise the welding parameters. A laser power of 4 kW
was selected and the welding speed varied from 0.84 m/min to 1.08 m/min. The laser
focal point was set 2 mm below the top surface of the plate. The top surface and under
side of the weld were protected using argon shielding gas with gas flow rates of 12
l/min and 8 l/min, respectively. The laser head was tilted by 8 ° to guard against
reflection.

After welding, the welds were cut and prepared as metallographic samples to assess the
integrity of the weld beads. The results for different welding parameters are presented in
Figure A.4.
Figure A.4 Weld bead cross sections in 6 mm SA508 sheets, using a laser power of 4 kW and different welding speeds: (a) 0.84 m/min, (b) 0.9 m/min, (c) 0.96 m/min, (d) 1.02 m/min and (e) 1.08 m/min.

Examination of the weld cross sections for the different sets of welding parameters revealed that acceptable weld bead profiles were achieved with welding speed of 0.84 m/min, 0.9 m/min and 0.96 m/min. Undercut at the top of the weld metal region was observed at a speed of 1.02 m/min, and a lack of penetration was observed at a speed of 1.08 m/min. After considering the effects of welding speed on weld quality, a speed of 0.96 m/min was selected. The optimised welding parameters are summarised in Table A.2. The autogenous laser butt welding of 6 mm SA508 steel was carried out using these optimised welding parameters.
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Table A.2 Autogenous laser welding parameters for 6 mm thick SA508 steel.

<table>
<thead>
<tr>
<th>Power (kW)</th>
<th>Welding speed (m/min)</th>
<th>Top shielding gas flow (l/min)</th>
<th>Back shielding gas flow (l/min)</th>
<th>Focal position (mm)</th>
<th>Heat input (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>4</td>
<td>0.96</td>
<td>12</td>
<td>8</td>
<td>-2</td>
<td>0.25</td>
</tr>
</tbody>
</table>

A Miller Dynasty 350 GTA welding power supply was used to weld the 2 mm thick SA508 steel plates. Manual autogenous GTA welding was carried out in order to provide the best comparison with the autogenous laser welds. Plates with a thickness of 2 mm were used owing to the inherently shallow penetration of GTA welding, and double sided welding was carried out. The welding parameters are summarised in Table A.3.

Table A.3 Parameters for manual GTA welding of SA508 steel.

<table>
<thead>
<tr>
<th>Welding pass</th>
<th>Voltage (V)</th>
<th>Current (A)</th>
<th>Welding speed (m/min)</th>
<th>Shielding gas flow (l/min)</th>
<th>Heat input (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>9.5</td>
<td>50</td>
<td>0.13</td>
<td>10</td>
<td>0.22</td>
</tr>
<tr>
<td>2</td>
<td>10</td>
<td>57</td>
<td>0.16</td>
<td>10</td>
<td>0.21</td>
</tr>
</tbody>
</table>

A.3 Results

A.3.1 Macrostructure characteristics

The macrostructure of the autogenous laser butt welded SA508 steel joint, using the optimised parameters, is shown in Figure A.5. It can be seen that the fusion lines on either side of the weld are almost parallel, which is a characteristic of keyhole welding. There is no evidence of defects such as porosity or undercut. The width of the weld is approximately 1.8 mm, and the width of the HAZ is approximately 0.8 mm. The joint can be split into several distinct metallurgical regions, such as the fusion zone (FZ) in the centre, the heat affected zone (HAZ) and the base material (BM). The fusion zone consists of coarse columnar dendritic grains which are aligned in a direction that is perpendicular to the fusion boundary. The direction of maximum heat flow is perpendicular to the fusion boundary, so grains tend to grow fastest in that direction, which results in columnar grains in the fusion zone [25, 26]. Under higher magnification in an optical microscope, it was observed that the grain size varied with distance from the weld centre line. The HAZ can be divided further into three different
zones: the coarse-grained HAZ (CGHAZ) (closest to the fusion line), the fine-grained HAZ (FGHAZ) and the intercritical HAZ (ICHAZ) adjacent to the BM.

![Macrograph section through the autogenous laser butt weld in 6 mm thick SA508 steel.](image1)

Figure A.5 A macrograph section through the autogenous laser butt weld in 6 mm thick SA508 steel.

A macrograph section through the manual autogenous GTA weld in 2 mm thick SA508 steel is shown in Figure A.6. Owing to the limited penetration depths achieved during GTA welding, double sided autogenous manual GTA welding was applied. The width of the fusion zone is approximately 2.4 mm, and the width of the HAZ is approximately 2.8 mm. The widths of the fusion zone and HAZ are much wider than those of the 6 mm thick laser weld.

![Macrograph section through the GTA weld in 2 mm thick SA508 steel.](image2)

Figure A.6 A macrograph section through the GTA weld in 2 mm thick SA508 steel.

### A.3.2 Microstructural characteristics

The microstructural evolution in each sub-zone of the welded joint is primarily determined by the peak temperature during the welding thermal cycle, and the cooling rate in each corresponding sub-zone [27, 28]. The as-welded microstructures within the 6 mm thick laser weld and the 2 mm thick manual autogenous GTA weld were
examined in the fusion zone and in each of the different sub-zones within the HAZ (CGHAZ, FGHAZ, ICHAZ) using an SEM and the results are presented in Figure A.7 and Figure A.8, respectively. The microstructures in the different sub-zones within the HAZ were similar for both welding processes. However, more fine precipitates are found in the HAZ of the GTA weld when compared to the laser welded joint. In the work of Kim et al. [29, 30], the fine precipitates were identified to be M$_2$C-type carbides with high Mo content. In both welds, the microstructures in the ICHAZ comprised bainite, marteniste and auto-tempered martensite. In the FGHAZ, the microstructures included fine grained martensite with auto-tempered martensite. In the CGHAZ, the microstructures comprised coarse grained martensite with auto-tempered martensite, while in the fusion zone, coarse martensite and auto-tempered martensite were also observed.

![Figure A.7 SEM micrographs showing the microstructures in the different sub-zones of the laser welded joint (a) ICHAZ, (b) FGHAZ, (c) CGHAZ, (d) fusion zone, (prior austenite grain boundaries are highlighted with arrows).](image-url)
Figure A.8 SEM micrographs showing the microstructures in the different sub-zones of the GTA welded joint (a) ICHAZ, (b) FGHAZ, (c) CGHAZ, (d) fusion zone, (prior austenite grain boundaries are highlighted with arrows).

**A.3.3 Microhardness**

The as-welded microhardness distributions across the laser weld and the manual GTA weld are presented in Figure A.9. It can be seen that the hardness in the weld and HAZ for both the laser (~ 430 HV$_{0.3}$) and GTA (~ 410 HV$_{0.3}$) welded samples is more than twice that of the base material (~ 200 HV$_{0.3}$). This is to be expected for welds remaining in the as-welded condition. The hardness in the fusion zone is slightly higher than that of the HAZ for laser welded specimens. The hardness in the fusion zone and HAZ for the GTA welded joint fluctuates around 410 HV$_{0.3}$, and the peak hardness occurs in the FGHAZ, being approximately 430 HV$_{0.3}$. The hardnesses in the fusion zone and HAZ of the laser welded joint (~ 430 HV$_{0.3}$) are higher than that in the fusion zone for the GTA welded joint (~ 410 HV$_{0.3}$).
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A.3.4 Tensile behaviour at room temperature

Cross-weld tensile data for parameters such as the apparent yield strength, the tensile strength and the apparent elongation, for both the 2 mm thick GTAW specimens and the 6 mm thick laser welded specimens are summarized in Table A.4, which includes both the average values and standard deviations. It should be borne in mind that the specimens are clearly not homogeneous, so the values recorded for the yield strength and the elongation are not truly representative of any particular microstructural zone, and they will also vary with the choice of gauge length (25 mm in this case). Nevertheless in this study the measured values are included to provide a qualitative comparison of each weld. The apparent yield strength (YS), ultimate tensile strength (UTS) and apparent elongation were estimated to be 494 MPa, 631 MPa and 26.3%, respectively, for the 6 mm thick laser welded specimens. All the tensile failures occurred in the BM away from the weld region. The YS, UTS and elongation for the 6 mm thick base material were 498 MPa, 632 MPa and 28.1%, respectively. In comparison, the apparent yield strength (YS), ultimate tensile strength (UTS) and apparent elongation for 2 mm thick GTAW specimens were estimated to be 498 MPa, 633 MPa and 17.1%, respectively. All the tensile failures occurred in the BM away from the weld region. The YS, UTS and elongation for the 2 mm thick base material were 501 MPa, 633 MPa and 19.3%, respectively.
Table A.4 Results of tensile testing on the welded SA508 specimen and the base material.

<table>
<thead>
<tr>
<th>Test specimens</th>
<th>Yield strength (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Base material (6 mm)</td>
<td>498.2 ± 1.5</td>
<td>632.1 ± 0.9</td>
<td>28.1 ± 0.4</td>
</tr>
<tr>
<td>Laser welded (6 mm)</td>
<td>493.6 ± 1.1*</td>
<td>630.6 ± 0.8</td>
<td>26.3 ± 0.8*</td>
</tr>
<tr>
<td>Base material (2 mm)</td>
<td>500.7 ± 2.9</td>
<td>633.1 ± 2.8</td>
<td>19.3 ± 0.4</td>
</tr>
<tr>
<td>GTAW (2 mm)</td>
<td>497.9 ±0.8*</td>
<td>633.1 ± 0.9</td>
<td>17.1 ± 0.3*</td>
</tr>
</tbody>
</table>

Note: *denotes apparent values for cross-weld specimens.

The fractured specimens are presented in Figure A.10. The recorded stress-strain curves for the base materials and the welded specimens with thicknesses of 2 mm and 6 mm are shown in Figure A.11. It can be seen from the tensile test results that the laser and GTA welded specimens have very similar tensile properties to those of the base material at the corresponding thickness. However, the apparent elongations of the welded specimens are slightly lower when compared with those for the corresponding base materials. The hardness profiles in Figure A.9 show that, in the as-welded condition, the material has been strengthened by the welding process, so it is likely that the weld region did not yield during the tensile test, thereby contributing to reduced elongation. In addition, it can be seen from the tensile test results that the thickness of the material has almost no effect on the YS and UTS, with the 2 mm thick and 6 mm thick materials presenting similar YS and UTS. Curiously, the thickness of the material did have a significant effect on the elongation, with the thinner material (2 mm thick) presenting lower elongation when compared with the 6 mm thick material.
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Figure A.10 Fractured specimens, (a) 6 mm thick BM, (b) 6 mm thick laser welded specimens, (c) 2 mm thick BM, (d) 2 mm thick GTAW specimens.

Figure A.11 The recorded stress-strain curves for 2 mm and 6 mm thick base and welded SA508 steel.

The fracture surfaces for both the base material and the cross-welded tensile specimens were examined in an SEM, and they are shown in Figure A.12. The welded specimens failed in the base material. The fracture surfaces of the welded specimens are therefore similar to those for the base material. Typical ‘cup-and-cone’ fracture surfaces can be seen from the low magnification SEM photographs of the 6 mm thick specimens. A typical fibrous region is located toward the centre and smaller shear lip regions were observed towards the edges. In comparison, the low magnification SEM photographs of the 2 mm thick specimens show no ‘cup-and-cone’ fracture surfaces. Under high magnification, all the fractographs are composed predominately of equiaxed dimples. This indicates that all the base material and welded specimens failed in a ductile manner under the action of tensile loading.
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Figure A.12 Fracture surfaces of the base material and laser welded specimens after tensile testing, (a) macrostructure of the failed 6 mm thick base material, (b) high magnification image toward centre of fractured surface for 6 mm thick base material, (c) high magnification image toward edge of fractured surface for 6 mm thick base material, (d) macrostructure of the failed 6 mm thick laser welded specimen, (e) high magnification image toward centre of fractured surface for 6 mm thick laser welded specimen, (f) high magnification image toward edge of fractured surface for 6 mm thick laser welded specimen, (g) macrostructure of the failed 2 mm thick base material, (h) high magnification image of fractured surface for 2 mm thick base material, (i) macrostructure of the failed 2 mm thick GTAW specimen, (j) high magnification image of fractured surface for 2 mm thick GTAW specimen.

A.3.5 Charpy impact toughness at different temperatures

The energy absorbed by the base metal and by the laser weld under impact is plotted as a function of temperature in Figure A.13. The sub-sized fractured samples after the Charpy impact test are shown in Figure A.14. It can be seen that the fracture paths of all the laser welded test samples start in the fusion zone and then deviate to the HAZ and base material. The test results for the base material were repeatable, while the test results for the laser welded samples showed significant scatter at low test temperatures (-40 °C and -20 °C), which may be attributed to the deviation of the crack during fracture. In order to highlight the scatter in the results for the laser welded specimens, the three tested results for each temperature are marked in Figure A.13 and Figure A.14. Many researchers have reported that the laser and electron beam welding processes may present difficulties for toughness testing owing to the narrow fusion zones, together with a higher degree of strength overmatching of the joint [31-34]. Elliott reported that the tendency for the fracture to deviate into the base metal rather than propagate through the fusion zone can lead to misleading results [35].
Figure A.13 Absorbed energies in Charpy impact tests on sub-sized specimens extracted from the 6 mm thick base material and the fusion zone of autogenous laser welded SA508 steel.
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Figure A.14 Fractured Charpy impact test samples, (a) base material (-40 °C), (b) laser welded specimens (-40 °C), (c) base material (-20 °C), (d) laser welded specimens (-20 °C), (e) base material (0 °C), (f) laser welded specimens (0 °C), (g) base material (23 °C), (h) laser welded specimens (23 °C).

The results for the base material show an overall trend: the absorbed energies increase with an increase in the test temperature. In contrast, the scatter in the results for the laser weld means that any such trend is not obvious. The base material achieves good toughness, with average absorbed energy values of approximately 70 J, 95 J, 97 J and 105 J at corresponding test temperatures of -40 °C, -20 °C, 0 °C and 23 °C, respectively.

It can be found from the Charpy impact test results that the average absorbed energies of the laser weld specimens were comparable to those of the base material. The average absorbed energies values for the laser welded specimens were approximately 92 J, 80 J, 100 J and 98 J at the corresponding test temperatures of -40 °C, -20 °C, 0 °C and 23 °C, respectively. However, there were isolated lower absorbed energy values of 66 J at -40 °C and 45 J at -20 °C for the laser welded specimens, despite the average values at each of these temperatures being approximately 100 J. These isolated low toughness values contributed to the large scatter in the absorbed energy values for laser welded specimens at low test temperatures.

The macroscopic fracture surfaces of the base material and the laser welded specimens after Charpy impact testing are presented in Figure A.15. For the base material tested at -40 °C (Figure A.15(a)), it can be seen that the cracks initially propagated from the notch in a ductile manner before continuing to propagate through the specimen in a brittle fracture. The boundary between the region of ductile fracture and the region of subsequent brittle fracture is clearly evident in Figure A.15(a). The brittle fracture region spanned approximately 60% of the fracture surface as a whole. The fractured
laser welded specimens at -40 °C (Figure A.15(b)) reveal two very different fracture surfaces: the sample on the left presents a completely ductile fracture surface and it achieved a high absorbed energy (102 J), while the sample on the right reveals that the crack initially propagated in a ductile manner before continuing to propagated in a brittle manner over the majority (~ 60 %) of the fracture surface, and the absorbed energy was significantly lower in this specimen (66 J). The fracture base material specimen tested at -20 °C presents completely ductile fracture surface in Figure A.15(c). The fractured laser welded specimens tested at -20 °C (Figure A.15(d)) again present two very different fracture surfaces: the sample on the left presents a completely ductile fracture surface (84 J), while the sample on the right reveals a completely brittle fracture surface (45 J). The base material and laser welded specimens tested at 0 °C and room temperature present completely ductile fracture surfaces in all remaining cases, as shown in Figure A.15(e)-(h).
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Figure A.15 Macroscopic fracture surfaces of the base material and laser welded specimens after Charpy impact testing, (a) base material (-40 °C), (b) laser weld (-40 °C), (c) base material (20 °C), (d) laser weld (20 °C), (e) base material (0 °C), (f) laser welded (0 °C), (g) base material (23 °C), (h) laser welded (23 °C).

The fracture surfaces of the base material and the laser welded specimens after Charpy impact testing are shown at higher magnification in Figure A.16. Cleavage fracture was confirmed in those base material and laser welded specimens which fractured with a low absorbed energy at -40 °C. The brittle regions on the fracture surfaces revealed a cleavage dominated fracture with a small proportion of dimples (Figure A.16(a) and (c)). In contrast, the laser welded specimens which achieved a higher absorbed energy at -40 °C revealed a ductile fracture surface with equiaxed dimples (Figure A.16(b)). At -20 °C, both the base material and laser welded specimens which achieved higher absorbed energies presented ductile fracture surfaces with equiaxed dimples in Figure A.16(d) and (e), while the laser welded specimen with a lower absorbed energy presented a fracture surface dominated by cleavage (Figure A.16(f)). All of the base material and laser welded specimens tested at 0 °C and at room temperature presented ductile fracture with equiaxed dimples in Figure A.16(g)-(j).
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Figure A.16 High magnification fracture surfaces of the base material and laser welded specimens after Charpy impact testing, (a) base material (-40 °C), (b) laser weld with high absorbed energy (-40 °C), (c) laser weld with low absorbed energy (-40 °C), (d) base material (-20 °C), (e) laser weld with high absorbed energy (-20 °C), (f) laser weld with low absorbed energy (-20 °C), (g) base material (0 °C), (h) laser welded (0 °C), (i) base material (23 °C), (j) laser welded (23 °C).

A.3.6 Formulation and procedures for three-dimensional finite element modelling of the autogenous laser welding process

It is important to understanding the temperature fields in the autogenous laser welding of SA508 steel when studying microstructural evolution during the welding process, and this is particularly so in the case of the HAZ. In constructing a numerical model to predict the thermal histories in different sub-zones during the welding process, the following assumptions were made in order to simplify the solution [36]:

(1) The material is isotropic, and the ambient temperature and initial specimen temperatures were both 20 °C.

(2) The convective flow of liquid metal in weld pool and the vaporization that occurs in keyhole laser welding can be ignored.

(3) Heat flow during the welding process is governed by the effects of conduction and convection only, i.e. the effects of radiation can be ignored. In addition, the convection coefficient at the interface between the specimens and the environment can be assumed to be constant.
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(4) Owing to the symmetry of welded joint, symmetry can be applied, so that it is only necessary to simulate one side of the welded joint.

The model dimensions were 50 mm × 50 mm × 6 mm. Figure A.17 shows the mesh configuration. There were 38337 nodes and 41040 elements in the 3D solid model which was generated using ANSYS software (Version 12.1). A fine mesh was employed in the vicinity of the fusion zone and the HAZ where steep temperature gradients can be expected, whereas a coarser mesh was used further away from the weld and HAZ where the gradients are likely to be less severe. In addition, the size of the elements increased gradually with distance from the weld centre line, with the smallest element size being 0.5 mm × 0.5 mm × 0.5 mm. In this model, the X-axis corresponds to the welding direction, the Y-axis is normal to the welding direction but within the plane of the plate, and the Z direction is the out-of-plane direction.

![Figure A.17 FEM mesh for 3D model used in ANSYS simulations.](image)

The thermal analysis was performed using temperature dependent thermal properties. The transient temperature, \( T \), was determined as a function of in time, \( t \), and space (x, y, z) by solving the following heat transfer equation [37, 38]:

\[
\frac{\partial}{\partial x} \left( k(T) \frac{\partial T}{\partial x} \right) + \frac{\partial}{\partial y} \left( k(T) \frac{\partial T}{\partial y} \right) + \frac{\partial}{\partial z} \left( k(T) \frac{\partial T}{\partial z} \right) + Q_v = \rho(T)C_p(T) \frac{\partial T}{\partial t} \]  
(A.2)

Here, \( k(T) \) is the thermal conductivity as a function of temperature in Wm\(^{-1}\)K\(^{-1}\), \( \rho(T) \) is the density as a function of temperature in kgm\(^{-3}\), \( C_p(T) \) is the specific heat at constant
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pressure as a function of temperature in Jkg$^{-1}$K$^{-1}$, and $Q_v$ is the volumetric heat flux in Wm$^{-3}$.

A high power laser beam is a highly concentrated heat source, and heat source models are often used with variations in power density at various penetration depths in the numerical analysis of laser beam welding. In many papers [39-41], the heat source is assumed to take the form of a Gaussian distribution, but it is usually modified on the basis of experimental studies. There is a well-recognized “keyhole” phenomenon [39, 42, 43] in which some of the laser power is absorbed by the ionized vapour in the keyhole and transferred to the weld pool surface, which is also the “keyhole” boundary. So a volumetric heat source model is often used for simulating the laser welding process. In the volumetric heat source model, a Gaussian heat flux distribution is often assumed in the radial direction and the “keyhole” is considered to be a cylinder or truncated cone [39]. In this investigation, a rotational paraboloidal volumetric heat source was used to simulate the temperature field. The power distribution followed a Gaussian heat flux distribution on each layer of the rotational paraboloid. The heat source can be described as [44]:

$$q = \frac{6\eta P (z_e - z_i) e^3}{\pi (e^3 - 1) H^2 r_e^2} \exp \left( -\frac{3r^2}{r_0^2} \right) r_0 = r_e \sqrt{\frac{z - z_i}{z_e - z}}, \quad (A.3)$$

where, $q$ is the power density for any point on the rotational paraboloidal volumetric heat source, and where the heat source efficiency, $\eta$, was assumed to be 80% in the thermal analysis [38], $z_e$ is the maximum possible value of the vertical coordinate on the paraboloid, $z_i$ is the minimum possible value of this vertical coordinate, $H$ is the height of the paraboloid, $r_e$ is the opening radius of the paraboloid, $r_0$ is the radius of any point of the paraboloid, $r$ is the distance from any point within the rotational paraboloidal volumetric heat source to the heat source centre, $P$ is the output laser power and $z$ is the coordinate in the out-of-plane direction, with respect to the plate, for any point in the model. The thermo-physical properties for the material used are given in literature [45].

During the thermal analysis, a convective boundary condition was applied to all free surfaces of the model except for the plane of symmetry, where an adiabatic boundary
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A condition was imposed. Equation (A.4) presents the thermal boundary condition during the simulation.

\[ q_{\text{conv}} = h_{\text{conv}}(T - T_0) \]

Here, \( T \) and \( T_0 \) are the temperature at the surface of the plate being welded and the ambient temperature, respectively. The convective heat transfer coefficient of air, \( h_{\text{conv}} \) was assumed to be \( 15 \text{ W K}^{-1} \text{ m}^{-2} \) [38].

In order to validate the simulation results, both the experimental measured thermal cycles and fusion zone morphology were compared with those predictions arising from the simulation. The thermal cycles within the laser welded specimens were measured continuously throughout the welding process using K-type thermocouples. A Squirrel-2040 series data logger was used to record the thermal histories during the welding process. The thermocouples were spot welded to the top surface of the plate, and they were located at different distances from the weld centre line, on a line perpendicular to the welding direction and half way along the length of the weld, as shown in Figure A.18.

![Figure A.18 Locations of the thermocouples on the top surface of the specimen.](image)

Based on the spatial distribution of peak temperature, the weld morphology and size can be predicted. The simulated transverse cross-section of the autogenous laser weld is shown in Figure A.19. If it is assumed that the fusion boundary corresponds to a temperature of approximately \( 1500 \) °C, then it can be seen that the fusion boundary is
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predicted to be approximately parallel to the through-thickness direction of the plate, and that the weld half-width is approximately 1 mm. The computed weld geometry and size agrees well with experimental results.

![Numerically predicted weld cross-section.](image)

Figure A.19 Numerically predicted weld cross-section.

Figure A.20 presents the computed thermal cycles at the locations of the two thermocouples that were spot welded on the top surface of the specimen, and compares the predictions with the experimental results. The peak temperatures at each location agree well. The predicted cooling rate also appears to be reasonable at a distance of 2.5 mm from the weld centreline, although the discrepancy between the predicted and measured cooling rate is greater at a distance of 3 mm. There appears to be a tendency to underestimate the cooling rate. Nevertheless, when the predicted weld temperature field in Figure A.19 is compared with the macrograph in Figure A.15, that there is a good correlation between the calculated and experimental weld bead shapes.
Figure A.20 Comparison of predicted temperature histories at different distance from the weld centre line with those measured using thermocouples.

A.4 Discussion

A.4.1 Microstructures transition in different metallurgical sub-zones

The results of the thermal analysis were validated and found to be in good agreement with the experimental results. The predictions arising from the model can be therefore used to infer the microstructures that are likely to be generated during the laser welding process. Predicted thermal cycles were extracted for locations half way through the plate thickness, but at different distances from the weld centre line, as shown in Figure A.21. The points at distances of 0 mm and 0.5 mm from the weld centre line fall within the fusion zone, while the point at 1 mm approximately coincides with the fusion line, and the point at a distance of 1.5 mm is expected to be within the HAZ, whereas the points at 2 mm and 2.5 mm are expected to coincide with the ICHAZ and base material, respectively. The predicted peak temperatures at distances of 0 mm, 0.5 mm, 1 mm, 1.5 mm, 2 mm and 2.5 mm are 2100 °C, 1900 °C, 1300 °C, 920 °C, 700 °C and 570 °C, respectively. Those points at which temperatures exceed ~ 1500°C are expected to be within the fusion zone, while at the distance of 2.5 mm (base material) no solid-state phase transformations are expected to occur, because the peak temperature at this location is lower than the AC₁ temperature (700 °C).
According to the continuous cooling transformation (CCT) diagram for SA508 steel [23], the critical cooling rate for the formation of martensite is 900 °C/min (15 °C/s). According to the simulation results, the average cooling rates over the temperature range between 900 and 420°C (the martensite-start temperature), at locations of 0 mm, 0.5 mm and 1.5 mm from the weld centre line are 675 °C/s, 608 °C/s and 246 °C/s, respectively. These cooling rates are much faster than the critical cooling rate for martensite formation. This means the fusion zone and HAZ will almost certainly transform to martensite.

According to the simulation results, at a distance of 2 mm from the weld centre line, the peak temperature is approximately 700 °C (i.e. the Ac₁ temperature). This region is likely to be close to the outer boundary of the ICHAZ. At distances just under 2 mm from the weld centre line, the temperature will be higher than 700 °C, but lower than 800 °C (the Ac₃ temperature). This region (the ICHAZ) will only partially transform to austenite during the welding thermal cycle. Any newly generated austenite will be quenched to form martensite in the subsequent fast cooling process. When the martensite transformation ceases, the temperature in this ICHAZ would still be high enough for the martensite self-temper. However, other un-transformed material (i.e. materials that does not undergo austenitisation) will be retained, and this may take the form of over-tempered ferrite or baininte. The final microstructure within the ICHAZ

Figure A.21 Extracted thermal cycles half way through the plate thickness at different distances from the weld centre line.
will be likely to comprise bainite and martensite mixed with some tempered martensite, as shown in Figure A.7(a).

At a distance of 1.5 mm from the weld centre line, the peak temperature is approximately 920 °C according to the simulation results. The peak temperature between the distances of 1.5 mm and approximately 1.8 mm will fall between 920 °C and 800 °C. This area corresponds to the FGHAZ. The peak temperature in the FGHAZ is slightly higher than the Ac₃ temperature (800 °C). The material is completely re-austenitized in this region, but there is limited austenite grain growth due to the relatively low peak temperatures and short periods of time at this temperature range [28, 46]. In the following rapid cooling process, this fine-grained austenite transforms completely to martensite, and some martensite will be self-tempered in the cooling process. The final microstructure in the FGHAZ will be martensite mixed with some auto-tempered martensite. The microstructure and the small grain size can be seen in Figure A.7(b).

At a distance of 1 mm from the weld centreline, located immediately adjacent to the fusion line, is a sub-zone called the CGHAZ. From the simulation results, the temperature in this region reaches approximately 1300 °C (lower than the melting point) for a sufficient period to ensure complete re-austenitisation and rapid grain growth [28, 47]. During the rapid cooling following welding, martensite forms in the CGHAZ, and some martensite will be auto-tempered in the cooling process, as seen in Figure A.7(c). The final microstructure of the CGHAZ is martensite mixed with some auto-tempered martensite, but the grain size in the CGHAZ is larger than that of the FGHAZ. The prior austenite grain boundaries are indicated by black arrows in Figure A.7(c).

The temperature in the weld pool is raised over the melting point of the material, as seen in the simulation results at distances of 0 mm and 0.5 mm, and after solidification, very large austenite grains are formed. Under the very fast cooling conditions following welding, these very large austenite grains again transform to martensite. When the martensitic transformation finishes, martensite is partly self-tempered during further cooling. The large prior austenite grain boundaries and the martensitic structure of the fusion zone can be seen clearly in Figure A.7(d).
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The microstructures in the fusion zone and each sub-zone of the HAZ for the GTA welded joint are almost the same as those in the corresponding sub-zones of the laser welded joint. However, there is much more tempered martensite in each of the sub-zones because of the higher heat input and the slower cooling rates associated with GTA welding when compared with laser welding.

A.4.2 Relationships between the mechanical properties and the microstructures in different sub-zones

The Ms (martensite start) temperature of SA508 steel is around 420 °C and the critical cooling rate for martensite formation is approximately 15 °C/s [23]. The Ms temperature for this steel is relatively high and the critical cooling rate for martensite formation is relatively low. This may have caused the fusion zone and HAZ of the GTA weld to transform to martensite. The cooling rates experienced by the fusion zone and HAZ in the laser weld were approximately 20 to 40 times higher than the critical cooling rate for the martensite transformation. This results in all of the fusion zone and HAZ transforming to martensite. Martensite with a high hardness was generated in the fusion zone and HAZ after laser welding, and in the as-welded condition, the microhardness of the fusion zone and HAZ for the laser welded joint more than doubled when compared to the base material. This also occurred in the manual autogenous GTA welded joint. This suggests that, for SA508 steel in the absence of preheating, GTA welding has the same effect as laser welding in hardening the welded joint.

As shown in Figure A.7, the microstructure of each sub-HAZ changes from a coarse grained region (CGHAZ) to fine grained regions (FGHAZ) and then a partially austenitized region (ICHAZ) with increasing distance from the fusion line. The grain size changes because of the different thermal cycles that experienced in the different sub-zones. In the as-welded condition, the microhardness in the CGHAZ and FGHAZ varied around 410 HV0.3, which was approximately double that of the base material (200 HV0.3), whilst the microhardness in the ICHAZ was significantly lower at ~ 300 HV0.3. There is an approximately proportional relationship between the strength and hardness in steels, with the harder material having higher strength, although this is not always the case [48]. The strengths of the fusion zone and each sub-zone of the HAZ were improved mainly by martensite transformation and fine carbides precipitation in
these sub-regions. The hardness results suggest that the yield strength of the weld metal could equal or even exceed that of the HAZ. The microhardness distribution in the sub-zones of the HAZ is consistent with the yield strength in the sub-zones of the HAZ before post weld heat treatment, as reported in work by Lee et al. [12] on SA508 steel. They demonstrated that the yield strengths were over 1100 MPa for the CGHAZ and FGHAZ, which were also approximately double yield strength of the base material (500 MPa), while the yield strength for ICHAZ was approximately 600 MPa [12]. Since the material in the ICHAZ is only partially transformed to martensite during the welding process and the other un-transformed material is retained, the yield strength of the ICHAZ is lower than for the CGHAZ and FGHAZ where the material fully transforms to martensite.

In addition, because of the higher heat input and slower cooling rates associated with GTA welding when compared with laser welding, much more martensite was auto-tempered during the cooling process, which makes the hardness in the fusion zone and HAZ of the GTA welded joint lower than that in the laser welded joint. Of course, it must be borne in mind that SA508 steels will always undergo post-weld heat treatment after welding and most, if not all, of the transformation hardening will be reversed. It is nevertheless worthwhile to establish the extent to which the steel might be embrittled by the laser welding process, and the potential for embrittlement will generally be greatest in the as-welded condition.

The hardening associated with the welding process resulted in the welded tensile test samples failing in the base material, without any loss of strength. In addition, a narrow fusion zone is a typical characteristic of laser welding. These two factors can combine to contribute to difficulties in the Charpy impact testing of laser welded samples, whereby cracks can deviate from the weld into the base material, thereby giving misleading impact toughness results. One of the laser welded Charpy specimens failed with a lower absorbed energy value (66 J) when tested at -40 °C. This may have occurred due to the crack initiating from the notch and deviating into the base material, and then continuing to propagate through the HAZ. The crack path for this specimen can be seen in Figure A.14(b). The base material can absorb some energy, but the brittle HAZ can absorb less energy before it fractures. Another of the laser welded specimens failed with a much lower absorbed energy value of 45 J when tested at -20 °C. This may...
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be attributable to the crack initiating from the notch and then propagating directly through the HAZ. The fracture path (through the HAZ) of this specimen can be seen in Figure A.14(d). The brittle HAZ cannot absorb as much energy before fracture. However, there were two other laser welded specimens fractured with higher absorbed energies (approximately 100 J) at -40 °C and -20 °C, respectively. This may be attributable to the cracks initiating from the notch and then propagating directly through the base material. The absorbed energies for these laser welded specimens were found to be even higher than those of the base materials at corresponding test temperatures. This may attribute to the curved propagation path for the cracks, which increased the area of the fracture surface in comparison to specimens extracted from the base material, thereby increasing the absorbed energy. The curved travelled paths can be seen in Figure A.14(b) and (d). All the specimens extracted from the base material fractured in a manner that was aligned with notch, with a relatively straight travelled path, as can be seen in Figure A.14.

A.5 Conclusions

The following conclusions can be drawn from this work:

(1) The laser welding process produced welds with acceptable weld profiles in 6 mm thick SA508 steel plates over a relatively wide range of welding parameters. The welds were free of macroscopic defects.

(2) In the as-welded condition, the mechanical performance of the laser weld in a 6 mm thick SA508 steel plate was similar to the performance of an autogenous GTA weld. Cross-weld tensile specimens fractured in the parent material away from the weld region.

(3) The absorbed energies of the fusion zone in the laser weld were found to be comparable to those of the parent material, based on sub-size Charpy specimens.

(4) The hardness in the fusion zone and HAZ for both the laser and GTA welded samples, in the as-welded condition, was approximately double that of the base
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material, with the measured values for the laser weld being slightly higher (~ 430 HV₀.₃) than those for the GTA weld (~ 410 HV₀.₃).

(5) FEM modelling established that the cooling rates during laser welding in the absence of preheat were between 20 and 40 times higher than the critical cooling rate for martensite formation. This suggests that martensitic microstructures will almost always form in SA508 steel as a consequence of laser welding. These findings were confirmed in experimental work, in which the as-welded microstructures in the fusion zone and HAZ of the laser welded joint were found to comprise martensite mixed with some self-tempered martensite.

(6) While these preliminary results are encouraging, further work is now required to assess the properties of laser welds in SA508 steel in the post-weld heat treated condition, and it is also essential that this work is extended to assess the performance of laser welded joints in thicker sections of material.

References

Appendix A

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