MODELLING AND ANALYSIS OF
LASER DIRECT METAL DEPOSITION
OF Ti-6Al-4V ALLOY

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A major strength of the laser direct metal deposition (LDMD) rapid manufacturing technique is its ability to manufacture freeform shapes, to directly create different surface coatings on a part, and to produce parts from graded porous to fully dense solid structures. This technique is gaining popularity in the fields of aerospace and biomedical manufacture due to its flexibility and cost effectiveness. Surface coating and repair are its biggest application in industry. Previous models of surface layer coating predicted heat flow in the substrate and the resulting temperature distributions successfully, however layer geometry predictions were absent or incomplete. Here, an analytical modelling method of surface layer coating has been presented which accounts for track interactions and uses a novel powder mass affinity factor in multiple track laser direct metal deposition. The model has been validated by a series of surface layer coating experiments using Ti-6Al-4V alloy. Surface layer characteristics in terms of layer profile, layer thickness and layer roughness have been compared. The model is, then, extended from multiple tracks to multilayer structure fabrication. The multilayer porous structures have been fabricated by using the laser in continuous-wave and pulsed-wave modes for potential use in biomedical applications. In the second part of the work, a coupled analytical-numerical solution has been developed for the single track deposition model described earlier. Laser direct metal deposition is a complex process involving multiple interdependent processes which can be best simulated using a fully coupled mass-energy balance solution and the model removes the shortcoming of previous models where energy and mass balance equations are solved in a decoupled manner. The model has been applied to find out temperature distributions, track profile and microstructure scale, its experimental validation makes it convincing. The model is quite efficient as compared to finite element methods and a useful industrial aid for selecting the parameters to use for laser direct metal deposition when separate geometric and microstructural outcomes are required. In the third part of the work, a comparative study of LDMD characteristics using GA (gas-atomised) and PREP (plasma rotating electrode process) Ti-6Al-4V powders has been presented. The LDMD characteristics in terms of geometric dimensions, surface finish, microstructure, micro hardness and any defects such as intralayer porosity are compared and it is concluded that PREP powder deposition has some potential benefits when high value components are to be built, for example, aerospace applications. This study has also exposed the scientific understanding on the causes of intralayer porosity generation behaviour in laser direct metal deposition. To minimize the intralayer porosity, optimum combinations of processing parameters/conditions and characteristics of the initial powder, have been proposed for practical engineering applications.
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Dedication

Dedicated to my great

Father & Mother,

for their continuous prayers, endless love, unconditional support

through thick and thin

and

in the memory of my grand father

Mian Abdul Aziz

who always wished for my success in life
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Nomenclature

\( T_{\text{amb}} \)  Ambient temperature (K)
\( T_m \)  Material melting temperature (K)
\( T_p \)  Peak melt pool temperature (K)
\( T_{\text{mean}} \)  Melt pool mean temperature (K)
\( P_l \)  Incident laser power (W)
\( P_m \)  Power required to sustain mass addition (W)
\( P_p \)  Laser power loss in powder stream (W)
\( P_{\text{evap}} \)  Laser power loss due to evaporation (W)
\( r_l \)  Half maximum beam radius (m)
\( r_p \)  Gaussian powder stream radius (m)
\( k \)  Thermal conductivity (W m\(^{-1}\) K\(^{-1}\))
\( v \)  Laser scan speed (m s\(^{-1}\))
\( q_{pr} \)  Powder mass flow rate (g s\(^{-1}\))
\( q_{pf} \)  Powder mass flux (g s\(^{-1}\) m\(^{-2}\))
\( Q_{\text{evap}} \)  Evaporation heat flux (W m\(^{-2}\))
\( m_e \)  Evaporation mass flux (kg m\(^{-2}\) s\(^{-1}\))
\( C \)  Specific heat capacity (J kg\(^{-1}\) K\(^{-1}\))
\( L_m \)  Latent heat of melting (J kg\(^{-1}\))
\( L_v \)  Latent heat of vaporization (J kg\(^{-1}\))
\( W \)  Melt pool width (m)
\( L_1 \)  Melt pool length in front of the origin (m)
\( L_2 \)  Melt pool length behind the origin (m)
\( D \)  Melt pool depth along the z-axis (m)
\( y_{\text{front}} \)  y position of front of melt pool relative to the origin (m)
\( y_{\text{rear}} \)  y position of rear of melt pool relative to the origin (m)
$y_{\text{front1}}$ y position of front of overlapped melt pool relative to the origin (m)

$y_{\text{rear1}}$ y position of rear of overlapped melt pool relative to the origin (m)

$H_a$ Melt pool height above the substrate level (m)

$A_a$ Melt pool area above the substrate level ($m^2$)

$V_a$ Melt pool volume above the substrate level ($m^3$)

$V_b$ Melt pool volume below the substrate level ($m^3$)

$Y$ Dimensionless y dimension

$y$ Dimension parallel to motion (m)

$X$ Dimensionless x dimension

$x$ Dimension perpendicular to motion (m)

$Z$ Dimensionless z dimension

$z$ Dimension parallel to depth of melt pool (m)

$m_o$ initial melt size (m)

$m_n$ melt pool size after $n$th iteration (m)

$h_z$ Track height (m)

$d_z$ Melt pool depth (m)

$d_{\text{max}}$ Maximum melt pool depth (m)

$K_1$ Deposition material constant used for microstructural scale calculations

$K_2$ Deposition material constant used for microstructural scale calculations

$A$ Deposition material constant used for evaporation heat losses calculations

$\alpha$ Absorptivity

$\beta$ Laser fibre efficiency

$\rho$ Deposition material density ($kg\ m^{-3}$)

$\sigma$ Dimensionless speed

$\mu$ Dimensionless time

$\omega$ Laser deposition constant dependent upon coaxial nozzle dimensions

$\tau$ Laser deposition constant dependent upon powder carrier gas flow rate
Nomenclature

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Acronyms

GA  Gas atomised
PREP  Plasma electrode rotating process
LDMD  Laser direct metal deposition
CAD  Computer aided design
RP  Rapid prototyping
RM  Rapid manufacturing
PR-MMC  Particulate-reinforced metal matrix composite
MicroCT  Microcomputed tomography
HPDL  High power diode laser
TEMs  Transverse electromagnetic modes
LENS  Laser Engineering Net Shaping
CVD  Chemical vapor deposition
PVD  Physical vapor deposition
HAZ  Heat affected zone
FEM  Finite element method
FGM  Functionally graded material
SEM  Scanning electron microscopy
TEM  Transmission electron microscopy
CNC  Computer numerical control
SE  Secondary electron
BSE  Back scattered electron
CCD  Coupled charged device
XRD  X-ray diffraction
HV  Vickers Hardness
Chapter 1

Introduction

1.1 Research rationale

In 1960 the first working ruby laser was invented by Maiman [1] and this was as the result of the considerable discovery of Einstein who demonstrated that lasing action should be possible [2]. Since its discovery the laser has been finding applications in many areas including agriculture, communication, metrology, biomedical, nuclear, chemical and material processing [1]. With the rapid growth of laser applications and reduced cost of laser systems, laser material processing has gained increased importance in fields such as material removal, joining, cutting, fabrication and layered manufacturing.

Layered manufacturing, i.e. additive manufacturing (AM), is also called rapid prototyping (RP) or rapid manufacturing (RM). With the laser based RP systems, the shift is taking place from prototyping of plastic parts to fully dense metallic parts made from metallic powders. This technology provides the single-step waste-free process to fabricate near-net-shape components directly from a CAD file. Since its inception, laser direct metal deposition has been used in various areas e.g. deposition of commercial alloy powders, deposition of carbides and intermetallics, particulate-reinforced metal matrix composite (PR-MMC) coatings, and alloy development by laser deposition [3]. Due to the flexibility of LDMD fabrication and its ability to manufacture the desired macro and microstructure [4], it has a great potential in the modern manufacturing industry.

Despite the various applications of the laser direct metal deposition, its main applications are in the surface coating and repair, where crack free repair and surface modification are possible. LDMD also has the potential to produce a structure with gradation in porosity and/or composition from different materials which makes it a favourable technique for biomedical implants manufacture. Due to complex nature of the process and involvement of multiple interdependent variables, experimental trials of the process can be fairly time consuming and expensive if high value
materials are to be used. This necessitates a reliable method of predicting the outputs of the process in advance using modelling techniques.

Different types of powder materials are now commercially available and are being used in laser direct metal deposition. Using different types of the powder can give different properties of the deposited material e.g. geometric accuracy, surface roughness, microstructural and mechanical properties, and defects of intralayer porosity etc. Amongst these, poor surface finish and the presence of intralayer porosity are still considered as major challenges in the field.

1.2 Aims and objectives of the project

The aims of the work reported in this thesis are to investigate the flexibility of LDMD theoretically and experimentally, from surface coating by multiple tracks deposition to porous structures fabrication by multilayer deposition. Ti-6Al-4V will be used as the deposition material because it has immense application in the aerospace and biomedical fields due its extra ordinary properties. The project will also explore the usability of different types of commercially available Ti-6Al-4V powders in terms of fabricated part characteristics. The causes of intralayer porosity generation in LDMD fabricated structures will also be explored.

To achieve the aforementioned aims the following objectives have been set.

- Mathematically model the multiple interacting laser deposition tracks needed to form surface coating and experimentally verify the model.
- Mathematically model multilayer laser deposition in continuous-wave and pulsed-wave modes for porous structures fabrication to be potentially used for biomedical applications and experimentally verify the model.
- Develop an analytical-numerical model to simulate the multiphase interdependent process variables in LDMD.
- Mathematically model microstructure scale and its variations with the processing parameters.
- Investigate the effects of different types of powder on a fabricated structure’s characteristics.
- Develop scientific understanding of intralayer porosity generation in LDMD and identify the process parameters / experimental conditions to minimize the intralayer porosity for engineering applications.
Contribute to the developments in laser direct metal deposition of Ti-6Al-4V for its applications in the aerospace and biomedical industries.

1.3 Thesis structure

This thesis presents the theoretical and experiment analysis of the laser direct metal deposition of Ti-6Al-4V alloy. The thesis is comprised of eight chapters which are described as follows.

Chapter 2 presents a fundamental review of laser direct metal deposition theory. Laser basics and working principles of laser direct metal deposition are covered here. This is followed by a detailed literature survey including mass and energy transfer by the powder stream, melt pool dynamics, track formation, and solidification microstructure. Finally the effects of processing parameters on the LDMD fabricated parts characteristics are discussed.

Chapter 3 gives a brief introduction to the materials and experimental equipment used in this research. The characterization techniques used are also discussed here.

Chapter 4 presents an analytical modelling method for prediction of the geometry of multiple, interacting laser direct metal deposition tracks for surface layer deposition. The model can accommodate single tracks, partially overlapping tracks suitable for surface cladding and multiple layers. The main model comprises sub-models of the powder stream, quasi-stationary conduction in the substrate, powder assimilation into the melt pool area and multiple tracks interactions. The model is verified by a series of experiments using a Laserline 1.5 kW diode laser along with a coaxial deposition head.

Chapter 5 describes laser deposition with a diode laser in continuous mode and with a CO$_2$ laser in pulsed mode to produce multi-layer porous structures for biomedical applications. The porous structures are compared with respect to their internal geometry, pore size, and part density using a range of techniques including microcomputed tomography. Analytical models of the processes are also developed by using Wolfram Mathematica software to solve interacting, transient heat, temperature and mass flow models. Measured and modelled results are compared and show good agreement.

Chapter 6 extends the single track deposition model described in chapter 4 and presents a coupled analytical-numerical solution of the energy-mass balance
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equations to simulate the multiple interdependent processes involved in LDMD. An iterative feedback loop is used to ensure mass and energy balances are maintained at the melt pool. Using modelled thermal gradients and cooling rates calculations, microstructure scale modelling has been carried out. The model is verified using Ti-6Al-4V single track deposition, produced with a coaxial nozzle and a diode laser.

Chapter 7 presents a systematic comparative investigation of LDMD using GA (gas-atomised) and PREP (plasma rotating electrode process) Ti-6Al-4V powders. Ti-6Al-4V powders are first analyzed using laser diffraction, scanning electron microscopy and microcomputed tomography. LDMD characteristics in terms of layer geometry, surface finish, microstructure and micro hardness, and internal porosity are compared under similar process conditions. The change of microstructure morphology, from long columnar to small equiaxed grains, with processing parameters and the factors that affect it are discussed. The causes of intralayer porosity generation during LDMD fabrication are investigated and discussed in details. An advanced three-dimensional porosity assessment method using microcomputed tomography (MicroCT) is used to quantify internal porosity in the powders and final LDMD thin-wall samples.

Chapter 8 gives a brief overview of the major conclusions and finding observed in this research and as well as the recommendations for future studies.
Chapter 2

Review of the fundamental theory of laser direct metal deposition

2.1 Introduction

This chapter presents a review of the fundamental theory of LDMD. Laser basics are covered including lasing phenomenon, laser types and characteristics of the laser beam. The basic operation of LDMD is described. Then a detailed literature review of the processes involved in LDMD is carried out. This review includes mass and energy transfer by powder stream, melt pool dynamics, track formation. Solidification and microstructure theory with reference to the LDMD is discussed. The effects of the LDMD process parameters on the manufactured parts characteristics e.g. porosity, surface finish, hardness, tensile strength and residual stresses are also covered

2.2 Laser basics

2.2.1 Lasing phenomenon

A laser beam is basically light amplified by the mechanism of stimulated emission (the process after which the laser is named- Light Amplification by the Stimulated Emission of Radiation). If an atom is placed in an electromagnetic field with frequency $f$, then the atom can absorb the energy $h_f$ and move to the higher energy level given by the equation

$$h_f = E_x - E_o$$  \hspace{1cm} (2.1)

where $h$ is Plank's constant, $E_x$ is higher level of energy and $E_o$ is ground level of energy. In this state a photon of energy $h_f$ can stimulate the atom to move to the ground level by emitting another photon of the same energy $h_f$. This process is called stimulated emission because this event is triggered by the external photon. The photons have the same energy, wavelength, polarization, phase and direction of travel. Therefore stimulated emission produces monochromatic, directional and coherent electromagnetic radiation. In normal conditions there are more atoms in ground level than at the higher energy level so if the photons are flooded in this state
then the probability of absorption is higher than emission. To produce laser light, emitted photons should be more numerous than the absorbed photons. To achieve this is to begin with, more atoms must be at the higher energy level than at the ground level. This phenomenon is called population inversion which is necessary to produce laser light.

2.2.2 Laser types

There are several ways to classify lasers, but the most common way is to classify them based on the physical state of the active material. So the lasers can be categorized as follows

- Gas lasers
- Excimer lasers
- Solid-state lasers
- Semiconductor lasers
- Liquid dye lasers
- Fibre lasers

Each class contains different lasers, but in the following pages only those that have applications in material processing are discussed.

*Gas lasers.* The CO$_2$ laser is most common of the molecular gas lasers. The main radiation from CO$_2$ is at 10.6 µm wavelength and the power output can be more than 45 kW. The wall plug efficiency (optical energy out/total electrical energy into the system) is only 12%. Regardless of the low efficiency, the CO$_2$ lasers have a good beam quality and focusability. This laser has been widely applied in many industrial applications such as welding, cutting, cladding, processing of glass and ceramics.

*Excimer lasers.* The name derives from the ‘excited dimer’ molecule which are the lasing species. Each molecule of the active medium is composed of an inert gas atom and a halogen gas atom such as krypton fluoride (KrF) and xenon fluoride (XeF) etc. Typical average output powers range from 1 W to approx 700 W. Excimer lasers are widely used in medical technology and in micromachining. With these lasers the ability to control depth in microns makes it favourable for removing excess material and removing oxide coatings.
**Solid-state lasers.** Neodymium-yttrium aluminium garnet (Nd-YAG) is most common amongst the solid-state lasers. The beam has a wavelength of 1.06 µm. The lasing material is distributed in a solid matrix and a pumping source is required to supply energy to the crystal rod. The pumping source can be a flash lamp, but these have very low operating efficiency: approx 2%. It is more effective to use a diode laser as the pumping system and the operating efficiency can then be raised to 8-10%. These lasers can be pulsed or continuous wave (CW) and have wider applications in drilling and cutting. The diode pumped Nd-YAG is an established tool for micro-cutting applications, however there are several disadvantages such as low wall plug efficiency, high running costs, and poor thermal stability [5].

**Semiconductor lasers.** Diode lasers are currently the most efficient devices for converting electrical energy into optical energy. Their wall plug efficiency may reach up to 50% and the wavelength range is 0.75-0.87 µm for a GaAs laser [1]. In diode laser the excited state is that of the electrons in the conduction band compared to those in the valence band. The two states can be separated at a p-n junction in a semiconductor material. Currently, their main applications are in pumping other solid-state lasers, but they are also being used for welding, cutting, drilling and rapid prototyping etc. The new generation of high-power diode laser (HPDL) can produce up to 4kW.

**Fibre lasers.** These lasers are doped plastic or glass fibres that are end or side pumped by diode laser. Diode pumped fibre lasers are being developed. IPG Photonics is marketing a 2 kW CW fibre laser based on Yb operating at 1085 nm wavelength. It gives spot size down to 50 µm providing a power density of 100 MW/cm² from a unit measuring 110x60x118 cm, which includes power supply and cooling system. This has a beam quality 10 times better than a standard Nd-YAG laser and the wall plug efficiency is 20%, while the lifetime of pumping diode is 100 000 hours, indicating a several years of maintenance free operation. The 700 W version of a fibre laser has been shown to be able to cut through 50 mm of steel [1]. With this sort of performance these lasers will much offer in the future of material processing.
2.2.2 Laser beam characteristics

Laser beam characteristics play a very important role in laser material processing. There are several parameters that characterise the beam such as laser beam mode, polarization and focusability.

The beam with low divergence angle produces a smaller focused spot and greater depth of focus [5]. The laser energy can be distributed in a uniform or Gaussian distribution over the laser beam spot area.

In order to achieve a good quality beam, it is necessary to resonate the beam in a chamber where certain distributions of amplitude and phases of electromagnetic field can be reproduced due to the repeated reflections between the mirrors [6]. These specific shapes produced in the resonator are called transverse electromagnetic modes (TEMs). Each TEM is a different energy distribution across the beam. TEM_{00} (Gaussian) and TEM_{01*} (created by oscillation between orthogonal TEM_{01} modes) are common in industrial lasers.

In the LDMD process, it is essential to provide the appropriate power density and interaction time between laser beam and the material. The LDMD process requires a power density from 70 -100 W/mm^2 and interaction time of 0.01 to 1 second [5] so any laser intended to be used for LDMD should be able to provide this level of power density.

Another important issue is the reflectivity from the surface of the metal. The reflectivity is a strong function of laser wavelength and temperature and it varies from metal to metal. As the temperature increases in the process zone, reflectivity decreases and absorptivity increases due to an increase in the photon population [1], which indicates the potential for more energy absorption by hot material. However, this is only true if the surface conditions remain constant. In practice, there is often oxidation or phase change which can alter this behaviour of absorptivity. Figure 2.1 shows the reflectivity of the smooth, polished surface of some common metals to unpolarised, perpendicularly incident electromagnetic radiations from different lasers.

A quantitative measure of the focusability or beam quality is $M^2$. It compares the divergence of a given beam with a pure Gaussian beam (i.e. $M^2=1$) with the same waist (i.e. minimum diameter of laser beam before any focusing optics) located at
the same position. The beam quality is a measure of the focusability (spot size and focal length) of a laser beam. It shows how closely a laser beam can be focused.

Figure 2.1. Reflectivity of some common metals to laser radiation of different wavelength [7]

2.3 Laser direct metal deposition basics

LDMD is a process which builds solid metallic parts by injecting powder into a laser-generated molten pool. A typical LDMD system is shown schematically in Figure 2.2. During the deposition process, a defocused laser beam creates a moving melt pool on the surface of the substrate and metal powder is conveyed to it from the nozzle via an inert gas stream (usually argon) called conveyance gas. The additional material increases the size of the melt pool, with a resultant raised track when the laser beam moves and the pool solidifies. Repeating the process, following the same or offset paths, allows multiple tracks to be deposited and walls or bulky parts to be built. The process is protected from harmful oxidation via a further inert gas stream (shroud gas) or by enclosing the process in a chamber filled with an inert gas.

Laser Engineering Net Shaping (LENS) patented by the Sandia National Laboratories USA, is an additive manufacturing process the same as LDMD. This process has several steps [8]. First, a solid CAD model is sliced electronically into a sequence of layers of a given thickness. Second, to build a metallic part, a solid substrate is used as a base and the laser beam is focused on the substrate to create a molten pool into which powder is simultaneously fed. Third, the substrate is moved beneath the laser beam, to trace out the pattern defined by the slice of CAD model. Starting from the bottom of the part, one layer is produced at a time. After formation
of a layer the deposition head (powder feeding nozzle and laser beam) is moved in the z-direction. Accordingly the part is built line by line and layer by layer.

![Figure 2.2. A schematic diagram of a coaxial laser direct metal deposition (LDMD) system](image)

### 2.4 Mass and energy transfer by the powder stream

#### 2.4.1 Powder feed

In the LDMD process there are different methods of mass transfer to the molten pool. Pre-placed powder is a cheap method compared to pneumatic delivery of powder to the substrate. It has some limitations as a binder paste is sometimes mixed with powder to ensure its cohesion, but this results in porosity due to its evaporation during the process [5]. Another shortcoming is lack of control of high dilution which increases with the interaction time [9]. Wire feeding and combined wire-powder feeding has also been investigated in LDMD [10, 11]. Wire feeding has higher material usage efficiency [10], but melting efficiency is low compared to powder deposition and the wire position and / or distance relative to the substrate controls the final quality of part [11]. Other material addition techniques such as thermal spray, chemical vapor deposition (CVD) and physical vapor deposition (PVD) are suitable for surface cladding but not for part fabrication due to low bonding strength and high heat affected zone (HAZ) and low thickness [5]. Laser powder deposition as compared to the plasma wire cladding has greater microhardness and refined grain...
structure and can produce specimens free of micro pores and micro cracks [12]. Elemental powder blends offers a variety of cost effective possibilities to manufacture novel alloys and functionally graded materials [13]. Hence pneumatic powder feed through delivery nozzle is an attractive method of mass transfer in LDMD.

Powder particles of size range 50-150 µm are normally used for the cladding process. With decreasing powder grain size, i.e. ultra fine powder with a size of less than 15 µm, the flowability of powder decreases, which causes problems in powder transport. For this reason different powder feeders are required for each type of powder [5]. Many researchers have used gas atomized powder, which has approximately spherical particles, because of its better flow characteristics. Powder shape has small effect on the process as investigated by Pinkerton and Li [14]. They found that material deposition rate with gas-atomized powder was higher, but with water-atomized powder, having irregular shapes, microstructure and surface finish was better. Figure 2.3 shows the morphologies of gas-atomized and water-atomized 316 L powder particles.

Figure 2.3. SEM images of 316 L powders: (a) gas-atomized; (b) water-atomized [14]

In another study Pinkerton et al [15], using a diode laser and lateral nozzle, discovered that higher proportions of water-atomized powder led to poorer surface finish, hardness and tensile strength. This is in contrast to the previous study [14] and the results are attributed to the lower heating of the water-atomized powder stream due to the short interaction time with the laser beam before the deposition.
point and an increase of melt pool size due to higher coupling efficiency between melt pool and diode laser radiations.

The plasma rotating electrode process (PREP) process is another powder production technique which can produce powder particles with highly spherical morphology [16]. The use of the PREP powder is receiving attraction in the LDMD field. Qi et al [17] and Zhao et al [18] investigated the use of GA and PREP powders in the LDMD of Inconel 718 (Ni alloy). These studies found that PREP powder deposition samples had better mechanical properties than those of GA powder deposition samples. Figure 2.4 shows optical micrographs of the GA and PREP powder samples; porosity is visible on the GA powder particle’s surface.

![Figure 2.4. Optical micrographs of Inconel 718 (Ni alloy) powders: (a) GA; (b) PREP [17](a)](image)

### 2.4.2 Delivery nozzles

Delivery nozzles play an important role in produced part quality. The early powder feed systems had off-axis powder supply nozzles called lateral nozzles. Figure 2.5 shows a diagram of a LDMD system with lateral powder delivery. In lateral powder feeding, as the powder jet incidence angle increases, the powder particle temperature increases and there is reduced beam intensity on the substrate [19]. They are both caused by the beam-powder attenuation and interaction time. So, the alignment of this type of nozzle with respect to the laser beam and the substrate is critical. The powder injection point at the melt pool is important to control the height of clad and for good surface quality the powder stream should be injected into the front part of the melt pool [20].
The coaxial nozzle is omnidirectional, which is an important feature, because it gives more flexibility in a manufacturing process than lateral powder feed. Figure 2.6 shows the configuration of a coaxial nozzle in a LDMD system.

The powder catchment efficiency of a coaxial nozzle, which is defined as the ratio of powder deposited on the substrate to the powder delivered by the powder feeder in a particular time, is typically less than that of a lateral nozzle [5]. The catchment of powder is decided by the bonding conditions of particle and substrate and the
catchment efficiency can be increased by an increase in the surface tension of the melt area or by decreasing the powder particle diameter and impact velocity [21]. In coaxial powder delivery a critical process variable is the stand-off distance between the nozzle and the deposition point. However, experimental studies reveal that under some conditions a good layer consistency can be obtained with no axial movement of substrate between the layers for more than 20 layers or 10 mm clad height [22]. In coaxial nozzles the angle of inclination does not affect the powder stream so it is possible to work in three dimensional space without any effect on the powder flow [23]. To get a good powder efficiency and high quality clad, the powder focus must be at the level of the melt pool and the powder stream should have laminar flow parallel to the laser beam profile [23].

Several forms of nozzles have been invented. Jeantette et al [24] invented a coaxial nozzle system for producing complex shape objects. The parts can be produced with varying material densities and functionally graded components made from more than two different materials are also possible. Keicher et al [25] patented a multiple beams and nozzles system to increase the deposition rate, the assignee is now Optomec Inc. This system uses a single laser beam to outline the features of a solid object and then uses a series of equally spaced laser beams to quickly fill in the region. Islam et al [26] patented a multiple nozzle head for manufacturing and repairing of turbine engines. In this system laser beams are orientated at an acute angle to the surface and spaced around the powder supply.

2.4.3 Powder stream flow and energy transfer

Detailed work has been done on the powder stream flow in the literature. Using lateral powder feed, Picasso et al [27] used a three dimensional analytical approach to model the cladding process, assuming that the laser beam and powder stream are two interacting cylinders. For a given laser power, beam width and geometry of powder-injection, this model gives traverse speed and powder feed rate and the results compare well with experiments. Toyserkani et al [28] proposed a 3-D transient FE model of laser cladding with a lateral powder feed. This model takes into account the laser beam attenuation due to the powder stream and absorption factor deviation during the process (Brewster effect). The results of the numerical model are in agreement with the experimental clad geometry. Neto and Vilar [29] modelled the temperature distribution of the powder stream through a lateral nozzle
as well as the attenuation of the laser beam by the powder jet. Laser beam attenuation increases linearly with powder feedrate while decreasing exponentially with increases in particle size and jet velocity. The average temperature of particles reaching the substrate showed an exponential decrease with an increase in particle size and velocity but a small decrease in temperature with an increase in powder feed rate.

Theoretical and experimental work by Lin and Steen [30] and Lin [31] shows a Gaussian distribution within the powder stream from a coaxial nozzle and this mode of powder concentration is further verified by Pinkerton and Li [32, 33]. An analytical model, and its validation by an experimental investigation using optical and image analysis techniques, shows that the powder stream from a coaxial nozzle has two regions, one prior to merging to a single jet and one after this point, and that both have Gaussian distribution perpendicular to the direction of flow [33]. This analytical model [33] describes the variations in powder stream concentration along the axis of the coaxial nozzle. It also relates the nozzle dimensions with material mass and volume flow rates and hence can be used for nozzle design.

Both focused and columnar powder streams can be generated by using outward and inward nozzle exit arrangements respectively [34]. A columnar nozzle arrangement tends to reduce the peak powder concentration by up to 50% of than that of the focused arrangement [34]. Zekovic et al [35] performed 3-D numerical simulations of the gas-powder flow from coaxial symmetrical nozzles using FLUENT software. This model examines the importance of shroud gas for protection of laser head optics and also confirms that blob formation is due to ricocheted powder particles hitting can produce high purity spherical pimage of the powder cloud and ricocheted particles during experiment. Effective deposition, with stable powder delivery, is only possible at proper standoff distance i.e. 4.75mm-6.50mm below the nozzle tip. This range is determined by their experiment and it may not applicable to all LDMD processes.

A three-way interactions between the powder stream, the laser beam and the substrate have recently been investigated by Medina and Pinkerton [36] using numerical stream flow modelling. They found that standoff distance between nozzle tip and the substrate strongly influences the mass concentration within the powder stream, the powder stream heating and the mass deposition rate.
Laser beam attenuation due to the powder stream and the resultant temperature distributions of powder particle have been analyzed by many researchers. Lin [37] performed a theoretical and experimental analysis of a coaxial powder stream and found that the insertion point of the stream into the laser beam determines the maximum possible temperature of the particles. The melting distance of powder particles was 20-30 mm from the nozzle exit. The particle size and powder stream velocity affect the final temperature of particles and a variation of 500°C was found due to these effects. Liu and Lin [38] showed that the powder temperature is mainly dependent on the absorbed laser energy, beam divergence angle, powder particles diameter, stream velocity and shielding gas velocity. High amount of powder evaporation due to high laser power may cause the mass loss of a powder particle to be as much as the 25% of the original particle [38]. Kovaleva and Kovalev [39] introduced the concept of light-propulsion force caused by the material-vapour recoil pressure from the laser illuminated area of the powder stream. They numerically simulated the heat-mass transfer mechanism and found that the light-propulsion force changes the acceleration of the particle and its velocity becomes significantly higher than the carrier gas velocity. The mass loss due to evaporation then becomes negligibly small which is attributed to the higher particle speed and lower interaction time with the laser beam under the influence of vapour recoil pressure.

More than 50% of the laser energy could be lost in powder stream and laser beam attenuation increases with increasing the standoff distance and powder concentration [40]. Huang et al [41] modelled the laser-powder interactions for lateral and coaxial
nozzles using a three-dimensional analytical model and concluded some important results. With increasing powder feed rate, the peak value of laser intensity greatly reduces and may not located at the centre of the laser spot but encircle it, creating an annular high temperature zone. Unlike this, the peak value of particle temperature decreases slightly with increases powder feed rate and the reason could be that the particles at low feed rate mostly use heat absorption for melting rather than increasing the temperature. With increasing feed angles the peak value of laser intensity decreases and the maximum temperature of a particle increases. Pinkerton [42] modelled the laser beam attenuation and powder heating during coaxial laser direct metal deposition using analytical techniques. This model is experimentally verified by using optical light sheet method for stream visualization and a thermal IR camera for temperature measurement. Beam attenuation increases rapidly below the consolidation plane; it is calculated as 7% at the consolidation plane, increasing to 28% 1mm below it and 39% a further 1mm below that. At the consolidation plane, where the annular streams merge to form a single stream, the average particle temperatures are maximum at the centre of the beam.

2.5 Melt pool dynamics and track formation

2.5.1 The effect of energy flow on the melt pool

In the LDMD process a melt pool is created on the substrate by the laser energy in the form of light radiation. The dimension of the melt pool is typically up to several millimetres or less and the temperature can reach up several thousand Kelvin in a very short time [3]. The melt pool can be moving or last for a few microseconds when produced by a continuous-wave or pulsed-wave laser respectively. The size of melt pool is determined by the peak melt pool temperature, which increases with peak laser power and pulse length but decreases with powder mass flow rate, in laser cladding [43]. Qi et al [44] showed that melt pool dimensions increases with laser power and Pinkerton & Li [45] showed that the melt pool temperature increases approximately linearly and melt pool length increases with a decreasing rate with the initial laser power.
Figure 2.8. A thermal image of moving melt pool (316 SS) with temperature map, the alloy liquidus and solidus are 1673 K and 1645 K respectively [46]

Figure 2.9. Modelled temperature contours on the top of a wall (substrate) (xy plane, z = 0) due to a moving Gaussian heat source of diameter: (a) full wall width; (b) half wall width; (c) quarter wall width [47]
In LENS processing the melt pool length ranges from 0.5-1.5 mm and the pool volume has approximately size of 0.5 mm$^3$ which is dependent on laser power, velocity and conduction path [46]. Figure 2.8 shows a thermal image of a moving LDMD melt pool. The melt pool shape depends on the distance between the laser head and substrate i.e. it changes with the change of laser spot diameter [47]. As shown in Figure 2.9, with a laser spot equal to the width of substrate the heat flow through the substrate is approximately two-dimensional, but for decreasing spot diameter, heat flow and temperature gradients develop in the y-direction until, for the smallest spot size, the temperature isotherms have an elliptical shape and heat flow as three-dimensional [47]. Practically, this means that the heat flow changes from three-dimensional to two-dimensional when the process proceeds from first layer deposition to subsequent layers deposition in the thin-wall LDMD.

A number of process maps (non-dimensionlized plots), based on thermo-mechanical 2-D finite element simulations and verified experimentally, have been developed by Vasinonta et al [48-50] and by Beuth and Klingbeil [51]. These process maps used Rosenthal solution of 2-D conduction heat transfer [52] and defined non-dimensionlized variables such as melt pool length, wall height and material melting temperature. The process maps show that the melt pool length increases with absorbed laser power and a considerable reduction in melt pool length is predicted for a very short wall (approaching to the substrate) [49] i.e. the heat transfer mode changing from two-dimensional to three-dimensional.

Substrate preheating can improve the LDMD process by reducing the settling time for the formation of a steady-state melt pool [53]. Labudovic et al [54] modelled the LDMD process by three-dimensional analytical and numerical simulations using FE software and verified the results experimentally. They found that the melt pool is deeper at the rear end than on the front because of heat accumulation, however an increase in laser energy or decrease in scanning speed results in a wider melt pool. Figure 2.10 shows model and experimental images of the melt pool boundaries.
Figure 2.10. Comparison of the melt pool boundaries obtained by the processing results of the edge detector with those obtained by the modelling: (a) original image acquired by high shutter-speed camera; (b) processing result of edge detector; (c) numerical modelling result; (d) analytical modelling result [54]

Peyre et al [55] performed analytical and numerical modelling of the LDMD process. They found that the melt pool size increases with growing layers, i.e. when additive manufacturing is carried out farther and farther away from the substrate, which is consistent with the process maps of Vasinonta et al [48-50] and Beuth and Klingbeil [51], and the melt pool enlargement is attributed to the lower heat dissipation in the substrate for higher layers. Figure 2.11 shows how melt pool size increases as the layer number increases.

Figure 2.11. Comparison between: (a) experimental (1000 Hz C-Mos fast camera): (b) simulated laser-induced melt-pools (600W, 0.006ms$^{-1}$), between the 2nd and the 8th manufactured layers [55]
In the LDMD process it is very important to know how much of the laser energy contributes to the process and how much is lost in surrounding environment by reflection, radiation, convection and vaporization. Hofmeister et al [46] showed that in the LENS process more than 90% of energy is conducted through the substrate and only 4% is used to heat and melt the new material. Gedda et al [56] examine the factors that affect the efficiency of CO\textsubscript{2} laser cladding and theoretically calculate the energy distribution. They found that 50% and 10% of the total energy are reflected from melt pool and powder stream respectively; 30% and 10% of the energy are responsible for heating and melting the substrate respectively. Pinkerton and Li [45] used analytical techniques to model the energy distribution in the LDMD process and a new energy balance model was created to demonstrate the effects of absorption and coupling efficiencies of CO\textsubscript{2}, Nd:YAG and high power diode lasers (HPDLs) for the LDMD applications. Figure 2.12 shows the distribution of initial laser power for different types of lasers and concludes that HPDL is the most efficient laser of the three.

![Modelled final distributions of LDMD input laser power for HPDL, Nd:YAG and CO\textsubscript{2} laser types](image)

**Figure 2.12.** Modelled final distributions of LDMD input laser power for HPDL, Nd:YAG and CO\textsubscript{2} laser types [45]

### 2.5.2 The effect of mass flow on the melt pool

The mass flow via the powder stream on the molten pool forms a clad, which is the basic output of the LDMD process. In this regard an important parameter is aspect ratio (AR) which is defined as the ratio between clad width and height. Clad width depends on traverse speed, specific energy, powder feed rate and laser beam diameter [57]. With other variables held constant, clad width decreases linearly with traverse speed and increases with specific energy [57]. Resch et al [58], Srivastava et
al [59] and Kaplan & Groboth [60] also found that clad width decreases with an increase in traverse speed. Pinkerton and Li [61] confirmed that multiple wall width shows the same behaviour with traverse speed. Yellup [62] showed that at higher laser power, traverse speed does not have a significant effect on clad width but at lower laser power it slightly decreases with traverse speed until the formation of a discontinuous clad due to insufficient heating. Peyre et al [55] and Hu et al [57] showed that clad width increased slightly with powder feed rate. In contrary, Srivastava et al [59] found that clad width decreased slightly and Pinkerton and Li [61] showed it remained unaffected with powder feed rate. The clad width is primarily dictated by the laser spot diameter [57] and increases with the laser power [55, 59, 61, 63]. The clad width increases after one or two layers in thin wall multi track deposition which is due to the lower heat dissipation and consequential increase of melt pool size [55].

The height of a deposited clad is inversely proportional to the traverse speed [57]; a finding that is supported by Resch et at [58], Kaplan & Groboth [60] and experimental investigation by Pinkerton and Li [61]. Srivastava et al [59] showed a different behaviour for multilayer cladding: that clad height increases with traverse speed to a certain point and then decreases with the traverse speed. Clad height is mainly dependent on powder feed rate and increases with an increase in powder feed rate [55, 57, 60, 61, 63]. However, Srivastava et al [59] showed that the rate of height increase increases with powder feed at low feed rate and decreases at high feed rate. Kerutz et al [64] showed that clad height increases linearly with the powder line mass ( powder feed rate divided by traverse speed). Qi et al [65] showed that clad height increases linearly with laser power and Keicher and Smugeresky [66] found that layer height was proportional to a function defined as: power density/traverse speed (J/cm$^3$). But some researchers showed that clad height is almost unaffected by the laser power [55, 61] and Srivastava et al [59] showed the opposite behaviour: that clad height decreases with the laser power. In LDMD, an initially high clad followed by a lower but wider clad might be due to the re-melting of the initial layer and increase in the size of the melt pool.

Alimardani et al [53, 67] used a three dimensional numerical approach to predict geometry and thermal distribution of multilayer laser cladding and found that clad height increases at the end point of a track as the temperature increases at this point.
due to the change in the heat transfer mechanism. The influence of heat convection/radiation is more at the end points of the deposited layer compared to the middle of it, but the lower conduction losses cause the increase in temperature. Figure 2.13 shows a comparison of experimental and numerical results and also shows the difference in wall height at the end points in longitudinal profile [53]. Kummailil et al [68] analyzed the LENS processing parameters using a two-level fractional factorial approach and found that deposition rate appears to be related to the product of the energy per unit area and mass flow rate. Mass flow rate and traverse speed have the biggest effect on deposition rate.

Pinkerton and Li [69] investigated the effects of laser beam shapes on deposition of a thin wall and concluded that LDMD is also feasible without complete circular symmetry of the laser beam. They found that mean clad width and height were different using a rectangular laser beam compared to a circular laser beam and also varied with beam orientation.

Figure 2.13. Geometrical comparison between experimental and simulation results [53]
2.5.3 Flow within the melt pool and its effects

In a LDMD melt pool, the two basic internal forces can be identified as; Marangoni forces, caused by surface tension gradients, and gravity-induced buoyancy forces [70]. The flows resulting from these forces are termed as Marangoni convection and gravity-induced buoyancy convection. The Marangoni convection is dominant and mainly dictates the overall flow in the melt pool [70, 71]. The Marangoni number is the measure of the extent of the Marangoni convection which is defined as below [70].

$$M_a = \frac{\partial \xi}{\partial T} \frac{1}{\eta \kappa} L \Delta T$$  \hspace{1cm} (2.2)

where, $\partial \xi / \partial T$ is surface tension temperature coefficient, $\eta$ is dynamic viscosity, $\kappa$ is thermal diffusivity, $L$ is radius of melt pool surface and $\Delta T$ is the temperature gradient between centre and edge of the melt pool.

Some researchers have also shown by three dimensional modelling that the effect of Marangoni convection is more pronounced than the natural convection in the melt pool [72, 73]. Kumar and Roy [74] showed that at lower scanning speed maximum and average melt pool temperatures reduce due to convective heat transfer but for higher scanning speeds these reductions are small. Surface tension or Marangoni force enhances the heat transfer in the melt pool by driving the molten liquid flow [71]. Limmaneevichitr and Kou [70] showed that surface tension gradient is induced by the temperature gradient along the pool surface and the temperature dependence of surface tension. Mathematically this is given as

$$\frac{\partial \xi}{\partial s} = \frac{\partial T}{\partial s} \frac{\partial \xi}{\partial T}$$  \hspace{1cm} (2.3)

where, $\xi$ is surface tension, $T$ is temperature and $s$ is distance along the pool surface.

The melt pool has a higher temperature under the laser beam, where a negative surface tension-coefficient causes the surface tension value to be lowest. Away from it the temperature decreases, surface tension increases, and due to this effect the liquid is pulled towards the rim of the pool creating a depression in the middle [75]. Voelkel and Mazumder [76] showed this depression in the centre of melt pool as shown in Figure 2.14. Lie et al [77] showed the spatial variations in Marangoni forces in a melt pool surface by numerical analysis. They showed that with a negative surface tension temperature coefficient, flow is outward from the pool.
centre to the pool periphery and results in a shallow and wide pool. This behaviour is also supported by Qi et al [65]. In contrast, with a positive surface tension temperature coefficient the flow is inward from the pool periphery to the pool centre causing a deep and narrow pool [77]. Figure 2.15 shows the outward melt pool flow and velocity distribution with a negative surface tension gradient causing a shallow and wide pool.

![Steel melt pool created with a CO2 laser and illuminated with both diffused and focused argon-ion laser light][76]

Surface active elements and alloying elements such as oxygen and sulphur have a major impact on surface tension [70, 78] and are responsible for a positive surface tension temperature coefficient [78, 79], increasing temperature and decreasing surface tension in the melt pool [80, 81]. Figure 2.16 shows the deep and narrow melt pool with high content of sulphur and positive surface tension temperature coefficient.
Figure 2.15. Velocity distribution at X-Z middle plane: (a) vector plot, (b) streamline plot [71]

Figure 2.16. Melt pool shape with positive surface tension temperature coefficient causing inward flow [79] tension:

Marangoni flow, whether it is caused by a positive or negative surface tension temperature coefficient, makes convection the dominant heat transfer mechanism within the melt pool [7]. Anthony and Cline [82] showed that Marangoni flow rippling can be avoided above a critical scanning speed because interaction time becomes too short to activate the Marangoni flow.

### 2.6 Solidification and microstructure

LDMD involves the interaction of a laser beam with a material and the consequential complex metallurgical and solidification processes. The melt pool is small and is in motion and its temperature depends on laser deposition parameters e.g. laser power, laser beam diameter, scanning speed and mass flow rate. As indicated previously that
the temperature of the laser induced melt pool can reach up several thousand Kelvin in a very short time [3]. The temperature gradient at the solid-liquid interface is very high and the melt pool quickly solidified with cooling rates in the range of $10^3$-$10^7$ Ks$^{-1}$.

The microstructure of the LDMD processed components is very dependent on the solidification behaviour and cooling rates [83]. The thermal gradient, cooling rate and solid-liquid interface velocity are related by the following equation [1, 7]

$$\frac{dT}{dt} = \frac{dT}{dx} \frac{dx}{dt}$$

(2.4)

where $dT/dt$ is cooling rate, $dT/dx$ is thermal gradient and $dx/dt$ is interface velocity or solidification rate. The ratio of thermal gradient to solidification rate is the criteria for supercooling and defines the mode of solidification [84]. If the temperature at a certain point maintained by the thermal gradient in the liquid is lower than the liquidus temperature at the same point, then the liquid will be constitutionally supercooled [84]. There is transition from cellular to dendritic microstructure as the degree of supercooling increases [7]. In LDMD, due to the direct contact of the melt pool and substrate, the solidification starts epitaxially without nucleation, and proceeds unidirectionally towards the surface. At the bottom of the melt pool solidification microstructure occurs with a plane interface but constitutional undercooling develops rapidly and leads the microstructure morphology to evolve from a plane interface to cellular and then finally dendritic [85]. However, the microstructure evolution depends on laser processing parameters. The resultant solidification microstructure can be predominantly equiaxed grain; or columnar to equiaxed transition (CET) can be reached at the end of solidification. Alternatively CET may be avoided and a purely columnar grain structure produced by changing processing parameters such as laser power, laser speed and preheat temperature [86].

Hofmeister et al [87] investigated the solidification microstructure of LENS processed 316 stainless steel and showed that microstructure near the melt pool interface had a columnar structure while the other regions had a cellular morphology. The dendrites and cells had a finer structure at lower laser power. In another study Hofmeister et al [46] showed that cooling rates during solidification are dependent on melt pool length, so by controlling the laser power a desired melt pool length and hence microstructure can be achieved. They also found solidification is not a quasi
steady state process due to Marangoni flows in the melt pool. The grain size decreases with traverse speed and increases with incident energy (laser power/traverse speed) which implies a finer grain size at higher cooling rates [83, 88]. The same behaviour of microstructure with scanning speed was also observed by other researchers [89, 90]. Frenk and Kurz [91] carried out a laser cladding experiment using Stellite 6 and the effect of changes in scanning speed on evolution of secondary dendritic arm spacing were investigated using a 2D heat conduction model. The results showed that a change in scanning speed from $1.67 \text{ mms}^{-1}$ to $167 \text{ mms}^{-1}$ led to considerable refinement in the microstructure with the secondary dendrite arm spacing decreasing from 2.5-2.6 $\mu$m to 0.5-0.8 $\mu$m. Using these results along with the conventional solidification theory there was an excellent agreement between the secondary dendrite arm spacing and the cooling rate. Figure 2.17 shows the microstructure refinement with scanning speed.

Figure 2.17. Effect of scanning speed on secondary dendritic arm spacing: (a) slow scanning speed of $1.67 \text{ mms}^{-1}$; (b) high scanning speed of $167 \text{ mms}^{-1}$ [91]

Mazumder et al [4] showed that large dendrites were the result of high specific energy as well as high deposition rate, when fine dendrites were obtained from low specific energy and low deposition rate. Wu et al [90] showed that microstructure is slightly coarsened at the bottom of the samples at higher powder feed rate but at the middle and top of the samples the effect of powder feed is negligible. Qian et al [92] investigated the microstructure of laser deposited Ti-6Al-4V samples and found that due to the repetitive heating and cooling cycles, the microstructure varies along the build direction i.e. the very top and very bottom layer of the samples have larger amount of martensite than other regions. However, Costa et al [93] showed that LDMD build parts of AISI 420 tool steel generally have a hard upper section containing fresh martensite and austenite and a softer lower section containing
tempered martensite. Kelly and Kampe [94] carried out detailed microstructure characterization of a multilayer Ti-6Al-4V laser deposit. The deposit shows layer bands of approximately 165 µm exhibiting the colony Widmanstätten α morphology. The microstructure between layer bands exhibits basketweave like Widmanstätten α outlined in retained β-phase with a gradient in α-lath thickness. By using microscopy, hardness and composition measurements they were able to conclude that the layer bands and gradient morphologies were the resultant of the complex thermal history that the deposit experience during multiple thermal cycles. The layer-band forms with a sequential pass of the heat source when a narrow region of a layer experiences heating of a critical time, temperature, and resultant cooling rate to form the colony morphology. Figure 2.18 shows the nominal and layer band Widmanstätten α morphologies.

![Figure 2.18. Laser deposited Ti-6Al-4V microstructure showing Widmanstätten α outlined in retained β: (a) nominal, basketweave Widmanstätten α; (b) layer band, colony Widmanstätten α [94]](image)

The solidification microstructure and growth morphologies change with laser power, laser speed and substrate preheat temperature [86], and also relate to the powder feed rate [95]. Marangoni flows and changes in nucleation density due to the powder feed rate may have effects on final microstructure [95]. The change in laser power and traverse speed can change both the cooling rates and the thermal gradients and hence have significant effect on the microstructure evolution [96]. Kumar and Roy [97] developed the process maps for identifying the form and scale of solidification microstructure using dimensionless quantities of thermal gradient, solidification velocity and cooling rate. They concluded that microstructure is dependent on these parameters.
2.7 Effects of LDMD process parameters

2.7.1 Porosity

Porosity is sometimes observed in the laser deposits, and can be categorized as interlayer porosity (lack-of-fusion porosity) and intralayer porosity (gas porosity). Kobryn et al [83, 88] investigated the porosity in laser direct metal deposition of Ti-6A-4V. The interlayer porosity is caused by the insufficient melting along the layer boundaries while the intralayer porosity is caused by the entrapment of gas in metal powder and/or release of the gas from powder particles. Both types of porosity tend to decrease with increasing traverse speed [88, 98] and laser power [88].

Steen et al [99] showed that lack-of-fusion porosity would develop in surface cladding if the track aspect ratio (width/height) becomes less than five. Wang and Felicelli [100] found that higher traverse speed can cause lack-of-fusion porosity, particularly in layers close to the substrate, and that more laser power is required at high traverse speeds in order to get a dense part.

The effects of processing parameters on generation of intralayer porosity are not yet very clear. Intralayer porosity occurs because the gases produced during cladding do not have enough time to surface out from the melt pool [101]. Porosity is reduced at high laser power because with longer solidification time less gas will become entrapped in the molten material [101]. This was also observed by Molian [102]; who concluded that if the rate of solidification is greater than the rate at which gas bubbles come out of the surface then they will be entrapped, causing porosity. Goswami et al [103] discovered that porosity can be avoided by the proper selection of scanning speed as at too high scanning speed the gas bubbles do not have enough time to be released from the melt pool and at too a low scanning speed the amount of gases dissolved increased substantially.

Susan et al [104] showed that powder with high porosity contents result in deposits with increased porosity. They also found that powder with large void/particle size ratios may result in less deposit porosity compared to powder with lower void/particle ratios and this was attributed to escape of gas from the melt pool due to the significant increase in the buoyancy with larger voids. However, Wang et al [105] found that the frequency of intralayer pores in the deposits did not show a correlation to the pores in the source powder. The intralayer porosity appears to
correlate with higher oxidation kinetics and atmospheric conditions in the laser deposition process and also show sensitivity to the scanning speed but not the laser power. Ng et al [106] showed that intralayer porosity depends on several processing parameters and that powder feed rate is the most influential amongst them. It can be minimized by using pore-free metal powder, no shielding gas flow and a stable melt pool, as shown in Figure 2.19. Hua and Choi [107] found that pore formation in the LDMD process is related to the dynamics of the melt pool, and irregular-shaped voids are caused by insufficient laser power. Choi and Chang [108] found that higher powder feed rate results in higher porosity, while higher laser power results in low porosity level regardless of powder feed rate.

Figure 2.19. Optical image of a transverse cross-sectioned laser deposit showing: (a) high level of intralayer porosity; (b) minimum level of intralayer porosity at optimized processing parameters [106]

Wu et al [90] showed that, for a given scan speed and powder feed rate, a porous microstructure is formed with connected pores up to 300 µm at lower laser power and a fully dense microstructure is formed at higher laser power as shown in Figure 2.20. Although not usually desirable, a porous microstructure can find applications such as catalysts, sensors and biomedical implants where a certain degree of porosity is crucial.
2.7.2 Surface finish

The surface quality of the parts produced by the laser direct metal deposition is still a challenge after thirty years of the research in this field. A review of existing literature shows that a good surface quality can be obtained by optimizing the process parameters. Oxidation and the adhesion of partially melted particles on the clad surface are the most important factors that determine the surface quality, as investigated by Li et al [20]. In the case of lateral powder feed, in order to minimize the powder adhesion, the interaction area between powder stream and the melt pool should be close to the front part of the pool. The powder adhesion in front of the melt pool will then become the part of the pool as the laser advances, making the surface smooth.

Resch et al [58] showed that surface roughness decreases with a decrease in powder feed rate and as increase in scanning velocity in multilayer laser deposition. They categorised the surface roughness into three modes of powder feed rate. For lower feed rate, thin layers of deposition and remelting of previous layers give a smooth surface. For medium feed rate, remelting is of a similar extent to the previous deposited layer so roughness increases slightly. For high feed rate, remelting is very shallow compared to the deposited layer and so gives circular cross section of geometry having high roughness. This is also supported by Pinkerton and Li [61]. Kaplan and Groboth [60] found that larger overlap decreases the surface roughness which is consistent with the findings of Mazumder et al [109]. The Marangoni forces generated by the surface tension gradient in the melt pool and vapour pressure above it causes surface ripples to form and become fixed as the result of rapid solidification.
Hua and Choi [107] found that the top surfaces of laser deposited samples with feedback control are flatter than those of samples without feedback control. Surface roughness also depends on which direction the measurements are taken. On a side wall the surface roughness is greater in the vertical direction than the horizontal direction, while on the top surface it is more in the direction perpendicular to cladding than in the parallel direction [4, 109]. The wall roughness increases as the layers become thicker and this is attributed to laser beam diameter variations associated with it [4, 109]. An adaptive slicing approach has been developed to recognize individual parts or part features before slicing and improves the part surface quality by eliminating ineffective slices [110].

2.7.3 Hardness

As indicated by the Hall-Petch relationship, hardness is a function of grain size [111] and can be given as below

\[
H = H_o + k d^{-\frac{1}{2}}
\]

(2.5)

where \( H, d \) are hardness, grain size respectively and \( H_o, k \) are constants. Hardness varies with the reciprocal square root of grain size. The Hall-Petch relation holds good for grain sizes above 0.1 µm but below this grain size the experimental results of hardness deviate from the relation [112]. This means the above relation is applicable to the LDMD process.

The grain size decreases with increasing cooling rate [113]; thus according to the Hall-Petch relationship hardness will increase with increasing cooling rate during solidification. The hardness is also increased by residual stresses due to the rapid cooling [114]. While using elemental powder with graded composition the hardness increases as the alloying content increases up to a peak point and then decreases; this behaviour is attributed to the significant changes in the microstructure scale [115]. The hardness decreases with layer thickness in LDMD due to the lower conduction rate in a thick layer and hence low cooling rate [107].

Wang and Felicelli [100] observed that the hardness of deposited layers increases with an increase in traverse speed, which is consistent with the earlier findings of Griffith et al [116]. The LDMD build parts have harder top surfaces and become softer toward the bottom because of slight annealing of the bottom layers each time
the laser beam passes while depositing successive layers [4, 93, 108, 116]. However, Pinkerton and Li [61] found that hardness can decrease on moving farther away from the substrate and this was attributed to the faster cooling rate during deposition closer to the substrate, which leads to the finer microstructure and hence greater hardness. The difference in results may be due to the different materials used for deposition.

2.7.4 Tensile and yield strengths

The grain size improves the mechanical properties, including tensile and yield strength, in the same manner as hardness [111]. Hence the mechanical properties of laser deposited materials are dependent on solidification. Tensile and yield strength increases with the scanning speed and decrease with the laser power [117].

The apparent benefit of the LDMD process is that melting and solidification occur at a higher rate than the conventional metallurgy processes [66]. A 85% increase in tensile strength along with 30% increase in ductility of laser deposited 316 stainless steel is reported [66]. Gao et al [118] investigated the mechanical behaviour of laser deposited Ti-6Al-4V and found that the tensile strength, yield strength and ductility of the samples were higher than those of cast or forged samples. The same behaviour in the mechanical properties of 316L stainless steel and 663 copper alloy was recorded by Zhang and Shi [119] and Zhang et al [120]. It was also found that the mechanical properties of the LDMD processed components show anisotropic behaviour i.e. these properties differ in the laser scanning orientation from the build orientation [118-120].

2.7.5 Residual stresses and distortion

Residual thermal stresses and distortion are frequently present in parts made by layered manufacturing techniques, including LDMD. The large thermal gradients can induce the warping through the build-up of thermal residual stresses [121] and the magnitude of this residual stress induced warping becomes larger as the part size increases [49]. Hence, an understanding of residual stresses and how to control them is important to get a consistent, defect-free build through LDMD. Dai and Shaw [122] et al found that the distortion in laser processed components is mainly caused by the transient thermal stresses rather than residual thermal stresses.
A holographic-hole drilling technique was used by Griffith et al. [116] to calculate the residual stresses in H13 tool steel parts built by the LENS process. The residual stress values were approximately 20% of the material yield strength and near the surface these values were low and tensile. Later Rangaswami et al. [121] measured the residual stress distributions in LENS-produced 316 stainless steel thin wall and pillar samples using neutron diffraction. They found that the stress was mostly uniaxial and directed along the grain growth axis, with the tensile stress at the edges and compressive stress at the centre. The maximum residual stresses were 50% and 80% of the 0.2% yield strength for the thin wall and pillar samples respectively. The thermal transients are the main cause for residual stress patterns. Dai and Shaw [122] showed that out-of–plane distortion of a laser deposited layer can be minimized by a proper selection of the laser scanning pattern. Figures 2.21 and 2.22 show the two different scanning patterns and the reduced distortion with scanning pattern II.

![Figure 2.21](image.png)

Figure 2.21. (a) Schematic of laser scanning pattern I and; (b) the final shape of the nickel plate after the laser scan using the scanning pattern-I [122]
The lower distortion in scanning pattern II is because the scanning direction keeps changing, causing the concave upward and concave downward distortion to partially cancel out [122]. Elperin and Rudin [123] investigate the effect of the composition of a functionally graded material (FGM) on the thermal stresses in a coating. They showed that variation in the volume content of ceramics slightly influences the temperature but strongly affects the thermal stresses and that the spatial distribution of ceramics in a coating also changes the thermal stresses.

A series of process maps developed by Vasinonta et al [48-50] and Beuth and Klingbeil [51] suggested that uniform preheating of the substrate would reduce the residual stresses. This preheating will not usually increase the melt pool size and if there is any change then it can be adjusted by a small decrease in laser power or small increase in scanning velocity. The decrease in residual stresses with substrate preheating is also supported by the other researchers [53, 54, 124]. Klingbeil et al [125] suggested that a combination of initial substrate preheating and part insulation would substantially reduce the residual stress induced warping. The part insulation would exploit the heating by the process itself, which would replace the active controlled part heating throughout the deposition process. Finnie et al [126] showed through numerical simulations that a non-preheated specimen showed high tensile residual stresses while a preheated specimen exhibited high compressive residual stresses.
2.8 Summary

This chapter summarized the major aspects of LDMD and concluded that it is a promising technique of the modern manufacturing era. The multiple processes involved in the LDMD and their effects on the final fabricated parts were discussed in details. Despite the fact that there has been an extensive research in the field for the last thirty years, there are still challenges. Poor surface finish and the presence of intralayer porosity in the deposited structures are two of the concerns. Tremendous efforts have been made to model and simulate LDMD but the multiple interdependent processes / parameters means that these models are not accurate enough to simulate the real process.
Chapter 3

Material and equipment for experimentation and characterization

3.1 Introduction

This chapter gives the brief introduction to the material and experimental equipment used for this research. The laser system, coaxial deposition head and powder feeder for mass transfer to the melt pool are introduced. The characterization techniques along with the equipment are also discussed here.

3.2 Material - Ti-6Al-4V (Ti alloy)

Titanium is a low density, having approximately 60% that of steel, and has one of the highest strength-to-weight ratios and corrosion resistances of the metals [127]. Titanium is extracted from rutile (TiO₂) and processed in multiple steps to obtain the finished material [127]. In its pure form, it is a ductile material and is alloyed with other materials to gain toughness [128]. Pure titanium is an allotropic element, that is it has more than one crystallographic structure. At room temperature titanium has a hexagonal close pack (hcp) crystal structure (α phase) and this is changed to body centred cubic (bcc) crystal structure (β phase) above the phase transformation temperature of 882°C [129], which is also called β-transus temperature. Figure 3.1 shows the unit cells of α and β phases of titanium crystal. The titanium alloys are categorised according to these phases as α alloys, β alloys and α + β alloys [127].

One of the most widely used titanium alloys is Ti-6Al-4V, which is an alpha + beta alloy containing 6% Aluminium (Al) and 4% Vanadium (V). Ti-6Al-4V accounts for 80% of the USA market share of the alpha + beta Ti alloys [129], which shows its importance to industry. It has an excellent combination of strength and toughness along with excellent corrosion resistance [127]. Because of these extraordinary properties, it finds extensive applications in the aerospace, biomedical and chemical industries. Table 3.1 shows a comparison of some physical properties of titanium and its alloys with other materials. The lower thermal conductivity makes it a
favourable material for the laser deposition as compared to the other metallic materials because conduction losses are lower.

![Figure 3.1](image1.png)

Figure 3.1. (a) Unit cell of α phase, (b) Unit cell of β phase [129]

Table 3.1 Some physical properties of titanium and its alloys as compared to other structural metallic materials [129]

<table>
<thead>
<tr>
<th>Material</th>
<th>Thermal expansion coefficient ($10^{-6}$ K$^{-1}$)</th>
<th>Thermal conductivity (W m$^{-1}$ K$^{-1}$)</th>
<th>Specific heat capacity (J kg$^{-1}$ K$^{-1}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>α Ti</td>
<td>8.4</td>
<td>20</td>
<td>523</td>
</tr>
<tr>
<td>Ti-6Al-4V</td>
<td>9.0</td>
<td>7</td>
<td>530</td>
</tr>
<tr>
<td>Fe</td>
<td>11.8</td>
<td>80</td>
<td>450</td>
</tr>
<tr>
<td>Ni</td>
<td>13.4</td>
<td>90</td>
<td>440</td>
</tr>
<tr>
<td>Al</td>
<td>23.1</td>
<td>237</td>
<td>900</td>
</tr>
</tbody>
</table>

Al is a strong α stabilizer element whereas V is a strong β stabilizer element. Ti-6Al-4V exhibits three different types of microstructure depending upon the thermo-mechanical process route; fully lamellar microstructure, fully equiaxed microstructure and bi-modal microstructure containing primary α in lamellar α + β matrix [129]. Figure 3.2 shows the bi-modal microstructure image obtained transmission electron microscopy (TEM). As seen, the recrystallized β phase penetrates into the recrystallized α lamellae causing separation into the individual primary α grains.
Figure 3.2. Bi-modal microstructure image (TEM) of Ti-6Al-4V at 950 °C [129]

Figure 3.3. Fully equiaxed microstructure of Ti-6Al-4V alloy; (a) optical microscope image; (b) transmission electron microscope image [129]

Figure 3.3 shows fully equiaxed microstructure of the Ti-6Al-4V recrystallized at 800 °C. Very fine α grains with size of about 2 µm are visible in TEM image.

The cooling rates affect the mechanical properties of the alloy. With an increase in cooling rate the α colony size is decreased with a corresponding increase in yield stress, when the tensile ductility reaches a maximum and then decreases [129]. Table 3.2 lists the thermo-physical properties of the Ti-6Al-4V and their temperature dependencies.
Table 3.2. Temperature dependent thermo-physical properties of Ti-Al-4V alloy [55]

<table>
<thead>
<tr>
<th>Temperature (K)</th>
<th>293</th>
<th>473</th>
<th>673</th>
<th>1073</th>
<th>1893</th>
<th>1933</th>
</tr>
</thead>
<tbody>
<tr>
<td>Thermal conductivity (W m(^{-1}) K(^{-1}))</td>
<td>10</td>
<td>12</td>
<td>15</td>
<td>17</td>
<td>20</td>
<td>60</td>
</tr>
<tr>
<td>Specific heat capacity (J kg(^{-1}) K(^{-1}))</td>
<td>580</td>
<td>680</td>
<td>760</td>
<td>800</td>
<td>930</td>
<td>950</td>
</tr>
<tr>
<td>Density (kg m(^{-1}))</td>
<td>4420</td>
<td>--</td>
<td>--</td>
<td>4300</td>
<td>4200</td>
<td>3800</td>
</tr>
</tbody>
</table>

### 3.3 LDMD setup

LDMD setup is comprises of a laser, a CNC motion system, a powder feeder, a coaxial deposition head and an enclosure box as shown in Figure 3.4. All the experiments in this research were carried out using this setup. A Laserline LDL160-1500 diode laser, with a maximum power of 1.5 kW, manufactured by Laserline GmbH is used in the LDMD apparatus. It operates in the near infrared wavelength range with an approximately equal intensities of 808 nm and 940 nm radiation. The output beam of the laser has an approximately “top hat” intensity distribution and is fitted with optics with a focal length of 300 mm. The wall plug efficiency of the diode lasers has been reported 30-40% [1].

A SIMATIC OP3 disc powder feeder manufactured by Flame Spray Technologies Ltd, UK is used for powder mass delivery in the deposition experiments. The powder feeder comprises two canisters of 1.5 litres capacity each and a mixing motor to distribute the powder evenly on the rotating disks. The mass flow rate can be controlled electronically using a Siemens controller by adjusting the speed of disks in revolutions per minute (rpm). Argon is used as a carrier gas for powder mass delivery to the coaxial deposition head.
Figure 3.4. LDMD experimental setup; (a) powder feeder, diode laser control panel and laser enclosure box; (b) coaxial deposition head; (c) coaxial nozzle with powder stream

For calibration purposes, the Ti-6Al-4V powder with 50-120 µm particle size was collected in a container at different rpm for known time periods. By weighing the delivered powder, the mass flow rates (g s\(^{-1}\)) were approximated for different speeds (rpm) of the disks of the powder feeder.

The coaxial deposition head consists of Laserline collimating and focusing optics and a coaxial nozzle made by DeBe Lasers Ltd. The coaxial nozzle consists of a central laser passage with a gas stream to protect the optics, a single powder outlet annulus and an annular water cooling passage. An Isel C116-4 CNC table controls
the motions in xy plane. The deposited layer patterns can be controlled by programming the CNC table through a PC attached with it. The coaxial deposition head and an Isel C116-4 CNC motion control table were housed in the enclosure box (Fig. 3.4).

For all LDMD experiments, the laser was used in continuous mode. The Ti-6Al-4V powder was of size 50-120 µm and the substrate blocks were of a nominal size 50 mm × 50 mm × 10 mm. The substrate blocks were sand blasted for laser absorption enhancement and degreased using ethanol prior to the experiment. The stand-off distance between the coaxial deposition head and the deposition point on the substrate was constant at the value of 7.5 mm which meant the focal plane was coincident with the deposition surface at all times. The laser spot size at substrate surface was 1.7 mm. Movements in the x-y plane were controlled by an Isel C116-4 CNC motion control table and each track within a layer was deposited with the substrate moving in same direction. The powder feeder with Argon conveyance gas with a flow rate of 5 litres per minute was used for powder mass delivery. Another flow of Argon gas, with a rate of 6 litres per minute, through the central passage of the coaxial deposition head was used to shield the objective lens and to minimize oxidation. The combined, targeted flow of 11 litres per minute of the Argon was sufficient to provide a local inert atmosphere to minimize oxidation in the deposition process.

3.4 Characterization equipment and methods

3.4.1 Metallography

For metallography, all LDMD samples were initially sectioned in two traverse planes near the middle of the wall about 10 mm apart and then hot mounted in conductive mounting resin (Struers ConduFast) using a Struers Labopress-3. The cross-sectioned planes were ground using a Presi Mecatec-334 polishing machine and then polished with diamond suspensions of 6 µm, 3 µm and 1 µm diamond grain sizes using the same machine. The samples were then loaded on an oxide suspension polishing (OSP) machine for two hours to produce a scratch-free and deformation-free surface. Kroll’s Reagent was used to etch the fine polished samples to reveal grain boundaries.
3.4.2 Optical microscopy and scanning electron microscopy

Optical microscopy and scanning electron microscopy are normally used for characterization of the laser metal deposition e.g. microstructure, interlayer and intralayer porosity. A visible light and a system of lenses that magnify the images, is the basic working principal of optical microscopy. But optical microscopy has a limited resolution which is defined by the diffraction limit of the visible light. A Polyvar-MET optical microscope, manufactured by Reichert/Leica Austria, along with 2x, 5x, 10x, 20x, 50x, and 100x objective lenses to capture quality images, was used for microstructure characterization of the LDMD samples. A 12 MP digital camera and an iSolution DT software for image post processing were coupled with microscope. Illumination on the sample was made by a tungsten halogen bulb and adjusted by changing voltage control. Figure 3.5 shows the optical microscopy setup.

![Optical microscopy setup at The University of Manchester](image)

A Hitachi (S-3400N) SEM manufactured by Hitachi High Technologies, Japan was used in this work as shown in Figure 3.6. The SEM works on the principle of exposing the sample with high acceleration electrons which determine the topography of the sample. The SEM works with Secondary Electron (SE) and Back Scattered Electron (BSE) detectors and requires the sample to be conductive. The SEM has tungsten filament featuring a magnification from 5X to 300kX and the acceleration voltage range of the system is 0.3-30 kV. The SEM used had a lateral resolution of 3-4 nm.
3.4.3 Microcomputed tomography (MicroCT)

In microcomputed tomography, an x-ray beam penetrates through the object and makes images of the object on an image detector. Microcomputed tomography is a powerful non-destructive testing technique able to produce 2D images of the sample being tested. A series of 2D images are taken around the single axis of rotation of the sample. The stacks of 2D images are further reconstructed to make 3D object. Usually, some visualization software is used to post process and analyze the tomography data.

In this research, an Xradia microXCT-400 system, as shown in Figure 3.7, was used to conduct MicroCT of powder samples. It features a x-ray source having a 7 µm spot size with a tungsten target, and operates with an accelerating voltage in the range 40-150 kV. The high resolution detector incorporates a scintillator coupled to a cooled CCD (coupled charged device) camera by a microscope-type objective, which provides optical magnification. The highest optical magnification (40x) gives a resolution of 0.7 µm.
Microcomputed tomography of the LDMD samples including thin-wall and porous structure were carried out using a Metris XTH 225 system as shown in Figure 3.8. The Metris XTH 225 kV x-ray source is housed in a standard radiation cabinet. The radiation cabinet incorporates a 5-axis specimen manipulator stage that is capable of supporting specimens of 15 kg. Within the enclosure it is possible to analyse specimens ranging in size from a few millimetres up to 170 millimetres in cross-section and also possible to image selected regions of interest within a larger specimen.
Avizo 6 software was used to visualization and post processing of the tomography data of the powder and LDMD samples. This includes the porosity quantifications, pore size distributions and visualization graphics.

### 3.4.4 Pyrometry

Pyrometry is a non-contact temperature measurement technique based on the working principle; every object above the absolute temperature emits radiations which are dependent on the temperature of that object. The radiated energy is used to determine the temperature of the material being measured. Two colours pyrometer measures the radiations using two different wavelengths, the ratios of the signal and finally calculate the temperature. The emissivity is eliminated as a part of calculations and temperature measurement becomes emissivity free.

In this research, a two colours pyrometry was employed to measure the substrate temperatures during the LDMD process. An IGAR 12-LO two colours pyrometer manufactured by Impac Infrared Thermometry, as shown in Figure 3.9, with the temperature measuring range 772 K to 2272 K was used.

![An IGAR 12-LO two colours pyrometer at The University of Manchester](image)
3.4.5 Surface roughness testing

Surface roughness testing was employed to the surface cladding layers in multiple track laser deposition and on the top surface in case of thin-wall deposition samples. Because of the different nature of samples analysed in this work two different types of surface roughness measuring equipment were used.

Laser Surface Profile Scanning System as shown in Figure 3.10, was used to measure surface roughness ($R_z$) of the surface cladding layers made by multiple track. The system consists of a LK031 laser distance sensor with 30 µm diameter beam spot, ±5 mm measuring range and 1 µm resolution. The system is designed for non-contact detection of the 3-D profile of objects and it converts surface topology into digital data points. While measuring the surface topology of the surface cladding layers, the scanning step resolution was set at 0.2 mm in x and y directions with a measuring area of 10 mm × 10 mm.

The surface roughness ($R_a$) on the top of the thin-wall LDMD samples was measured using Wyko NT1100 White Light Interferometer as shown in Figure 3.11.

Figure 3.10. Laser surface profile scanning system at The University of Manchester
White light interferometry is a fast non contact surface topography measurement technique. The working principle of the interferometer is that a light beam is split into two beams with one reflected from the sample and other reflected from a reference image. The two beams are then recombined and the interference data of the sample is used to generate the topography of the sample. The surface roughness was checked on the top of the samples with scanning area of 1000 µm × 500 µm.

### 3.4.6 Micro hardness testing

Vickers micro hardness testing was undertaken using a Micromet® 5114 micro-indentation hardness tester manufactured by Buehler Instruments, as shown in Figure 3.12. The micro hardness tests were performed on the fine polished mounted LDMD samples. The hardness tester forces a small pyramid into the surface of the sample with a known load and the Vickers hardness number is then obtained from the diameter of the impression.
Figure 3.12. Buehler micro hardness testing machine at The University of Manchester

The indentation impression is measured diagonally in millimetres. The diagonal lengths of indentations were measured under the magnification of 10X. The Vickers hardness number can be calculated using the following relation

$$HV = \frac{1.8544W}{d^2}$$

(3.1)

where $d$ is the mean value of the diagonal in millimetres and $W$ is the imposed load in kilogram. The hardness testing were conducted according to the EN ISO 6507-1 standard with a 0.5 kg load which is considered as low load Vickers hardness and referred as “HV 0.5”.
3.4.7 X-ray diffraction

In 1917 A. W Hull [130] gave a paper where he indicated that

“... every crystalline substance gives a pattern; the same substance always gives the same pattern; and in a mixture of substances each produces its pattern independently of the others”

The x-ray diffraction (XRD) pattern of a pure material can therefore, be considered as a fingerprint of the material. It is a powerful technique to identify the crystalline materials. A lot of information about a material e.g. crystal structure (size, phase etc.), lattice strain, grain orientation (texture) can be extracted from the XRD data. Bragg’s Law is the theoretical base of the x-ray diffraction which is given as below.

\[ n \lambda = 2d \sin \theta \]  

(3.1)

where, \( \lambda \) is wavelength of incident x-ray beam, \( \theta \) is the Bragg’s angle, \( d \) is crystal d-spacing and \( n \) is an integer.

The x-ray diffraction was employed on thin-wall LDMD samples to carry out phase analysis of the microstructure. A Pananalytical Xpert-MPD diffractometer system with a Phillips, PW 3040 Goniometer, as shown in Figure 3.13, was used for the analysis. The system uses Cu radiations of 1.54060 Å wavelength. Diffraction peaks were recorded at 2\( \theta \) angle of about 90° starting from 30 to 120 degrees. The respective diffraction peaks were shown with respect to the crystal d-spacings which were calculated from 2\( \theta \) angle using the relation given in Equation (3.1).
3.4.8 Laser diffraction

Particle size analysis of the Ti-6Al-4V powder was conducted on Malvern Mastersizer laser diffraction equipment. Figure 3.14 shows the laser diffractometer used in this work. Laser diffraction is a non-destructive and non-intrusive technique that can be used to derive particle size using fundamental scientific principle. Laser diffraction based particle size analysis works on the principle that particles passing through a laser beam will scatter light at an angle which is directly related to their size. As particle size decreases the observed scattering angle increases. Therefore, large particles scatter light at narrow angles with high intensity whereas small particles scatter at wider angle with low intensity. A typical system consists of a laser to provide a coherent light of fixed wavelength; a series of detectors to measure the light pattern produced over wide range of angles; and some kind of sample holding system that ensure that sample material passes through the laser beam as homogenous stream of particles.

In the laser diffraction, particle size distribution is calculated by comparing it with an appropriate optical model e.g. Fraunhofer Approximation or Mie Theory. Mie theory provides a more rigorous solution for calculation of the particle size distributions from the light scattering date and hence used in modern laser diffractometer.
Figure 3.14. Malvern Mastersizer laser diffractometer system at The University of Manchester
Chapter 4

Analytical modelling of multiple track laser direct metal deposition for surface layer cladding

4.1 Introduction

Despite being increasingly widely adopted for rapid tooling and manufacture, the biggest application of laser deposition or multiple track laser cladding is surface coating and repair. It has been shown that crack free repair or modification of parts can be affected. However, due to the large number of variable involved in LDMD, results are very dependent upon the process parameters used and experimental optimization of them is time consuming. Modelling the process and resultant deposition characteristics would be a more efficient way of setting the correct parameters. Although numerical software has proved highly successful at modelling aspects of the deposition process such as heat flow in the substrate, the multi-phase nature of the process makes it difficult to establish an easily applied model of track geometry. Analytical models meanwhile have tended to address single tracks or thin walls.

In this chapter, an analytical modelling method for prediction of the geometry of multiple, interacting laser direct metal deposition tracks for surface layer deposition is proposed. The model can accommodate single tracks, partially overlapping tracks suitable for surface cladding and multiple layers, as required when replacing areas of worn material during a repair operation. The main model comprises the sub-models of the powder stream, quasi-stationary conduction in the substrate, powder assimilation into the melt pool area and the multiple tracks interactions.

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A Laserline 1.5 kW diode laser along with a coaxial deposition head is employed to verify the model. Ti-6Al-4V, an aerospace alloy that is currently used in turbine blade tip repair is used as deposition and substrate material. The model predictions show good agreement with deposition layer profiles, layer thickness and layer roughness \( (R_z) \). The model uses novel methods to predict interacting track geometries and is useful as a tool for selecting the correct parameters for surface modification by multiple track laser metal deposition.

4.2 Previous work

Melt pool size and shape modelling has so far been done by various researchers in the literature. Picasso et al [27] approximated the melt pool shape as an ellipse using a 3-D analytical model and it was assumed that the laser beam and powder particles were two intersecting cylinders. Toyserkani et al [131] presented a transient finite element model to predict the track geometry and its relation with processing parameters. This model calculated the interaction between powder and melt pool by decoupling heat and mass flows and the melt pool boundary was obtained first in the absence of a powder spray. In a similar manner, Alimardani et al [67] published a morphological model for laser cladding accounting for the powder efficiency in the thermal and geometrical calculations. Pinkerton and Li [63] analytically modelled the moving melt pool and deposition track; the model addressed the issue of a track width greater than the melt pool width at higher powder flow rates.

A combination of analytical and finite element models was used by Peyre et al [55] to model multiple additive tracks with full overlap to build a thin wall, and process maps for laser deposition of thin wall structures were developed by Vasinonta et al [48-50] using finite element modelling alone. Qi et al [65] numerically simulated the heat transfer mechanism in coaxial laser cladding using the level-set method to track the free surface movement of the melt pool. Wang et al [132] numerically modelled thermal behaviour in the laser deposition process and also described the effect of Marangoni convection on the melt pool temperature. Kaplan and Groboth [60] analytical modelled the melt pool size and track geometry for a single track, assuming the laser was a moving Gaussian source of heat. For a multiple track layer with overlapping tracks the same melt pool geometry was assumed throughout whatever the overlap ratio and changes in the melt pool size and track geometry due to the interaction were not modelled.
He et al [133] numerically modelled the temperature and composition profiles of doubled-track laser deposition; it was found that the peak temperature location shifted 0.34 mm above the substrate level for the second track, but the effect of track overlap on the melt pool and resulting deposition layer geometry was not modelled. Roberts et al [134] presented the 3D finite element model of multiple additive layers manufacturing model which primarily described the change in thermal distribution in subsequent deposited layers.

In summary, the previous researches described thermal histories in single and multiple track laser deposition very well. Some of them predicted the geometries of the thin-wall depositions, but the surface layer geometry modelling comprising of multiple track interaction was not modelled.

4.3 Mathematical model

4.3.1 Model assumptions

The model assumes the substrate as semi-infinite for the governing equations to remain valid. Although any substrate would actually be finite in size this is a reasonable approximation as the laser heat input is highly localised and moving. The laser moves at a constant speed parallel to the y-axis. The process is quasi-stationary, meaning it is stationary in a reference frame attached to the laser beam. The laser beam is perpendicular to the substrate.

Deposition material and substrate are of the same material and considered as isotropic and homogeneous. To simplify the model, and as is customary in analytical heat flow models, all thermophysical properties of the material are assumed to be temperature independent. Ti-6Al-4V thermophysical properties as given in Chapter 3 (Table 3.2) are used and values at half way between the ambient and the melt point temperatures are taken. Powder flows at constant speed with a Gaussian concentration distribution across the powder stream [42]. All powder impinging upon the melt pool is absorbed by it and the rest is considered as lost [37].

Convection and radiation heat losses are ignored, as justified by previous research [45]. However, the model accounts for the power losses due to the vaporization heat flux and the power losses in the powder stream.
4.3.2 Powder stream losses

The laser beam and powder stream interaction causing beam attenuation and powder heating is beyond the scope of this chapter. But the powder stream cloud injecting in the way of laser beam causes energy loss tends to decrease the melt pool size as compared to the melt pool size obtained in de-coupled mass and heat flow calculations. This inaccuracy could be partially overcome if the laser power loss in powder stream subtracted from the initial laser power giving a reasonable melt pool size. Power loss in the powder stream, \( P_p \), can be found from the following equation

\[
P_p = q_{pr} \left( 1 - \text{Exp} \left( -\frac{2r_i^2}{r_p^2} \right) \right) \left[ C(T_m - T_{amb}) + L_m \right]
\]

where \( C, T_m, T_{amb}, L_m, q_{pr} \) and \( r_p \) stand for specific heat capacity, material melting temperature, ambient temperature, latent heat of melting, powder mass flow rate and Gaussian powder stream radius respectively. The Equation (4.1) considers the powder stream with Gaussian distribution and gives the laser power loss due to the attenuation of the laser beam.

4.3.3 Melt pool geometry and losses

To determine the melt pool size, a method to de-couple heat and mass flow and then find the boundaries at which the substrate material reaches to its solidus temperature is used. Cline and Anthony’s [135] equation for a moving Gaussian source of heat is used to calculate the temperature profiles on the substrate and hence melt pool size

\[
T(x, y, z) = \frac{ap_{l}(P_f + P_{evap})}{kr_l} \int_0^\infty \frac{\text{Exp}(-H)}{(2\pi^2)^\frac{3}{2}(1+\mu_x)} d\mu
\]

Where the temperature distribution function \( f \) is

\[
f(x, y, z, \nu) = \int_0^\infty \frac{\text{Exp}(-H)}{(2\pi^2)^\frac{3}{2}(1+\mu_x)} d\mu
\]

Here \( k \) is thermal conductivity; \( \nu \) is scanning speed; \( r_l \) is laser beam radius; \( P_f \) is laser power; \( \alpha \) is laser absorptivity; \( \mu \) is dimensionless time and \( \sigma \) is dimensionless speed. \( \kappa \) is thermal diffusivity. \( X, Y \) and \( Z \) are the linear dimensions made dimensionless by dividing by the laser beam radius, i.e. \( \mu = \frac{\sqrt{2\kappa t}}{r_l}, \sigma = \frac{rtw}{\kappa}, X = \frac{x}{r_l}, Y = \frac{y}{r_l} \) and \( Z = \frac{z}{r_l} \).
Equation (4.2) gives the three-dimensional temperature distributions in the substrate while the laser is moving in the y-direction with a uniform speed. This equation is solved in Wolfram *Mathematica* and the melt pool divided into two regions, one representing the pool in front of the heat source centre point and one the pool behind the heat source centre point. The boundaries of these regions are approximated as quarter ellipsoids defined by the melt pool width, $W$, the melt pool length in the forward direction, $L_1$, the melt pool length in the rear direction, $L_2$, and the melt pool depth at the Z-axis, $D$, as shown in Equations (4.5), (4.6) and Figure 4.1.

For $y > 0$, $z > 0$:

$$\frac{y^2}{L_1^2} + \frac{4x^2}{W^2} + \frac{z^2}{D^2} = 1$$

(4.5)

For $y < 0$, $z > 0$:

$$\frac{y^2}{L_2^2} + \frac{4x^2}{W^2} + \frac{z^2}{D^2} = 1$$

(4.6)

Where:

$$T\left(\pm \frac{W}{2}, 0, 0\right) = T_m$$

(4.7)

$$T(0, L_1, 0) = T_m$$

(4.8)

$$T(0, -L_2, 0) = T_m$$

(4.9)

$$T(0, 0, d_{max}) = T_m$$

(4.10)

For calculations of evaporation heat flux losses the overall evaporation model of Choi et al [136] is used. The evaporation heat flux, $Q_{evap}$, is given in terms of evaporation mass flux and latent heat of vaporization as:

$$Q_{evap} = m_e h_v$$

(4.11)

Where $m_e$ is of the form

$$\log m_e = A + 6.1210 - \frac{18836}{T} - 0.5 \log T$$

(4.12)

Where, $T$ is the melt pool temperature as calculated by the Cline equation (4.2). The $A$, is a constant dependent on the material e.g. iron 4.82, nickel 4.834 and chromium 4.807 for $m_e$ in kgm$^{-2}$s$^{-1}$ and temperature in Kelvin [16]. The value of $A$ for nickel is taken as good approximation for non-ferrous metals.

Combining Equation (4.11) and (4.12) gives:

$$Q_{evap} = \left(\frac{455.32 \exp\left[4.384 - \frac{18836}{T}\right]}{\sqrt{T}}\right) h_v$$

(4.13)
Chapter 4: Analytical modelling of multiple track LDMD for surface layer cladding

The above equation gives the heat flux per unit area. The laser power loss due to evaporation heat flux, $P_{evap}$, from the melt pool can be calculated by applying it to the melt pool surface area. This power loss is fed back to Equation (4.2).

$$P_{evap} = \frac{\pi}{4} W(L_1 + L_2) \left( \frac{455.32 \exp\left[ \frac{4.934 \cdot 18836}{T} \right]}{\sqrt{T}} \right) h_v$$ (4.14)

4.3.4 Mass addition

After defining the melt pool limits, mass can be added within these limits which results in a deposition profile. To determine the deposition height, $h_z$, at any $x$ position the model integrates the powder mass flux over the distance between the front and rear limits of the melt pool. If $y_{front}$ and $y_{rear}$ are these limits on the substrate

![Figure 4.1 Schematic diagram of a moving melt pool in y-direction with powder mass addition: (a) plan view; (b) side elevation](image)
surface \((z = 0)\) then at any \(x\) position \(|x| \leq W/2\), as illustrated in Figure 4.1, the melt pool can assimilate powder that falls within these limits. Equations (4.5) and (4.6) can be re-arranged at \(D = 0\) to give these values:

\[
y_{\text{front}} = L_1 \sqrt{1 - \frac{4x^2}{W^2}} \tag{4.15}
\]

\[
y_{\text{rear}} = -L_2 \sqrt{1 - \frac{4x^2}{W^2}} \tag{4.16}
\]

The powder distribution in coaxial laser direct metal deposition has been shown to have a Gaussian distribution \([42]\). The coaxial powder stream with powder mass flux \(q_{pf}\) at any point \((x,y)\), with the laser beam centred at origin, can be written as:

\[
q_{pf} = \frac{2q_{pf}}{\pi r_p^2} \exp \left[ -2 \frac{x^2 + y^2}{r_p^2} \right] \tag{4.17}
\]

The deposition height, \(h_z\), is thus given by

\[
h_z = \frac{1}{pv} \int_{y_{\text{front}}}^{y_{\text{rear}}} q_{pf} \, dy \tag{4.18}
\]

Where, \(\rho\) is density of the deposition material and \(v\) is velocity. Equation (4.18) applies to the full extent of the melt pool, \(-W/2 \leq x \leq W/2\).

### 4.3.5 Track interaction

Multiple track laser deposition to create a surface layer involves tracks overlapping and thus interacting with a previously deposited track as well as the substrate. To model this, the simulated melt pool size and powder assimilation will be modified. Parallel tracks are considered as this is the most common deposition pattern used industrially. During deposition of a second, parallel, overlapping track, the melt pool will become partially inclined due to the difference in height levels of the substrate and the first track on which it is overlapped. This inclination effect will reduce the melt pool area projected normal to the surface in the overlapped region compared to the non-overlapped region. The melt pool for the second track is modelled to incorporate this effect. Consider Figure 4.2(a), in which the melt pool is moving in the \(y\)-direction with a uniform speed and partially overlapping the previous track. The track spacing is given by the variable \(x_{\text{centre}}\), which represents the centre to centre distance between the consecutive overlapped tracks. The tracks thus overlap and interact if \(x_{\text{centre}} < W\). The area of the melt pool when over the previous track is taken by assuming the boundaries of the melt pool are straight lines rather than arcs
of ellipses. Using this method, a simplified diagram of the geometry of the overlapped region is shown in Figure 4.2(b), which will be used to calculate the modified melt pool limits.

The $y_{\text{front}}$ and $y_{\text{rear}}$ are the modified melt pool limits in forward and rear direction respectively and can be written as:

$$y_{\text{front}} = \frac{L_1}{W-x_{\text{centre}}} \sqrt{1 - \frac{4(x_{\text{centre}} - \frac{W}{2})^2}{w^2}}$$

Figure 4.2. Schematic diagram of multiple tracks; (a) melt pool overlapping a previous track surface; (b) overlapped melt pool region used for calculation of the modified melt pool limits

$$y_{\text{front}} = \frac{L_1}{W-x_{\text{centre}}} \sqrt{1 - \frac{4(x_{\text{centre}} - \frac{W}{2})^2}{w^2}}$$

(4.19)
\[ y_{rear1} = -L_2 \frac{1 - \frac{4(x_{centre} - \frac{W}{2})^2}{W^2}}{w - x_{centre}} (x - x_{centre} - \frac{W}{2}) \]  

The deposition height, \( h_a \), in the overlapped region can be calculated using the modified melt pool limits in Equation (4.18).

\[ h_a = \frac{1}{pv} \int_{y_{front1}}^{y_{rear1}} q_{pf} \, dy \]  

The deposition height in non-overlapped region can be obtained using Equation (4.18).

### 4.3.6 Powder affinity

In multiple track deposition, another important factor is how the change in surface geometry affects the powder affinity. Powder affinity is defined as a higher probability of the powder assimilation at a particular substrate/clad area due to its surface topology as compared to other areas in that vicinity. The effect of non-planar surfaces on laser beam absorptivity has been introduced [67], but how these surfaces change powder absorption has not been mentioned in previous research. Consider Figure 4.3, the track corner shows a highly concave surface and has high affinity for powder as compared to other portions of the track. This may be related to powder having lesser tendency to ricochet from the surface at this position, the area tending to attract stray powder or other reasons.

The coaxial nozzle produces a Gaussian distribution of powder, giving a maximum powder flux under the central axis of the nozzle. The effect of this high powder affinity is therefore particularly pronounced at overlap ratios of about 50% (\( x_{centre} = W/2 \)) because the nozzle central axis is near to the limit of the previous track. As the overlap ratio increases the nozzle moves away from the high powder affinity zone reducing the effect. The powder affinity factor, \( \varepsilon \), can be written in the form:

\[ \varepsilon = 1 + \left( 1 - \psi \right) \omega \tau \left( \frac{w}{2} - x \right)^2 \]  

Where \( \psi \) is the overlap ratio defined as \( (1 - x_{centre}/W) \). \( \omega \), and \( \tau \), are laser deposition constants which are dependent upon coaxial deposition nozzle dimensions and powder gas flow rates respectively. For modelling with the coaxial deposition nozzle used, the empirical values of \( \omega = 5 \) and \( \tau = 7 \times 10^7 \) are taken.
After incorporating the effect of the powder affinity factor, the modified deposition height, $h_{ap}$, for the overlapped region, and $h_{bp}$, for non-overlapped region, can be written as:

$$h_{ap} = \frac{\varepsilon}{\rho_{ PV}} \int_{y_{front}}^{y_{rear}} q_{PF} \, dy$$

$$h_{bp} = \frac{\varepsilon}{\rho_{ PV}} \int_{y_{front}}^{y_{rear}} q_{PF} \, dy$$

The final deposition height for both regions, i.e. the total track, $h_z$, becomes:

$$h_z = h_{ap} + h_{bp}$$

Equation (4.25) defines the deposition profile for two tracks with the second track overlapping the first. Subsequent tracks are dealt with in the same way as above, each time allowing for interaction with the previously deposited track, leading to a defined layer.

### 4.3.7 Model solution and results

The overall operation of the multiple tracks deposition model is shown in flow chart form in Figure 4.4. First heat flow calculations are performed without accounting for laser power losses to find approximate temperature distribution and melt pool size. These are used to calculate the heat flow losses and then on the basis of this more accurate heat flow calculations are performed to quantify melt pool size and temperature. The process parameters and the thermo-physical properties of the Ti-6Al-4V, given in Table 4.1, are used for the model calculations. The powder absorption efficiency was taken as 80% of the powder mass flow rate.
A routine has been written in Wolfram Mathematica software to realize the model and post-process and display the results. Numerical integration, polynomial least-squares fit and logic selection techniques are used. Figure 4.5 shows the ability of the model to predict surface layer deposition at different overlap ratios.
Chapter 4: Analytical modelling of multiple track LDMD for surface layer cladding

Table 4.1. Process and material parameters for model deposition profile calculations

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Symbol</th>
<th>Unit</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Laser absorptivity</td>
<td>$\alpha$</td>
<td>------</td>
<td>0.5</td>
</tr>
<tr>
<td>Laser fibre efficiency</td>
<td>$\beta$</td>
<td>------</td>
<td>0.7</td>
</tr>
<tr>
<td>Melting temperature $^a$</td>
<td>$T_m$</td>
<td>K</td>
<td>1933</td>
</tr>
<tr>
<td>Ambient temperature</td>
<td>$T_{amb}$</td>
<td>K</td>
<td>293</td>
</tr>
</tbody>
</table>

$^a$ Values taken at 1073 K temperature

4.4 Experimental verification procedure

The model was verified with a number of surface layer deposition experiments using Ti-6Al-4V (an aerospace grade Ti alloy) as substrate and deposition material (powder). Two levels of laser power, three levels of powder flow rate and four levels of track overlap ratio at a constant scanning speed of 10 mm.s$^{-1}$ were tested, as shown in Table 4.2, giving a total of twenty four experimental runs. Track overlap ratio is defined as the percentage width of the previous track overlapped by the consecutive track. The detail of the LDMD experiment is given in Section 3.3 of Chapter 3.

Table 4.2. Processing parameters for laser direct metal deposition

<table>
<thead>
<tr>
<th>LDMD Variables</th>
<th>Levels</th>
</tr>
</thead>
<tbody>
<tr>
<td>Laser power (W)</td>
<td>800 1100</td>
</tr>
<tr>
<td>Powder flow rate (g.s$^{-1}$)</td>
<td>0.066 0.089 0.112</td>
</tr>
<tr>
<td>Overlap ratio (%)</td>
<td>10 20 40 60</td>
</tr>
</tbody>
</table>

Samples for microscopy were prepared using standard metallography techniques, details can be found in Section 3.4.1 of Chapter 3. Images of the samples were captured and after calibration the deposition layer thickness of each sample was recorded. The layer thickness was checked at four different locations of each samples and a variation of the data was estimated to ± 5% of the mean value.
Chapter 4: Analytical modelling of multiple track LDMD for surface layer cladding

Figure 4.5. Modelled multiple track surface layer deposition of Ti-6Al-4V at 800 W, 0.089 gs$^{-1}$, 10ms$^{-1}$; (a) 0% overlap; (b) 10% overlap; (c) 20% overlap. The substrate level is at Z=0
A Laser Surface Profile Scanning System was used to measure the surface roughness ($R_Z$) of the deposition layer on each sample. Surface roughness ($R_Z$) was calculated across the deposition tracks, perpendicular to the laser traverse direction on a 10 mm × 10 mm scanning area. The layer surface roughness was checked at three different locations of each samples and a variation of the data was estimated to ± 20% of the mean value.

### 4.5 Results and Discussion

A typical deposited layer profile is shown in figure 4.6; columnar prior beta grains can be seen. Melt pool boundaries of individual overlapping tracks are faintly visible. Figure 4.7 compares the experimental and modelled surface layer cladding profiles at three different overlap ratios, two laser power levels at constant powder mass flow rate and scanning speed. The modelled and experimental layer profiles look reasonably similar at equivalent process parameters.

Figure 4.8 shows the comparison of modelled and experimental results of mean deposition layer thickness at two laser power levels and three mass flow levels. In all cases, the model follows the experimental trends with good accuracy up to 40% overlap ratio. The model slightly under estimates the layer thickness at higher overlap ratios at almost at all processing parameters. This can be attributed the fact that laser beam absorptivity characteristics changes at higher overlap ratio due to the rough surface of the previous deposited track. The higher absorptivity results in a longer melt pool and hence a thicker deposited layer. Effects such as change in absorptivity are not catered for in the model.

Surface roughness ($R_Z$) values of experimental deposition layers produced at different laser powers and powder mass flow rates are compared with the model in Figure 4.9. At 800 W laser power and 20 % overlap ratio the roughness values changes from approximately 200 µm to approximately 300 µm when powder flow rate changes from 0.066 gs⁻¹ to 0.089 gs⁻¹. Overall, a decrease in surface roughness
with overlap ratio has been seen. There is an increase in roughness with powder flow rate and laser power. The model reasonably follows these experimental trends.

It is found from the experiments that deposition layer thickness increases with overlap ratio and that surface roughness increases with increases in the powder mass flow rate at a constant overlap ratio. The model predicts the same behaviour of the layer thickness and surface roughness, as evident from Figures 4.8 and 4.9, which are also consistent with previous published work [60, 137]. Track overlap ratio plays an important role in determining the deposited layer’s thickness and surface roughness.
Figure 4.8. Experimental and model results comparison of deposition mean layer thickness at scanning speed 10 mms\(^{-1}\) and: (a) 800 W and 0.066 gs\(^{-1}\); (b) 800W, 0.089 gs\(^{-1}\); (c) 800 W, 0.112 gs\(^{-1}\); (d) 1100 W, 0.066 gs\(^{-1}\); (e) 1100 W, 0.089 gs\(^{-1}\); (f) 1100 W, 0.112 gs\(^{-1}\)
Figure 4.9. Experimental and model results comparison of surface roughness ($R_z$) at scanning speed $10\text{mms}^{-1}$ and; (a) 800 W and $0.066\text{gs}^{-1}$; (b) 800 W, $0.089\text{gs}^{-1}$; (c) 1100 W, $0.066\text{gs}^{-1}$; (d) 1100 W, $0.089\text{gs}^{-1}$

4.6 Conclusions

This chapter explained the development of a 3D analytical model of multiple overlapping laser direct metal deposition tracks for surface layer deposition. The laser power losses in powder stream are catered for in this model to partially overcome inaccuracies of the previous conventional models in case of de-coupled heat and mass flow calculations. The model accounts for track interaction in deposition of a single clad layer or multiple layers and its effects on the melt pool. A novel factor of ‘powder mass affinity’ is introduced to capture the effect of surface geometry on deposition rate for particular powder flow conditions. Powder mass affinity is a function of the track profile geometry, overlap ratio, nozzle dimensions, carrier gas flow rate and possibly other factors not examined in this work.

The surface layer profile, layer thickness and layer surface roughness of a number of surface deposition layers, produced using multiple track laser direct metal deposition
of Ti-6Al-4V alloy, show good agreement with the model simulations. Modelled and experimental results show the layer thickness to increase with mass flow rate and overlap ratio. The surface roughness of the multiple track surface layers decreases with increasing overlap ratio in the range 10-60%.
Chapter 5

Multilayer porous structures fabrication by continuous and pulsed-wave laser direct metal deposition (LDMD) for biomedical applications

5.1 Introduction

The use of engineered porous structures is gaining popularity in biomedical implant manufacture due to its ability to promote increased osseointegration and cell proliferation. LDMD is one of the few rapid manufacturing techniques that have the potential to produce a structure with graded porosity and/or composition from different biomaterials including titanium, Ti-6Al-4V, stainless steel and shape memory alloys. It can produce the implant with a greater degree of geometrical flexibility than other methods, however predicting the outcome from a particular combination of LDMD input parameters can be difficult because of the complex nature of the process and adoption of a ‘trial-and-error’ method is both slow and expensive when medical grade alloys are being used. There is thus a need for both a reliable method to produce structures with controlled porosity and a way of predicting their nature through advances in modelling.

In this chapter, LDMD with a diode laser in continuous mode and with a CO2laser in pulsed mode are used to produce multi-layer porous structures. Gas-atomized Ti–6Al–4V and 316L stainless steel powders are used as the deposition materials. The porous structures are compared with respect to their internal geometry, pore size, and part density using a range of techniques including microcomputed tomography.

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Results show that the two methods produce radically different internal structures, but in both cases a range of part densities can be produced by varying process parameters such as laser power and powder mass flow rate. Prudent selection of these parameters allows the interconnected pores that are considered most suitable for promoting osseointegration to be obtained. Analytical models of the processes are also developed by using Wolfram Mathematica software to solve interacting, transient heat, temperature and mass flow models. Measured and modelled results are compared and show good agreement.

5.2 Literature survey of biomedical implant manufacture

There is growing demand for biocompatible implants in the world, therefore the manufacturing technology required to produce them should be developed at the same pace. The estimated market for dental and orthopaedic implants is already over $4.8 billion and it is expected to grow briskly in the future. Thus, there are both humanitarian and economic incentives to optimize implant design and manufacture.

The development of a bone implant starts with an understanding of the bone characteristics. The bone tissues in an adult skeleton are arranged in two architectural forms. The first is trabecular, also called cancellous or spongy, bone and the second is cortical or compact bone [138]. Cortical bone is almost solid having only 10% porosity, when the trabecular bone has a much higher porosity of 50-90% making its modulus and compressive strength 20 times lower than that of cortical bone [138]. The trabecular bone is surrounded by the cortical bone [138]. Thus the bone can be considered as a composite material with graded porosity. Natural bone has density 1.8-2.1 g/cc, compressive strength 130-180 MPa and Young’s modulus 3-20 GPa [139].

The main requirements for implant-materials are bio-functionality (adequate mechanical properties), bio-corrosion resistance, biocompatibility, bio-adhesion, process-ability and availability as referred by Black and Hastings [140]. The possible material options include - metals, polymers and ceramics. Compared to other biomaterials, metals are widely used for many biomedical applications, including - hip and knee endoprostheses, plates, screws, pins and dental implants, due to outstanding tensile strength. A major problem with metallic implants in orthopaedic surgery is the mismatch of the mechanical properties between bone and bulk metallic
materials. Due to this mechanical mismatch, bone is insufficiently loaded and becomes stress-shielded, which eventually leads to bone resorption. This impairs the clinical performance of the prosthesis and is responsible for implant migration, aseptic loosening and fractures around the prosthesis as investigated by Kroger et al [141]. The relationship between implant flexibility and the extent of the bone loss has been established; it has been confirmed that the changes in bone morphology are an effect of the stress shielding and subsequent adaptive remodelling process as observed by Sychterz et al [142]. Therefore, there should be a suitable balance between strength and stiffness of the implants to best match the behaviour of bone for prolonged trouble-free performance.

One consideration to achieve a strong interfacial bond and also to reduce the modulus mismatch has been the development of implants with engineered porous materials as investigated by Kirshna et al [143]. Use of engineered porous materials effectively reduces the modulus mismatch and also provides pathways for bone ingrowth through the pores for stable long-term anchorage or biological fixation of the implant as explained by Pilliar [144]. Conventional powder metallurgical processing has been used to fabricate surface-treated or fully porous implants for biomedical applications as shown by Ik-Hyun [145]. However, conventionally sintered metals are very brittle and pore size, shape, volume fraction, and distribution are difficult to control. These factors all have a major influence on mechanical and biological properties.

Other fabrication techniques that use foaming agents, as elaborated by Wen et al [146], suffer from limitations such as contamination, impurity and limited part geometry. The applications of selective laser sintering in bone tissue engineering have been demonstrated using polymeric and ceramic composites [147], glass-ceramic materials [148], and metals (e.g. titanium [149]). The particular issues associated with selective laser sintering are material spheroidization under the laser beam, interparticle wetting and dimensional stability [150]. Fused deposition modelling has been applied to manufacture scaffolds for tissue engineering applications but the process has weaknesses such as the need to produce filament material and the inability to manufacture metallic materials [151]. Electron beam melting, on other hand, has limitations such as the need for a vacuum, production of gamma rays and the requirement of for the material to be conductive [138].
Pore size is also very important in the porous structures to be used for biomedical applications. Ingrowth of bone tissues into the pores is critical for proper osseointegration. Karageorgiou and Kaplan [152] found that a 100 µm pore was the minimum required for the migration requirements while Götz et al [153] showed that 200 µm is an optimal pore size for osseointegration in laser textured Ti-6Al-4V implants. Xue et al [154] showed that a pore size greater than 200 µm was required for cell ingrowth into the pores and for pore size less than 150 µm the cells span directly across the pores in spite of going into the pores. The bone cell responses were evaluated *in vitro* with human osteoblast cell (OPCI). So, proper osseointegration and cell proliferation depends upon a reasonable pore size selection in biomedical implants. LDMD is a rapid manufacturing technique capable of producing such a structure with graded porosity, pore size and/or composition from different biomaterials.

5.3 Porous structures fabrication using continuous-wave LDMD

5.3.1 Experimental procedure

For the continuous-wave LDMD porous structure experiment, three levels of each powder flow rate and track overlap ratio at a constant laser power of 1 kW and scanning speed of 6 mm.s\(^{-1}\) were tested, as listed in Table 5.1, giving a total of nine experimental runs. Track overlap ratio is defined as the percentage of the track width that overlaps a previous track. Mathematically it is equal to \((1 - x_{centre}/W)\), where \(x_{centre}\) is the track spacing and \(W\) is the track width. Nine multilayer porous structures of approx size 20 x 20 x 6 mm were produced with 8 to 12 orthogonal deposition layers in each structure. The variation in number of layers was due to change in the powder flow rate. Because layer height increases with powder flow rate, a lesser number of deposition layers were required with a higher powder flow rate compared to a lower powder flow rate to produce the same height.

<table>
<thead>
<tr>
<th>Parameters</th>
<th>Level 1</th>
<th>Level 2</th>
<th>Level 3</th>
</tr>
</thead>
<tbody>
<tr>
<td>Powder flow rate (g.s(^{-1}))</td>
<td>0.033</td>
<td>0.066</td>
<td>0.089</td>
</tr>
<tr>
<td>Track overlap ratio (%)</td>
<td>-10</td>
<td>0</td>
<td>10</td>
</tr>
</tbody>
</table>
All porous samples were initially detached approximately 1 mm above the substrate and then cut to form a cube of approx size $12 \times 12 \times 3$ mm. All surfaces were ground to 1200 grit and then examined under optical microscopy to find the pore sizes. Due to variation in pore sizes in different planes, all measurements were taken on $x$-$y$ plane mentioned in Figure 5.5(a). The variations of $\pm 12\%$ to the mean value were estimated. These pores were not of circular shape so the diameter of a circle of equal area was taken as a representative dimension of each pore. There was some variation in pore sizes within each sample so four random pores per sample were selected and an average value used. The micro computed tomography (MicroCT) technique was also employed to examine the internal morphologies of a selection of the samples. The Metris X-Tek 225 kV MicroCT machine was used to conduct the analysis.

### 5.3.2 Analytical model

The analytical model of multiple tracks interaction to form surface cladding layers, presented in Section 4.3 of the previous chapter, has been used here. The model is extended from surface layer deposition to multilayer deposition structures. An orthogonal deposition pattern, as shown in Figure 5.1, is used for each layer and no interaction between layers is considered. By optimizing the track overlap ratio and interaction between multiple tracks, pore formation can be expected during deposition of multilayer structures.

The deposition profile of a layer, with two multiple tracks, is given below, recalling equation (4.25) of Chapter 4.

$$h_{z(x)} = h_{z(y1)} + h_{z(y2)} + \ldots + h_{z(yn)}$$

(5.1)

Where $h_{zp}$ is deposition height for the overlapped region and $h_{zp}$ is deposition height for non-overlapped region. The deposition layer profile in the $x$-direction, $h_{z(x)}$, with $n$ tracks can be written as

$$h_{z(x)} = h_{z(y1)} + h_{z(y2)} + \ldots + h_{z(yn)}$$

(5.2)

In the same manner, the deposition layer profile in orthogonal direction i.e. $y$-direction can be written as

$$h_{z(y)} = h_{x(y1)} + h_{x(y2)} + \ldots + h_{x(yn)}$$

(5.3)
The overall deposition profile of two orthogonal layers thus can be found by adding the layers as given in equation (5.4).

\[ h_{z(x+y)} = h_{z(x)} + h_{z(y)} \]

Figure 5.1. Schematic diagram of the orthogonal deposition pattern used to fabricate the porous builds

A routine has been written in Wolfram Mathematica software to realize the model. Numerical integration, polynomial least-squares fit and logic selection techniques are used. Figure 5.2 shows the modelled multilayer porous structures. The results can be interpreted from the figure; the model predicts changes in porosity, pore size and pore shape with overlap ratio and change in deposition parameters such as laser power, traverse speed and powder mass flow rate.

5.3.3 Results and discussion

The produced structures were well-bonded with the substrate and had little visible oxidation. Figure 5.3 (a) shows good bonding at clad-substrate interface while the shiny surface in Figure 5.3 (b) depicts the presence of oxide. Figure 5.4 shows a 3D MicroCT image of the porous sample showing the regular location of the pores. Figure 5.5 shows three orthogonal images (a), (b) and (c) at y-z plane, x-y plane and x-z plane respectively. The MicroCT post-processing software can take images at discrete slice positions through the sample in each orthogonal plane. The pictures shown are taken at the mid of the pores. The pores are interconnected throughout the thickness of sample and partially melted powder particles can also be seen (Fig 5.5)
Figure 5.2. Modelled multilayer porous structures of Ti-6Al-4V at 1000W laser power, 6mm/s scanning speed and 0.089 g/s powder flow rate with; (a) 10% overlap ratio; (b) 0% overlap; (c) 10% overlap ratio.

The pore size is mainly dependent on overlap ratio while pore shape changes with the powder flow rate. It is found that pores are nearly circular at lower powder flow rate and become elongated at higher powder flow rate. Figure 5.6 shows that pore size decreases linearly with the increases in overlap ratio at all powder flow rates. There is a slight decrease in pore size with the increases in powder flow rate at all overlap ratios.
Chapter 5: Multilayer porous structures fabrication by continuous and pulsed-wave LDMD

Figure 5.3. Continuous-wave deposition sample at 1000W, 6mm s\(^{-1}\), 0.089 g s\(^{-1}\) and 10% overlap: (a) Optical micrograph showing clad-substrate interface region; (b) clad area showing presence of oxidation.

Figure 5.4. 3D Micro CT image of the porous structure at laser power 1000 W, scanning speed 6 mm s\(^{-1}\), powder flow rate 0.033 g s\(^{-1}\) and overlap ratio 0%.

The analytical model, for multilayer porous structures deposition described in Section 5.3.2, is verified by comparing with experimental results. Experimental and modelled cross-sections from two samples with different overlap ratios are compared in Figure 5.7. The model accurately captures the pore morphology. Melt pool profiles can be faintly seen in the experimental samples and the modelled melt pool shapes correspond quite well to these, especially at 0% overlap. The comparison of modelled pore sizes with the experimental values is shown in Figure 5.8. The model shows good agreement with experimental values of pore sizes in an overlap ratio operating range -10 to 10%.
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Figure 5.5. Orthogonal Micro CT images of the sample shown in figure 2: (a); Top, x-y plane (b) Right, y-z plane; (c) Front, x-z plane

Figure 5.6. Effect of overlap ratio and powder flow rate on pore size at laser power 1000W, scanning speed 6 mm.s⁻¹

The slight variations in experimental and modelled data for pore sizes can be attributed to the fact that during multiple layer deposition the substrate surface becomes rough due to the deposition of the previous layer. This can slightly change the clad formation characteristics during the subsequent clad.
Figure 5.7. Experimental and modelled porous cross-sections at 1000W, 6 mm.s\(^{-1}\) and 0.066gs\(^{-1}\). (a) overlap ratio 0% experimental, (b) overlap ratio 0% modelled, (c) overlap ratio 10% experimental, (d) overlap ratio 10% modelled

Figure 5.8. Experimental and model results comparison; effect of overlap ratio on pore size (1000W, 6 mm.s\(^{-1}\), 0.089gs\(^{-1}\))
Chapter 5: Multilayer porous structures fabrication by continuous and pulsed-wave LDMD

5.4 Porous structures fabrication using pulsed-wave LDMD

5.4.1 Experimental procedure

For the pulsed wave deposition experiment, a 3.5 kW CO$_2$ laser system, operating in pulse mode, integrated with a beam delivery system consisting of water-cooled gold coated plane mirrors (PM) and a concave mirror (CM) of 600 mm radius of curvature, a co-axial powder feed system and a 5-axis laser workstation, was used. At the fabrication point, a defocused laser spot of 1.5 mm diameter was used to create a melt pool. Stainless steel 316L was used as the substrate and deposition material. The 316L powder with size range 45-106 µm was fed into the pool coaxially using a volumetric controlled powder feeder. Argon was used as the shielding and powder carrier gas. A series of single layer test samples were first produced at a scan speed of 20 mms$^{-1}$ and powder mass feed rate of 0.133 gs$^{-1}$. The laser pulse power and pulse period were 1.6 kW and 500 ms respectively with a corresponding pulse frequency of 2 Hz.

5.4.2 Analytical model

For modelling of the pulsed-wave laser deposition, the laser is considered as a stationary source of heat during the length of a pulse. The Cline and Anthony [135] equation for a moving Gaussian source of heat is used with a scanning speed term equal to zero. Equation (5.5) is used to calculate the temperature profiles on the substrate and hence melt pool size

$$T(x,y,z) = \frac{\alpha P_l}{kr_l} f(x, y, z)$$

where the temperature distribution function $f$ is

$$f(x,y,z) = \int_0^\infty \frac{\text{Exp}(-H)}{(2\pi)^2(1+\mu^2)} d\mu,$$

$$H = \frac{y^2+x^2}{2(1+\mu^2)} + \frac{z^2}{2\mu^2}$$

Here $k$ is thermal conductivity; $r_l$ is laser beam radius; $P_l$ is laser power; $\alpha$ is laser absorptivity; $\mu$ is dimensionless time. $\kappa$ is thermal diffusivity.

* The pulsed-wave LDMD experiment using 316L stainless steel was conducted by Christ P. Paul and L.M. Kukreja at RRCAT (India).
If a laser pulse begins at $t = 0$ and continues for a pulse duration $t_p$, then initially there will be heating but no melting, and then a melt pool of radius $R$ will be generated. $R$ will then continue to increase over the length of the pulse. If $t_R$ is the time after the pulse started then $R = f(t_R)$. For the purposes of the model this is inverted and by curve fitting $t_R$ expressed as a fifth order polynomial function of $R$. For example if 1.6 kW laser power and a 316L stainless steel substrate material are used, then the polynomial becomes:

$$t_R = -0.107035 + 416.338 R - 511652 R^2 + 1.28132 \times 10^8 R^3 + 1.36412 \times 10^{11} R^4 - 4.30024 \times 10^{13} R^5$$  \hspace{1cm} (5.7)$$

The powder mass flux $q_{pf}$ within coaxial powder stream with the laser beam centred at origin can be written as:

$$q_{pf} = \frac{2q_{pr}}{\pi r_p^2} E x p \left[ -2 \frac{r^2}{r_p^2} \right]$$  \hspace{1cm} (5.8)$$

where, $q_{pr}$ and $r_p$ stand for powder mass flow rate and Gaussian powder stream radius respectively. The deposition height can be found by integrating equation (5.8) over the effective pulse duration.

$$h_z = \frac{1}{\rho} \int_{t_R}^{t_p} q_{pf} \, dt$$  \hspace{1cm} (5.9)$$

Equation (5.9) defines the deposition profile in a pulse ($0 < t < t_p$), where $\rho$ is the deposition material density.

In practice, surface tension and gravity forces become dominant, changing the shape of the profile. The shape modified in this way is assumed to have an elliptical profile, with dimensions determined by the extent of the melted area (giving the width) and the total mass of material in the ball.

The total deposition cross sectional area $A_d$, can be calculated as

$$A_d = \int_0^R h_z \, dR$$  \hspace{1cm} (5.10)$$

The maximum ball height $H_b$, assuming its shape as elliptical, is given by the following equation
and the ball deposition profile \( h_z \), can be calculated as

\[
h_b = H_b \sqrt{\frac{1-x^2}{R^2}}
\]

Equation (5.12) defines the ball deposition profile, incorporating the effects of surface tension and gravity forces.

A routine was written in Wolfram Mathematica software to realize the model and post-process and display the results. Numerical integration, polynomial least-squares fit and logic selection techniques are used. Table 5.2 presents the 316 L parameters used for the modelling calculations. Figure 5.9 shows that model ability to predict the deposition ball shape and size at a given set of processing parameters.

![Figure 5.9. Model results at 1.kW laser power and 150 ms pulse duration; (a) 2D ball image; (b) 3D ball image.](image)

<table>
<thead>
<tr>
<th>Parameters</th>
<th>Unit</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Laser absorptivity, ( \alpha )</td>
<td></td>
<td>0.25</td>
</tr>
<tr>
<td>Melting temperature, ( T_m )</td>
<td>K</td>
<td>1673</td>
</tr>
<tr>
<td>Thermal conductivity, ( k )</td>
<td>W m(^{-1}) K(^{-1})</td>
<td>15</td>
</tr>
<tr>
<td>Specific heat, ( C )</td>
<td>J Kg(^{-1}) K(^{-1})</td>
<td>500</td>
</tr>
<tr>
<td>Density, ( \rho )</td>
<td>Kg m(^{-3})</td>
<td>8000</td>
</tr>
</tbody>
</table>
5.4.3 Results and discussion

Each laser pulse was confirmed to produce a ‘ball’ like deposit on the plane substrate that was circular in horizontal section and elliptical in vertical section. Laser pulses of various durations were found to give balls of different sizes; this was primarily attributed to the gravity and surface tension effects. Table 5.3 presents the size of balls created using different pulse lengths.

Porous structures with overlap ratios of 14 %, 23 % and 32 % were manufactured by depositing multiple layers of overlapping balls. Similarly to the definition for continuous tracks in Section 5.3.1, overlap ratio is defined as the percentage of a ball overlapping a previous one divided by the ball diameter and can be applied parallel to the motion of the laser or perpendicular to it, across tracks. Mathematically it is equal to $(1 - x_{centre}/D)$ where $x_{centre}$ is the ball spacing and $D$ is the ball diameter.

Figure 5.10, shows a typical porous structure fabricated by pulsed laser deposition and Figure 5.11, shows the results of the density measurement for the determination of the porosity using Archimedes principle. A structure with porosity of more than 25% is created with an overlap index of 14%. The porosity then decreases with increasing overlap ratio.

Table 5.3. Laser pulse duration and average ball dimensions

<table>
<thead>
<tr>
<th>Pulse duration (ms)</th>
<th>Average ball diameter (mm)</th>
<th>Average ball height (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>50</td>
<td>1.25</td>
<td>0.56</td>
</tr>
<tr>
<td>150</td>
<td>1.49</td>
<td>0.72</td>
</tr>
<tr>
<td>250</td>
<td>1.75</td>
<td>0.87</td>
</tr>
<tr>
<td>350</td>
<td>1.89</td>
<td>0.94</td>
</tr>
<tr>
<td>450</td>
<td>2.12</td>
<td>1.02</td>
</tr>
</tbody>
</table>

Experimental and modelled results of deposited ball’s diameter and height are shown in Figure 5.12. There is a good agreement between measured and modelled ball diameter values at all pulse durations. The model also predicts the deposited ball heights quite well but slightly underestimates them at lower pulse durations and overestimates them at higher pulse durations. This may be due to the fact that at higher pulse durations the surface tension forces become less significant causing the
ball shape to become more elliptical and hence decreasing its height. Overall the model follows the experimental trends.

Figure 5.10. A typical porous structure of 316L stainless steel fabricated by LDMD at 1.6 kW laser power, 250 ms pulse duration and 0.133 g s\(^{-1}\) powder flow rate; (a) sample; (b) MicroCT image
Chapter 5: Multilayer porous structures fabrication by continuous and pulsed-wave LDMD

Figure 5.11. Effect of overlap index on porosity of laser deposited structures (1.6 kW laser power, 250 ms pulse duration, 0.133 gs⁻¹ powder flow rate)

![Bar chart showing porosity for different overlap indices.]

Figure 5.12. Experimental and model results comparison: (a) change in ball diameter with pulse duration; (b) change in ball height with pulse duration

![Graphs showing experimental and model results for ball diameter and height against pulse duration.]
5.5 Conclusions

The chapter showed the flexibility of laser direct metal deposition both in continuous and pulsed mode to produce different levels of porosity, pore sizes and pore morphologies. The pore size is very important for proper osseointegration in biomedical implants and it was shown that this can be controlled to some extent by optimizing the deposited track geometry and offset distance. MicroCT analysis of continuous-wave deposited samples showed that pores with excellent interconnectivity were produced, which is an important requirement for cells migration. A 3D analytical model for deposition of multilayer porous structures has been developed. The model accounts for the multiple track interaction in deposition of a single layer and then multiple layers to build the porous structures. The model results are compared with experiments and show a good agreement.

Using a pulsed beam method to create interacting balls rather than deposition tracks allowed structures with a greater level of controlled porosity to be created than with a continuous beam. It is also possible to control final part characteristics such as pore size and overall porosity by selection of deposition parameters and overlap ratio using this method. A problem with this method is that the final structures are not regular and may consequently be liable to premature failure. An analytical model for the pulsed deposition of the individual balls has been developed. The model accurately predicts the deposited ball diameters and also follows the experimental trends in the ball height.

Both methods are able to produce the porous structures with pore size of the order of 100 microns but with different sets of processing parameters for each method. A major process variable affecting pore sizes is laser spot diameter. With the same laser spot diameter, pulsed-wave laser deposition can produce smaller pore sizes compared to those of continuous-wave laser deposition.
Chapter 6

Analytical-numerical modelling of laser direct metal deposition including microstructure formation

6.1 Introduction

Multiple analytical and numerical models of the laser metal deposition process have been presented, but most rely on sequential solution of the energy and mass balance equations or discretization of the problem domain. LDMD is a complex process involving multiple interdependent processes which can be best simulated using a fully coupled mass-energy balance solution.

This chapter extends the single track deposition model described in Sections 4.3.3 and 4.3.4 of Chapter 4 and presents a coupled analytical-numerical solution of the energy-mass balance equations. Sub-models of the powder stream, quasi-stationary conduction in the substrate, and powder assimilation into the area of the substrate above the liquidus temperature are combined. An iterative feedback loop is used to ensure mass and energy balances are maintained at the melt pool. Using modelled thermal gradients and cooling rates calculations, microstructure scale modelling has been carried out. The model is verified using Ti-6Al-4V single track deposition, produced with a coaxial nozzle and a diode laser. The model predictions of local temperature history, track profile and microstructure scale show good agreement with the experimental results. The model is a useful industrial aid and alternative to finite element methods for selecting the parameters to use for laser direct metal deposition when separate geometric and microstructural outcomes are required.

Some parts of this chapter have been published by M. N. Ahsan and A. J. Pinkerton in the Journal *Modelling and Simulation in Materials Science and Engineering*. The material is reproduced by permission of IOP Publishing Ltd under copyright agreement. Full details are given in the *Publications* section.
6.2 Modelling strategy

LDMD is a complex multiphase nature of process involving different variables that are interrelated. Figure 6.1 describes the simultaneously occurring laser deposition processes showing their effects on each other.

![Figure 6.1. Schematic diagram of simultaneously occurring coupled processes in laser direct metal deposition](image)

In such a scenario, the sequential solution or discretization of the problem domain can produce inaccurate results. For example, a much-used method has been to firstly calculate melt pool size based on energy balance and secondly calculate track dimensions based on mass added to this melt pool [63]. In reality, this neglects the energy losses due to this mass addition itself (marked ‘*’ in Figure 6.1) and so can give an inaccurate melt pool size and thus final track geometry. The error remains whatever method is used to solve the separate energy and mass flow equations. A coupled analytical-numerical approach, accounting for the coupled nature of the process, avoids introducing these errors and thus provides a more accurate way to simulate the deposition processes.

Sequential analytical solutions are flexible and work fine for simple problems, but become unsuitable for very complex problems. On the other hand, deposition track geometry prediction is quite difficult using finite element models and it is not easy to integrate multi-phase complex problems. Previous literature surveys describe the various analytical and finite element model of laser direct metal deposition. Han et al [155] using numerical simulation computed the melt pool temperatures and powder heating in laser deposition, but these models did not reveal any information about the deposition track profiles. Some researchers such as Labudovic et al [54] and Qian et al [92] have proposed transient finite element models for laser deposition process...
using the element death-and-birth function in the available commercial software. Those models addressed the thermal and mechanical behaviour of the process, but all calculations were made on predefined deposition geometries.

All the above described models and the modelling literature survey presented in Section 4.2 of Chapter 4, used sequential solutions of the energy-mass balance equations, when in reality a fully coupled solution is more desirable because all the laser deposition processes are interdependent. This chapter addresses this weakness in the previously described models.

### 6.3 LDMD process modelling

#### 6.3.1 Model assumptions

The model assumes the substrate as semi-infinite to allow well known governing equations to be used to describe conduction. Although the substrate block is actually finite in size this is reasonable as the heat input is both highly localised and transient. Due to those heat source characteristics and to it moving at constant velocity, the process is considered as quasi-stationary.

Deposition material and substrate are of the same material and considered as isotropic and homogeneous. To enable more efficient solution of the analytical equations, the thermophysical properties are taken as temperature independent. This is clearly an approximation, but to reduce any possible error the common technique of using properties at an intermediate temperature, in this case halfway between the ambient and the material melting temperature (1073 K), has been used. This will to some extent cancel out the errors from both extreme conditions.

Powder flows at constant speed with a Gaussian concentration distribution across the powder stream [42]. All powder impinging upon the melt pool is absorbed by it and the rest is considered as lost.

Convection and radiation heat losses are ignored as justified by previous research [45]. However, the model accounts for the power losses due to the vaporization heat flux from the melt pool. Any power loss above the melt pool in the form of absorption of laser energy by particles in the stream (attenuation) is taken as a contribution to that required to sustain mass addition, which is the power to melt all the incoming powder to build a clad.
6.3.2 Energy and mass balance calculations

For energy balance calculations on the substrate level, the melt pool generation model is developed using Cline and Anthony’s \[135\] equation for a moving Gaussian source of heat. The detailed model is given in Section 4.3.3 of Chapter 4. The Cline equation gives the temperature distribution on the substrate.

\[
T(x, y, z) = \frac{aP_{l1} - (P_{m} + P_{evap})}{kr_{l}} f(x, y, z, v) \tag{6.1}
\]

where the temperature distribution function \( f \) is

\[
f(x, y, z, v) = \int_{0}^{\infty} \frac{\text{Exp}(-H)}{(2\pi)^{\frac{1}{2}}(1+\mu^2)} \, d\mu \tag{6.2}
\]

\[
H = \frac{(y+\frac{\sigma^2}{2}\mu^2)^2 + x^2}{z(1+\mu^2)} + \frac{z^2}{2\mu^2} \tag{6.3}
\]

The difference in Equation (6.1) and Equation (4.2) is that the power loss in the powder stream, \( P_{p} \), (Eq. 4.1) has been replaced with the power required to sustain mass addition, \( P_{m} \), (Eq. 6.6). The laser power loss in the powder stream above the melt pool level is considered as a contribution to the laser power required to melt the all incoming powder. The other difference is in the calculation of evaporation heat flux losses, \( P_{evap} \), described by Equation (4.14) which uses the temperature value as defined by the Cline equation. This is modified by using the mean melt pool temperature calculated by the enthalpy balances in the whole melt pool area as given in Equation (6.11) in Section 6.3.4.

The mass addition within the melt pool limits forms the clad and its profile, \( h_{z} \), can be calculated by using Equation (4.18) from Section 4.3.4 of Chapter 4.

\[
h_{z} = \frac{1}{\rho v} \int_{y_{\text{front}}}^{y_{\text{rear}}} q_{pf} \, dy \tag{6.4}
\]

Where \( q_{pf} \) is powder mass flux in the coaxial powder stream which is given by the following equation.

\[
q_{pf} = \frac{2q_{p} r_{p}^{2}}{\pi r_{p}^{2}} \text{Exp} \left[ -2 \frac{x^2 + y^2}{r_{p}^2} \right] \tag{6.5}
\]

Where, \( \rho \) is density of the deposition material and \( y_{\text{front}} \) & \( y_{\text{rear}} \) are the melt pool front and rear limits.
6.3.3 Power required to sustain mass additions

Mass addition tends to reduce the melt pool size at the substrate level. It is necessary to calculate the power required to melt the all incoming powder and subtract this from the initial laser power in order to avoid overestimation of the modelled melt pool size. The track profile can then be used to calculate mass of powder added to the melt pool per unit length and time and thus the power required to sustain mass addition can be written as

\[
P_m = \rho v \left[ C (T_m - T_{\text{amb}}) + L_m \right] \int_{-w/2}^{w/2} h_z \, dx
\]  

(6.6)

Where, \( W \) is the melt pool width. The power \( P_m \) is then fed back to equation (6.1) and melt pool limits are recalculated.

6.3.4 Mean melt pool temperature

In a quasi-stationary state, when mass is added in the melt pool, it can be considered as divided into two portions according to the heat pathways and temperature distributions. The first portion, which is constrained by the mass balance, is above the substrate surface, and the second portion, which is constrained by the energy balance, is below the substrate surface as shown in Figure 6.2. The mass addition means the first portion is initially around the material melting temperature while the temperature of the second portion is initially as defined by Equation (6.1). Flows and mixing within the pool serve to create rapid dispersal of the energy and the mean melt pool temperature, \( T_{\text{mean}} \), is calculated by the enthalpy balance of the whole system. For the enthalpy balance, the volumes of the two melt pool portions needs to be known.

![Figure 6.2. Schematic diagram showing upper and lower melt pool regions](image)

First considering the upper melt pool portion, the melt pool height, \( H_a \), at any \((x,y)\) position above the substrate level can be found as
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\[ H_a = \frac{1}{\rho_p \gamma} \int_{y_{\text{front}}}^{y_{\text{rear}}} q_{pf} \, dy \]  

(6.7)

and the area of that melt pool portion, \( A_a \), can be written as

\[ A_a = \int_{y_{\text{rear}}}^{y_{\text{front}}} H_a \, dy \]  

(6.8)

The melt pool volume in this portion, \( V_a \), when \(-W/2 \leq x \leq W/2\) is given as

\[ V_a = \int_{-W/2}^{W/2} A_a \, dx \]  

(6.9)

The volume of the melt pool portion below the substrate level, \( V_b \), can be immediately written by considering this portion as semi-ellipsoid.

\[ V_b = \frac{1}{2} \left[ \frac{4}{3} \pi \left( \frac{W}{2} \right)^2 \left( \frac{L_{\text{front}}}{2} + \frac{L_{\text{rear}}}{2} \right) \right] \]  

(6.10)

By maintaining the enthalpy balances in the whole melt pool, the mean melt temperature, \( T_{\text{mean}} \), can be found

\[ T_{\text{mean}} = \frac{V_a T_m + V_b T}{V_a + V_b} \]  

(6.11)

Where, \( T \) is the temperature calculated by Equation (6.1). The Equation (6.11) gives mean melt pool temperature which is used to calculate evaporation heat losses.

### 6.3.5 Cooling rates during LDMD solidification

The thermal gradient, cooling rate and solid-liquid interface velocity are related by the following equation [1]; assuming the melt pool is moving in the y-direction and the solid-liquid interface is moving to the direction of motion.

\[ \frac{dT}{dt} = \frac{dT}{dy} \times \frac{dy}{dt} \]  

(6.12)

Where \( dT/dt \) is cooling rate, \( dT/dy \) is thermal gradient and \( dy/dt \) is interface velocity or solidification rate. The thermal gradients can be found by differentiating the temperature field with respect to distance. The thermal gradient and hence the cooling rates are calculated at the solidification front at the rear most part of the pool i.e. \( y = -L_r \), where gradient is calculated by differentiating with respect to the negative y-direction.

### 6.3.6 Model solution

The overall operation of the coupled deposition model is shown in flow chart form in Figure 6.3. First, heat flow calculations are performed without accounting for laser
power losses to find approximate temperature distribution and melt pool size. These are used to calculate heat flow losses, deposition track profile and the power required to sustain mass addition.

The model works on a feedback loop system. After each iteration the output of the model is fed back as input for the next iteration. For the first iteration, the model over-estimates the losses and under-estimates melt pool size and conversely on the second iteration the melt pool size is over-estimated and the losses are under-estimated. This process goes on until a stabilized melt pool size is obtained. The
model has been realised using Wolfram Mathematica software. Numerical integration, polynomial least-squares fit, logic selection and structured programming techniques are used. Figure 6.4 shows typical melt pool width variations as the model proceeds to reach a solution that balances all equations. Typical processing time is of the order of 100 seconds on an Intel® Core 2 Duo™ desktop computer. Table 6.1 shows the Ti-6Al-4V properties used for modelling calculations.

![Figure 6.4. Melt pool variations with number of iterations as model proceeds to get a stabilized melt pool](image)

Table 6.1. Ti-6Al-4V properties used in the modelling calculations [55].

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Unit</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Laser absorptivity, $\alpha$</td>
<td>------</td>
<td>0.55</td>
</tr>
<tr>
<td>Melting temperature, $T_m$</td>
<td>K</td>
<td>1933</td>
</tr>
<tr>
<td>Thermal conductivity, $k$</td>
<td>W m$^{-1}$ K$^{-1}$</td>
<td>17</td>
</tr>
<tr>
<td>Specific heat, $C$</td>
<td>J Kg$^{-1}$ K$^{-1}$</td>
<td>800</td>
</tr>
<tr>
<td>Density, $\rho$</td>
<td>Kg m$^{-3}$</td>
<td>4300</td>
</tr>
<tr>
<td>Latent heat of melting, $L_m$</td>
<td>KJ Kg$^{-1}$</td>
<td>370</td>
</tr>
<tr>
<td>Latent heat of vaporization, $h_v$</td>
<td>KJ Kg$^{-1}$</td>
<td>9830</td>
</tr>
</tbody>
</table>
6.4 Experimental verification

A number of experiments were performed for verification of the model. Figure 6.5 shows the experimental setup used for deposition and local temperature measurement. The detail of LDMD experiment is given in Section 3.3 of Chapter 3.

![Experimental setup diagram](image)

**Table 6.2. Processing parameters for the laser direct metal deposition experiment**

<table>
<thead>
<tr>
<th>Parameters</th>
<th>Level 1</th>
<th>Level 2</th>
<th>Level 3</th>
</tr>
</thead>
<tbody>
<tr>
<td>Laser power (W)</td>
<td>800</td>
<td>1000</td>
<td>1200</td>
</tr>
<tr>
<td>Powder flow rate (gs(^{-1}))</td>
<td>0.033</td>
<td>0.066</td>
<td>0.089</td>
</tr>
</tbody>
</table>

An Impac Infrared Thermometry IGAR 12-LO two colours (wavelength) pyrometer with temperature measuring range 772 K to 2272 K was used to monitor the substrate surface temperature along the entire length of each track. The measurements were taken 1.60 mm to the side of the laser spot and outside the melt pool, thus avoiding shielding by the powder stream or any phase change effects. A measurement was taken every 2 ms during the deposition, and the mean value of all readings taken as the representative value for that track. Although an ideal temperature monitoring method would be to directly measure the 3D temperature distributions in the melt-pool and substrate during processing, multi wavelength
Chapter 6: Analytical-numerical modelling of LDMD including microstructure formation

Pyrometry provides a reliable method for emissivity-independent non-contact temperature measurement and a representative point away from interference provides a reasonable basis for verification of the quantitative temperature predictions of the model.

Samples for microscopy were prepared using standard metallography techniques, details can be found in Section 3.4.1 of Chapter 3. After etching, all samples were examined using a Polyvar optical microscope and Hitachi scanning electron microscope. Images were captured and after calibration the respective deposition dimension were recorded.

6.5 Results

6.5.1 Local temperature history

The local temperature history on the substrate surface at different mass flow rates are shown in Figure 6.6. The deposition process starts at the 2nd sec and finishes at the 10th sec, thus the Figure only shows temperature history during that time period. It can be seen that the temperature increases slightly as time elapses due to heat accumulation in the substrate.

Experimental and modelled results of local temperature history on the substrate surface are compared in Figure 6.7. The results show the temperature variations with mass flow rates at three laser power levels. Some trials temperature readings were noted at different processing parameters before the experiment and the error for experimental values was estimated as about ±8% of the measured value. The trend of a decrease in temperature with an increase in mass flow rate is evident at all power levels. The model follows the experimental trends.
Figure 6.6. Local temperature history on the substrate surface parallel to the deposition tracks: (a) 800 W; (b) 1000 W
Figure 6.7. Experimental and model results comparison of local temperature history with mass flow rate at 1.60 mm side of the deposition track at laser power: (a) 800 W; (b) 1000 W; (c) 1200 W
6.5.2 Deposition track geometry

Track deposition geometry is a major characteristic of the LDMD rapid manufacturing technique. It is one of the important factors that determine the surface morphology of the manufactured part. Figure 6.8 shows the laser deposition track geometries at three laser power and three mass flow rates at a constant scanning speed. Track profile changes with laser power and mass flow rate. Experimental and model comparison of track width is presented in Figure 6.9 showing excellent agreement at all mass flow and laser power levels. There is a good agreement between experimental and model results of the track height as shown in Figure 6.10.

Figure 6.8. Laser deposition track profile at 800 W laser power, mass flow rates: (a) 0.033 gs\(^{-1}\); (b) 0.066 gs\(^{-1}\); (c) 0.089 gs\(^{-1}\) and 1000 W laser power, mass flow rate; (d) 0.033 gs\(^{-1}\); (e) 0.066 gs\(^{-1}\); (f) 0.089 gs\(^{-1}\) and 1200 W laser power, mass flow rate; (g) 0.033 gs\(^{-1}\); (h) 0.066 gs\(^{-1}\); (i) 0.089 gs\(^{-1}\)
Figure 6.9. Experimental and modelled track widths with mass flow rate at laser power: (a) 800 W; (b) 1000 W; (c) 1200 W
Figure 6.10. Experimental and model results comparison of track height with mass flow rate at laser power: (a) 800 W; (b) 1000 W; (c) 1200 W
6.5.3 Cooling rate and solidification microstructure

Laser deposition of Ti-6Al-4V gives a columnar prior-β grains microstructure which is dependent on the laser processing parameters that dictate the cooling rates on the solidification front. Figure 6.11 shows the columnar prior-β grains at three different laser powers. Widmanstätten α lathes can also be seen as well as the boundaries of β grains. Figure 6.12 shows the scanning electron microscopy (SEM) image of a typical deposition sample. Figure 6.12(a) shows the basketweave like Widmanstätten α lathes; the boundaries of prior-β grains are faintly visible. Figure 6.12(b) shows a higher magnification image of 6.12(a) showing the α lathe structure. The prior-β grain size has been measured experimentally using image processing software. The average prior-β columnar grain size increases from 200 μm to 500 μm as laser power increases from 800 W to 1200W in the single layer deposition.

The experimental grain sizes are compared with the model using the microstructure scale relation [156] below

$$\varphi = K_1|\gamma|^{-K_2}$$

Where $\varphi$ is microstructure scale, $\gamma$ is cooling rate and $K_1$ and $K_2$ are material dependent constant. The cooling rate is extracted from the model for the point at the rear edge of the melt pool as described in Section 6.3.5.

From Ti-6Al-4V experimental results; the values of $K_1$ and $K_2$ are found to be $1.18 \times 10^{20}$ & $4.90$ respectively. The size of columnar prior-β grains tend to increase with decrease in the cooling rates as shown in Figure 6.13.
Figure 6.11. Optical micrograph showing columnar prior-β grain microstructure of single layer Ti-6Al-4V deposition structures at 5mms$^{-1}$ scanning speed, 0.89 gs$^{-1}$ mass flow rate and laser power: (a) 800 W; (b) 1000 W; (c) 1200 W
Figure 6.12. SEM image of a typical single layer Ti-6Al-4V deposition structures Widmanstätten α lathes at 5mm$^{-1}$ scanning speed, 0.89 gsm$^{-1}$ mass flow rate and 1000W laser power: (a) 1200 magnification; (b) 5000 magnification

Figure 6.13. Ti-6Al-4V prior-β columnar grains size as a function of cooling rates
6.6 Discussion and applications

6.6.1 Temperature distribution and pool size

It is evident from Figures 6.6 and 6.7 that as the mass flow rate increases, the temperature decreases at the substrate surface. This is attributed the fact that at higher mass flow rates beam attenuation is increased resulting in less energy being available at the substrate. The model follows the experimental trend, but the experimental change is steeper than that of the model. This may be due to the powder cloud intensity at higher mass flow hindering the pyrometer measuring area. Figure 6.14 shows the modelled mean melt pool temperature of Ti-6Al-4V with respect to laser power and mass flow rate at scanning speed 5mms⁻¹. Mean melt pool temperature increases with increase in laser power and/or decrease in mass flow rate at a constant scanning speed. The change of mean melt pool temperature is not linear because it is obtained by the enthalpy balance of the upper and lower melt pool portions as described by Equation (6.10). The substrate temperatures distribution as a function of laser power and mass flow rate at a constant scanning speed is shown in Figure 6.15. The temperatures on the substrate show an almost linear increase with laser power and a slight linear decrease with mass flow rate, which is in agreement with previous research [45]. The model has been applied to obtain melt pool depths as a function of laser power and mass flow rate at a constant scanning speed. The melt pool depths results also show linear increases with laser power and a slight decrease with mass flow rate as shown in Figure 6.16.

6.6.2 Deposition track geometry

The track width is mainly dependent upon the laser energy available at the substrate as shown in Figure 6.9. The trend is for width to decrease slightly as mass flow increases at 800 W and 1000 W, but sharper width decreases can be seen at 1200 W. The increased power required to sustain mass addition at higher mass flow rates, as revealed in Equation (6.6), is responsible for the trends seen in Figures 6.9(a), (b) and (c). The increased power required results in less energy being available at substrate level and hence tends to decrease the melt pool width. The sharp width decrease at 1200 W can attributed to the fact that at higher power evaporation heat flux losses become dominant making less energy available at substrate.
Figure 6.14. Modelled mean melt pool temperatures as a function of laser power and mass flow rate at scanning speed 5mms$^{-1}$ for Ti-6Al-4V

The track height is mainly dependent on mass flow rate at all laser power levels as shown in Figure 6.10. There is a slight increase in track height with laser power at all mass flow rates because with an increase of incident energy the melt pool size increases so more of the incoming mass becomes the part of deposition.
Figure 6.16. Modelled melt pool depths as a function of laser power and mass flow rate at scanning speed $5\text{mms}^{-1}$ for Ti-6Al-4V

Figure 6.17 shows modelled track widths with respect to laser power and mass flow rate. There is an increase of track width with laser power and slight decrease with mass flow rate. Model simulations of track height are shown in Figure 6.18. This confirms the track widths and heights trends reported by Pinkerton and Li [63].
6.6.3 Cooling rates and solidification microstructure

The microstructure and hence the mechanical properties of the laser deposited structure is very much dependent on the solidification behaviour and cooling rates [83]. Figure 6.19 shows modelled cooling rates of Ti-6Al-4V at the rear of melt pool, for experiment conditions. The cooling rates change depending upon substrate temperature and the laser input energy. Figure 6.11 shows how microstructure grain size changes with laser energy or indirectly with cooling rates because; the
calculated cooling rates for these microstructures for 800 W, 1000W, 1200 W were 4981 Ks\(^{-1}\), 4858 Ks\(^{-1}\) and 4448 Ks\(^{-1}\). The experimental results of grain size show good agreement with modelled grain size at higher cooling rate, but the model slightly under estimates the grain sizes at lower cooling rates (Fig. 6.13). This can be attributed to the fact that these cooling rates would have been produced when the melt pool was bigger [157] and thus had a larger Marangoni Number and more significant Marangoni flow. The model is based purely on heat conduction so the effects of these Marangoni flows are not included and the results indicate that the flow and mixing in the melt pool tend to reduce the solidification rate and hence lead to a slight increase in the grain size. Overall, the model follows the experimental trends.

This model found the cooling rates for the Ti-6Al-4V to be in the range of \(10^3\) Ks\(^{-1}\) which is consistent with the previous modelled [92, 158] cooling rates in the range of \(10^3\) to \(10^4\) Ks\(^{-1}\). It is found that mass flow has little effect on cooling rates and these are mainly dictated by the laser input energy. Figure 6.19 shows a decrease in cooling rates with laser power and a slight increase with mass flow rate. The cooling rates decreases as melt pool length increases, which is in agreement with the work reported by Hofmeister et al [157].

### 6.7 Conclusions

In this chapter, a 3D analytical-numerical model of LDMD based on solving energy and mass balance equations within a quasi-stationary system has been presented. It simulates the multiple interdependent laser deposition processes by taking advantage of simple analytical problem construction and an iterated numerical solution. This analytical-numerical coupled approach makes the model not only accurate but computationally efficient as compared to finite element models. The model has been applied to melt pool generation, track formation and microstructure formation. Comparison with the experimental results of local temperature history, track profile geometry and microstructure grain size shows good agreement. Modelled cooling rates are found to be of the order of \(10^3\) K for Ti-6Al-4V; they decrease with laser power but are not substantially affected by mass flow rate. The average prior-\(\beta\) columnar grain size increases from 200 \(\mu m\) to 500 \(\mu m\) as laser power increases from 800 W to 1200W in the single layer deposition. There is no meshing or similar and
typical model processing time is of the order of 100 seconds, which makes the model a fast and efficient aid for setting accurate deposition processing parameters for a required track profile and microstructural scale.
Chapter 7

A comparative study of laser direct metal deposition characteristics using gas and plasma atomized Ti-6Al-4V powders

7.1 Introduction

Chapters 4, 5 and 6 of this thesis cover modelling of LDMD for applications from surface layer cladding to porous volume fabrication and also encapsulate the solidification microstructure formation. This chapter emphasises on the experimental and characterizations of LDMD where the causes of intralayer porosity generation are specifically figured out which is still a concern in all applications of LDMD. This chapter also applies the previous developed models where necessary to explain the physical phenomena.

The availability of alternative powder types and demand for high quality LDMD parts and repairs necessitates a comparative study using different types of Ti-6Al-4V powder to see if either gives advantages in terms of geometric accuracy, surface finish, microstructural and mechanical properties, or prevalence of defects such as intralayer porosity. In this chapter, a systematic comparative investigation of LDMD using GA (gas-atomised) and PREP (plasma rotating electrode process) Ti-6Al-4V powders has been carried out. Ti-6Al-4V powders prepared using the gas-atomization and the plasma rotating electrode processes are first analyzed using laser diffraction, scanning electron microscopy and microcomputed tomography.

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LDMD characteristics in terms of layer geometry, surface finish, microstructure and micro hardness, and internal porosity are compared under similar process conditions. The change of microstructure morphology, from long columnar to small equiaxed grains, with processing parameters and the factors that affect it are discussed.

Defects of interlayer and intralayer porosity are sometime observed in laser deposited structures and repaired components. Interlayer porosity can be controlled to some extent by adjusting processing parameters, but there is still disagreement as to the source of intralayer porosity and whether process conditions, process parameters or initial powder materials are the dominant causal factor. In this chapter, the causes of intralayer porosity generation during the LDMD fabrication are investigated and discussed in details. An advanced volumetric porosity assessment method using microcomputed tomography (MicroCT) is used to quantify internal porosity in the powders and final LDMD thin-wall samples because it is capable of performing analysis on a designated volume and is hence a more reliable method than microscopy. The chapter shows a new way to quantitatively assess the differences between the two types of powder that are dominant in the production of high quality titanium alloy components by LDMD.

7.2 Previous work
GA powders are mostly used for the LDMD process; gas-atomization is very common process for metals powder production, but has the inherent weakness of entrapped gases during rapid solidification [159]. The use of the PREP powder is increasing; the PREP process is an advanced powder production technique, able to produce particles with near perfect spherical morphology [16]. The GA and PREP powder types dominate; other powder production methods exist (e.g. the Hydride-DeHydride process) but these currently do not have significant market share.

Comparative studies of LDMD using GA and water-atomized (WA) metallic powder of 316L stainless steel and H13 tool steel have been conducted previously by Pinkerton and Li [160, 161] to characterize microstructure, surface finish and deposition rate. There have been previous investigations on Inconel 718 (Ni alloy) GA and PREP powder deposition [17, 18]. The studies showed that PREP powder deposition structures exhibited minimal porosity [17, 18] and superior mechanical properties to those of the GA powder structures [18], but this research did not
provide any comparative information on the microstructures of the deposited samples. Wang et al [162] studied the microstructure and mechanical properties of samples produced by hot isostatic press (HIP) forming of Ti-6Al-4V using GA and PREP powders at one set of parameters. The microstructures were similar, consisting of platelet \( \alpha \) and fine transformed \( \beta \) structure, with fine equiaxed \( \alpha \) distributed uniformly, but the strength of the GA samples were higher than that of the PREP samples. In summary, there have been few published studies in this area, particularly regarding laser deposited materials using GA and PREP Ti-6Al-4V powders.

7.3 Experimental procedures

7.3.1 Powder characterization

The two types of Ti-6Al-4V powders, GA and PREP, used in the experiment both had a particle size range of 45-105 \( \mu \)m. Particle size analysis of the powders samples was performed on a Malvern Mastersizer laser diffractometer and the surface morphologies of the powder samples were examined by scanning electron microscopy to reveal any irregularities and/or pores.

Microcomputed tomography of the powder samples was carried out using an Xradia microXCT-400 system, which operates in point projection geometry. Scans of the gas-atomized and PREP powder samples were carried out using the 10x objective with 1001 images being taken over 184 degrees for source voltages of 80 and 65 kV respectively. The final reconstructed voxel (3D pixel) dimensions were 2.3450 \( \mu \)m and 2.3120 \( \mu \)m for GA and PREP samples respectively.

7.3.2 Laser deposition using GA and PREP powder

For LDMD experiment, two levels of laser power and five levels of mass flow rate for each powder type were tested at a constant scanning speed of 5 mms\(^{-1}\), which gave a total of twenty experimental runs, as listed in Table 7.1. For each laser power and powder type, thin-wall samples were produced with 15, 12, 10, 8 and 8 deposition layers at 0.033, 0.066, 0.089, 0.112, and 0.135 gs\(^{-1}\) mass flow rate respectively. The experimental detail can be found in Section 3.3 of Chapter 3.
Table 7.1. Processing parameters for laser direct metal deposition

<table>
<thead>
<tr>
<th>Parameters</th>
<th>Level 1</th>
<th>Level 2</th>
<th>Level 3</th>
<th>Level 4</th>
<th>Level 5</th>
</tr>
</thead>
<tbody>
<tr>
<td>Laser Power (W)</td>
<td>800</td>
<td>1000</td>
<td>---</td>
<td>---</td>
<td>---</td>
</tr>
<tr>
<td>Mass flow rate (gs⁻¹)</td>
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<td>0.066</td>
<td>0.089</td>
<td>0.112</td>
<td>0.135</td>
</tr>
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</table>

7.3.3 LDMD characterization

All deposition samples were characterized using optical microscopy, scanning electron microscopy and microcomputed tomography, white light interferometry and micro hardness testing. Samples for microscopy were prepared using standard metallography techniques, details can be found in Section 3.4.1 of Chapter 3. Optical microscopy was used to characterize the microstructure of the deposition samples. Scanning electron microscopy was performed on selected samples. Average microstructure grain sizes were measured using an intercept method. Because of the relatively large grains, for each sample an average value of seven consecutive grains along a straight line, with random sampling, was taken as a representative value.

X-ray diffraction was employed at ‘top’ and ‘bottom’ positions of a thin-wall sample using Pananalytical x-ray diffractometer. The samples for x-ray diffraction were prepared by cutting the thin-wall parallel to the deposition direction, 2 mm and 6 mm above the substrate level for the bottom and top sample respectively, as shown in Figure 7.22. The samples were ground and then polished using oxide polishing suspension for four hours to remove any deformation introduced during the cutting operation.

Surface roughness on the top of the thin-wall samples was measured using a white light interferometer (Wyko NT1100 Optical Profiling System). In the focused position, 1000 µm × 500 µm of the clad top surface area was scanned by the equipment to find surface variations. Micro hardness measurement was also employed on deposition samples using a Buehler Hardness Testing Machine. Micro hardness was measured from top to bottom of thin-wall sample at discrete locations.
2 mm apart from each other along a straight line. An average value of all the discrete readings was taken as representative value.

For MicroCT sample preparation, the thin-wall samples were cut in two traverse planes near the middle of the thin-wall in order to make their widths and breadths approximately equal as shown in Figure 7.1(a). Microcomputed tomography of the deposition samples was carried out using a Metris XTH 225 CT machine as shown schematically in Figure 7.1(b). Scans of the each sample were performed with 2880 projections over 360 degrees of rotation. The final reconstructed resolution was achieved in the range of 5.5 µm to 10.8 µm depending upon the field of view in each sample. A 0.1 mm thick copper filter was used to adjust the photon intensity ratio on sample and background areas. The source voltage, current, x-ray (no. of photons) and final reconstructed resolution in terms of voxel sizes for all the samples are given in Table 7.2. The difference in source voltages, for the samples mentioned in Table 7.2 and the powder samples mentioned in Section 7.3.1, is only an adjustment to the MicroCT system to get the best image resolution and is not expected to have an effect on the results.
Chapter 7: A comparative study of LDMD characteristics using GA and PREP powders

Figure 7.1. Microcomputed tomography setup: (a) schematic diagram of MicroCT sample preparation; (b) schematic diagram of MicroCT measurement
Table 7.2. X-ray tomography source variables and sample reconstructed resolutions

<table>
<thead>
<tr>
<th>Sample ID</th>
<th>Powder type</th>
<th>Laser power (W)</th>
<th>Mass flow rate (gs⁻¹)</th>
<th>Source voltage (kV)</th>
<th>Source current (µA)</th>
<th>No of photon on sample</th>
<th>Reconstructed voxel size (µm)</th>
</tr>
</thead>
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<td>Gx1</td>
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<td>0.033</td>
<td>80</td>
<td>116</td>
<td>14000</td>
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<td>GA</td>
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<td>0.066</td>
<td>80</td>
<td>116</td>
<td>13225</td>
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<td>Gx3</td>
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<td>78</td>
<td>116</td>
<td>12750</td>
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<td>0.112</td>
<td>79</td>
<td>112</td>
<td>13000</td>
<td>8.6</td>
</tr>
<tr>
<td>Gx5</td>
<td>GA</td>
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<td>0.135</td>
<td>79</td>
<td>112</td>
<td>12750</td>
<td>9.8</td>
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<td>PREP</td>
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<td>0.033</td>
<td>79</td>
<td>112</td>
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<td>79</td>
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<td>GA</td>
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<td>Py1</td>
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<td>0.066</td>
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<td>12000</td>
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<tr>
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<td>PREP</td>
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</tr>
<tr>
<td>Py5</td>
<td>PREP</td>
<td>1000</td>
<td>0.135</td>
<td>79</td>
<td>112</td>
<td>10500</td>
<td>10.6</td>
</tr>
</tbody>
</table>

7.4 Results

7.4.1 Powder characteristics

The particle size distribution of the powder samples is given as the lognormal graphs in Figure 7.2.
Figure 7.2. Particle size distribution of the powder samples: (a) GA powder; (b) PREP powder
There is an additional concentration of particles around 300 µm in the GA sample which may be attributed to inadequate dispersal by the laser diffraction equipment. The mean particle diameter was found to be 94 µm for GA powder and 72 µm for PREP powder.

Figure 7.3. SEM micrograph showing surface morphologies of: (a) GA powder; (b) PREP powder

Surface morphologies of the two powder samples are shown in Figure 7.3. GA powder particles are generally spherical with rough surfaces and satellite particles as shown in Figure 7.3(a). Some pores are also visible on the surface due to the atomization process. Figure 7.3(b) shows the morphology of PREP powder which
contains no visible pores on the surface analyzed. Particles show a highly spherical shape and fine surface.

Figure 7.4 shows the results of the microcomputed tomography analysis of the powder samples. Internal morphologies in terms of two slices through the GA and PREP powder samples are shown. In Figure 7.4(a), more pores are evident in the GA samples as compared to the SEM micrograph in figure 7.3(a). The pores within the particles are spherical and assumed to contain atomizing gases [104]. A typical tomography slice of the PREP powder sample shown in Figure 7.4(b) reveals that there are some internal pores irrespective of the fact that the SEM micrograph in Figure 7.3(b) shows a completely smooth surface. A detailed volumetric analysis of the powder samples is needed to find the exact proportion of porosity.

Figure 7.4. MicroCT images at a particular slice of: (a) GA powder; (b) PREP powder
Comprehensive image processing analysis was carried out using image segmentation tools on commercial tomography data post-processing software. Figure 7.5 shows porosity volume inside the powder particles in the whole powder volume under analysis. Each yellow colour surface is a representation of the pore volume inside a powder particle. It is found that the PREP powder sample has internal porosity but much less than in the gas-atomized powder sample.

Table 7.3 shows the quantitative porosity results of these samples. The porosity in the GA powder sample is found to be three times higher than in the PREP powder sample.

Table 7.3. Quantitative porosity results of the powder volume used in analysis

<table>
<thead>
<tr>
<th>Sample</th>
<th>Powder Volume (mm$^3$)</th>
<th>Pores Volume (mm$^3$)</th>
<th>% Porosity</th>
</tr>
</thead>
<tbody>
<tr>
<td>GA</td>
<td>3.877</td>
<td>0.002117</td>
<td>0.055</td>
</tr>
<tr>
<td>PREP</td>
<td>2.329</td>
<td>0.000399</td>
<td>0.017</td>
</tr>
</tbody>
</table>
Figure 7.5. MicroCT images of two orthogonal slices and the porosity (indicated by yellow spheres) in the whole volume ($x \times y \times z$) under analysis: (a) GA powder; (b) PREP powder.
7.4.2 LDMD characteristics

7.2.2.1 Intralayer porosity

In the deposition samples, interlayer (lack of fusion) porosity was mostly found at the deposition substrate interface and only on higher mass flow rates. Figure 7.6(a) shows a typical SEM image of interlayer porosity at higher mass flow rate; unmelted powder particles can also be seen. Figure 7.6(b) shows a typical SEM micrograph of PREP sample showing intralayer porosity.

Figure 7.6. SEM micrograph showing interlayer and intralayer porosity: (a) GA deposition sample at 1000W and 0.066 gs\(^{-1}\); (b) PREP deposition sample at 800W and 0.135 gs\(^{-1}\)
In this analysis only values for intralayer (gas) porosity are presented because it is assumed that is affected by initial powder porosity [104]. Figure 7.7 shows the intralayer porosity results obtained using optical microscopy. Figure 7.8 shows the 3D images of two full thin-wall deposition samples reconstructed using surface generation techniques applied on the tomography data. The clad surface is not generated deliberately and the two orthogonal slices of thin-wall deposition are shown in order to clearly visualize the porosity and deposition area. The samples were deposited using GA and PREP powders when all the other deposition parameters are the same. The red and purple spheres show the location and size of intralayer pores present in the samples. As seen that the GA deposition sample has more intralayer pores compared to the PREP deposition samples with the same processing parameters. Figure 7.8(c) to (e) shows the enlarged images of the first two deposition layers in each sample with the surface generation of intralayer porosity only in the deposition area under analysis. The results show more porosity in GA deposition samples than in PREP deposition samples. Porosity is mostly observed in the first deposition layer.

A quantitative porosity analysis is performed on tomography data of all the deposition samples using image segmentation tools and the results are presented as a bar chart in Figure 7.9. Overall, results show that the porosity percentage is higher at 0.033 gs\(^{-1}\) mass flow and porosity tend to decrease with mass flow rate and reach its minimum value at 0.066 gs\(^{-1}\). Further increase of mass flow increases the intralayer porosity. The same trend is observed for both laser power levels and for both types of powders. PREP deposition samples show lower intralayer porosity at all mass flow and laser power levels.
Figure 7.7. Intralayer porosity measured using optical microscopy using GA and PREP deposition samples: (a) laser power 800W; (b) laser power 1000W

The quantitative analysis is further extended to find pore size distribution in the deposition samples. Figure 7.10 shows the pore size distribution of four samples deposited at 0.033 gs$^{-1}$ mass flow rate with two powder types and two laser power levels. Results describe that at 800 W laser power the GA deposition sample maximum pore diameter frequency is in the range 40-60 µm while the PREP deposition sample has maximum pore diameter frequency in the range 60-80 µm.
Figure 7.8. 3D surface generation of intralayer porosity (red and purple spheres) of microcomputed tomography data of full thin-wall deposition samples: (a) GA, 800 W, 0.089 $\text{gs}^{-1}$; (b) PREP, 800 W, 0.089 $\text{gs}^{-1}$ and thin-wall with only two first layers: (c) GA, 800 W, 0.066 $\text{gs}^{-1}$; (d) PREP, 800 W, 0.066 $\text{gs}^{-1}$; (e) GA, 1000 W, 0.112 $\text{gs}^{-1}$; (f) PREP, 1000 W, 0.112 $\text{gs}^{-1}$

The GA deposition sample pore diameter frequency is higher (twenty two) as compared to PREP deposition samples (ten). At 1000 W laser power the same trend of higher frequency of smaller pore diameters is observed for GA and PREP deposition samples. In general PREP deposition samples have more large diameter pores and fewer small diameter pores than GA samples. Figure 7.11 shows the pore size distribution of GA and PREP deposition samples at 0.135 $\text{gs}^{-1}$ mass flow rate. A similar trend as at 0.033 $\text{gs}^{-1}$ mass flow rate is observed: higher frequency for smaller pore diameters and lower frequency for larger pore diameters.
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Figure 7.9. Intralayer porosity measured using micro computed tomography data of GA and PREP deposition samples: (a) laser power 800W; (b) laser power 1000W

There is less porosity with diameter less than 40 µm in all cases and a marked difference for GA deposition at 1000 W. Figure 7.12 shows the maximum pore diameter in the deposition samples as a function of mass flow rate, laser power and powder type. A general trend of a decrease in maximum pore diameters with mass flow rate up to 0.066 gs$^{-1}$ can be seen and further increase in mass flow increases the maximum pore diameters.
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Figure 7.10. Pore diameter frequency distribution in deposition samples at 0.033 gs\(^{-1}\) mass flow rate and: (a) GA, 800W; (b) PREP, 800W; (c) GA, 1000W; (d) PREP, 1000W

The only exception is the PREP deposition sample at 1000W laser power and 0.112 gs\(^{-1}\) mass flow rate. PREP deposition sample shows the highest pore diameter 230 µm at 0.089 gs\(^{-1}\) mass flow rate and 800 W laser power. The lowest value of maximum pore diameter, 98 µm, is observed with the PREP deposition sample at 0.112 gs\(^{-1}\) mass flow rate and 1000 W laser power. Figure 7.13 shows the variations in mean pore diameter in the deposition samples as a function of mass flow rate, laser power and powder type. For GA samples, the behaviour of mean pore diameters is the same as that of maximum pore diameters. PREP deposition samples generally show a decrease in mean pore diameter with mass flow rate for both laser power levels.
Figure 7.11. Pore diameter frequency distribution in deposition samples at 0.135 gs\(^{-1}\) mass flow rate and: (a) GA, 800W; (b) PREP, 800W; (c) GA, 1000W; (d) PREP, 1000W

Figure 7.12. Change of maximum pore diameter as a function of mass flow rate in the deposition samples
7.4.2.2  Geometric dimensions and surface roughness

A comparison of mean layer thickness of the GA and PREP deposition thin-wall samples at two laser power levels is presented in Figure 7.14. As expected, the layer thickness increases with mass flow for both types of powders. PREP powder deposition shows a slightly higher layer thickness at each mass flow rate and laser power level than that of GA powder deposition. A maximum difference of 0.22 mm in average layer thickness is observed between GA and PREP powder depositions. A comparison of mean layer widths shows a slight decrease with mass flow rate at both laser power levels, as shown in Figure 7.15. PREP deposition samples are generally found to have a lower layer width than GA deposition samples. Figure 7.16 shows the surface roughness ($R_a$) trends for GA and PREP powder deposition as a function of mass flow rate and laser power levels. Surface roughness tends to increase with mass flow rate; this increase is quicker at 800 W than 1000 W. There is also a slight decrease in roughness when laser power increases from 800 W to 1000 W. The surface roughness of the PREP powder deposition increases from 4.4 µm to 18.1 µm as mass flow rate changes from 0.033 gs$^{-1}$ to 0.135 gs$^{-1}$ at 800 W laser power. PREP powder deposition samples show lower surface roughness than the GA deposition samples at all mass flow rates and laser power levels.
Figure 7.14. Comparison of mean layer thickness of GA and PREP deposition samples as a function of mass flow rate and laser power: (a) laser power 800 W; (b) laser power 1000 W
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Figure 7.15. Comparison of mean layer width of GA and PREP deposition samples as a function of mass flow rate and laser power: (a) laser power 800 W; (b) laser power 1000 W
7.4.2.3 Microstructure characteristics and microhardness

Laser direct metal deposits of Ti-6Al-4V exhibit a microstructure of large columnar prior beta grains, as shown in Figures 7.17 and 7.18, which compares fifteen-layer, thin walls deposited from GA (Figs. 7.17(a) & 7.18(a)) and PREP (Figs. 7.17(b) & 7.18(b)) powder. Approximately equally spaced layer bands are visible through the whole thin-wall except the last three deposition layers. The prior beta grains boundaries are continuous across the layer bands. A comparison of the
microstructure of the prior beta grains in GA and PREP deposition samples at two mass flow rates and two laser power levels is shown in Figures 7.19 and 7.20.

Figure 7.17. Ti-6Al-4V laser deposited thin-wall structures showing columnar prior beta grains and layer band morphology at 0.033 gs\(^{-1}\) mass flow rate: (a) GA, power 800 W; (b) PREP, power 800 W
Figure 7.18. Ti-6Al-4V laser deposited thin-wall structures showing columnar prior beta grains and layer band morphology at 0.033 gs$^{-1}$ mass flow rate: (a) GA, power 1000 W; (b) PREP, power 1000 W
Figure 7.19. Comparison of prior beta grains morphology of GA and PREP deposition samples at 800 W
Figure 7.20. Comparison of prior beta grains morphology of GA and PREP deposition samples at 1000 W
Figure 7.21. Comparison of average prior beta grain size of GA and PREP deposition samples as a function of mass flow rate: (a) laser power 800 W; (b) laser power 1000 W

It can be seen that the size of prior beta grains tends to increase with laser power and decrease with mass flow rate. Figure 7.21 shows the changes in average prior beta grain sizes for GA and PREP deposition samples with respect to mass flow rate and laser power. There is a trend of increase in grain size with laser power, while a decrease in grain size with mass flow rate. Average grain size increases from 277 µm to 386 µm when laser power changes from 800 W to 1000 W at 0.066 gs$^{-1}$ for GA deposition samples. PREP powder deposition samples exhibit larger prior beta grains.
size than that of GA powder deposition samples at all mass flow rates and laser power levels.

The x-ray diffraction of a GA thin-wall deposition sample (800 W and 0.033 gs\(^{-1}\) mass flow rate) at top and bottom positions reveals that the sample contains both alpha and beta phases (Fig. 7.22). At the top, the sample contains more beta phase than at the bottom. A previous study [163] showed that the absence of the beta phase might be an indication of a martensitic microstructure. The presence of beta phase at both positions of the sample thus confirms that microstructure is \(\alpha + \beta\) rather than a form of martensitic microstructure. The deviations of peak intensity from that of an ideal structure indicate that the material analysed is textured; this is the case for both alpha and beta.

Figure 7.22. X-ray diffraction results of a thin-wall deposition sample at two positions i.e. 2 mm and 6 mm above the substrate level. Arrows indicates the x-ray diffraction measurement planes

Scanning electron microscopy was employed to characterize inside the prior beta grains which exhibit a fully lamellar \(\alpha + \beta\) microstructure, as shown in Figure 7.23. The microstructure exhibits basketweave like \(\alpha\) Widmanstätten grains outlined in retained \(\beta\) grains. The lamellar \(\alpha + \beta\) phase spacing (\(S_{\alpha + \beta}\)) tends to increase with laser
power. PREP powder deposition samples show slightly higher phase spacing than GA powder deposition samples produced at equivalent process parameters.

![SEM images of deposition samples](image)

Figure 7.23. SEM images of the deposition samples showing variations in lamellar alpha-beta phase spacing ($S_{\alpha+\beta}$) at 0.033 g s$^{-1}$ mass flow rate: (a) GA, power 800 W; (b) PREP, power 800 W; (c) GA, power 1000 W; (d) PREP, power 1000 W

Multiple factors can affect a material’s micro hardness, including phase, solute concentration and microstructural scale. However, most of these factors were common to the two types of powder and the powders and substrate were of the same material, negating any dilution effects. Therefore, analysis focussed on quantitative metallographic measurements of lamellar $\alpha+\beta$ phase spacing ($S_{\alpha+\beta}$); its effects on mean micro hardness are given in Table 7.4. The table shows that mean micro hardness tends to decrease with increases in lamellar $\alpha+\beta$ phase spacing ($S_{\alpha+\beta}$). Vickers micro hardness (HV 0.5) decreases from 363 to 343 as the lamellar $\alpha+\beta$ phase spacing ($S_{\alpha+\beta}$) changes from 0.95 µm to 1.56 µm. The effects of mass flow rate and laser power on mean micro hardness of GA and PREP deposition samples are shown in figure 7.24. Micro hardness tends to decrease with laser power for both GA and PREP deposition samples; mass flow rate does not have a significant effect in either case.
Table 7.4. Variations of micro hardness with lamellar α+β phase spacing ($S_{α+β}$)

<table>
<thead>
<tr>
<th>Powder type</th>
<th>Laser power (W)</th>
<th>lamellar $α+β$ phase spacing, (µm)</th>
<th>Mean micro hardness (Hv500)</th>
</tr>
</thead>
<tbody>
<tr>
<td>GA</td>
<td>800 W</td>
<td>0.95 ± 0.15</td>
<td>363</td>
</tr>
<tr>
<td>PREP</td>
<td>800 W</td>
<td>1.20 ± 0.22</td>
<td>354</td>
</tr>
<tr>
<td>GA</td>
<td>1000 W</td>
<td>1.32 ± 0.26</td>
<td>354</td>
</tr>
<tr>
<td>PREP</td>
<td>1000 W</td>
<td>1.56 ± 0.25</td>
<td>343</td>
</tr>
</tbody>
</table>

Figure 7.24. Mean micro hardness (HV 0.5) as a function of laser power and mass flow rate.


7.5 Discussion

7.5.1 Powder analysis

Microcomputed tomography analysis of both powders shows that PREP powder has internal porosity, despite the fact that GA powder sample does not show any surface porosity in scanning electron microscopy characterization. These findings are contrary to previous research [18] that PREP powders are free from porosity. The PREP technique can produce high purity spherical powder particle, but still with this technique it is not possible to get rid of all pores inside the powder particles. However, image analysis of the microcomputed tomography data reveals that the GA sample has three times more porosity than the PREP sample.

Scanning electron microscopy and laser diffraction analysis of the powder samples reveals that GA powder has a larger mean particle size than PREP powder (94 µm compared to 72 µm equivalent spherical diameter). This leads to the particles interacting differently with the laser beam above the melt pool. Firstly, it is established that laser beam attenuation decreases with increase in powder particle size [42] and this will result in more laser energy available at substrate level in the case of GA powder deposition. Secondly, smaller particles heat up more quickly in a laser beam and are more likely to become fully molten in the powder stream [164]. Thus, they are more likely to be molten or near-molten particles in the stream in the case of PREP deposition. In addition to these thermal effects it is possible that the different sizes and morphologies of the powders led to slightly different flow characteristics in the powder stream.

7.5.2 Intralayer porosity generation and its causes

Comparison of Figure 7.7 and Figure 7.9 reveals that the choice of characterization method can produce entirely different values for the measured value of intralayer porosity. Microscopy measurements (Fig. 7.7) indicates that GA and PREP deposition samples have no porosity at 0.033 gs⁻¹ mass flow rate and 800 W laser power while Figure 7.9(a) shows that both samples have maximum porosity at this mass flow rate. Optical microscopy and scanning electron microscopy are very good techniques for microstructure analysis but their drawback in intralayer porosity characterization is that they can only provide information about discrete planes. So, it cannot be said with confidence that intralayer porosity in a single plane obtained
by microscopy techniques is a true representation of the whole deposited volume. In contrast, microcomputed tomography can reveal information about the internal morphology of the whole volume under analysis. It is evident from Figure 7.8 that intralayer pores are at different discrete locations while some volumes are free of pores. The cross-section plane of samples for microscopy analysis might have been selected at a location which was free of pores. Microcomputed tomography analysis disagrees with the previous work of Zhao et al [18], which asserted that PREP deposition samples were free of porosity, and this may be because characterization in that work was conducted using microscopy. It also shows no evidence of the microporous coalescence that these authors attributed the fracture mechanism to.

Figure 7.9 gives an interesting insight into the porosity generation behaviour. Porosity generation is clearly complex and does not show a linear relation with the primary processing parameters of laser power and mass flow rate. This indicates that intralayer porosity is governed by more than one mechanism. A clear source of intralayer porosity formation in deposition is the initial porosity present in powder samples, which is evident from Figures 7.8-7.11. The quantitative analysis presented in Figures 7.9-7.11 reveals that PREP deposition samples have less porosity at all mass flow rates and laser power levels which provide evidence that it relates to the initial powder porosity. However, Table 7.3 showed that PREP powder has approximately three times less porosity than GA powder, but PREP deposition samples do not show three times less porosity than GA deposition samples and porosity does not scale with mass flow rate. Therefore, either conditions for porosity release vary or there is an additional source of porosity.

A second influential factor in the final level of porosity may be the incident energy level and its relation with melt pool stability. Deposition samples from either type of powder have a minimum porosity at 0.066 gs$^{-1}$ mass flow rate. This can be attributed to the fact that at this particular mass flow rate enough energy was available to melt all incoming powder and the melt pool was stable enough to allow the gas bubbles to escape before solidification. As mass flow rate decreases from 0.066 gs$^{-1}$, the porosity starts increasing and reaches a maximum value at 0.033 gs$^{-1}$ for both laser power levels. Porosity also tends to appear again when mass flow rate increases above 0.066 gs$^{-1}$ for both laser power levels. The complex behaviour of porosity generation with processing parameters can perhaps be explained by considering
Marangoni flows in the melt pool. Previous research [106] shows that Marangoni flows become stronger and make the melt pool unstable at higher laser power. When incident energy level becomes higher than the optimal level, strong Marangoni flows appear and these may help retain entrapped gas bubbles, which further appear as intralayer porosity. When mass flow increases, causing less energy to be available at substrate, the melt pool again becomes stable and allows gas bubbles to escape, thus reducing the porosity (Fig. 7.9). Further increases in mass flow decreases the incident energy and when it goes below an optimal energy level the intralayer porosity increases again, as is apparent in Figure 7.9(a) and Figure 7.9(b) at 0.089 gs\(^{-1}\) to 0.135 gs\(^{-1}\) mass flow rates. This can be attributed to the fact that at lower energy levels, less powder is melted in the powder stream and the powder does not fully melt as quickly in the melt pool thus leaving less time for the pores to become exposed and gas to escape. Laser power shows a clear influence on the porosity for both powders, as shown in Figure 7.9(a) and (b). When the scanning speed is constant and laser power changes so does the temperature of the melt pool and thus the Marangoni convection, melt pool flow, behaviour of the pore nucleation, growth and movement will all be different. Figure 7.9(a) and (b) show maximum intralayer porosity at 0.033 gs\(^{-1}\), when maximum energy is available at substrate hence suggesting the melt pool’s instability.

Increased probability of conveyance gas entrapment at higher mass flow rates could be another factor that increases overall intralayer porosity. The maximum pore diameters variation with mass flow rate in each deposition sample, as shown in Figure 7.12 is similar to the overall porosity levels behaviour (Fig. 7.9) and could be due to gas entrapment adding larger pores. The variation in mean pore diameters with mass flow rates for GA deposition samples is same as it is for maximum pore diameters. However, the decrease in mean pore diameters with mass flow rate for PREP deposition samples (Fig. 7.13) may be attributed to a lower probability of gas entrapment because of its smooth surface, as shown in Figure 7.3(b). The pores diameter frequency distribution and its relations with laser power and mass flow rate are evident in Figures 7.10 and 7.11. The higher pore frequency in GA deposits than PREP deposits suggests that there will be more gas entrapment in GA deposits than PREP deposits, which further reflects in terms of overall lower volume porosity in PREP powder deposition samples (Fig. 7.9).
7.5.3 Geometric properties and surface finish

The greater energy at the substrate with GA powder leads to more of the initial substrate being melted and a wider and deeper melt pool than with PREP powder. Because powder is assimilated over the width of the pool, this gives wider deposits with GA powder than PREP powder (Fig. 7.15). It is the amount of powder assimilated along the length of the melt pool that gives the final track height. This is initially greater for the GA powder, but the higher temperature of the PREP powder means it is more easily assimilated towards the rear ‘mushy’ zone of the pool, thus increasing the pool length and the deposit height for this type of powder. So, the melt pool generated during PREP powder deposition will be slightly shallower and longer and the deposition layer thicker than during GA powder deposition at the same parameters (Fig. 7.14).

The lower surface roughness ($R_a$) of PREP deposition samples can be attributed to the particles being both smaller and hotter and there thus being fewer cases of only partial-assimilation into the substrate surface, which is in agreement with previous studies [160, 165].

7.5.4 Microstructure and micro hardness analysis

The microstructure of the GA and PREP powder deposits exhibits long, columnar prior beta grains, which grow across the layer bands and are mainly inclined perpendicular to the plane of laser scanning direction. The layer bands, which appears from the first deposition layer and disappear before the last three deposition layers, are the results of the complex thermal history experienced during multiple layers deposition, as explained by Kelly and Kampe [94]. The increase in prior beta grains size with laser power can be related to the lower cooling rates at higher laser powers. The prior beta grains of PREP powder deposition samples are larger than those of GA powder deposition samples at the same process parameters, which can be explained by considering the longer melt pool generated by the PREP powder. A longer melt pool decreases the temperature gradient at the solidification front, resulting in decreased cooling rates [46].

The analytical-numerical model developed for LDMD which is explained in Chapter 6 (Section 6.3.5) is used here to explain the relationship between cooling rates and solidification microstructure.
Figure 7.25 shows the modelled cooling rates for Ti-6Al-4V as a function of mass flow rate. It can be seen that there is a slight, but not steep, increase in cooling rate with mass flow rate. Conversely, an abrupt decrease in the grain size with mass flow rate is shown in Figures 7.19-7.21. This suggests that there is another important factor, along with cooling rates, controlling microstructural scale and this factor is most likely nucleation density during microstructure formation. It is possible that an increased mass flow rate changes solidification conditions in the upper melt pool portion by increasing nucleation density. The increased nucleation density then changes the microstructure morphology from long columnar to small equiaxed grains, as is evident from Figures 7.19 and 7.20 [86, 95]. So, there are two reasons why the grain size tends to reduce with increasing mass flow rate: the first is the increase in cooling rates due to less energy being available at the substrate and the second is the increase in nucleation density.

The lamellar $\alpha+\beta$ phase spacing ($S_{\alpha+\beta}$) increases with laser power, which can also be explained by relating it to the change in cooling rates discussed above. The result presented in Table 7.4 show that the micro hardness of the deposited samples is primarily related to lamellar $\alpha+\beta$ phase spacing ($S_{\alpha+\beta}$). The less the lamellar $\alpha+\beta$ phase spacing ($S_{\alpha+\beta}$), the greater the phase boundary reinforcement thus increasing the hardness of the material. The larger lamellar $\alpha+\beta$ phase spacing ($S_{\alpha+\beta}$) in the case of PREP powder deposition may be attributed to the slower cooling rates. Micro
hardness results indicate that the lamellar $\alpha+\beta$ phase spacing ($S_{\alpha+\beta}$) is not affected significantly by mass flow rate.

7.2 Conclusions

This chapter explains a systematic investigation of laser deposited thin-wall structures produced using gas-atomized (GA) and plasma rotating electrode process (PREP) Ti-6Al-4V powders. GA and PREP Ti-6Al-4V powders are investigated using scanning electron microscopy and microcomputed tomography. The PREP powder particles are highly spherical, have pore free surfaces but, contrary to some previous investigations, do have some internal pores. However, image processing of the tomography data reveals that the GA powder had three times more porosity than the PREP powder. Comparisons of porosity results obtained from previous microscopy studies and this work indicate that microcomputed tomography is the more reliable method of characterization of porosity in powder and laser deposited structures.

GA powder deposition samples show higher intralayer porosity at all mass flow rate and laser power levels than the PREP powder deposition samples. This evidence directly relates intralayer porosity in deposition samples with initial powder porosity; however, the ratio of porosities in the powders is not equal to that of porosities in the deposits which indicates that porosity is not simply transferred from the powder to the deposit and that other factors are involved. Nevertheless it can be concluded that PREP powder should be used to obtain minimum intralayer porosity. The relationship between volumetric porosity in the deposition samples and the laser processing parameters of laser power and powder mass flow rate is more complicated. Other factors which may be at work include strong Marangoni flows favouring pore nucleation, growth and buoyancy at high laser powers, melt pool instability at low mass flow rates and the influence of the surrounding atmosphere. In this experiment, the intralayer porosity was reduced to a minimum level of 0.0025% using PREP powder at 1000 W, 0.066 gs$^{-1}$ and 5 mms$^{-1}$. The main trends with process parameters are a reduction in porosity with increased power and high porosity at very low mass flow rates. So, for high value parts, PREP powder, higher power and moderate powder mass flow rate, as dictated by other process constraints, appears to be a practical combination.
The comparative study of LDMD using these two types of powder has considered the potential benefits of using PREP powder. The investigation revealed that PREP powder deposits show lower surface roughness, lower intralayer porosity and a higher deposition rate. Laser deposited Ti-6Al-4V exhibits long columnar prior beta grains with a fully lamellar $\alpha+\beta$ microstructure. Prior beta grain size in the deposits increases with laser power and tends to decrease with mass flow for both types of powders. Nucleation density also plays an important role during microstructure formation and at higher powder mass flow rates increased nucleation density in the melt pool tends to change the microstructure morphology from long columnar to small equiaxed grains. The lamellar $\alpha+\beta$ phase spacing ($S_{\alpha+\beta}$) increases with laser power, but seems unaffected by variation in mass flow rate. Micro hardness of the laser deposited Ti-6Al-4V is dependent on the lamellar $\alpha+\beta$ phase spacing ($S_{\alpha+\beta}$): Vickers micro hardness (HV 0.5) decreases from 363 to 343 as the lamellar $\alpha+\beta$ phase spacing ($S_{\alpha+\beta}$) changes from 0.95 µm to 1.56 µm.
Chapter 8

Conclusions and future recommendations

8.1 Conclusions

This thesis has expanded some aspects of laser direct metal deposition of Ti-6Al-4V alloy, ranging from modelling investigations to experimental analysis. Although detailed conclusions for each part of the work have been given in the relevant chapter some important results and conclusions are summarized as follows:

Analytical modelling of multiple track LDMD for surface layer cladding

- A 3D analytical model of multiple overlapping laser direct metal deposition tracks for surface layer deposition has been developed. The model caters for the laser power losses in the powder stream to partially overcome inaccuracies of previous conventional models that rely on de-coupled heat and mass flow calculations.
- The model accounts for track interaction in deposition of a single clad layer or multiple layers and its effects on the melt pool. A novel factor of ‘powder mass affinity’ is introduced to capture the effect of surface geometry on deposition rate for particular powder flow conditions.
- Powder mass affinity is a function of the track profile geometry, overlap ratio, nozzle dimensions, carrier gas flow rate and possibly other factors not examined in this work.
- The surface layer profile, layer thickness and layer surface roughness of a number of surface deposition layers, produced using multiple track laser direct metal deposition of Ti-6Al-4V alloy, show good agreement with the model simulations.
- Modelled and experimental results, show the layer thickness to increase with mass flow rate and overlap ratio. The surface roughness of the multiple track surface layers decreases with increasing overlap ratio in the range 10-60%.
Multilayer porous structures fabrication by continuous and pulsed-wave LDMD for biomedical applications

- MicroCT analysis of continuous-wave deposited samples showed that pores with excellent interconnectivity were produced, which is an important requirement for cells migration. The pore size is very important for proper osseointegration in biomedical implants and it can be controlled by optimizing the deposited track geometry and offset distance.

- Pulsed-wave LDMD to create interacting balls rather than deposition tracks allowed structures with a greater level of controlled porosity to be created than continuous-wave LDMD. It is possible to control final part characteristics such as pore size and overall porosity by selection of deposition parameters.

- Both methods are able to produce porous structures with pore sizes of the order of 100 microns, but with a different set of processing parameters required for each method. A major process variable affecting pore sizes is laser spot diameter and at same laser spot diameter, pulsed-wave laser deposition can produce smaller pore sizes compared to those of continuous-wave laser deposition.

- Analytical models of both processes have been developed and can be used as a reliable way to predict the nature of porous structures.

Analytical-numerical modelling of LDMD including microstructure formation

- A 3D analytical-numerical model of laser direct metal deposition to simulate the multiple interdependent laser deposition processes has been developed.

- Comparison with the experimental results of local temperature history, track profile geometry and microstructure grain size shows good agreement. Modelled cooling rates are found to be of the order of $10^3$ K for Ti-6Al-4V; they decrease with laser power but are not substantially affected by mass flow rate.

- The analytical-numerical coupled approach makes the model not only accurate but computationally efficient as compared to finite element models. Typical model processing time is of the order of 100 seconds, which makes the model an efficient aid for setting accurate deposition processing parameters for a required track profile and microstructural scale.
A comparative study of LDMD characteristics using gas and plasma-atomized Ti-6Al-4V powders

- Microcomputed tomography is more reliable method than microscopy for characterization of porosity in powder and laser deposited structures due to its ability to characterize a designated 3D volume.

- The PREP powder particles are highly spherical, have pore-free surfaces but, contrary to some previous investigations, do have internal porosity. MicroCT characterization reveals that the GA powder has three times more porosity than the PREP powder.

- Intralayer porosity in deposition samples directly relates to the initial powder porosity, however its relationship with the laser processing parameters is more complicated. Other factors which may be at work include strong Marangoni flows favouring pore nucleation, growth and buoyancy at high laser powers, melt pool instability at low mass flow rates and the influence of the surrounding atmosphere. To minimize the intralayer porosity, PREP powder, higher power and moderate powder mass flow rate, as dictated by other process constraints, should be used.

- The investigation reveals some potential benefits of using PREP powder e.g. lower surface roughness, lower intralayer porosity and a higher deposition rate.

- Laser deposited Ti-6Al-4V exhibits long columnar prior beta grains with a fully lamellar $\alpha+\beta$ microstructure. Prior beta grain size in the deposits increases with laser power and tends to decrease with mass flow for both types of powders. At higher powder mass flow rates increased nucleation density in the melt pool tends to change the microstructure morphology from long columnar to small equiaxed grains.

- The lamellar $\alpha+\beta$ phase spacing ($S_{\alpha+\beta}$) increases with laser power, but seems unaffected by variation in mass flow rate. Micro hardness of the laser deposited Ti-6Al-4V increases with decrease in the lamellar $\alpha+\beta$ phase spacing ($S_{\alpha+\beta}$).
8.2 Future work recommendations

*Multiple track LDMD model improvement*

The current multiple track interaction model (Chapter 4) needs improvement to predict deposition layers at higher overlap ratios e.g. more than 60% overlap. The model can be further extended to develop a universal track interaction model that would encapsulate from the surface layer coating (40-60% overlap) to thin-wall fabrication (100% overlap ratio). The current multilayer model (Chapter 5) does include the tracks interactions within a layer but, does not include the interactions between orthogonal layers that need to be incorporated in future.

*Image-based modelling of porous structures fabricated using LDMD*

Porous structures fabricated by LDMD and characterized by MicroCT (Chapter 5) opens up further opportunities for image-based modelling of such structures. The tomography data in the form of a stack of 2D images can be reconstructed to 3D models using available commercial software. The constructed 3D models will fully describe the porous structures and can be analysed using finite element software for structural analysis.

*Further exploitation of analytical-numerical coupled approach*

The analytical-numerical approach (Chapter 6) can be further exploited to perform detailed energy distributions modelling of the LDMD process. Due to its capability to balance energy-mass equations at the melt pool level, it could give more accurate energy distributions of the process than previous energy distribution models.

The second possible extension of the analytical-numerical approach is in detailed microstructure modelling to prepare microstructure process maps of Ti-6Al-4V and other common deposition materials e.g. 316 L stainless steel, Inconel 718 etc. The important considerations may be the effect of mass flow rate on cooling rates, and the influence of the Marangoni Number, which changes the flow behaviour within the melt pool and hence solidification behaviour.
References


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Appendix A

Mathematica code developed for surface cladding model

*Heat flow model without power losses for calculations of melt pool size and temperature*

```mathematica
absbeampower=1100;
velocity=0.01;
qtotal=0.000112;
absorptivity=0.5;
halfmaxbeamrad=0.0012;
rpowder=0.00325/2ambt=20;
conductivity=17;
density=4300;
specificheat=800;
meltpoint=1660;
Lm=370000;
Lv=9830000;

diffusivity:=conductivity/(specificheat× density)
dimlessx:= x/halfmaxbeamrad
dimlessy:= y/halfmaxbeamrad
dimlessz:= z/halfmaxbeamrad
σ:=(halfmaxbeamrad ×velocity)/diffusivity

beampower1=absbeampower×fibreefficiency-beampowerlosspowderstream/.fibreefficiency → 0.7

integrand1:=(beampower1×absorptivity)/(conductivity×halfmaxbeamrad)×Exp[-{((dimlessy+σ/2
µ^2)+dimlessx^2)(2(1+µ/2))+dimlessz^2(2,2))]/(2π)^1/2(1+µ/2)/.z → 0
/.
x →{0,0.0005,0.001,0.0015,0.002,0.0025,0.003}
integrand2:=(beampower1×absorptivity)/(conductivity×halfmaxbeamrad)×Exp[-{((dimlessy+σ/2
µ^2)+dimlessx^2)(2(1+µ/2))+dimlessz^2(2,2))]/(2π)^1/2(1+µ/2) /.x → 0/.y →{0,-0.0005,-0.001,-0.0015,-0.002,-0.0025,-0.003}

integrandy1:=(beampower1×absorptivity)/(conductivity×halfmaxbeamrad)×Exp[-{((dimlessy+σ/2
µ^2)+dimlessx^2)(2(1+µ/2))+dimlessz^2(2,2))]/(2π)^1/2(1+µ/2) /.x → 0/.y →{0,0.0005,0.001,0.0015,0.002,0.0025,0.003}
integrandy2:=(beampower1×absorptivity)/(conductivity×halfmaxbeamrad)×Exp[-{((dimlessy+σ/2
µ^2)+dimlessx^2)(2(1+µ/2))+dimlessz^2(2,2))]/(2π)^1/2(1+µ/2) /.y → 0/.x →{0,-0.0005,-0.001,-0.0015,-0.002,-0.0025,-0.003}

integrandz1:=(beampower1×absorptivity)/(conductivity×halfmaxbeamrad)×Exp[-{((dimlessy+σ/2
µ^2)+dimlessx^2)(2(1+µ/2))+dimlessz^2(2,2))]/(2π)^1/2(1+µ/2) /.x → 0/.z →{0,0.0005,0.001,0.0015,0.002,0.0025,0.003}

temperaturesx1=NIntegrate[integrandx1,{µ,0,20}]+{ambt}

```

192
meltmeantempx1 := Mean[Take[temperaturesx1[[1]], 6], {1, x, x^2, x^3, x^4, x^5}, x] = meltpoint, x ]
meltmeantempx2 := Mean[Take[temperaturesx2[[1]], 6], {1, x, x^2, x^3, x^4, x^5}, x] = meltpoint, x ]
meltmeantempy1 := Mean[Take[temperaturesy1[[1]], 6], {1, x, x^2, x^3, x^4, x^5}, x] = meltpoint, x ]
meltmeantempy2 := Mean[Take[temperaturesy2[[1]], 6], {1, x, x^2, x^3, x^4, x^5}, x] = meltpoint, x ]
meltmeantemp = Mean[{meltmeantempx1, meltmeantempx2, meltmeantempy1, meltmeantempy2}]
s1 := Solve[Fit[Partition[Riffle[{0, 0.0005, 0.001, 0.0015, 0.002, 0.0025, 0.003}, temperaturesx1[[1]], 2], {1, x, x^2, x^3, x^4, x^5}, x], meltmeantempx1, x ]
s2 := Solve[Fit[Partition[Riffle[{0, 0.0005, 0.001, 0.0015, 0.002, 0.0025, 0.003}, temperaturesy1[[1]], 2], {1, x, x^2, x^3, x^4, x^5}, x], meltmeantempy1, x ]
s3 := Solve[Fit[Partition[Riffle[{0, 0.0005, 0.001, 0.0015, 0.002, 0.0025, 0.003}, temperaturesy2[[1]], 2], {1, x, x^2, x^3, x^4, x^5}, x], meltmeantempy2, x ]
s4 := Solve[Fit[Partition[Riffle[{0, 0.0005, 0.001, 0.0015, 0.002, 0.0025, 0.003}, temperaturesx2[[1]], 2], {1, x, x^2, x^3, x^4, x^5}, x], meltmeantempx2, x ]
s1a := Sort[# & /@ Cases[x /. s1, _Real], Less]
s1b := TakeWhile[s1a, # < 0 &]
s1c := Complement[s1a, s1b] // Sort
halfmeltwidth1 := s1c[[1]]
s2a := Sort[# & /@ Cases[x /. s2, _Real], Less]
s2b := TakeWhile[s2a, # < 0 &]
s2c := Complement[s2a, s2b] // Sort
meltfront1 := s2c[[1]]
s3a := Sort[# & /@ Cases[x /. s3, _Real], Less]
s3b := TakeWhile[s3a, # < 0 &]
s3c := Complement[s3a, s3b] // Sort
meltrear1 := s3c[[1]]
s4a := Sort[# & /@ Cases[x /. s4, _Real], Less]
s4b := TakeWhile[s4a, # < 0 &]
s4c := Complement[s4a, s4b] // Sort
meltdepth1 := s4c[[1]]
meltw1 = 2 Round[halfmeltwidth1, 0.0001]
meltlf1 = Round[meltfront1, 0.0001]
meltlr1 = Round[meltrear1, 0.0001]
meltd1 = Round[meltdepth1, 0.00001]

**Accurate heat flow model with power losses using melt pool size and temperature**

massflux := Solve[Log[m] == A + 6.1210 - 18836/Log[temp]/(temp)^1/2, temp] => meltmeantemp
evaporationheatflux := mm × Lv
surfacearea := Pi/4 × meltw × (meltlf + meltlr)
evaporationheatfluxloss = evaporationheatflux × surfacearea
beampower = beampower1 - evaporationheatfluxloss

integrand1 := (beampower × absorptivity × conductivity × halffmaxbeamrad) × Exp[-((dimless + µ2/2)² + dimless²)/(2(1 + µ2) + dimless²)/(2π)^(1/2) (1 + µ2)/z] × 0
/x -> {0, 0.0005, 0.001, 0.0015, 0.002, 0.0025, 0.003}
y -> 0
integrand2 := (beampower × absorptivity × halffmaxbeamrad) × Exp[-((dimless + µ2/2)² + dimless²)/(2(1 + µ2) + dimless²)/(2π)^(1/2) (1 + µ2)/z] × 0
/y -> {0, 0.0005, 0.001, 0.0015, 0.002, 0.0025, 0.003}
z -> 0
integrand3 := (beampower × absorptivity × halffmaxbeamrad) × Exp[-((dimless + µ2/2)² + dimless²)/(2(1 + µ2) + dimless²)/(2π)^(1/2) (1 + µ2)/x] × 0
/x -> {0, 0.0005, 0.001, 0.0015, 0.002, 0.0025, 0.003}
y -> 0
integrand4 := (beampower × absorptivity × halffmaxbeamrad) × Exp[-((dimless + µ2/2)² + dimless²)/(2(1 + µ2) + dimless²)/(2π)^(1/2) (1 + µ2)/y] × 0
/y -> {0, 0.0005, 0.001, 0.0015, 0.002, 0.0025, 0.003}
x -> 0

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Appendix A

\[ \text{integrand}\ z_1 := \frac{(\text{beampower} \times \text{absorptivity})}{(\text{conductivity} \times \text{halfmaxbeamrad})} \times \exp\left[-\frac{(\text{dimlessy} + \sigma/2 \times \mu^2 + \text{dimlessx}^2)}{(2(1 + \mu^2))}\right] / (2\pi)^{1/2} (1 + \mu^2) / y \rightarrow 0 / x \rightarrow 0 / z \rightarrow \{0, 0.0005, 0.001, 0.0015, 0.002, 0.0025, 0.003\} \]

\[ \text{temperaturesx1} = N\text{Integrate}[\text{integrandx1}, \{\mu, 0, 0.2\}] + \{\text{ambt}\} \]
\[ \text{temperaturesx2} = N\text{Integrate}[\text{integrandx2}, \{\mu, 0, 0.2\}] + \{\text{ambt}\} \]
\[ \text{temperaturesy1} = N\text{Integrate}[\text{integrandy1}, \{\mu, 0, 0.2\}] + \{\text{ambt}\} \]
\[ \text{temperaturesy2} = N\text{Integrate}[\text{integrandy2}, \{\mu, 0, 0.2\}] + \{\text{ambt}\} \]
\[ \text{temperaturesz1} = N\text{Integrate}[\text{integrandz1}, \{\mu, 0, 0.2\}] + \{\text{ambt}\} \]

\[ s1 := \text{Sort}[\# &/\@ \text{Cases}[x, \text{s1} . \_\text{Real}] , \text{Less}] \]
\[ s1b := \text{TakeWhile}[s1a, \# < 0 &] \]
\[ s1c := \text{Complement}[s1a, s1b] /\@ \text{Sort} \]
\[ \text{halfmeltwidth} := s1c[[1]] \]
\[ s2 := \text{Sort}[\# &/\@ \text{Cases}[x, \text{s2} . \_\text{Real}] , \text{Less}] \]
\[ s2b := \text{TakeWhile}[s2a, \# < 0 &] \]
\[ s2c := \text{Complement}[s2a, s2b] /\@ \text{Sort} \]
\[ \text{meltfront} := s2c[[1]] \]
\[ s3 := \text{Sort}[\# &/\@ \text{Cases}[x, \text{s3} . \_\text{Real}] , \text{Less}] \]
\[ s3b := \text{TakeWhile}[s3a, \# < 0 &] \]
\[ s3c := \text{Complement}[s3a, s3b] /\@ \text{Sort} \]
\[ \text{meltrear} := s3c[[1]] \]
\[ s4 := \text{Sort}[\# &/\@ \text{Cases}[x, \text{s4} . \_\text{Real}] , \text{Less}] \]
\[ s4b := \text{TakeWhile}[s4a, \# < 0 &] \]
\[ s4c := \text{Complement}[s4a, s4b] /\@ \text{Sort} \]
\[ \text{meltdepth} := s4c[[1]] \]
\[ \text{meltw} := 2 \times \text{Round}[\text{halfmeltwidth}, 0.0001] \]
\[ \text{meltlf} := \text{Round}[\text{meltfront}, 0.0001] \]
\[ \text{meltlr} := \text{Round}[\text{meltrear}, 0.0001] \]
\[ \text{meltd} := \text{Round}[\text{meltdepth}, 0.0001] \]

**Deposition profile calculations**

\[ \text{qp} := (2 \times q_{\text{total}} / (\pi \times \text{rpowder}^2)) \exp[-2 ((x - k \times \text{xcentre})^2 + y^2) / \text{rpowder}^2] \]

\[ \text{forwardlimit := Abs}[\text{Re}[\text{meltlf} \sqrt{1 - \frac{4 (x - k \times \text{xcentre})^2}{\text{meltw}^2}]] \]

\[ \text{rearlimit := Abs}[\text{Re}[\text{meltlr} \sqrt{1 - \frac{4 (x - k \times \text{xcentre})^2}{\text{meltw}^2}]} / \text{forwardlimit}] \]

\[ \text{depheight0 := 1 / (density \times \text{velocity})} \]
\[ \text{plotdepheight0 = plotdepheight0 / (1 - overlapratio) \times \text{meltw} / k \rightarrow 0;} \]
\[ \text{trackplot = Plot[plotdepheight0, \{x, -0.002, 0.004\}, PlotRange -> \{-0.002, 0.004\}, \{-0.002, 0.002\}], AspectRatio -> 0.25];} \]
\[ \text{dilution := meltdepth} / (\text{density} \times \text{velocity}) \]
\[ \text{dilutionplot = Plot[dilution, \{x, -0.002, 0.004\}, PlotRange -> \{-0.002, 0.004\}, \{-0.002, 0.002\}], AspectRatio -> 0.25];} \]
\[ \text{Show[dilutionplot, trackplot];} \]
baselayerplot=Plot3D[plotdepheight0, {x, -0.002, 0.004}, {y, -0.002, 0.004}, PlotRange -> {{-0.002, 0.004}, {-0.002, 0.004}, {-0.002, 0.002}}];
dilutionlayerplot=Plot3D[dilution, {x, -0.002, 0.004}, {y, -0.002, 0.004}, PlotRange -> {{-0.002, 0.004}, {-0.002, 0.004}, {-0.002, 0.002}}];
firstlayerplot=Plot3D[plotdepheight2+0.001, {x, -0.002, meltw/2}, PlotRange -> {{-0.002, 0.004}, {0.002, 0.004}, {0.001, 0.003}}], Filling -> 0.001, FillingStyle -> {Red, Brown}, AspectRatio -> 0.20];
Show[dilutionlayerplot, baselayerplot];

**Surface layer deposition**

forwardlimit1 := (meltlf)/(meltw - xcentre) (x - k×xcentre + meltw/2)
rearlimit1 := -(meltlr)/(meltw - xcentre) (x - k×xcentre + meltw/2)

\[
depheight1 := \frac{1}{\text{density} \times \text{velocity}} 
\]

\[
xcentre := (1 - \text{overlapratio}) \times \text{meltw}
\]

\[
i := \frac{\text{meltw}/2 + (k-1) \times \text{xcentre}}{2}
\]

\[
\text{Powderabsorptionfactor} := 1 + \frac{1 - \text{overlapratio}}{5} \times \exp\left(-70000000 \left(x - i\right)^2\right)
\]

\[
\text{Powderabsorptionfactor1} := 1 + \frac{1 - \text{overlapratio}}{5} \times \exp\left(-70000000 \left(y - i\right)^2\right)
\]

\[
\text{overlapratio} := 0.40;
\]

\[
k := 1;
\]

plotdepheight 1a=dephight1×Powderabsorptionfactor;
plotdepheight 1ar=If[(k×xcentre - meltw/2) < x < i, plotdepheight1a, 0];
plotdepheight 1b=dephight0×Powderabsorptionfactor;
plotdepheight 1br=If[x > i, plotdepheight1b, 0];
Plot[plotdepheight0+plotdepheight1ar+plotdepheight1br, {x, -0.002, 0.0008}, PlotRange -> {{-0.002, 0.0008}, {0, 0.002}}, AspectRatio -> 0.20, AxesOrigin -> {0, 0}];
baselayerplot1=Plot3D[plotdepheight0+plotdepheight1ar+plotdepheight1br, {x, -0.002, 0.0004}, {y, -0.002, 0.0004}, PlotRange -> {{-0.002, 0.0004}, {0, 0.0003}, {0, 0.0003}}, Filling -> Bottom, FillingStyle -> {Red, Brown}, AspectRatio -> 0.20];
Show[{baselayerplot1, firstlayerplot1}];

k := 2;
plotdepheight 2a=dephight1×Powderabsorptionfactor;
plotdepheight 2ar=If[(k×xcentre - meltw/2) < x < i, plotdepheight2a, 0];
plotdepheight 2b=dephight0×Powderabsorptionfactor;
plotdepheight 2br=If[x > i, plotdepheight2b, 0];
Plot[plotdepheight0+plotdepheight1ar+plotdepheight1br+plotdepheight2ar+plotdepheight2br, {x, -0.002, 0.0008}, PlotRange -> {{-0.002, 0.0008}, {0.00025}}, AspectRatio -> 0.20, AxesOrigin -> {0, 0}];
baselayerplot2=Plot3D[plotdepheight0+plotdepheight1ar+plotdepheight1br+plotdepheight2ar+plotdepheight2br, {x, -0.002, 0.0006}, {y, 0.0003}, PlotRange -> {{-0.002, 0.0006}, {0.0003}, {0, 0.0003}}, Filling -> Bottom, FillingStyle -> {Red, Brown}, AspectRatio -> 0.20];
plotdepheight 2=dephight1/x−y;
plotdepheight2jj = plotdepheight2j \times \text{Powderabsorptionfactor1};
plotdepheight2jjr = \text{If}[(k \times \text{xcentre} - \text{meltw}/2) < y < i, \text{plotdepheight2jj}, 0];
plotdepheight2ii = \text{depheight0}/x \rightarrow y;
plotdepheight2iir = \text{If}[y > i, \text{plotdepheight2ii}, 0];
plotdepheight2ii = \text{plotdepheight2i} \times \text{Powderabsorptionfactor1};
plotdepheight2iir = \text{If}[y > i, \text{plotdepheight2ii}, 0];

\text{firstlayerplot2} = \text{Plot3D}[
\text{plotdepheight2} + \text{plotdepheight1jjr} + \text{plotdepheight1iir} + \text{plotdepheight2jjr} + \text{plotdepheight2iir} + 0.001, \{x, -0.002, 0.006\}, \{y, -(\text{meltw}/2), \text{meltw}/2 + k \times \text{xcentre}\}, \text{PlotRange} \rightarrow \{-0.002, 0.006\}, \{-0.002, 0.006\}, \{0.001, 0.003\}\}, \text{Filling} \rightarrow \text{Bottom}, \text{FillingStyle} \rightarrow \{\text{Red}, \text{Brown}\}, \text{AspectRatio} \rightarrow 0.20\]
\text{Show}[[\text{baselayerplot2}, \text{firstlayerplot2}]]
Appendix B

Mathematica code developed for continuous-wave porous structures deposition model

**Multilayer continuous-wave deposition**

\[
\text{forwardlimit1} := \left(\frac{1 - \frac{4 \left(x\text{centre} - \frac{\text{meltw}}{2}\right)^2}{\text{meltw}^2}}{(\text{meltw} - x\text{centre}) (x - k \times x\text{centre} + \text{meltw}/2)}\right) \ 
\text{rearlimit1} := -\left(\frac{1 - \frac{4 \left(x\text{centre} - \frac{\text{meltw}}{2}\right)^2}{\text{meltw}^2}}{(\text{meltw} - x\text{centre}) (x - k \times x\text{centre} + \text{meltw}/2)}\right) \ 
\text{depheight1} := \frac{1}{(\text{density} \times \text{velocity})} \ 
\text{overlapratio} = 0.00000001; \ 
\text{xcentre} = (1 - \text{overlapratio}) \times \text{meltw}; \ 
i = (\text{meltw}/2 + (k - 1) \times x\text{centre}); \ 
m = \text{overlapratio} \times \text{meltw}; \ 
\text{mr} = \text{If}[m > 0, m, 0] \ 
\text{dilution} := -\text{Abs}[\text{Re[meldx]}] \ 
\text{Powderabsorptionfactor} := 1 + (1 - \text{overlapratio})/5 \times \text{Exp}[-70000000 (i - x)^2]; \ 
\text{Powderabsorptionfactor1} := 1 + (1 - \text{overlapratio})/5 \times \text{Exp}[-70000000 (i - y)^2]; \ 
k = 1; \ 
\text{plotdepheight0} = \text{depheight0} \times \text{Powderabsorptionfactor}; \ 
\text{plotdepheight1} = \text{If}[x < i, \text{plotdepheight1a}, 0]; \ 
\text{plotdepheight1b} = \text{depheight0} \times \text{Powderabsorptionfactor}; \ 
\text{plotdepheight1br} = \text{If}[x > i, \text{plotdepheight1b}, 0]; \ 
\text{trackplot1} = \text{Plot}[	ext{plotdepheight0} + \text{plotdepheight1a} + \text{plotdepheight1br}, \{x, -0.002, 0.004\}, \text{PlotRange} \to \{-0.002, 0.004\}, \{y, -0.002, 0.002\}, \text{AspectRatio} \to \{0.20, \text{AxesOrigin} \to \{0, 0\}\}; \ 
\text{dilution1} = \text{dilution}; \ 
\text{dilutionplot1} = \text{Plot}[	ext{dilution0} + \text{dilution1}, \{x, -0.002, 0.004\}, \text{PlotRange} \to \{-0.002, 0.004\}, \{y, -0.002, 0.002\}, \text{AspectRatio} \to \{0.20, \text{AxesOrigin} \to \{0, 0\}\}; \ 
\text{Show[trackplot1, dilutionplot1]} \ 
\text{baselayerplot1} = \text{Plot}[	ext{plotdepheight0} + \text{plotdepheight1a} + \text{plotdepheight1br}, \{x, -0.002, 0.004\}, \{y, -0.002, 0.004\}, \{z, -0.002, 2 \times \text{maxtrackheight}\}, \text{Filling} \to \text{Axis}, \text{FillingStyle} \to \{\text{Brown}\}, \text{AspectRatio} \to \{0.20, \text{AxesOrigin} \to \{0, 0\}\}; \ 
\text{baselayerplot1} = \text{Plot}[	ext{plotdepheight0} + \text{plotdepheight1a} + \text{plotdepheight1br}, \{x, -0.002, 0.004\}, \{y, -0.002, 0.004\}, \{z, -0.002, 0.004\}, \{\text{AxesStyle} \to \{\text{Bold}\}, \text{AspectRatio} \to \{0.20, \text{AxesOrigin} \to \{0, 0\}\}; \ 
\text{Show[baselayerplot1, baselayerplot1]} \ 
\text{plotdepheight1} = \text{depheight1}/x \times y; \ 
\text{plotdepheight1j} = \text{plotdepheight1j} \times \text{Powderabsorptionfactor1}; \ 
\text{plotdepheight1jj} = \text{If}[k \times x\text{centre} - \text{meltw}/2] < y < i, \text{plotdepheight1jjr}, 0]; \ 
\text{plotdepheight1i} = \text{depheight0}/x \times y; \ 
\text{plotdepheight1ii} = \text{plotdepheight1ix} \times \text{Powderabsorptionfactor1}; \ 
\text{plotdepheight1ii} = \text{If}[y > i, \text{plotdepheight1ii}, 0]; \ 
\text{Plot}[	ext{plotdepheight2} + \text{plotdepheight1j} + \text{plotdepheight1jj}, \{y, -0.002, 0.008\}, \text{PlotRange} \to \{-0.002, 0.004\}, \{0, 0.002\}, \text{AspectRatio} \to \{0.20, \text{AxesOrigin} \to \{0, 0\}\};
Appendix B

firstlayerplot1=Plot3D[plotdepheight2+plotdepheight1iir+plotdepheight1jjr+maxtrackheight, {x,-0.002,0.004}, {y, -(meltw/2), meltw/2+k×xcentre}, PlotRange→{-0.002,0.004},{-0.002,0.004},{-0.002,2×maxtrackheight}], Filling→maxtrackheight,FillingStyle→{Brown},AspectRatio→0.20];
dilution1=dilution1/.x→y;
firstdilutionplot1=Plot3D[dilution0+dilution1+maxtrackheight, {x,-0.002,0.004}, {y, -(meltw/2), meltw/2+k×xcentre}, PlotRange→{-0.002,0.004},{-0.002,0.004},{-0.002,2×maxtrackheight}], Filling→maxtrackheight,FillingStyle→{Brown},AspectRatio→0.20];
Show[{baselayerplot1,basedilutionplot1,firstlayerplot1,firstdilutionplot1}]
k=2;
plotdepheight2a=depheight1×Powderabsorptionfactor;
plotdepheight2ar=If[(k×xcentre-meltw/2)<x<i,plotdepheight2a,0];
plotdepheight2b=depheight0×Powderabsorptionfactor;
plotdepheight2br=If[x>i,plotdepheight2b,0];
trackplot2=Plot[plotdepheight0+plotdepheight1ar+plotdepheight1br+plotdepheight2ar+plotdepheight2br,{x,-0.002,0.006},PlotRange→{{-0.002,0.006},{-0.002,0.002}},AspectRatio→0.20,AxesOrigin→{0,0}];
dilution2=dilution;
dilutionplot2=Plot[dilution0+dilution1+dilution2, {x,-0.002,0.006},PlotRange→{{-0.002,0.006},{-0.002,0.006}}],AspectRatio→0.20,AxesOrigin→{0,0};
Show[trackplot2,dilutionplot2];
baselayerplot2=Plot3D[plotdepheight0+plotdepheight1ar+plotdepheight1br+plotdepheight2ar+plotdepheight2br, {x,-0.002,0.006}, {y,-0.002,0.006},PlotRange→{{-0.002,0.006},{-0.002,0.006},{-0.002,2×maxtrackheight}}, Filling→Axis,FillingStyle→{Brown},AspectRatio→0.20];
basedilutionplot2=Plot3D[dilution0+dilution1+dilution2, {x,-0.002,0.006}, {y,-0.002,0.006},PlotRange→{{-0.002,0.006},{-0.002,0.006},{-0.002,2×maxtrackheight}}, Filling→Axis,FillingStyle→{Brown},AspectRatio→0.20];
Show[baselayerplot2,basedilutionplot2];
plotdepheight2j=depheight1/.x→y;
plotdepheight2jj=plotdepheight2j×Powderabsorptionfactor1;
plotdepheight2jjr=If[(k×xcentre-meltw/2)<y<i,plotdepheight2jj,0];
plotdepheight2i=depheight0/.x→y;
plotdepheight2ii=plotdepheight2i×Powderabsorptionfactor1;
plotdepheight2iir=If[y>i,plotdepheight2ii,0];
firstlayerplot2=Plot3D[plotdepheight2+plotdepheight1iir+plotdepheight1jjr+plotdepheight2jjr+plotdepheight2iir+maxtrackheight, {x,-0.002,0.006}, {y, -(meltw/2), meltw/2+k×xcentre}, PlotRange→{{-0.002,0.006},{-0.002,0.006},{-0.002,2×maxtrackheight}}, Filling→maxtrackheight,FillingStyle→{Brown},AspectRatio→0.20];
dilution2=dilution2/.x→y;
firstdilutionplot2=Plot3D[dilution0+dilution1+dilution2+maxtrackheight, {x,-0.002,0.006}, {y, -(meltw/2), meltw/2+k×xcentre}, PlotRange→{{-0.002,0.006},{-0.002,0.006},{-0.002,2×maxtrackheight}}, Filling→maxtrackheight,FillingStyle→{Brown},AspectRatio→0.20];
Show[{baselayerplot2,basemistplot2,firstlayerplot2,firstdilutionplot2}]

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Appendix C

Mathematica code developed for pulsed-wave porous structures deposition model

Multilayer pulsed-wave deposition

absbeampower=1600;
velocity=0.006;
qtotal=0.0001333;
absorptivity=0.15;
halfmaxbeamrad=0.00075/2;
rpowder=0.0035/2;
ambt=293;
conductivity=21.5;
density=8000;
specificheat=500;
meltpoint=1672;
Lm=370000;
Lv=9830000;

diffusivity:=conductivity/(specificheat× density);
dimlessx:= x/halfmaxbeamrad;
dimlessy:= y/halfmaxbeamrad;
dimlessz:= z/halfmaxbeamrad;
σ:=(halfmaxbeamrad ×velocity)/diffusivity;

t1=0.01; t2=0.015; t3=0.02; t4=0.025; t5=0.03; t6=0.035; t7=0.04; t8=0.05;
integrand1:=(absbeampower×absorptivity)/(conductivity×halfmaxbeamrad)×Exp[-{(dimlessy
2+dimlessx2)/2(1+µ2) +dimlessz2/(2µ2)}/(1+µ2)]/.z→0/.y→0/.x→{0,0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.00225,0.0025}
integrand2:=(absbeampower×absorptivity)/(conductivity×halfmaxbeamrad)×Exp[-{(dimlessy
2+dimlessx2)/2(1+µ2) +dimlessz2/(2µ2)}/(1+µ2)]/.x→0/.y→0/.z→{0,0.0001,0.0002,0.0003,0.0004,0.0005,0.0006,0.0007,0.0008,0.0009,0.001}

temperaturesx1=NIntegrate[integrandx1,{µ,0,√2× diffusivity × t1
/halfmaxbeamrad}] +{ambt};

temperaturesx2=NIntegrate[integrandx1,{µ,0,√2× diffusivity × t2
/halfmaxbeamrad}] +{ambt};

temperaturesx3=NIntegrate[integrandx1,{µ,0,√2× diffusivity × t3
/halfmaxbeamrad}] +{ambt};

temperaturesx4=NIntegrate[integrandx1,{µ,0,√2× diffusivity × t4
/halfmaxbeamrad}] +{ambt};

temperaturesx5=NIntegrate[integrandx1,{µ,0,√2× diffusivity × t5
/halfmaxbeamrad}] +{ambt};

temperaturesx6=NIntegrate[integrandx1,{µ,0,√2× diffusivity × t6
/halfmaxbeamrad}] +{ambt};

temperaturesx7=NIntegrate[integrandx1,{µ,0,√2× diffusivity × t7
/halfmaxbeamrad}] +{ambt};
temperaturesx8=NIntegrate[Integrandx1,\{\mu,0,\sqrt{2 \times \text{diffusivity} \times t8} /\text{halfmaxbeamrad}\}]+\{\text{ambt}\};

\text{temperaturesz}=NIntegrate[Integrandx2,\{\mu,0,\sqrt{2 \times \text{diffusivity} \times tpulse} /\text{halfmaxbeamrad}\}]+\{\text{ambt}\};

s1a:=\text{Sort}[\#&/@\text{Cases}[x/.s1,\_\text{Real}],\text{Less}]
s1b:=\text{TakeWhile}[s1a,\#<0&]
s1c:=\text{Complement}[s1a,s1b]/\text{//Sort}
meltradius1=s1c[[1]]

s2a:=\text{Sort}[\#&/@\text{Cases}[x/.s2,\_\text{Real}],\text{Less}]
s2b:=\text{TakeWhile}[s2a,\#<0&]
s2c:=\text{Complement}[s2a,s2b]/\text{//Sort}
meltradius2=s2c[[1]]

s3a:=\text{Sort}[\#&/@\text{Cases}[x/.s3,\_\text{Real}],\text{Less}]
s3b:=\text{TakeWhile}[s3a,\#<0&]
s3c:=\text{Complement}[s3a,s3b]/\text{//Sort}
meltradius3=s3c[[1]]

s4a:=\text{Sort}[\#&/@\text{Cases}[x/.s4,\_\text{Real}],\text{Less}]
s4b:=\text{TakeWhile}[s4a,\#<0&]
s4c:=\text{Complement}[s4a,s4b]/\text{//Sort}
meltradius4=s4c[[1]]

s5a:=\text{Sort}[\#&/@\text{Cases}[x/.s5,\_\text{Real}],\text{Less}]
s5b:=\text{TakeWhile}[s5a,\#<0&]
s5c:=\text{Complement}[s5a,s5b]/\text{//Sort}
meltradius5=s5c[[1]]

s6a:=\text{Sort}[\#&/@\text{Cases}[x/.s6,\_\text{Real}],\text{Less}]
s6b:=\text{TakeWhile}[s6a,\#<0&]
s6c:=\text{Complement}[s6a,s6b]/\text{//Sort}
meltradius6=s6c[[1]]

s7a:=\text{Sort}[\#&/@\text{Cases}[x/.s7,\_\text{Real}],\text{Less}]
s7b:=\text{TakeWhile}[s7a,\#<0&]
s7c:=\text{Complement}[s7a,s7b]/\text{//Sort}
meltradius7=s7c[[1]]

s8a:=\text{Sort}[\#&/@\text{Cases}[x/.s8,\_\text{Real}],\text{Less}]
s8b:=\text{TakeWhile}[s8a,\#<0&]
s8c:=\text{Complement}[s8a,s8b]/\text{//Sort}
meltradius8=s8c[[1]]

zd:=\text{Solve}[\text{Fit}[\text{Partition}[\text{Riffle}[\{0.0,0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.002,0.0025,0.00025\},\text{temperaturesx8}[[1]]]],[2],\{1, x, x^2, x^3, x^4, x^5\}],[x]=\text{meltpoint},x ]

zdepth=\text{Select}\{\text{Sort}[\#&/@\text{Cases}[z/.zd,\_\text{Real}],\text{Less}],\#>0&\}

tradius=\text{Fit}[\text{Partition}[\text{Riffle}[\{\text{meltradius1},\text{meltradius2},\text{meltradius3},\text{meltradius4},\text{meltradius5},\text{meltradius6},\text{meltradius7},\text{meltradius8}\},\{t1,t2,t3,t4,t5,t6,t7,t8\}],[2],\{1,r,r^2,r^3,r^4,r^5\},r]
Appendix C

\[
\text{Plot}[\text{tradius},\{r,0,0.002\}]
\]

\[
\text{powderflux} := (2\times q_{\text{total}})/(\pi \times \text{powder}^2) \ \text{Exp}[2 \ r^2/\text{powder}^2]
\]

\[
\text{beadheight} = 1/\text{density}
\]

\[
\text{beadheightplot} = \text{Plot}[\text{If}[r>0,\text{beadheight},0],\{r,0,0.002\},\text{PlotRange} \to \{0,0.002\},\{0,0.002\},\text{AspectRatio} \to 1]
\]

\[
\text{br} := \text{Solve}[\text{beadheight} = 0, r]
\]

\[
\text{bra} := \text{Sort}[\# &/\@\text{Cases}[r/\text{br}_, \text{_Real}, \text{Less}]]
\]

\[
\text{brb} := \text{TakeWhile}[\text{bra}, \#<0&]
\]

\[
\text{brc} := \text{Complement}[\text{bra}, \text{brb}] /\@\text{Sort}
\]

\[
\text{beadradius} = \text{brc}[1]
\]

\[
\text{bh} = \text{beadheight}/r \to 0
\]

\[
\text{beadarea} := \text{Integrate}[\text{beadheight}, \{r,0,\text{beadradius}\}]
\]

\[
\text{beadheighteq} = (\text{beadarea} \times 4)/(\pi \times \text{beadradius})
\]

\[
\text{beadprofile} := \text{beadheighteq} \sqrt{1 - \frac{x^2}{\text{beadradius}^2}}
\]

\[
\text{plotbeadprofile} = \text{Plot}[\text{beadprofile}, \{x,-\text{beadradius},\text{beadradius}\},\text{PlotRange} \to \{-0.002,0.002\},\{0,0.001\},\text{AspectRatio} \to 0.25,\text{Filling} \to \text{Axis},\text{FillingStyle} \to \text{Red},\text{AxesOrigin} \to \{-0.001,0\}]
\]

\[
\text{RevolutionPlot3D}[\text{beadprofile}, \{x,-\text{beadradius},\text{beadradius}\},\text{ColorFunction} \to \text{Automatic}]
\]
Mathematica code developed for analytical-numerical model of LDMD

Start

ClearAll;

Power available after fibre

\text{fibreLoss}\text{=absbeampower\_fibreefficiency}\_\text{\_\_mod}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text{\_\_array}\_\text
0.0005/x→0/z→\{0.0,0.0025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.00225,0.0025,0.00275,0.003\};

temperaturesx1=NIntegrate[integrandx1,\{\mu,0,20\}\}+\{ambt\};
temperaturesx2=NIntegrate[integrandx2,\{\mu,0,20\}\}+\{ambt\};
temperaturesy1=NIntegrate[integrandy1,\{\mu,0,20\}\}+\{ambt\};
temperaturesy2=NIntegrate[integrandy2,\{\mu,0,20\}\}+\{ambt\};
temperaturesz1=NIntegrate[integrandz1,\{\mu,0,20\}\}+\{ambt\};
temperaturesz2=NIntegrate[integrandz2,\{\mu,0,20\}\}+\{ambt\};
meltmeantempx1:=Mean[TakeWhile[temperaturesx1[[1]],#\faltermeltpoint\}];meltmeantempx2:=Mean[TakeWhile[temperaturesx2[[1]],#\faltermeltpoint\}];
meltmeantempy1:=Mean[TakeWhile[temperaturesy1[[1]],#\faltermeltpoint\}];
meltmeantempy2:=Mean[TakeWhile[temperaturesy2[[1]],#\faltermeltpoint\}];
meltAvgTemp=Mean\{meltmeantempx1,meltmeantempx2,meltmeantempy1,meltmeantempy2\};

s1:=Solve[Fit[Partition[Riffle[{0,0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.00225,0.0025,0.00275,0.003},temperaturesx1[[1]]\}],2,\{1,x,x^2,x^3,x^4\},x\}==meltpoint,x ];
s2:=Solve[Fit[Partition[Riffle[{0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.00225,0.0025,0.00275,0.003},temperaturesx2[[1]]\}],2,\{1,x,x^2,x^3,x^4\},x\}==meltpoint,x ];
s3:=Solve[Fit[Partition[Riffle[{0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.00225,0.0025,0.00275,0.003},temperaturesy1[[1]]\}],2,\{1,x,x^2,x^3,x^4\},x\}==meltpoint,x ];
s4:=Solve[Fit[Partition[Riffle[{0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.00225,0.0025,0.00275,0.003},temperaturesy2[[1]]\}],2,\{1,x,x^2,x^3,x^4\},x\}==meltpoint,x ];
s5:=Solve[Fit[Partition[Riffle[{0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.00225,0.0025,0.00275,0.003},temperaturesz1[[1]]\}],2,\{1,x,x^2,x^3,x^4\},x\}==meltpoint,x ];
s6:=Solve[Fit[Partition[Riffle[{0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.00225,0.0025,0.00275,0.003},temperaturesz2[[1]]\}],2,\{1,x,x^2,x^3,x^4\},x\}==meltpoint,x ];
s1a:=Sort[#\&/@Cases[x/.s1,\_Real\},Less\};
s1b:=TakeWhile[s1a,\#<0\};
halfmeltwidth1:=s1c[[1]]\};
s2a:=Sort[#\&/@Cases[x/.s2,\_Real\},Less\};
s2b:=TakeWhile[s2a,\#<0\};
s2c:=Sort[Complement[s2a,s2b\}];
meltfront1:=s2c[[1]]\};

s3a:=Sort[#\&/@Cases[x/.s3,\_Real\},Less\};
s3b:=TakeWhile[s3a,\#<0\};
s3c:=Sort[Complement[s3a,s3b\}];
meltrear1:=s3c[[1]]\};

s4a:=Sort[#\&/@Cases[x/.s4,\_Real\},Less\};
s4b:=TakeWhile[s4a,\#<0\};
s4c:=Sort[Complement[s4a,s4b\}];
meltd1:=Round\[s4c[[1]]\};

s5a:=Sort[#\&/@Cases[x/.s5,\_Real\},Less\};
s5b:=TakeWhile[s5a,\#<0\};
s5c:=Sort[Complement[s5a,s5b\}];

meltw1=2×Round[halfmeltwidth1,0.0000001\};
meltlf1=Round[meltfront1,0.0000001\};
meltlr1=Round[meltrear1,0.0000001\};
meltd1=Round[meltd1,0.0000001\};

Return[{meltAvgTemp,meltw1,meltlf1,meltlr1,meltd1\}};]
Heat flow model with power losses for calculations of melt pool size and temperature

heatFlowWithLosses[newpower_, velocity_, qtotal_, absorptivity_, halfmaxbeamrad_, rpowder_, ambt_, conductivity_, density_, specificheat_, meltpoint_] := Module[{diffusivity, dimlessx, dimlessy, dimlessz, dimlessx, dimlessy, dimlessz, µ, x, y, z, integrandx1, integrandx2, integrandy1, integrandy2, integrandz1, integrandz2, temperaturesx1, temperaturesy1, temperaturesz1, temperaturesx2, temperaturesy2, temperaturesz2, melttempx1, melttempy1, melttempz1, meltmeanx, meltmeanz, newtempx1, newtempy1, newtempz1, meltdepth1, meltdepth2, meltw, meltlf, meltlr, meltm, meltdepth1, meltdepth2, meltw, meltlf, meltlr, meltm},

diffusivity := conductivity/(specificheat× density);
dimlessx := x/halfmaxbeamrad;
dimlessy := y/halfmaxbeamrad;
dimlessz := z/halfmaxbeamrad;
σ := (halfmaxbeamrad × velocity)/diffusivity;

integrandx1 := (newpower × absorptivity)/(conductivity × halfmaxbeamrad) × 
Exp[-{(2 μ^3)^(1/2) (1+μ^2)/2)]/.
         x → {0, 0.000025, 0.0005, 0.00075, 0.001, 0.00125, 0.0015, 0.00175, 0.002, 0.00225, 0.0025, 0.00275, 0.003};

integrandx2 := (newpower × absorptivity)/(conductivity × halfmaxbeamrad) × 
Exp[-{(2 μ^3)^(1/2) (1+μ^2)/2)]/.
         x → {0, -0.000025, -0.0005, -0.00075, -0.001, -0.00125, -0.0015, -0.00175, -0.002, -0.00225, -0.0025, -0.00275, -0.003};

integrandy1 := (newpower × absorptivity)/(conductivity × halfmaxbeamrad) × 
Exp[-{(2 μ^3)^(1/2) (1+μ^2)/2]}/.
         y → {0, 0.000025, 0.0005, 0.00075, 0.001, 0.00125, 0.0015, 0.00175, 0.002, 0.00225, 0.0025, 0.00275, 0.003};

integrandy2 := (newpower × absorptivity)/(conductivity × halfmaxbeamrad) × 
Exp[-{(2 μ^3)^(1/2) (1+μ^2)/2]}/.
         y → {0, -0.000025, -0.0005, -0.00075, -0.001, -0.00125, -0.0015, -0.00175, -0.002, -0.00225, -0.0025, -0.00275, -0.003};

integrandz1 := (newpower × absorptivity)/(conductivity × halfmaxbeamrad) × 
Exp[-{(2 μ^3)^(1/2) (1+μ^2)/2)}/.
         z → {0, 0.000025, 0.0005, 0.00075, 0.001, 0.00125, 0.0015, 0.00175, 0.002, 0.00225, 0.0025, 0.00275, 0.003};

integrandz2 := (newpower × absorptivity)/(conductivity × halfmaxbeamrad) × 
Exp[-{(2 μ^3)^(1/2) (1+μ^2)/2)}/.
         z → {0, -0.000025, -0.0005, -0.00075, -0.001, -0.00125, -0.0015, -0.00175, -0.002, -0.00225, -0.0025, -0.00275, -0.003};

temperaturesx1 = NIntegrate[integrandsx1, {µ, 0, 20}]} + {ambt};
temperaturesx2 = NIntegrate[integrandsx2, {µ, 0, 20}]} + {ambt};
temperaturesy1 = NIntegrate[integrandsy1, {µ, 0, 20}]} + {ambt};
temperaturesy2 = NIntegrate[integrandsy2, {µ, 0, 20}]} + {ambt};
temperaturesz1 = NIntegrate[integrandsz1, {µ, 0, 20}]} + {ambt};
temperaturesz2 = NIntegrate[integrandsz2, {µ, 0, 20}]} + {ambt};

melttempx1 = Select[temperaturesx1[[1]], meltpoint];
meltmeanx = If[melttempx1 == {}, meltmeanx, Mean[melttempx1]];
melttempx2 = Select[temperaturesx2[[1]], meltpoint];
meltmeanx = If[melttempx2 == {}, meltmeanx, Mean[melttempx2]];
melttempy1 = Select[temperaturesy1[[1]], meltpoint];
meltmeanx = If[melttempy1 == {}, meltmeanx, Mean[melttempy1]];
melttempy2 = Select[temperaturesy2[[1]], meltpoint];
meltmeanx = If[melttempy2 == {}, meltmeanx, Mean[melttempy2]];
meltmeantempy2=If[melttempy2={},meltpoint
 Mean[melttemp2]]; melttempz1=Select[temperaturesz1[[1]],meltpoint
 Flatten[Select[temperaturesz2[[1]],meltpoint
 Flatten[Select[temperaturesy1[[1]],meltpoint
 Flatten[Select[temperaturesy2[[1]],meltpoint
 Flatten[Select[temperaturesz1[[1]],meltpoint
 Flatten[Select[temperaturesz2[[1]],meltpoint
 newmeltAvgTemp=Mean[{meltmeantempx1,meltmeantempx2,meltmeantemy1,meltmeantemy2}];
 newtempx1=If[melttempx1={},ReplacePart[temperaturesx1[[1]],(1→meltpoint+20)],temperaturesx1[
 newtempy1=If[melttempy1={},ReplacePart[temperaturesy1[[1]],(1→meltpoint+20)],temperaturesy1[
 newtempy2=If[melttempy2={},ReplacePart[temperaturesy2[[1]],(1→meltpoint+20)],temperaturesy2[
 newtempz1=If[melttempz1={},ReplacePart[temperaturesz1[[1]],(1→meltpoint+20)],temperaturesz1[
 newtempz2=If[melttempz2={},ReplacePart[temperaturesz2[[1]],(1→meltpoint+20)],temperaturesz2[

 s1:=Solve[Fit[Partition[Riffle[{0,0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.002
 25,0.0025,0.00275,0.003},newtempx1[[1]],2]],{1,x^2,x^3,x^4,x^5},x]=meltpoint,x];
 s2:=Solve[Fit[Partition[Riffle[{0,0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.002
 25,0.0025,0.00275,0.003},newtempy1[[1]],2]],{1,x^2,x^3,x^4,x^5},x]=meltpoint,x];
 s3:=Solve[Fit[Partition[Riffle[{0,0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.002
 25,0.0025,0.00275,0.003},newtempy2[[1]],2]],{1,x^2,x^3,x^4,x^5},x]=meltpoint,x];
 s4:=Solve[Fit[Partition[Riffle[{0,0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.002
 25,0.0025,0.00275,0.003},newtempz1[[1]],2]],{1,x^2,x^3,x^4,x^5},x]=meltpoint,x];
 s5:=Solve[Fit[Partition[Riffle[{0,0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.002
 25,0.0025,0.00275,0.003},newtempz2[[1]],2]],{1,x^2,x^3,x^4,x^5},x]=meltpoint,x];

 s1a:= Sort[#[[1]]&/@Cases[x/.s1,_Real],Less];
 s1b:= TakeWhile[s1a,#[<0&];
 s1c:=Sort[Complement[s1a,s1b]]; meltfront1:=s1c[[1]]; halfmeltwidth1:=s1c[[1]]; s2a:= Sort[#[[1]]&/@Cases[x/.s2,_Real],Less]; s2b:= TakeWhile[s2a,#[<0&]; s2c:=Sort[Complement[s2a,s2b]]; meltrear1:=s2c[[1]]; s3a:= Sort[#[[1]]&/@Cases[x/.s3,_Real],Less]; s3b:= TakeWhile[s3a,#[<0&]; s3c:=Sort[Complement[s3a,s3b]]; meltdepth1:=s3c[[1]]; s4a:= Sort[#[[1]]&/@Cases[x/.s4,_Real],Less]; s4b:= TakeWhile[s4a,#[<0&]; s4c:=Sort[Complement[s4a,s4b]]; meltdepth2:=s4c[[1]]; s5a:= Sort[#[[1]]&/@Cases[x/.s5,_Real],Less]; s5b:= TakeWhile[s5a,#[<0&]; s5c:=Sort[Complement[s5a,s5b]]; meltw=2×Round[halfmeltwidth1,0.0000001]; meltlf=Round[meltfront1,0.0000001]; meltlr=Round[meltrear1,0.0000001]; meltld=Round[meltdepth1,0.00000001]; meltld2=Round[meltdepth2,0.00000001];

 Return[{newmeltAvgTemp,meltw,meltlf,meltlr,meltld}];]
Calculate mean melt pool temperature

calcMeltPoolTemp[qtotal_, rpowder_, powderCatchEff_, density_, meltw1_, meltlf1_, meltlr1_, meltAvgTemp_, meltpoint_,_] := Module[{qp, forwardlimit, rearlimit, meltpoolheight, meltpoolarea, meltUpperVolume, meltBottomVolume, meltmeantemp},
  qp := (2 x qtotal)/(π × rpowder²) Exp[- 2 (x²+y²)/rpowder²];
  forwardlimit := Abs[Re[meltlf1] \[1 - \frac{4 \times x²}{meltw1²} \] ];
  rearlimit := - Abs[Re[meltlr1] \[1 - \frac{4 \times x²}{meltw1²} \] ];
  meltpoolheight := powderCatchEff/(density × velocity);
  meltpoolarea := \[\int_{\text{forwardlimit}}^{\text{rearlimit}} \text{meltpoolheight} \, dy\];
  meltUpperVolume := NIntegrate[meltpoolarea, {x, -(meltw1/2), meltw1/2}];
  meltBottomVolume := 1/2 × (4/3 × π × (meltw1/2)² × (meltlf1/2 + meltlr1/2));
  meltmeantemp := (meltBottomVolume × meltAvgTemp + meltUpperVolume × meltpoint)/(meltBottomVolume + meltUpperVolume);
  Return[meltmeantemp];
]

Calculate deposition profile and power to sustain mass addition

calculateDepositionProfile[qtotal_, rpowder_, powderCatchEff_, density_, velocity_, meltw1_, meltlf1_, meltlr1_, meltpoint_, ambt_, specificheat_, Lm_]:= Module[{qp, forwardlimit, rearlimit, depheight, powerM},
  qp := (2 x qtotal)/(π × rpowder²) Exp[- 2 (x²+y²)/rpowder²];
  forwardlimit := Abs[Re[meltlf1] \[1 - \frac{4 \times x²}{meltw1²} \] ];
  rearlimit := - Abs[Re[meltlr1] \[1 - \frac{4 \times x²}{meltw1²} \] ];
  depheight := powderCatchEff/(density × velocity) \[\int_{\text{forwardlimit}}^{\text{rearlimit}} \text{depheight} \, dy\];
  powerM := velocity × density × (specificheat (meltpoint-ambt) + Lm) × NIntegrate[depheight, {x, -(meltw1/2), meltw1/2}];
  Return[powerM];
]

Calculate evaporation heat losses

calcEvaporationLosses[A_, meltmeantemp_, meltw1_, meltlf1_, meltlr1_, Lw_]:= Module[{mm, evaporationheatflux, surfacearea, evaporationheatfluxloss, massflux},
  massflux := Solve[Log[m] = A × 6.1210 - 18836/meltmeantemp - Log[meltmeantemp]/2, m];
  mm := (455.32 × Exp[4.834 - 18836/meltmeantemp]) / \sqrt{\text{meltmeantemp}};
  evaporationheatflux := mm × Lw;
  surfacearea := Pi/4 × meltw1 × (meltlf1 + meltlr1);
  evaporationheatfluxloss := evaporationheatflux × surfacearea;
  Return[evaporationheatfluxloss];
]

Perform Meltpool Calculation

absbeampower=1000;
velocity=0.005;
qtotal=0.000033;
absorptivity=0.55;
halfmaxbeamrad=0.0017/2;
rpowder=0.0035/2;
Appendix D

powderCatchEff=0.8;
ambt=293;
density=4300;
specificheat=800;
meltpoint=1933;
Lm=370000;
Lv=9830000;
A = 4.834;
fibreefficiency=0.7;

fibrePowerLoss=fibreLoss[absbeampower,fibreefficiency];
beampower=absbeampower-fibrePowerLoss;
{meltAvgTemp,meltw1,meltlf1,meltlr1,meltd1} = initialHeatFlow[beampower,velocity,qtotal,absorptivity,halfmaxbeamrad,rpowder,ambt,conductivity,density,specificheat,meltpoint];
meltmeantemp=calcMeltPoolTemp[qtotal,rpowder,powderCatchEff,density,velocity,meltw1,meltlf1,meltlr1,meltAvgTemp,meltpoint];
Print[ {meltAvgTemp,meltmeantemp,beampower,meltlf1,meltlr1,meltw1,meltd1} ];

powerM = powerToMeltPowder[qtotal,rpowder,powderCatchEff,density,velocity,meltw1,meltlf1,meltlr1,meltpoint,ambt,specificheat, Lm];
evaporationheatfluxloss=calcEvaporationLosses[A,meltmeantemp,meltw1,meltlf1,meltlr1,Lv];
newpower=beampower-(evaporationheatfluxloss+powerM);
{newmeltAvgTemp,meltw,meltlf,meltlr,meltd} = heatFlowWithLosses[newpower,velocity,qtotal,absorptivity,halfmaxbeamrad,rpowder,ambt,conductivity,density,specificheat,meltpoint];
newmeltmeantemp=calcMeltPoolTemp[qtotal,rpowder,powderCatchEff,density,velocity,meltw,meltlf,meltlr ,meltAvgTemp,meltpoint];
Print[ {newmeltAvgTemp,newmeltmeantemp,newpower,meltlf,meltlr,meltw,meltd} ];

{newmeltAvgTemplist,newmeltmeantemplist,meltwlist,meltflist,meltlrlist,meltdlist} = Reap[ While[Abs[meltw1-meltw]>0.00001||Abs[meltlf1-meltlf]>0.00001||Abs[meltlr1-meltlr]>0.00001,
Sow[newmeltAvgTemp,"newmeltAvgTemp"];
Sow[newmeltmeantemp,"newmeltmeantemp"];
Sow[meltw,"meltw"];
Sow[meltlf,"meltlf"]; 
Sow[meltlr,"meltlr"];
Sow[meltld,"meltld"]; ];
meltAvgTemp=newmeltAvgTemp;
meltmeantemp=newmeltmeantemp;
meltw1=meltw;
meltlf1=meltlf;
meltlr1=meltlr;
meltd1=meltld;
powerM = powerToMeltPowder[qtotal,rpowder,powderCatchEff,density,velocity,meltw1,meltlf1,meltlr1,meltpoint,ambt,specificheat, Lm];
evaporationheatfluxloss=calcEvaporationLosses[A,meltmeantemp,meltw1,meltlf1,meltlr1,Lv];
newpower=beampower-(evaporationheatfluxloss+powerM);
{newmeltAvgTemp,meltw,meltlf,meltlr,meltd} = heatFlowWithLosses[newpower,velocity,qtotal,absorptivity,halfmaxbeamrad,rpowder,ambt,conductivity,density,specificheat,meltpoint];
newmeltmeantemp=calcMeltPoolTemp[qtotal,rpowder,powderCatchEff,density,velocity,meltw,meltlf ,meltlr,newmeltAvgTemp,meltpoint];
Print[ {newmeltAvgTemp,newmeltmeantemp,newpower,meltlf,meltlr,meltw,meltd} ];
]
ListLinePlot[newmeltmeantemplist,PlotRange→All]
ListLinePlot[meltwlist,PlotRange→All]
Print[ {newmeltmeantemp,evaporationheatfluxloss,powerM,meltw,meltd} ]
Calculate Single Track Deposition Profile

calcSingleDepProfile[qtotal_,rpowder_,powderCatchEff_,density_,velocity_,meltw_,meltlf_,meltlr_,
meltd_,k_,kld_]:=Module[{qp,forwardlimit,rearlimit,xcentre,depheight0,dilution0},
qp:=(2×qtotal)/(π×rpowder²) Exp[-2 ((x-k×xcentre)²+y²)/rpowder²];
forwardlimit:=Abs[Re[meltlf]/diffdiffd];
rearlimit:=-Abs[Re[meltlr]/diffdiffd];
depheight0:=powderCatchEff/(density×velocity) ;
dilution0:=-Abs[Re[meltd]/diffdiffd];
Return[{depheight0,dilution0}];

Perform Single Track Deposition Calculation

powderCatchEff=0.8;
k=0;
{depheight0,dilution0}=calcSingleDepProfile[qtotal,rpowder,powderCatchEff,density,velocity,meltw,
.meltlf,meltlr,meltd,kld];
trackplot=Plot[depheight0,{x,-0.002,0.004},PlotRange→{{-0.002,0.004},{-0.002,0.002}}];
dilutionplot=Plot[dilution0,{x,-0.002,0.004},PlotRange→{{-0.002,0.004},{-0.002,0.002}}];
Show[dilutionplot,trackplot];

Temperature distributions and cooling rates calculations

temperatureFunction[newpower_,velocity_,absorptivity_,halfmaxbeamrad_,ambt_,conductivity_,density_,specificheat_,
]=Module[{diffdiffd,diffdiffd,diffdiffd,diffdiffd,diffdiffd},
diffusivity:=conductivity/(specificheat× density);
dimlessxx:= x/halfmaxbeamrad;
dimlessyy:= y/halfmaxbeamrad;
dimlesszz:= z/halfmaxbeamrad;
σ:=(halfmaxbeamrad ×velocity)/diffusivity;

integrand1:=(newpower×absorptivity)/(conductivity×halfmaxbeamrad)×Exp[-((dimlessy^2+dimlessx^2)/(2 (1+µ²)+dimlessz^2/(2/µ²)))/((2 π²)½ (1+µ²))/z→0
/./y→0,x→0,0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.00225,0.0025,0.0
0275,0.003];

integrand2:=(newpower×absorptivity)/(conductivity×halfmaxbeamrad)×Exp[-((dimlessy^2+dimlessx^2)/(2 (1+µ²)+dimlessz^2/(2/µ²)))/((2 π²)½ (1+µ²))/z→0
/./x→0,y→0,0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.00225,0.0025,0.00275,0.0
0275,0.003];
\[
\text{Appendix D}
\]

\[
/ x \rightarrow 0 / . z \rightarrow \{0.0,0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.00225,0.0025,0.00275,0.003\};
\]

\[
\text{integrandz2}:=(\text{newpower} \times \text{absorptivity})/(\text{conductivity} \times \text{halfmaxbeamrad}) \times \text{Exp}[-\{(\text{dimlessy} + \sigma/2
\mu^2)^2 + \text{dimlessx}^2)(2(1+\mu/2) + \text{dimlessz}^2/(2\mu/2))/((2 \pi^3)^{1/2}(1+\mu/2))/y \rightarrow
0.0005/\text{x} \rightarrow 0 / . \text{z} \rightarrow \{0.0,0.00025,0.0005,0.00075,0.001,0.00125,0.0015,0.00175,0.002,0.00225,0.0025,0.00275,0.003\};
\]

\[
\text{temperaturesx1}=\text{NIntegrate}[\text{integrandx1},\{\mu,0,20\}]+\{\text{ambt}\};
\]

\[
\text{temperaturesx2}=\text{NIntegrate}[\text{integrandx2},\{\mu,0,20\}]+\{\text{ambt}\};
\]

\[
\text{temperaturesy1}=\text{NIntegrate}[\text{integrandy1},\{\mu,0,20\}]+\{\text{ambt}\};
\]

\[
\text{temperaturesy2}=\text{NIntegrate}[\text{integrandy2},\{\mu,0,20\}]+\{\text{ambt}\};
\]

\[
\text{temperaturesz1}=\text{NIntegrate}[\text{integrandz1},\{\mu,0,20\}]+\{\text{ambt}\};
\]

\[
\text{temperaturesz2}=\text{NIntegrate}[\text{integrandz2},\{\mu,0,20\}]+\{\text{ambt}\};
\]

\[
\text{Return}[\{\text{temperaturesx1},\text{temperaturesx2},\text{temperaturesy1},\text{temperaturesy2},\text{temperaturesz1},\text{temperaturesz2}\}];
\]

\[
\text{temperaturesx1}[[1,8]]\]

\[
\text{ListLinePlot}[\text{temperaturesx2},\text{PlotRange}\rightarrow\text{Full}]
\]

\[
\text{riffleTempDist}=\text{Riffle}[\{0.0,-0.00025,-0.0005,-0.00075,-0.001,-0.00125,-0.0015,-0.00175,-0.002,-0.00225,-0.0025,-0.00275,-0.003\},\text{temperaturesy2}[[1]]];
\]

\[
\text{partitionTempDist}=\text{Partition}[\text{riffleTempDist},2];
\]

\[
\text{polynomTempDist}=\text{Fit}[\text{partitionTempDist},\{1,\text{x},\text{x}^2,\text{x}^3,\text{x}^4,\text{x}^5\},\text{x}];
\]

\[
\text{tempGrad}=\text{D}[\text{polynomTempDist},\text{x}];
\]

\[
\text{Plot}[\text{tempGrad},\{\text{x},0,-\text{meltlr}\}];
\]

\[
\text{thermalGrad}=\text{tempGrad}/1000;
\]

\[
\text{coolingRate}=\text{thermalGrad}\times\text{velocity}
\]