Laser metal deposition materials-process-performance characterization studies

Tan, Eddie Zhi'En

2021

Tan, E. Z. (2021). Laser metal deposition materials-process-performance characterization studies. Doctoral thesis, Nanyang Technological University, Singapore.
<https://hdl.handle.net/10356/152892>

<https://hdl.handle.net/10356/152892>

<https://doi.org/10.32657/10356/152892>

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LASER METAL DEPOSITION MATERIALS-PROCESS-PERFORMANCE CHARACTERIZATION STUDIES

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Eddie Tan Zhi'En

School of Mechanical and Aerospace Engineering

In collaboration with the Advanced Remanufacturing Technology Centre

LASER METAL DEPOSITION
MATERIALS-PROCESS-PERFORMANCE
CHARACTERIZATION STUDIES

Eddie Tan Zhi'En

Eddie Tan Zhi'En 2021

Supervisor: Prof. John Pang Hock Lye (NTU, MAE)

Co-Supervisor: Dr. Jacek Kaminski (A*STAR, ARTC)

Thesis submitted to the Nanyang Technological University

in partial fulfilment of the requirements for the degree of

Doctor of Philosophy

Statement of Originality

I hereby certify that the work embodied in this thesis is the result of original research, is free of plagiarised materials, and has not been submitted for a higher degree to any other University or Institution.

04 Feb 2021

Date

Eddie Tan Zhi'En

Supervisor Declaration Statement

I have reviewed the content and presentation style of this thesis and declare it is free of plagiarism and of sufficient grammatical clarity to be examined. To the best of my knowledge, the research and writing are those of the candidate except as acknowledged in the Author Attribution Statement. I confirm that the investigations were conducted in accord with the ethics policies and integrity standards of Nanyang Technological University and that the research data are presented honestly and without prejudice.

26 Mar 2021

Date

John Pang Hock Lye

Authorship Attribution Statement

The thesis contains material from 2 paper(s) published in the following peer-reviewed journal(s), 1 paper(s) accepted at conferences, and 1 international patent(s) in which I am listed as an author.

Chapter 3 is published as Tan, E. Z. E., Pang, J. H. L., Kaminski, J., & Pepin, H. (2019). Characterisation of porosity, density, and microstructure of directed energy deposited stainless steel AISI 316L. Additive Manufacturing, 25, 286-296, <https://doi.org/10.1016/j.addma.2018.11.014>.

Chapter 4 is published as Tan, E. Z. E., Pang, J. H. L., & Kaminski, J. (2018). Laser metal deposition in-situ process control for different build strategies. In 3rd International Conference on Progress in Additive Manufacturing (Pro-AM 2018), 291-296 <https://doi.org/10.25341/D4JS3T> and Tan, E. Z. E., Pang, J. H. L., & Kaminski, J. (2021). Directed energy deposition build process control effects on microstructure and tensile failure behaviour. In Journal of Materials Processing Technology, 294, 117139, <https://doi.org/10.1016/j.jmatprotec.2021.117139>.

Chapter 5, 6, and 7 are published as an international patent WO2020167249A1. 2020. Method of fabricating an interfacial structure and a fabricated interfacial structure.

The contributions of the co-authors are as follows:

- Prof John pang Hock Lye and Dr Jacek Kaminski provided the initial project direction and oversight and edited the manuscript drafts.
- I performed all laboratory work and analyses at the Advanced Remanufacturing Technology Centre.
- I drafted the manuscript and made all amendments based on the advice of Prof John Pang Hock Lye and Dr Jacek Kaminski.

04 Feb 2021

Date

Eddie Tan Zhi'En

Acknowledgements

I would like to offer my sincerest gratitude to my supervisor, Prof. John Pang Hock Lye, and co-supervisor, Dr. Jacek Kaminski for their advice, guidance, and mentorship. This research would not be possible without their relentless support and knowledge. They have created a highly conducive and vibrant environment to discuss new ideas, to learn precise analytical methods, and to learn how to be a professional researcher. It has been an incredible experience and privilege to learn under their supervision.

I would also like to thank the technical staff, scientists, and engineers at A*STAR, ARTC, as well as the laboratory staff at NTU, MAE for their assistance in providing the necessary equipment training, conducting various experiments, and analyses.

I would like to acknowledge NTU, MAE and A*STAR, ARTC for awarding the NTU-ARTC scholarship for this PhD study and providing an opportunity for me to contribute to ARTC's industry projects, that has served as a crucial training and exposure to industry research.

Abstract

This research addresses the development of aspects regarding materials-process-performance characterization of laser metal deposition process as well as the development of a novel joining method using AM techniques. This work presents an establishment of a systematic materials-process-performance framework in characterising LMD in the form of a series of studies.

A study of the porosity, defect, and microstructure of LMD built Stainless Steel AISI 316L specimens (SS316L) was performed. Defects were measured and analysed using optical microscopy techniques. Types of defects were identified using metrological quantities of circularity and roundness. The microstructure of the LMD parts were also analysed using SEM/EDX. Calculations were made based on the EDX mapped elemental compositions using Schaeffler diagram in order to identify the solidification mode based on the quantity of δ -ferrite present in the specimen. This study provides insight into the fundamentals of solidification characteristics and the effects of process parameters on defect formations. This study was published in the Additive Manufacturing journal: Tan, E. Z. E., Pang, J. H. L., Kaminski, J., & Pepin, H. (2019). Characterisation of porosity, density, and microstructure of directed energy deposited stainless steel AISI 316L. *Additive Manufacturing*, 25, 286-296,

A detailed study was done to investigate the effects of build strategy on LMD Stainless Steel 316L specimens and provide a framework for characterizing the anisotropic properties and behaviour arising from build strategy pattern effects. A series of tensile tests were used to characterise the mechanical anisotropy of LMD specimens that were produced using long unidirectional, bidirectional, and short unidirectional raster scan build strategies. Differences in the thermal history between the three build strategy raster scan patterns resulted in differences in tensile mechanical properties, microstructural grain sizes, fracture surface topology, and hardness properties. This indicates that there is an anisotropic property that arises from different build strategies used in AM-built specimens. This study was accepted as a publication in the Journal of Materials Processing Technology: Tan, E. Z. E., Pang, J. H. L., & Kaminski, J. (2021). Directed energy deposition build process control effects on microstructure and tensile failure behaviour. In *Journal of Materials Processing Technology*, 294, 117139, and

was presented at the 3rd International Conference on Progress in Additive Manufacturing 2018: Chapter 4 is published as Tan, E. Z. E., Pang, J. H. L., & Kaminski, J. (2018). Laser metal deposition in-situ process control for different build strategies. In 3rd International Conference on Progress in Additive Manufacturing (Pro-AM 2018), 291-296

A study on the fabrication of novel interfacial joint designs for select design parameters was conducted. Two levels for three design parameters: Step angle, step height, and feature geometry were selected to understand their effects on the build quality. Unique forms of defects were found based on the interfacial design parameter combination. By setting an optimisation goal of lowest degree of large defects formed, the most effective combination of interfacial joint design would be to lower the Step Height and Step Angle. On top of this, tighter tolerances can be made for the step joint features in order to ensure that more precise initiation point can be made for the LMD process that produces a more consistent process and a higher quality build.

The study investigates two key process variables: step angle and feature geometry, were used to fabricate Stainless Steel AISI 316L specimens. Charpy impact samples were extracted from these specimens such that the notch is located at the interfacial joint. An instrumented Charpy impact test was done to evaluate the impact strength of the interfacial joint designs. The Charpy impact test results show that the V-shaped and stepped interfacial joint designs produce a significant improvement in the mechanical properties of the interfacial joints. This and the previous mentioned study are published as an international patent: “Method of fabricating an interfacial structure and a fabricated interfacial structure” with WO2020167249A1 designation.

A numerical study using finite element modelling was done using ABAQUS software to simulate the Charpy impact testing of specimens designed to improve the interfacial strength between a substrate and an LMD built part. Parametric input for material properties was taken from the tensile test studies, interfacial designs were taken from the previous study with variation in step angle and feature geometry. The results for the FEM show a similar trend of impact strength per interfacial joint design to the instrumented empirical Charpy test results, providing a tool for future prediction models for future work other forms of interfacial designs.

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Chapter 1: Introduction

1.1 Overview

Additive manufacturing (AM) enables the manufacture of metallic parts from metallic powder. Laser metal deposition (LMD) is one method of Directed energy deposition (DED) technology that utilises a laser to form near net shape functional metallic components. It is a technique that has garnered considerable interest from various industries.

Addressing the material-process-performance (MPP) relationship is complex due to the interdependency of the LMD parameters. Process characteristics like laser source, melt pool characteristics, powder delivery system, build pattern are all key contributors to the material-mechanical behaviour of the LMD component. Publications characterising either the materials, process, or performance domains, an insufficient are needed to push AM fabricated parts toward testing and certification.

Structural optimisation of LMD-built functional components is vital area of interest for the AM and remanufacturing. One essential engineering feature in functional structures are joints or interfaces that connect two or more bodies. This basic feature can be found in numerous applications in the engineering field like rotor blades from the aerospace industry and exhaust manifolds from the automotive industry. These applications require an extremely high quality forms of joints or interface in order to withstand the loads experienced in their respective environments. DED techniques like LMD differ from other AM techniques like selective laser melting (SLM) and electron beam melting (EBM) in its capability to easily add or modify features on existing components. Its compatibility with hybrid manufacturing, in the form of subtractive and additive capabilities, hasten the speed of the art-to-part process that is identified as an important industry enabling feature.

Although LMD possess these unique traits that enable the application of feature adding or modifying, the interface which has areas of weaknesses that requires further investigation. The existing method of layer by layer deposition

and fusion joining using LMD techniques is to deposit metallic material over a flat substrate. Although this is an efficient way of adding or modifying features, further improvements to the strength of the interfacial joint is needed to enable the adoption of the technique. This study explores the fundamentals of the MPP relationship in LMD processes as well as methods to enhance the interfacial strength between the LMD build-up and interface with the substrate.

1.2 Objectives

The objectives of the research work include the following:

1. Characterize the material and defect characteristics for LMD of AISI 316L stainless steel. This involves (a) defect and porosity characterization using optical microscopy and x-ray tomography (b) material characterization techniques like SEM/EDX analyses and optical microscopy, Schaeffler diagram methods, solidification mode analyses, and (c) hardness properties using Vickers microhardness indentation techniques and Hall-Petch analysis.
2. Characterize the material, process, and performance parameters for LMD of AISI 316L stainless steel and investigate the effects of different build strategies on each characteristic (a) defect and porosity characterization using techniques like optical microscopy, (b) process analyses techniques like powder efficiency calculation, process monitoring, process mapping, (c) material characterization techniques like SEM/EDX analyses, optical microscopy, grain size measurement, Schaeffler diagram methods, and solidification mode analyses, (d) mechanical property characterization techniques using hardness measurement techniques like Vickers microhardness indentation, Berkovich tip microhardness indentation, Hall-Petch analysis, and instrumented tensile tests.
3. To develop a methodology for a novel interfacial joining technique for LMD that yields an improvement to mechanical strength at the interface. This section encompasses studies that investigate the optimisation and performance characteristics of the novel technique. This includes a preliminary study exploring the feasibility of several variations of the interfacial joint design (a) by characterizing the defects and porosity that arise from different designs using optical microscopy techniques, ultrasound measurement, and x-

ray tomography, and (b) measuring the dilution and heat affected zone using optical microscopy and hardness measurement techniques, and (c) using statistical methods to evaluate the effects of each parameter on amount of defects formed in LMD part.

4. To investigate the performance differences among interfacial joint designs by (a) analysing the melt pool size and melt pool temperature process monitoring data via process mapping, (b) measuring the amount of defects using optical microscopy techniques, and (c) measuring the impact toughness using an instrumented Charpy impact test and analysing the fracture surface topology using optical microscopy techniques.
5. To model the novel interfacial joint designs and the conventional flat interfacial joint design and investigate their differences in the simulation results. Use ABAQUS to model the interfacial joint designs explored in the previous study, using parameters taken from the tensile test study. The simulation covers the model of the novel interfacial design and its performance characteristics under an impact load. The results from the model will serve as a theoretical verification of the results obtained from the empirical interfacial joint design study and serve as a prediction tool for future iterations and modifications to the interfacial joint design.

1.3 Scope

This thesis reports on the Materials-Process-Performance (MPP) characterization approach for LMD AISI Stainless Steel 316L. AISI Stainless Steel 316L was selected for this thesis because of its excellent combination of toughness, ductility and strength as well as its relevance in numerous manufacturing applications such as marine engineering, potable water systems, food preparation equipment, pharmaceutical manufacturing and medical implants due to its exceptional corrosion and oxidation resistance, high work hardening, and good formability [1].

The research scope includes conducting laser metal deposition experiments and obtaining MPP results to establish crucial parametric input to be used in the next phase of work. A fundamental investigation into the LMD's build strategy effect on MPP properties were done using process monitoring techniques, defect analysis, microstructural

analysis, hardness measurements, and tensile testing. The result from this section serves as crucial parametric input for the next section where a novel interfacial joint technique is developed and simulated in finite element modelling. An overview of the scope and structure of the PhD work is illustrated in Figure 1.1.

In the novel interfacial joint design study, several interfacial designs were compared against the conventional flat interfacial joint design using process and performance characterization techniques. An in-depth look into defects that arise from the unique interfacing technique and melt pool data from the process monitoring analysis was presented alongside an instrumented Charpy impact test. The Charpy test was conducted such that the anvil was struck at the interface location between the substrate and the LMD build-up to evaluate and compare the mechanical strength of each interfacial joint design type. The experimental specifications used in this study and the results from the previous MPP studies were subsequently fed to the finite element model of the novel interfacial joint designs. To simulate the impact test, the Johnson-Cook plasticity model was used, and the simulation results were examined comparatively between each interfacial joint design type.

References [1-179] is cited within the main body of the thesis and references [180-228] are cited in the Appendix. The Appendix includes additional background information and review about AM and LMD technologies which is separated from the main body of the thesis for easier reading.

Figure 1.1: Flowchart of PhD scope

1.4 Outline of the thesis

The thesis is organised into eight chapters. Chapter 1 provides an introduction, broad overview, and the motivation for the research. Chapter 2 provides a literature review of additive manufacturing, laser metal deposition, details about mass transfer and heat transfer at the melt pool, laser processing parameters, characteristics of LMD parts like porosities, microstructure, and mechanical properties, as well as definitions and methods of hybrid manufacturing. Chapter 3 covers the first LMD study where a fundamental investigation into the defect and microstructure of LMD built parts are examined and characterized. LMD build was evaluated by measuring the consistency of the deposition track width and heights, measuring the part density and porosities using microscopy techniques, and examining the microstructure, composition, and morphology using SEM and EDX techniques. Chapter 4 covers an investigation into build strategy's effect on MPP properties and study any form of anisotropy in its results. In this chapter, three types of deposition raster scan patterns were explored, and several MPP measurements were conducted to evaluate its properties. The LMD process was evaluated by analysing its process monitoring results, measuring the powder efficiency, part density, conducting an instrumented tensile test on several specimens built using different build strategies and evaluating the mechanical anisotropy, examining the microstructure and fracture surface topologies, and verifying the results using micro-indentation techniques. Chapter 5 covers an investigation into a novel interfacial joining method and measure its effect on the mechanical strength of the joint. Defects were examined using ultrasound, x-ray tomography, and microscopy techniques, and the dilution and heat affected zone were analysed, and based on the measurements, an optimal set of design parameters were identified. Chapter 6 covers a comparative study into the optimal design from the previous study, the conventional flat interfacial joint design, and an alternative interfacial joint design type. In this study, the process monitoring data was analysed, defects were measured using microscopy techniques, and an instrumented Charpy test was conducted to measure and evaluate the mechanical strength of each interfacial joint type. Chapter 7 covers the finite element modelling of the Charpy test of the three interfacial joint design types using a Johnson-Cook plasticity model. The design specifications from Chapter 6 were fed to the model and parametric input for the model were taken from Chapter 2 and 3. Chapter 8 provides an overall conclusion and recommendation for future work.

Chapter 2: Literature Review

2.1 Additive Manufacturing

Additive Manufacturing (AM) is defined by ISO/ASTM as “a process of joining materials to make objects from 3D model data, usually layer upon layer, as opposed to subtractive manufacturing methodologies” [2]. AM is an emerging technology that has the potential to be a disruptive and revolutionary manufacturing technique. Its ability to take computer aided drawings (CAD) and data to manufacture complex structures has garnered much attention from the industry as a means for direct manufacture of complex metallic parts at low volume.

Today’s AM systems have the ability to produce functional parts in small quantities. Typically, parts produced have a dimensional accuracy and surface finishing inferior to those produced by conventional manufacturing methods. When coupled with the suitable post-processing techniques, AM can achieve material characteristics and properties that are comparable and alike to conventional manufacturing methods.

Research on Material-Process-Performance (MPP) characterization of AM fabricated parts are needed for testing and subsequently certification for its fitness for purpose in service. The MPP characteristics of an AM built part depends on the LMD parameters considered at the materials, process, and performance evaluation phases of the research. Process characteristics like laser source, melt pool characteristics, powder delivery system, build pattern are all key contributors to the material-mechanical behaviour of the LMD component. Despite the various publications characterising either the materials, process, or performance domains, an insufficient number of published works utilises standard samples and methods.

The fundamental working principle behind most AM techniques is essentially a layer-by-layer fabrication method. The AM process typically starts with a digital design of the component via CAD software. This design is subsequently converted into a stereolithography file format (STL), whereby the component surfaces are tessellated into triangles, and a database of the triangular nodes are stored. Next, layers comprising of the deposition path per layer are generated from the database. These settings are defined by the user requirements of the AM process. After a layer is deposited, the next layer is fabricated accordingly according to the deposition scan implementation

and is repeated until the entire component is built. Today, there are numerous in-depth studies on laser based additive manufacturing techniques [3-7]

Figure 2.1: Metal Additive Manufacturing Classification

There are three dominant metal AM techniques, namely, Selective Laser Melting (SLM), Electron Beam Melting (EBM), and Laser Metal Deposition (LMD). The SLM and EBM techniques are explored further in detail in Appendix A.

2.2 Laser Metal Deposition

LMD is one variant of AM technology that has numerous industry applications, including: application of coatings, repair, and 3D build ups [8, 9]. LMD uses a combination of technologies: CAD, computer-aided manufacturing (CAM), lasers, powder metallurgy, and sensors, in its process. LMD is considered beneficial in medium-to-large sized, small batch manufacturing by employing a near net shape manufacturing process. In contrast to SLM, which employs a powder bed method whereby a laser selectively melts a bed of metal powder feed, LMD directly feeds

the metal powder material to a target location while using a focused laser beam to melt the powder feed. In LMD, a Computer Numerical Control (CNC) system is used to control the laser beam path from the CAM model input to manufacture a three-dimensional component build layer-by-layer. Control of parameter input such as: laser spot size, laser power, scanning velocity and powder feed rate, are crucial in the control of the build's dimensional geometry and properties [10]. A schematic of the LMD process for its coaxial nozzle and system set-up is shown in Figure 2.2. The materials-process-performance LMD map is shown in Figure 2.3.

Figure 2.2: Laser metal deposition schematic coaxial nozzle

Figure 2.3: Laser metal deposition materials-process-performance map

The LMD process deposits metal over a solid substrate like an existing part, and hence does not encounter issues with the molten metal flowing through a powder bed. This enables the use of most of the commercially available metallic powders to be used as the LMD powder feed. Various types of metal powder blends can be used to create multiple types of graded materials and alloys [11].

LMD has been developed and explored by numerous companies and research groups and is hence known by many different names. LMD is also known as Laser Cladding [12, 13], Laser Direct Casting (LDC) [14], Direct Metal Deposition (DMD) by the University of Michigan [15], Directed Light Fabrication (DLF) by the Los Alamos national Laboratories [16], Laser Engineered Net Shaping (LENS) by Optomec [17]. Other known alternative

names for LMD include laser powder deposition [18], direct laser powder deposition [19], and direct laser casting [20].

Other areas of interest for LMD include the repair and remanufacturing of components that have undergone severe wear, in the form of laser cladding. LMD has significant advantages over traditional repair techniques like welding or thermal spraying, due to its smaller heat affected zone, less distortions, strong bonding between metallic bodies, and ease of process automation [21, 22]. Various industries have incorporated laser cladding into their process line for the repair and remanufacturing purposes of high value components. Some examples of this include the repair of turbine blades by the aerospace industry [23], the refurbishment of drilling rods by the marine offshore industry, repair and modification of injection mould tooling by the tooling industry, and coating of valves by the automotive industry [22]. The process has been demonstrated to be able to produce numerous types of pure metals, alloys, and functionally graded materials (FGMs).

The LMD process is a highly complex process due to the numerous interactions between the various input parameters like the laser beam, powder flow rate, and scanning strategies. Numerous attempts have been made to optimize the LMD process and identify a suitable process window. Despite these attempts, complex and larger geometries introduce further difficulties and uncertainties in the process quality. Hence, further studies have been done to integrate forms of in-situ control measures for surface temperature, particle velocity, melt pool temperatures monitoring by photodiodes and pyrometers, melt pool height triangulations, and thermal imaging, into the LMD process, and various implementation types of controllers like laser power controls, tool path controls, proportional-integral-derivation (PID) controls, fuzzy logic controls, motion controls, and powder flow rate controls. A detailed background overview of other types of AM technologies like electron beam melting (EBM) and selective laser melting (SLM) can be found in Appendix A.

LMD is a process in which metal material is deposited layer by layer using a high-power laser beam. The laser beam is focused over the surface of a substrate as deposition material is fed to the target location, creating a melt pool. A detailed review of laser fundamentals can be found Appendix B. A stream of inert gas is directed at the melt pool to protect the deposition zone from oxidation effects. The deposition head moves across the substrate with accordance to a user-defined deposition scan path by motion systems like CNC or robot arms. The deposited

material solidifies rapidly and fusion bonds to the previous layer, enabling the manufacture of fully dense geometries like multiple tracks, walls, or bulky components [15, 24, 25]. A compilation of common materials used in LMD processes is detailed in Appendix C.

2.2.1 Mass Transfer at Melt Pool

A powder feed system typically comprises of a powder feeder, a powder supply subsystem, powder delivery tubes, and a nozzle. Methods of mass transfer for LMD includes powder feed and wire feed techniques [26]. The two techniques are fundamentally different in the manner in which the material is fed to the melt pool. The differences between these two LMD material supply techniques are summarised in Table 2.1 [27-32].

Table 2.1: Powder feed and wire feed comparison

Parameter	Powder feed	Wire feed
Feed rate control	Flexible feed rate control Wide range of feed rates possible	Limited feed rate control
Build Rate	High build rates not possible due to powder instability	Higher build rates possible due to stability of wire feed
Part complexity	Complex and intricate geometries possible – free-form, flat, and curved surfaces.	Complex and intricate geometries possible, suitable for rotationally symmetric geometries
Energy utilization	Efficient power utilization. Preheated powder feed possible.	Higher power supply required
Deposition track control	Relatively better control over dilution and surface quality Clad dimension control flexible	Layer thickness control more inflexible Coarser part details
Material utilization efficiency	Between 30% to 80% depending on coaxial or lateral nozzle configuration.	100% material utilization.

There are two types of nozzle designs to deliver powder feed to the melt pool, a lateral nozzle, a series of radially aligned nozzles, and a coaxial nozzle. The differences between the coaxial and lateral nozzle types are summarised in Table 2.2 [30, 33-38].

Table 2.2: Coaxial and lateral nozzle comparison

Parameter	Coaxial	Lateral
Positioning accuracy	Concentric powder flow provides accurate positioning	One-side powder flow difficult to align with laser beam
Powder utilization efficiency	Up to 80% Independent of traverse speed	Up to 30% - 40% Influenced by traversing speed
Part complexity	Suitable for simple and geometric parts – Flat, curved surfaces Bidirectional motion allows free-form fabrication	Suitable for linear motions and simple part geometries
Oxidation effects	Allow effective implementation of shielding gas Deposited metal protected from most of the oxidation effects	Ineffective implementation of shielding gas Deposited metal not protected from oxidation effects
Part quality	Smooth with uniform and consistent deposition tracks Minimal porosity and control dilution	Rough deposition tracks

The coaxial nozzle delivers metallic powder to the melt pool concentrically. A major advantage of the coaxial nozzle design is that the powder delivery is independent from the deposition scan motion, ensuring a continuous stream of powder feed to the melt pool. Hence, the coaxial nozzle design allows for the deposition pathing to be independent of orientation, and is omnidirectional in terms of deposition pathing and has a stable track formation. [39]. Powder efficiencies measured for coaxial nozzle design has been shown to be significantly lower than the lateral nozzle design [40]. The fundamental design is essentially a configuration of two cones, with one cone positioned over the other. The cones are concentrically aligned with the laser beam.

The coaxial nozzle design features three passages, namely, the central passage, the powder passage, and the secondary external gas passage. The laser beam is directed through the central passage. The powder flow is directed through the powder passage, which is a space between the two cones, that forms a coaxial stream of powder. The secondary external gas passage directs additional shielding gas to the melt pool, to provide additional shielding from the atmosphere [41]. Portions of the coaxial nozzle most susceptible to energy reflection from the melt pool are typically designed to be made of copper due to the material's high reflectivity. This ensures that the powder particles do not bond to the coaxial nozzle using the reflected energy from the melt pool, which can cause nozzle obstructions by particle clogging [42].

LMD involves a complex non-equilibrium physical and metallurgical process that embodies multiple consecutive heat and mass transfer modes [43-46]. LMD is achieved by directing a stream of metal powder particles to the melt pool location. Maintaining a continuous and stable powder flow is crucial in producing a quality LMD build. The stream of powder converges at a distance from the coaxial nozzle. The delivery of the carrier gases and nozzle geometry play a key role in the powder flow. For instance, the delivery of shielding gases from the central passage has been shown to produce a spreading effect, whereas shielding gases that are delivered via the other passages have been shown to have an effect on the jet impingement toward the substrate [47, 48].

The inner and outer conical geometry of the coaxial nozzle have also been shown to be essential in controlling the characteristics of the powder stream. An outward and inward nozzle tip design was proposed as shown in Figure 2.4. In the case of an outward nozzle tip design, the powder stream exhibited a converging consolidating effect further from the nozzle tip than that of the inward nozzle tip design. This is due to the difference in the size of the effective powder passage. A smaller powder passage accelerates the carrier gas that subsequently improves the powder stream consolidation [49]. Other studies have also shown that the width of the outer diameter of the powder outlet passage have a stronger effect on the powder stream structure than the inner diameter [50].

Figure 2.4: Nozzle tip configurations (a) inward (b) onward [38]

In LMD, mass transfer to the melt pool location is influenced by several process parameters such as, powder deposition rate, scanning velocity, laser spot size, and laser power. Studies have shown that the deposition track's width, height, and dilution have a linear correlation with the laser power input [51-53]. Other studies have reported

that increasing the scanning velocity has a reduction in deposition track height and width [52, 54, 55]. Furthermore, studies have shown that the beam divergence and standoff distance, which in turns affects the laser spot size and specific energy, has a direct effect on the deposition track width [54, 56]. The powder flow rate has also been reported to have a direct linear relationship with the deposition track's height [57-59]. Hu et al. also demonstrated that a small deviation in powder flow rate will lead to a significant variation in the clad geometry. A notable contribution in Hu's study at the time of publication was the proposal of an improved powder feeding system design where a precision fluidized-bed powder feeder, a powder flow splitter, and a coaxial nozzle as used, which is a crucial design implementation used in many DED systems today [60]. At higher deposition rates, the track height increment plateaus out due to a powder clouding effect [61, 62].

A combination of the laser power, scanning velocity, and powder flow rate parameters can also have a collective effect on the mass transfer at the melt pool [52, 63]. The number of permutations of combinations for the LMD parameters are numerous and result in a complex definition of the process input to achieve a desired build. Studies have explored ideal process windows for LMD processes as a function of laser power spot diameter and powder feed, and their effect on dilution and porosities [64].

The following equation is used to calculate the total mass of powder used during the LMD:

$$m_{total} = \frac{v_{powder} t}{\rho} \quad (2.1)$$

Where m is the total mass of powder used in the process, v_{powder} is the powder deposition rate, t is the deposition duration, and ρ is the density of the powder feed material. Subsequently, to calculate the powder efficiency, the following equation is used:

$$\eta_{powder} = \frac{m_{deposited}}{m_{total}} \times 100\% \quad (2.2)$$

Where η_{powder} is the powder efficiency and $m_{deposited}$ is the mass of powder that is deposited at the substrate.

2.2.2 Heat Transfer at Melt Pool

During the LMD process, a melt pool is formed due to the laser's irradiation interaction with the deposited powder. The melt pool size is a crucial characteristic that determines the quality of the build process. There are numerous factors that affect the dimensions of the melt pool, which in turn influences the mechanical and material properties. Some of these factors include the laser and substrate interaction, melt pool and powder interaction, powder delivery, powder and shielding gas force effects on melt pool, Marangoni and thermos(s)capillary flow, and thermal conduction [65-68]. The effects of the LMD process parameters have been widely studied empirically and by modelling. Liu et al. found that different raster scan types produce differences in the grain growth structures in Inconel 718 [69]. Although Liu et al. provided a comprehensive look into the microstructural grain growth of Inconel 718, little is actually known about the thermal history during the production of the samples. Hofmeister et al. fills the information gap of the thermal history of the DED process by using an in-situ high-speed thermal imaging sensor to monitor the deposition process and correlating the in-situ results to the mechanical and material property results [70].

The Marangoni convection effect is a surface tension driven convection that circulates the melt pool. The surface tension and temperature are inversely proportional, whereby the melt pool fluid flows toward and along the surface from the centre when the temperature is high and surface tension is low, and towards the edge where the temperature is low and surface tension is high [71-75]. The phenomenon is governed by the following equation (2.3).

$$\frac{\partial \gamma}{\partial s} = \frac{\partial \gamma}{\partial T} \times \frac{\partial T}{\partial s} \quad (2.3)$$

Where γ is surface tension, T is temperature, and s is the distance along the pool surface from the centre of the melt pool. The melt pool exhibits the highest temperature and lowest surface tension characteristics toward the laser beam. The inverse is true toward the boundary of the melt pool. This effect forms a depression toward the centre of the melt pool. A pressure difference is formed as a result of the depression and induces a flow within the

melt pool. This flow circulation in the melt pool essentially the force driving the thermo-capillary convection. The Marangoni number is defined as [76]:

$$\text{Ma} = \frac{\partial \gamma}{\partial T} \frac{1}{\mu \alpha} L \Delta T \quad (2.4)$$

Where, Ma is the Marangoni number, $\frac{\partial \gamma}{\partial T}$ is surface tension temperature coefficient, μ is the dynamic viscosity, α is the thermal diffusivity, L is the melt pool radius, and ΔT is the temperature difference between the melt pool's centre and edge. Studies have shown that in the case of a melt pool with a negative surface tension gradient, the flow moves outward from the centre of the melt pool. The inverse is true for a positive surface tension gradient, whereby the flow moves inward from the edges of the melt pool [77]. This phenomenon is illustrated in Figure 2.5. Factors that affect the surface tension gradient behaviour have been shown to be the presence of surface elements like oxygen or sulphur that drive the positive surface tension temperature coefficient, that in turn decreases the surface tension at the melt pool [76-78]. Studies have shown that by increasing the laser power input, the Marangoni convection and melt pool size increases. This occurs due to the thermocapillary convection that causes the physics of the melt pool to transit from a conduction to a convection dominated process.

Figure 2.5: Marangoni effect in melt pool for (a) inward and (b) outward flow [77]

2.2.3 Laser Process Parameters

Selecting the right LMD process parameters is crucial in ensuring part homogeneity with desirable material and mechanical properties. There are numerous key process parameters that interact during the deposition process namely, laser power, scanning velocity, laser beam diameter, powder feed rate, powder efficiency, and deposition track overlap.

An important process parameter in an LMD process is the laser power. The energy incident at the substrate location is dependent on the peak intensity of the laser emitted from the LMD machine. When choosing the appropriate laser power required for the LMD process, numerous factors need to be considered. The laser power

is directly proportional to the amount of dilution that is formed at the substrate. Using a laser power above the optimal input would result in a high dilution rate that results in drastic changes in the thermal history, producing a low thermal gradient and poor dimensional accuracy of the part or poor surface roughness which is further explored in Appendix D. In extreme cases, it could result in evaporation of the substrate material. On the other hand, using a laser power below the optimal input would result in multiple defects such as incomplete melting, porosity, inclusions, and delamination. These defects, depending on its type, size, location, and geometry, serve as crack initiation and susceptible to failure.

The scanning velocity is another key process variable of the LMD process that has a significant effect on the microstructural properties of the LMD part. Scanning velocity is defined as the rate of travel of the deposition across the substrate. A lower scanning velocity results in a lower thermal gradient and a higher dilution rate and vice versa. A closer look at residual stresses and distortion can be found in Appendix E.

Another key process variable is the laser spot size. This variable is defined as the diameter of the laser beam spot at its focal point. The focal point is commonly described as the working distance from the substrate to the deposition nozzle. During the LMD process, the powder feed and shielding gas is directed at the focal point, and the deposition track width is consequently defined based on the laser spot size used. The relationship between the laser power, scanning velocity, and laser spot size variables is expressed in the following equation:

$$E = \frac{p}{dv} \quad (2.5)$$

Where E is the laser energy density, p is laser power, d is laser beam diameter, and v is the scanning velocity respectively. From equation (2.5), it can be seen that adjusting some of the key process variables in LMD have a correlation in the effects of other variables. For instance, increasing the scanning velocity would result in a lower net laser energy density, or increasing the laser power input would result in a higher net laser energy density. Despite this correlation, the equation does not take into account other variables to come into play during the LMD process. For example, increasing the scanning velocity while increasing the laser power in theory can result in overall time savings of the additively manufactured part, but may result in issues with the powder flow stability

and consequently a defect-ridden part. Adrita and Atieh [79] established a process map for DED produced alloys from the compilation of various sources of literature.

Numerous attempts have been made to optimize the LMD process and identify a suitable process window. For example, Liu et al. looked into the process variables that influence the formation of thin metallic wall structures and developed a model where wall thickness, powder efficiency, and formation speed is defined as a function of the process variables [80]. Liu et al. found that the powder primary efficiency and formation speed are both parameters that are dependent on an exponential function involving the ratio of melt pool width, which is a consequential function of process variables to powder flow diameter. The identification of this relationship informs us about the process parameters that affect the melt pool characteristics which has a direct impact on the thermal history and microstructural grain growth. The application of this findings is essential to challenging AM geometries like thin-walled structures where control over the thermal history is a crucial necessity to achieve a quality build.

Manthiram et al. looked into the LMD process optimisation of Ti6Al4V/TiC composites and found that an FGM produced by optimised process parameters produced superior wear-resistant behaviour as compared to a fixed-process parameter sample [81]. The investigation into the FGM Ti6Al4V/TiC composite showed that by adjusting the process parameters, specific mechanical characteristics can be targeted and improved. In doing so, the qualification process can be expedited for industries that require a particular standard for their applications. Kong et al. presented a demonstration of the LMD process of an Inconel 718 deposited boss on a curved plate. A low heat input and high-rate deposition was used for this work. Small and scattered porosities with no cracking was achieved [82]. Balu et al. investigated the LMD process optimisation of Hastelloy C-276. Graphical optimisations of the process window were developed in order to access the suitable process parameter input to achieve a desired set of properties. The set of optimised parameters were identified to be a 285 W laser power, 11.3 mm/s scanning velocity, 0.176 g/s powder flow rate, and deposition track overlap of 34.1%. The optimised parameters were able to produce a minimal dilution of 1.24%, mean microhardness of 254 HV, with fine equiaxed microstructure.

2.3 Characteristics of LMD parts

2.3.1 Porosity

The formation of porosities and void defects are persistent problems in LMD produced parts. Many factors come into play during the deposition process and the powder feed material that gives rise to a number of porosity types. In general, the source of porosities in LMD parts are categorised into two broad categories, poor fusion, or gas entrapment. Poor fusion occurs when there is incomplete melting of the deposited metallic powders during the process. This causes an uneven surface on the deposited layer, and when a new layer is deposited over the previously deposited layer, it results in pores that get entrapped between the two layers. Gas entrapments occur when shielding gas, delivery gas, or atmospheric air gets entrapped in the melt pool during the deposition process or when there are pre-existing porosities in the powder feed particles that are delivered to the melt pool. These gaseous pockets subsequently get locked into the LMD part due the rapid solidification that occurs in AM processes. Lewis et al. [83] conducted an examination on porosity and defects on LMD produced SS316L samples showing the gas entrapment and interlayer porosity types as seen in Figure 2.6.

Figure 2.6: Micrographs of defects from
(a) gas entrapment and (b) interlayer porosities [83]

Numerous methods have been proposed to control and reduce the formation of these porosities and defects. For instance, Li et al. [84] applied a form of pre-heating to the powder feed material via baking and found that performing this step prior to the deposition process produced a significant decrease in porosities. Li et al. also added that using a higher laser power would ensure a more complete melt that would reduce gas entrapment types of porosities because a lower thermal gradient and a lower solidification rate allows more time for the gaseous phase to migrate to the atmosphere.

Susan et al. [85] investigated the effects of pre-existing porosities in powder feed particles and their effects on the degree of porosities in the LMD part. It was shown that powder feed particles with a large void-to-particle size ratio generally resulted in a lower porosity content compared to powder feed particles with the same amount of porosity but a lower void-to-particle size ratio. This observation is attributed to the higher buoyancy of the gaseous phase from the larger powder feed particle's pores during the deposition process that allows the gas to escape to the atmosphere easier. Susan et al.'s study into the pre-existing porosities in powder feed particles sets a precedent for common practices of AM characterization and quality checks, whereby the necessity to qualify the powder feed particles is an important step prior to metal deposition.

Steen et al. [86] demonstrated that adjusting the deposition track overlap ratio also has a significant effect on controlling the amount of interrune porosities. Interrune porosities are illustrated in Figure 2.7. In the mentioned study, the formation of these interrune pores were associated with a geometrical problem related to the aspect ratio of the deposited tracks. For deposition tracks with an aspect ratio of >5 and a 50% overlap, the edge slope was found to be $<46^\circ$, allowing the powder feed to access these corners when depositing the next deposition track. It was also shown that using an overlap of $>70\%$ results in more interrune porosities due to the resultant steep contact angle.

Figure 2.7: Interrune porosities through a series of overlapping deposition tracks

Choi and Chang [87] showed that a higher powder flow rate resulted the formation of more porosities, while the inverse trend is true for a higher laser power regardless of the powder feed rate variation. Choi and Chang developed a feedback height control system where the LMD variables were adjusted in-situ, establishing a significant development in DED technologies where geometrical accuracies become a significant challenge for larger AM builds or thin wall structures. Furthermore, Wang and Felicelli [88] demonstrated that a higher scanning speed generally had a higher degree of porosities due to poor fusion and incomplete melting. This was observed to be especially the case for the first deposition layer at the substrate.

Although porosities are broadly categorised as a type of AM defect, there are specialised cases where porosities are a desired feature of an AM-built part. For example, Ahsan et al. [89] used LMD techniques to produce porous structures of Inconel 625 using a cross-thin-wall fabrication build strategy. The objective of that study was to

show that if porous structures can maintain an adequate level of mechanical strength and performance, the range of applications can be extended to numerous areas like lightweight structures, transportation materials, interior channels, and even biomedical applications like prosthetics.

There are numerous techniques that are commonly utilized for the characterization of defects in AM parts. It is crucial for the furtherance of the AM field that manufacturing processes have to be constantly improved and minimize defects which can have a catastrophic effect on mechanical properties on the final AM part. With the emergence of x-ray tomography and ultrasound measurement for the inspection of defects in parts, it becomes a crucial enabler for the inspection of parts and reject them based on defect size, shape, and location with accordance to manufacturing standards and criteria. This form of non-destructive testing has become a standardized routine form of inspection for industrial inspection of various types of metallic parts including castings, injection moulds, and composites [90, 91]. The time and cost benefits of tomographic techniques for the purposes of screening AM parts are far superior compared to conventional destructive metallographic techniques that involve sectioning the AM part and inspecting the specimen using optical microscopy or dye penetrant testing.

X-ray computer tomography (CT) is an image-based, three-dimensional evaluation of defects using either a pencil, fan, or cone beam methods. The pencil beam x-ray translates linearly opposite an x-ray detector to capture solid density data along each beam. In the fan beam method, a two-dimensional fan of x-ray is used with a one-dimensional detector array that corresponds to the outer edges of the fan beam. The cone beam method utilizes a full 3D cone of x-rays with a two-dimensional detector. The CT technique contrasts the image and the resolution that determines the solidity of the parts measured. CT systems can be categorized into macro, micro, or nano CTs based on their focal spot size where high quality micro CT can detect voxel sizes $10\text{ }\mu\text{m}$ – $100\text{ }\mu\text{m}$, and nano CT can detect voxel sizes down to 1 nm [91]. Although CT systems can reduce the measurement time of defect detection, or even the examination of inner and outer geometries within an AM part, problems can arise when attempting to scan a multi-material AM part or when attempting to detect defects beyond the scope of voxel sizes equipped by the system.

Another form of tomography is ultrasound based measurement technique where a signal is transmitted through a material and received in a variety of ways, and the amplitude or time-of-flight are evaluated. Fundamentally, the

principle of ultrasound test (UT) operation is similar such that an acoustic signal is emitted and then received in either a pulse-echo or through-transmission mode. Generally, the response signal between a defective and non-defective area are different and can be differentiated by analysing the receiving signal and can detect defects down to 15 μm [92]. The ultrasound used in NDT are acoustic elastic waves that use frequencies higher than the audible range to humans, above 16 kHz to 20 kHz. The type, size, and depth of the damage or defect can be examined by processing and interpreting the processed signals. One limitation of the UT method is the necessity for coupling agents as well as difficulties in examining parts that have very thin, thick, or very complex geometries [93].

The metallographic method of measurement typically entails the examination of metallic materials using visible light to capture a magnified image of the micro- and macrostructure. 2D images are obtained using a microscope and sectioned images are stitched and processed to record a whole cross-sectional image of the sectioned sample. Light microscopes, scanning electron microscopy (SEM), and transmission electron microscopy (TEM), are examples of metallographic techniques that can be helpful in the detection of defects, microstructure, and macrostructure. Optical microscopy, SEM, and TEM are used to characterize structure by revealing not only microstructural characteristics like grain boundaries, phase boundaries, and inclusion distributions, but also defect characteristics like defect density, size, shape, and clusters. Sectioning the metallographic sample is a destructive necessity in excessing the cross-sectional view of the sample before mounting, grinding, polishing, etching, and examining the sample through a microscope [94].

Metallographic methods like optical microscopy are generally considered to have a lower cost setup and is a highly accessible method of material and defect characterization due to its simplicity compared to more sophisticated techniques like CT methods. Nonetheless, CT has its place in metrology because it allows for quick evaluation into material volumes and measurement of defects in critical sections of the AM part without destroying it. Romano et al. investigated the quality control in AM parts between metallography versus CT scanning techniques and found that metallographic inspection requires a larger amount of man hours per sample, but becomes more cost-effective when evaluated in bulk, whereas CT scanning becomes more cost-effective with increase in measured build volume [95].

2.3.2 Microstructure

The microstructure of AM-built parts is largely dependent on numerous factors like the material's solidification phases, solidification conditions like the thermal gradient, rate of solidification, and residual stresses. Kurz [96] expounds on this by using a combination of the latest solidification theories for various types of metals and generating process maps from analytical and numerical heat flux models to obtain local solidification parameters. Kurz goes on to explain that combining these analytical methods allows users to have a high level of control over the microstructural evolution and becomes a powerful tool for many industrial applications and processes. LMD processes broadly produce microstructures that are much finer than conventional manufacturing methods due to its rapid solidification and solid-state phase transformations. The rate of solidification, thermal gradient, and rate of solidification cooling are related by the following equation [97]:

$$R = \frac{1}{G} \frac{\partial T}{\partial t} \quad (2.6)$$

Where R is solidification velocity, G is thermal gradient, and $\frac{\partial T}{\partial t}$ is solidification cooling rate. Kurz and Trivedi [98] explains that the core interest of rapid solidification processing in the manufacture and production of alloys is to improve the homogeneity and properties. It was also established that rapid solidification covers an interfacial growth velocity range of 10^{-2} m/s to 10^3 m/s based on the localization of diffusion with respect to the size of the microstructure, the localization of diffusion with respect to the width of the interface, and the collision limit of atom attachment. The same study showed that the effects of the solidification velocity and the cooling rates outweigh the temperature gradient in the molten metal ahead of the solid-liquid interface, in terms of microstructure evolution.

Gaumann et al. [99] generated microstructural process maps based on analytical models that correlate a number of process parameters with the microstructure evolution and growth morphologies. For instance, a process map depicting the predicted range of solidification microstructure for two pre-heating temperatures as a function of laser power and scanning speed, as well as a function of laser power and laser beam diameter.

There are numerous empirical expressions that have been developed to correlate primary dendritic arm spacings (DAS), secondary DAS, cooling rates, and local solidification times. There are corresponding advantages to each empirical model, although the accuracy of each empirical model to various types of steel is generally unclear. El-Bealy and Thomas [100] developed simplified expressions to predict primary and secondary DAS as a function of the carbon content and thermal conditions for different types of steels. The general relation for primary DAS is expressed in equation (2.7).

$$\lambda_1 = K(C_R)^m(C_0)^n \quad (2.7)$$

Where λ_1 is the primary DAS, K , m and n are constants, C_R is the cooling rate, C_0 , is the nominal carbon content. The general relation for secondary DAS for steels of $<0.53\%$ wt. C is expressed in equation (2.8)

$$\lambda_2 = A_1(C_R)^{-f} \quad (2.8)$$

Where λ_2 is the secondary DAS, A_1 and f is a constant. Similarly, the general relation for secondary DAS for steels of $>0.53\%$ wt. C is expressed in equation (2.9):

$$\lambda_2 = A_2 \left(\frac{T_{liq} - T_{sol}}{C_R} \right)^d \quad (2.9)$$

Where A_2 is a constant, T_{liq} is the liquidus phase temperature, and T_{sol} is the solidus phase temperature, and d is a constant.

Yin and Felicelli [101] produced analytic models of dendritic structure growth using finite element modelling and cellular automation for the solidification of Fe-C alloy using DED process. Several process conditions were selected to investigate their effects on the solidification microstructure. For example, the effect of laser scanning speed's effect on the dendrite growth in the solidification microstructure is shown in Figure 2.8.

Figure 2.8: Solidification microstructure for the dendrite growth with varying laser scan speeds (a) 2 mm/s, (b) 10 mm/s, and (c) 20 mm/s and their respective cooling rates and solute concentration of C wt. % [101]

A strength of this study is the multiscale model that simulates the dendritic structure during the solidification process in the molten pool of LMD. Although this model serves as a strong foundation for future models for microstructural prediction, the extrapolation of this model to larger LMD builds with complex multi-layer geometries poses many challenges. The microstructural evolution during the deposition process and the thermal cycles needs to be considered in future iterations of the model.

2.3.3 Mechanical properties

In literature, the mechanical properties of LMD-built parts have focused a lot on tensile and compression testing, hardness testing, and more recent publications have focused on fracture-related studies. In this review, the build orientations of the LMD samples are described with accordance to the ASTM standard [102]. The test coupons require three axes X, Y, and Z, to designate their orientation. Using this terminology, the Z axis designation describes the build orientation, the X axis designation describes the direction that is parallel to the front of the AM machine and is parallel to the Z axis, and the Y axis designation describes the direction that is perpendicular to the Z and X axes. The three axes designation applies to cuboid specimens like the flat tensile specimen type, whereas only one axis designation is required for cylindrical specimens. These designations are illustrated in Figure 2.9.

Despite the established standards for specimen orientations, many studies do not abide by these rules, which makes a comprehensive review of the mechanical properties of AM materials challenging. Furthermore, the ASTM standards for AM materials and processes are constantly evolving as the technologies and applications progress. Moving the scientific body and forward as a conglomerate alongside the changes in ASTM standards is another challenge. A compilation of mechanical properties of various alloys produced by DED is listed in Table 2.3.

Figure 2.9: Orientation designations for mechanical testing of AM materials [102]

Table 2.3: Summary of Tensile properties of alloys produced by DED

DED Type	Material	Post-treatment	Orientation	Young's Modulus (GPa)	Yield Strength (MPa)	Ultimate Tensile Strength (UTS)	Elongation (%)	Ref
LSM	Ti-6Al-4V	As-built	XYZ	-	994.2	1068.1	5.4	[103]
		Heat treated	YXZ	-	860.7	962.3	11.7	
LDMD	Ti-6Al-4V	As-built	XYZ	-	922.6	992.8	8	[104]
			ZXY	-	868	956.8	13.8	
LNSM	IN718	Heat treated	X	-	1034	1276	12	[105]
DLD	IN718	Heat treated	-	-	1097.6	1321	9.8	[106]
LMD	IN718	As-built	XYZ	-	525	847	29	[107]
		HIP + Heat treated		-	1120	1301	18	
LRM	IN625	As-built	-	-	571	915	49	[108]
LMD	IN625	As-built	XYZ	-	656	1000	24	[109]
			ZXY	-	480	882	36	

Table 2.3 (Continued): Summary of Tensile properties of alloys produced by DED

DED Type	Material	Post-treatment	Orientation	Young's Modulus (GPa)	Yield Strength (MPa)	Ultimate Tensile Strength (UTS)	Elongation (%)	Ref
DED	SS304L	As-built	XZY	-	337	609	48.2	[110]
			ZXY	-	314	606	56.4	
DLD	SS316L	As-built	XYZ	199	900	580	6.5	[111]
			ZXY	197	770	415	4	
LMD	SS316L	As-built	ZXY	3.5	229	414	37.8	[1]
DMD	4340 steel	As-built	XYZ	-	-	1398.65	1.665	[112]
DMD	Cu-30Ni	As-built	-	-	229.08	258.72	3.37	[113]
LMD	HSLA-100	As-built	XYZ	-	783	841	20	[114]

2.4 Materials-Process-Performance

Materials-Process-Performance characterization is a wholistic approach to studying AM applications. The relationship between the materials, process, and performance domains are deeply intertwined and are crucial to the determination of the quality of the product. There are several key processing parameters that are important influencing criteria to the morphology and mechanical performance. Parametric studies are crucial in the prediction of part quality and performance and have the ability to save cost and time at the manufacturing stage. The necessity for process maps for functionally graded materials are increasingly important as the adoption of AM technologies grow. Developing process maps for materials is a complex process that requires the combination of extensive parametric studies. The current state of the art for Materials-Process-Performance characterization utilizes various tools in its development like statistical methods, in-situ monitoring techniques, finite element modelling, SEM/EDX methods, non-destructive testing, instrumented testing, qualification etc.

One method to simplify the extensive nature of parametric studies is to first define the scope of the study by identifying the key processing parameters and focusing on the main effects. Kersten et al. identified laser power, scanning speed, and powder mass flowrate as the key processing parameters in a study on optimization of laser DED of SS316L material [115]. Felix et al. used a 2^3 -full factorial design to determine the optimal processing window, and used the optimal process parameters to determine the influence of alternative parameters on mechanical properties [116]. Bax et al. developed process maps for Inconel 718 and showed that dilution and build rate do not directly depend on scanning speed, whereas powder efficiency and convexity of single track deposition clads do [117].

Parametric studies in DED have been demonstrated to be a useful tool in charting relationships between characteristics in the process. For instance, Yadav et al. showed that track geometry has a direct and inverse dependence with laser power and scan speed with increasing Cu in Cu-Ni composition due to an improvement in wettability [118]. Benarji et al. found that microhardness reduces with an increase in scanning speed mainly due to an increase in lack of fusion and porosity in the deposited material [119].

The above cited studies demonstrate a range of relationships between individual parameters within the process of DED, and its effect on the mechanical performance of the part. These studies form a strong foundation for the laser deposition of functionally graded materials and provide a generalized understanding of the process. However, one drawback of these empirical studies is that the conclusions are highly contextualized to the conditions of the experiments. For instance, the study conducted by Bearji et al. constructed a process map, charting the parametric relationship for DED, but is localized to the DMG MORI LaserTec 65 machine. Using the same AM parameters in a different machine is likely to yield different results due to a variety of reasons like cladding head design, gas flow rate tube, powder feed canister design, laser type, etc. Furthermore, the conclusions obtained from the study conducted by Yadav et al. are highly localized to the single-track LMD track beads and cannot be extrapolated to multi-layer, or larger build geometries where the effects of the thermal history play a more significant role in the quality of the part.

A compilation of materials-process-performance characterization studies are shown in Table 2.4.

Table 2.4: Compilation of materials-process-performance findings across selected literature

Material	Powder Feed	Laser Source	Findings	Ref
Ti-6Al-4V	Particle size: –100+325 mesh	CO ₂ laser 6 kW	As deposited Yield Strength: 1105 ± 19 MPa UTS: 1163 ± 22 MPa Elongation: 4 ± 1 %	[120]
Ti-6Al-4V	Particle size: 25 – 45 µm	Nd : YAG laser (pulsed)	Yield Strength: 1100 ± 12 MPa UTS: 1211 ± 31 MPa Elongation: 6.5 ± 0.6 %	[121]
Ti-4Al-1.5Mn	Particle size: 45 – 420 µm	CO ₂ laser 5 kW	As deposited Toughness 599 ± 57 kJ m ⁻²	[122]
Inconel 625	Particle size: 45 – 135 µm	CO ₂ laser 6 kW	Defect free Columnar dendrites Hardness 254 ± 6 HV	[123]
Inconel 718	Particle size: 44 – 150 µm	CO ₂ laser 5 kW	As deposited Yield Strength: 590 MPa UTS: 845 MPa Elongation: 11 %	[123]
Rene 88DT	Particle size: 44 – 150 µm	CO ₂ laser 5 kW	HIP + heat treated Yield Strength: 1010 – 1030 MPa UTS: 1400 – 1440 MPa Elongation 16.5 % - 17.5 %	[124]
Tool Steel H13	Particle size: – 70 mesh	CO ₂ laser 4.5 kW	Yield Strength: 1505 MPa UTS: 1820 MPa Strain: 6% Hardness: 690 HK	[15]
Tool Steel H13	Particle size: – 70 mesh	CO ₂ laser 4.5 kW	Yield Strength: 1505 MPa UTS: 1820 MPa	[15]

			Strain: 6% Hardness: 690 HK	
AISI 4340 Steel	Particle size: – 140/+325 mesh	Diode laser 1 kW	Microhardness 681 – 480 HV Porosity 4.13%	[125]
Stainless Steel 316L	-	CO ₂ laser 5 kW	1000 mm/min scanning velocity Yield Strength: ~400 MPa UTS: ~450 MPa Elongation: ~20%	[126]
CoCrMo	Particle size: – 100/+325 mesh	Nd : YAG laser 500 W	Hardness 40 HRC	[127]
Cu-30Ni alloy	Particle size: – 100/+325 mesh	CO ₂ laser 5 kW	Yield Strength: 316.16 MPa UTS: 240.49 MPa Elongation: 13.9%	[128]

The correlation between the LMD process parameters and the materials-process-performance characteristics are deeply intertwined. Numerous studies have explored the optimization of parameters in the use of LMD process for a variety of applications. However, the boundary conditions for many of these studies often vary widely from experiment to experiment. In order to narrow down the range of parameters to explore for this study, a few studies were examined with boundary conditions as close as possible to the existing set up to identify an approximate set of parameters for the study of defect and microstructure. Kledwig et al. explored the use of a DMG MORI LaserTec 65 3D system to investigate various process conditions focusing on parameters like the energy intensity distribution [129]. This study provided useful information for the use of a similar LMD machine model to this study with an added focus on the melt pool conditions as a consequence of the energy input. By using the built-in CCD camera sensor in the machine, the melt pool size was evaluated for varying values of laser power, scanning speed, and powder feed rate. Although the study identified parametric values that resulted in the lowest deviation from the target melt pool size, it does not give insight into the final quality of the AM-built part. Yang et al. used a similar equipment to investigate the influences of LMD processing parameters on the surface morphologies relating to heat treatment. [130]. The samples produced by the parameters employed in this study yielded results that had non-melted particles on the surface and had irregular microstructural characteristics. Lim et al. utilised a similar LMD set up and varied printing parameters like laser power, printing patterns like zig-zag and spiral, and closed-loop control to measure the effects of the parameters on deposition height [131]. The study offers a useful tool that approximates the geometrical height of the deposited volume and the process parameters that influence the result. However, the study does not provide any information on the defects, microstructural characteristics,

and mechanical characteristics of the build, and hence cannot speak to its quality. The studies cited in this section utilise very similar processing conditions in terms of the LMD equipment but have varied experimental objectives. Nonetheless, these studies provide useful information into the use of the DMG MORI LaserTec 65 3D equipment and it is possible to extract an approximation of suitable process parameters for future applications

2.5 Hybrid Manufacturing (Subtractive and Additive)

2.5.1 Introduction

The modern manufacturing industry is responsible for the mass production of a myriad of products for many industrial sectors. Key technologies that play a crucial role in the industry include CNC machining, joining, and transformative processes, and with the fourth industrial revolution, additive manufacturing. Despite the immense progress made in recent years, there are several drawbacks that require novel solutions. Due to technological limitations, there are certain types of complex part geometries that cannot be produced by CNC machining due to restrictions in tool path access. Certain materials also are difficult to machine due to the hardness properties, that cause overheating as well as high tool wear. On the other hand, AM technologies have their own limitations like low part accuracy and poor surface finish.

In light of these challenges, the combination of two or more manufacturing processes in one step becomes an idea of interest because of the potential of using an alternative manufacturing process to cover for the weakness of another.

2.5.2 Definition

The definition of hybrid processes in literature is a non-precise term used by numerous researchers to describe a combination of more than one manufacturing process. For instance, Rajukar et al. describes hybrid machining as a combination of two or more machining processes to remove material [132]. Rivette et al. describes hybrid processes as ‘different forms of energy or forms of energy caused in different ways respectively are used at the

same time at the same zone of impact' [133]. Lauwers et al. describes hybrid as a 'combination of processes having a large influence on the process characteristics' [134].

Hybrid manufacturing, in the case of this PhD work, refers to the combination of subtractive CNC processes and the additive LMD process. LMD's use of a coordinate system control allows it to be compatible with conventional 5-axis CNC systems. Hybrid manufacturing machines leverage on this compatibility by combining additive and subtractive processes in a single machine, hence streamlining, and quickening the manufacturing process. The interchanging of additive and subtractive functions enables the direct machining of locations, like internal features, within the component that is not previously accessible by conventional AM means.

2.5.3 Method

Jeng et al. utilised LMD in combination with CNC milling to repair and modify an injection mould component [135]. Zhang *et al.* configured a 5-axis LMD with a 5-axis milling machine to demonstrate the ability to produce AM features without the need for support structures [136]. Hur *et al.* offers a manufacturing solution that combines AM and CNC systems in order to produce AM parts of a required surface finishing quality in the shortest time possible [137]. An illustration of a two-step additive-subtractive hybrid manufacturing process for a cylindrical LMD build-up is shown in Figure 2.10. Further information about existing hybrid manufacturing machines can be found in Appendix F.

Figure 2.10: Hybrid manufacturing process (a) Step 1: Additive, LMD (b) Step 2: Subtractive, CNC milling

2.6 Interfacial Joints

The interfacial structural strengthening in LMD built components is critical in field of AM, repair, and remanufacturing of functional parts. When components have been in service under working conditions that undergo thermal cycling, fatigue, corrosion, and localised impacts, the component can result in different forms of damage like defects, fractures, and cracks. In the industry, there are numerous scenarios in advanced manufacturing technologies where high-value and high-performance components are often discarded after the component has undergone significant that result in failure. When a component is damaged, the extent of the damage is first assessed from the repair and remanufacturing perspective and evaluated if it is suitable for the component to be repaired. To recover the component from its unserviceable state, the right repair and remanufacturing methods need to be employed followed by quality control and reliability evaluations.

The development of LMD is a strong enabler for repair and remanufacturing applications due to its lower heat input, lower dilution rate, less distortion, combined with very strong metallurgical bonding and mechanical performance. Despite these advantages over traditional methods of additive repair methods like tungsten inert gas welding, electro spark welding, and thermal spray processes, the joining feature between the added material and

the substrate continues to serve as a persistent challenge to the structural quality of the remanufactured part. For instance, the presence of defects due to poor fusion between the deposited material and the substrate is a common issue found in laser cladding applications and has been extensively discussed in Section 2.3.1.

A common problem found in LMD applications are cracking at the interfacial area between two dissimilar materials. These defects typically occur in the repair and remanufacturing sector or in the production of FGM metals where the deposited material consist of brittle intermetallic phases between elements that form during the gradient transition of its chemical composition. Khodabakhshi et al. studied the possibility of additively manufacturing a dissimilar metal wall (A410-L stainless steel, A316-L stainless steel, and zirconium) with different crystallographic structures and found disintegration of the structure with enlarged cracking occurring at the interfacial joint [138]. By optimizing the deposition constitution of elements like Zr and Ni, the authors were able to reduce the amount of interlayer cracks that form between the dissimilar metallic layers. Chan et al. studied the feasibility of additively manufacturing TA15 to an Inconel 718 bimetallic structure via a Nb/Cu multi-interlayer strategy [139]. The study demonstrated the possibility of producing a defect-free, bimetallic structure that utilized an Nb/Cu multi-interlayer to prevent the generation of brittle Ti-Ni and Ti-Cu phases effectively. Bonny and Amit demonstrated that the use of DED for the deposition of Inconel 718 on Ti64, the compositional gradation of the two alloys with the use of NiCr as an intermediate bond layer was attempted but resulted in a variety of severe defects including delamination, cracking and debonding due to thermal cracking, residual stresses, and the immiscibility of dissimilar materials caused by the differences in thermophysical and metallurgical properties [140]. The authors were able to optimize the quality of the build by using a compositional bond layer mixture of VC, Inconel 716, and Ti64 to fabricate a crack-free bimetallic structure.

Although a lot of progress have been made in material studies field on the optimization of additively manufactured FGM materials by altering the elemental composition or the use of intermediary layers of compatible materials, little to no effort has been made on the mechanics of additively manufactured joints. Additional effort could be made to supplement existing work by employing concepts of mechanics to improve the joint strength by incorporating new concepts of geometrical optimization into the design for additive manufacturing.

Chapter 3: Defect and Microstructure Characterization (Study 1)

3.1 Introduction

The objective of this study is to investigate and characterize the porosity, density, defect, and microstructure of an LMD built Stainless Steel AISI 316L component. LMD was used to form a Stainless Steel AISI 316L steel block component on a Mild Steel S235JR substrate. Porosities, density, and defects were characterised at four localities within an LMD component by microscopy. The element composition was analysed at the fusion lines and deposition body sites to verify the presence of oxidation during the build process using SEM and EDX techniques. Lastly, an element mapping analysis was conducted to determine the solidification mode for the LMD component. Sources for defects were proposed based on the characteristics of the porosity analysis and conclusions were made about the solidification behaviour of the LMD component.

Defects in LMD-manufactured structures such as foreign particles, porosity, interface delamination, have been shown to be crucial contributors to a build's propensity to rupture, by acting as stress concentrators and crack growth locations. For example, porosities can affect the overall build quality in LMD built components as the conjoining of adjacent discontinuities have a severe effect on mechanical properties like tensile, fatigue, and fracture strengths [141-143]. Extensive research has been conducted on mechanical properties [107, 144-148]. Detrimental defects in LMD-created structures such as foreign particles, porosity, interface delamination, have been proven to be crucial contributors to a build's propensity to rupture, by acting as stress concentrators and crack growth locations. For example, porosities can affect the overall build quality in LMD built components as the conjoining of adjacent discontinuities have a severe effect on mechanical properties like tensile, fatigue, and fracture strengths [141-143].

3.2 Methodology

3.2.1 Laser Metal Deposition Process

Stainless Steel 316L powder material with a 45 μm – 105 μm particle size range was used as feedstock material. LaserTec 65 3D hybrid was used to fabricate the LMD samples. An image of the LaserTec 65 3D hybrid is shown in Figure 3.1. The build parameters employed in this build were selected based on the review done in Section 2.4 and are specified in Table 3.1. The chemical composition of the Stainless Steel 316L powder feed is shown in Table 3.2

Figure 3.1: DMG MORI LaserTec 65 3D

Table 3.1: Build parameters employed in LMD process

	Parameter	Value
1.	Laser Spot Diameter	3 mm
2.	Laser Power	1.4 kW
3.	Deposition Rate	14 g/min
4.	Scanning velocity	1000 mm/min
5.	Track Overlap	60%

Table 3.2: Nominal chemical composition of SS316L powder feed (wt.%)

Material	C	Mn	P	S	Si	Cr	Ni	Mo	Fe
SS316L	0.03	2.00	0.045	0.03	1.00	16.00-18.00	10.00-14.00	2.00-3.00	Bal.

A diode laser was used as a laser source, and argon gas as the shielding gas. A bidirectional deposition pattern (raster scan) and bidirectional layer orientation (90° rotation per layer) build strategy was used to additively

manufacture a Stainless Steel AISI 316L block component of 30 mm × 50 mm × 50 mm dimensions, built on a Mild Steel S235JR substrate of 100 mm × 100 mm × 12 mm dimensions, as illustrated in Figure 3.2.

Figure 3.2: LMD Stainless Steel AISI 316L
(a) build dimensions and (b) bidirectional raster scan path strategy

3.2.2 Porosity and Density Analysis

The LMD Stainless Steel AISI 316L build was sectioned using a bench saw, extracting four specimens at four target surface locations: Top Zone: YZ-Plane, Top Zone: XZ-Plane, Bottom Zone: YZ-Plane, Bottom Zone: XZ-Plane. These specimens were subsequently hot mounted in resin under heat and high pressure, and polished for density and porosity analysis by microscopy. Microscope images were captured at 25x magnification, 200 nm resolution, stitched, and analysed using a Zeiss Light Microscope as seen in Figure 3.3. Specimen density was measured using Zeiss software and porosity characteristics were analysed using ImageJ software.

Figure 3.3: LMD Stainless Steel AISI 316L
(a) Specimen extraction locations and (b) stitched microscopy images

3.2.3 Microstructure and Morphology Characterisation

Specimens were chemically etched using an electrolyte solution of 10% perchloric acid, 90% methanol composition (% vol.), with a voltage input of 5.5V, for a 5 second duration. Microscope images were captured at 50x magnification using a Zeiss Microscope Stemi 2000-C and analysed using ImageJ software. SEM images were captured at 300x magnification to capture and distinguish the fusion line and deposition zone locations using Zeiss EVO SEM equipment and analysed with Inca software. Aztec software was further used to conduct EDX analysis.

3.3 Results and Discussion

3.3.1 Build Evaluation

After chemical etching, the deposited track boundaries can be visibly identified due to the etching of oxide films at SS316L surfaces and can be measured by microscopy [149]. The cross-sectional morphology of the specimens

at the YZ-Plane and XZ-Plane, middle region of the LMD Stainless Steel AISI 316L block is displayed in Figure 3.4.

From these microscopy images, the bidirectional layer orientation, 90° deposition orientation rotation per layer, build strategy can be observed. This is seen from the alternating fusion lines patterns between adjacent layers and a similar deposition pattern seen in both the XZ-plane and YZ-plane images. Layer thickness was measured to be approximately 1.1 mm, track width was measured to be 3.0 mm, and track overlap was measured to be 1.8 mm. From which, the track overlap percentage can be derived by $overlap/width=60\%$. These results are consistent with the input build parameters.

Figure 3.4: Layer thickness, track width, track overlap measurements for (a) YZ-Plane cross-sectional surface, near middle region, (b) XZ-Plane cross-sectional surface, near middle region

In order to evaluate the LMD Stainless Steel AISI 316L build, track widths were measured across the Top Zone: YZ-Plane, Bottom Zone: YZ-Plane, Top Zone: XZ-Plane, Bottom Zone: XZ-Plane cross-sections as designated in Fig. 2 (a). 10 track width measurements were taken for each of the 8 measurement layers per 4 specimen cross-sections. The mean track width measured for each of measurement layers is displayed in Figure 3.5. The Top Zone and Bottom Zone's mean and standard deviation track width measurement values are collated in Table 2.

Figure 3.5: Mean track width measurement for Bottom Zone and Top Zone locations

Table 3.3: Track width measurements

Measurement Zone	Mean	Standard Deviation
Bottom Zone	3.22 mm	0.24 mm
Top Zone	3.57 mm	0.26 mm

From Fig. 5 and Table 2, the Bottom Zone locations of the specimen exhibited a lower track width mean value than the Top Zone locations. This is largely due to the differences in the thermal gradient and cooling rates between the Bottom Zone and Top Zone. A smaller thermal gradient and a faster cooling rate at the start of the process compared to toward the end results in a larger melt pool size toward the Top Zone of the LMD Stainless Steel AISI 316L specimen. This is consistent with the larger track width size measured at the Top Zone compared to the Bottom Zone zones. Manvatkar et al. developed a transient heat transfer and fluid flow model for laser assisted deposition of stainless steel and found that the melt pool dimensions increases while the cooling rate decreases toward the upper layers [17]. Aggarangsi et al. found that in order to control for the track width size consistency within a build, the process scaling of laser powers need to be monitored and controlled [18].

3.3.2 Density and Porosity Evaluation

The LMD Stainless Steel AISI 316L's object count distribution per porosity size class at four internal target measurement areas are compiled in Figure 3.6, and overall density and object count per target measurement area is consolidated in Table 3.4. Most of the objects detected concentrated within the Class 1 (0.0 μm – 20.0 μm) and Class 2 (21.0 μm – 40.0 μm) range. All four target zones exhibited a high degree of part density of more than 99.80%. Top zone measurement areas exhibited a slightly higher degree of part density, and a larger object count distribution of small pore sizes, and larger object count, than bottom zone measurement areas.

Figure 3.6: Object count distribution per pore size class for each target zone surface

Table 3.4: Density and object count detection per measurement area

	Measurement Area	Object Count	Density
1.	Top Zone: XZ-Plane	730	99.87%
2.	Top Zone: YZ-Plane	1313	99.91%
3.	Bottom Zone: XZ-Plane	220	99.80%
4.	Bottom Zone: YZ-Plane	101	99.80%

A reason for the difference in top and bottom zone measurements for part density and porosity sizes is the dissimilar rate of melting and solidification experienced at the two said zones. No pre-heating was applied to the substrate prior to deposition of the first few layers, hence, air entrapment due to a degree of incomplete melting can be expected at the bottom zone area. Larger porosities and higher object count have a higher tendency to cause unpredictable failures than smaller porosities and lower object counts. In this measurement, a lower object count

was observed at the bottom zone where larger pore sizes were detected, compared to a higher object count observed at the top zone where smaller pore sizes were detected.

A method to narrow down the possible sources of defect for the porosity measurement is to characterise the form of the detected objects. By determining the objects' form, and taking into account the object size detected, an inference of the source of defect can be made. There are two quantifiable characteristics for the sphericity measurement of objects measured in a 2D plane like microscopy: circularity and roundness according to ImageJ [150]. Circularity takes into account the object's form and smoothness, while roundness takes into account the object's ovality and aspect ratio. This correlation is illustrated in Figure 3.7. The closer these two characteristic values are to 1.0, the greater the resemblance to a sphere. Using ImageJ software, each detected pore was measured for its roundness and circularity. Roundness and circularity are calculated based on equations (3.1) and (3.2). Results of the findings are subsequently graphed and compiled in Figure 3.8.

$$\text{Circularity} = 4\pi \times \frac{\text{Area}}{\text{Perimeter}^2} \quad (3.1)$$

$$\text{Roundness} = 4 \times \frac{\text{Area}}{\pi \times \text{Major Axis}^2} \quad (3.2)$$

Figure 3.7: Circularity and roundness chart [150]

Figure 3.8: Object count distribution for (a) circularity and (b) roundness

In order to correlate the circularity and roundness measurements, a high (+) and low (-) criterion threshold was set at 0.8 in Figure 3.8, where a strong spheroid resemblance can be detected. Here, it can be observed that majority of the objects detected fall within the high circularity with low roundness, and high circularity and high roundness categories. Various research papers have reported that different defect types have different microscopic forms [151-153]. Based on Figure 3.7, for objects that exhibit both strong degrees of circularity and roundness, there is a strong likelihood that these are voids that originate from defects like, moisture evaporation, local voids after powder-layer deposition, gas entrapment, solidification porosities, and key holing depending on the size of the defects. For objects that exhibit strong degrees of circularity and low degrees of roundness, a possible source of these fabrication problems are bonding defects due to incomplete melting due to lack of fusion, thermal cracking, or interrune porosities.

The results from Figure 3.8 indicate that the bulk of the object population exhibited strong degrees of circularity and roundness. When examined in combination with the pore size measured in Figure 3.6, where the bulk of the object population are also very small ($< 20 \mu\text{m}$), it can be inferred that the voids detected in the LMD Stainless Steel 316L build mostly originate from issues arising from the deposition process like gas entrapments and moisture evaporation. Monroy et al. studied the porosity features and correlated round pores with smooth features with gas entrapments [154]. Hu et al. found that when the scan velocity is high, gaseous bubbles get trapped within

the melt pool and settle to form round pores, and that poor interlayer bonding produces pore defects that are more irregular in form [155]. Ziolkowski et al., in their porosity study, concluded that by diagnosing the porosities' forms is useful in determining the course of cracks and pin point locations for crack initiation sites, especially in fatigue [156].

Figure 3.9: X-ray tomography (a) image slice at first deposition layer, (b) 3D image sequence reconstruction of porosities at first deposition layer, (c) image slice of large interlayer porosities, (d) 3D image sequence reconstruction of large interlayer porosities

The porosities measured by X-ray tomography at the first deposition layer are distinctly identifiable by the irregular profile, as seen from Figure 3.9 (a). The cooler substrate at the deposition initiation result in a lack of fusion, producing a lower part density measurement toward the first layer. Other large interlayer porosities and located toward the middle of the block were also measured, as seen from Figure 3.9 (b). These large pores exhibit a flat and spheroid profile and are $>150\ \mu\text{m}$ in width and are attributed to the interlayer air entrapments that result in this specific form of pore shape. Smaller porosities are not able to be captured with this technique due to

the spatial resolution limitations of X-ray tomography. Maskery et al. found that larger pores measured by X-ray tomography, exhibited a similar disc-like profile, with aspect ratios extending to high values, and proposed that the layer-by-layer manufacture of AM parts is responsible for the large disc-like pore profiles observed [157]. Van Bael et al. also cited similar limitations in the measurement of smaller sized pores due to the spatial resolution of the X-ray tomography [158].

3.3.3 Morphology, Microstructure, and Composition Analysis

An optical micrograph of the microstructure morphology is shown in Figure 3.10, where the transition points between the fusion line and track body locations are indicated. The microstructure morphology primarily comprises of columnar grains that grow epitaxially from the base of the LMD build. This is due to the direction of heat flow through previously deposited layers and the substrate that results from the layer-by-layer deposition. At the fusion line locations, the epitaxially grown columnar grains transit to a fine equiaxed grain growth, where cooling rates are expected to be highest. The rapid flow of heat out of the melt zone into surround metal creates a thin layer of fine equiaxed grains at the fusion line locations that trail the laser spot scan movement.

Figure 3.10: Micrograph of microstructure morphology within LMD build. Dashed lines indicate transition between track body columnar

Vickers microhardness test was conducted to characterise the mechanical differences at the fusion line and track body locations. Images of some of the indent locations are shown in Figure 3.11. Sixteen indentation points were measured at the fusion line locations, and another 16 indentation points were measured at the track body locations. Results of the microhardness measurements for the fusion line and deposition body are shown in Figure 3.12.

Figure 3.11: Vickers microhardness indentation locations at (a) fusion line, (b) track body

Figure 3.12: Vickers microhardness measurements for track body and fusion line

Figure 3.13: Vickers microhardness versus $d^{-1/2}$, where d is the mean grain width size adjacent to microindentation location

The fusion line locations exhibited a higher Vickers hardness than the track body locations as seen in Figure 3.12.

The mean microhardness value at the track body and fusion locations were measured to be 182.24 HV and 187.10 HV respectively. This observation is attributed to the Hall-Petch relation:

$$\sigma_y = \sigma_0 + \frac{k}{\sqrt{d}} \quad (3.1)$$

Where σ_y is the yield strength, σ_0 is a materials constant that governs the stresses for grain dislocation movement, k is the Hall-Petch slope coefficient, and d is the mean grain size of the material. The Hall-Petch relates the material's yield strength with its grain size where a smaller grain size would produce a higher yield strength and hardness. The higher microhardness measured of the fine equiaxed grains at the fusion line locations is reciprocal of the square root of the grain width.

In order to correlate the grain sizes with the microhardness measurements, a Hall-Petch plot was used to chart the two sets of data as seen in Figure 3.13. From the Hall-Petch plot, the σ_0 and k constants are approximated by applying a linear fit and using the general relationship between hardness and yield stress [159]:

$$\Delta\sigma_y = 3.03 \Delta H_v \quad (3.2)$$

Where H_v is Vickers hardness. The σ_0 and k constants of at the fusion lines are hence approximated to values of $\sigma_0 = 534.63$ MPa and $k = 345.40$ MPa and the track body to values of $\sigma_0 = 494.76$ MPa and $k = 334.89$ MPa. These σ_0 and k constants are consistent with stainless steel's Hall-Petch data [160]. Slight differences observed between the 2 sets of data can be attributed to the uneven distribution and variation in the grain sizes morphologies, typical of an additively manufactured part. Huang et al. found that the hardness distribution across a section of a AM part produced a periodical fluctuation in the hardness values that was attributed to the bonding surface locations [161].

From Figure 3.14, distinct white deposition fusion lines were observed under microscopy, after chemical etching. To understand what constitutes these white boundaries, SEM measurements were taken at the fusion lines' location and deposition body locations. Two target areas were measured in total, with one target area taken from each area and characterised by EDX as seen in Figure 3.14. SEM was set at Backscatter Electron (BSE) mode, 300 times magnification, 9 mm working distance and an Extra High Tension (EHT) voltage level of 15kV. The measured element composition for the target deposition body and fusion lines areas can be seen in Figure 3.14. The elemental mapping of carbon and oxygen is possible mainly because of a Super Atmospheric Thin Window (SATW) configuration, which has a greater sensitivity to lighter elements.

(a)

(b)

Figure 3.14: SEM location from (a) microscope image and (b) EDX sites at deposition body and fusion lines

Figure 3.15: Characteristic X-ray peaks for (a) fusion lines and (b) deposition body sites

The element Oxygen (O) was observed in the characteristic X-ray peaks for the fusion lines' location, not seen at deposition body location. This indicates that there is a chromium oxide formation occurring as each LMD track is deposited. Chromium reacts with atmospheric oxygen to form a protective oxide layer. Sporadic chromium oxide deposits can also be observed within deposition body locations. This is due to same reaction occurring at the powder feed particle level during the deposition process.

An EDX element mapping of a deposition body location was conducted and is displayed in Figure 3.16, with (a) showing the layered element image, and (b)–(h) showing the detected elemental constituents. From this measurement, columnar microstructures were observed in Molybdenum, Chromium, and Iron element maps. To understand the composition of these columnar microstructures, 6 target points were characterised for their elemental compositions, three at the columnar boundary and three at the columnar body, as seen in Molybdenum's

element map. The results for the element composition measurements at each point are collated in Table 3.5, Table 3.6, Table 3.7.

Figure 3.16: Element mapping for deposition body area: (a) Layered element image, (b) Molybdenum and target body, boundary element composition points, (c) Chromium, (d) Silicon, (e) Iron, (f) Nickel, (g) Carbon, (h) Manganese

Table 3.5: Element compositions for (a) columnar boundary (b) columnar body areas (c) overall scanned area

Element	1	2	3	Mean wt. %
C	0.03	0.04	0.03	0.04
Si	0.45	0.50	0.35	0.43
Cr	26.06	25.71	25.34	25.70
Mn	1.28	1.31	1.60	1.40
Fe	59.60	58.63	59.73	59.32
Ni	10.28	11.73	10.71	10.91
Mo	2.31	2.08	2.24	2.21
			N_{eq} :	12.66
			Cr_{eq} :	28.56

Table 3.6: Element composition for columnar body area

Element	4	5	6	Mean wt. %
C	0.05	0.05	0.04	0.05
Si	0.48	0.35	0.47	0.43
Cr	17.22	16.42	16.15	16.60
Mn	1.50	1.33	1.41	1.41
Fe	69.34	70.20	70.49	70.01
Ni	10.05	10.38	10.02	10.15
Mo	1.37	1.27	1.41	1.35
			N_{eq} :	12.24
			Cr_{eq} :	18.60

Table 3.7: Element composition for overall scanned area

Element	Overall wt. %
C	0.05
Si	0.85
Cr	16.59
Mn	1.55
Fe	69.3
Ni	10.79
Mo	0.87
N_{eq} :	13.07
Cr_{eq} :	18.74

The element compositions were measured and calculated for their Nickel and Chromium equivalents using equations (3.3) and (3.4), and their ferrite composition was derived using Schaeffler's diagram as seen in Figure 3.17a. [162]. From Schaeffler's diagram, the target columnar boundary area has a δ -ferrite composition of approximately 40%, the columnar body area has a lower δ -ferrite composition of approximately 5%. This suggests that the columnar body area is predominantly austenitic, while the columnar boundary area composition is δ -ferritic. From the overall scanned area element composition, the Cr_{eq}/Ni_{eq} ratio was calculated to be 1.43, and

hence falls within the upper range of the Austenitic-Ferritic (AF) solidification mode, as determined in Table 5 [163, 164]. In an AF solidification mode, austenitic steels form a microstructure constituency of primarily austenite with columnar δ -ferrite formations, along the solidification direction as seen in Figure 3.17b [165, 166]. This is consistent with the columnar δ -ferrite formations observed, parallel to the deposition direction. Darker localities in Figure 3.16's element mapping occur in element maps that contribute to Cr_{eq} and are deduced to be ferritic deposits.

$$Ni_{eq} = Ni + (30 \times C) + (0.5 \times Mn) \quad (3.3)$$

$$Cr_{eq} = Cr + Mo + (1.5 \times Si) + (0.5 \times Nb) \quad (3.4)$$

Figure 3.17: (a) Schaeffler's diagram (b) Austenitic steel solidification diagram [162]

Table 3.8: Solidification mode classification

Solidification Mode	Cr_{eq}/Ni_{eq} ratio
Ferritic (F) Mode	$1.95 \leq Cr_{eq}/Ni_{eq}$
Ferritic-Austenitic (FA) Mode	$1.48 \leq Cr_{eq}/Ni_{eq} \leq 1.95$
Austenitic (A), Austenitic-Ferritic (AF) Mode	$Cr_{eq}/Ni_{eq} \leq 1.14$

3.4 Summary

A Stainless Steel AISI 316L block was built by LMD. The part's porosity, density, defect, and microstructure were characterised by microscopy, SEM, and EDX methods. Conclusions about the resultant track dimensions, likely sources for defects, fusion lines and body compositions, and austenitic and ferritic characterisations are offered.

Track widths toward the Top Zone of the specimen were measured to be larger than the Bottom Zone locations. This is due to higher thermal gradient and lower cooling rates at the Top Zone that result in the larger melt pool and track width sizes compared to the Bottom Zone location. The majority of pores detected fell within the 0.0 μm – 20.9 μm range, and part densities were measured to be 99.8% and above. A larger population of small pore sizes, and slightly higher part densities were measured in the top zone area than the bottom zone area. To achieve a more consistent build between the bottom and top zones, further optimisation can be employed in the dynamic control of laser power input, or a pre-heating to the substrate prior to the build can be applied. The majority of pores detected exhibited a high degree of circularity and a mixed degree of roundness. This suggests that the primary source of defects in this build is air entrapment from moisture evaporation, and local voids after powder layer deposition. X-ray tomography image sequencing and 3D reconstruction techniques were used to characterise areas within the LMD block that exhibited severe porosity defects. Large irregular profile porosities due to a lack of fusion were detected at the first layer, and large disc-like interlayer porosities due to air entrapments were detected toward the middle of the LMD block. Differences in the Vickers microhardness measurement values are attributed to the larger grain sizes found at the track body than fusion line sites. The Hall-Petch relation's constants are found to be consistent with Stainless Steel's data.

However, this finding should be caveated by declaring that the dual-phase austenite and ferrite microstructural observation in this study should be evaluated in further detail in future studies. The measurement of the amount of δ -ferrite concentrations are highly location dependent and have a range of variability within the LMD component. The scope of the present study encompasses a differentiation between the microstructural compositions at the fusion lines and the deposition track body locations. However, future studies should attempt to compensate for this degree of variation within the LMD build and establish a broader statistical analysis into

multiple localities within the LMD part and evaluate the suitability of the LMD properties to targeted engineering applications. Depending on the conditions and requirements of the application, the use of dual-phase stainless steel can be beneficial and exhibit better corrosion resistance and toughness compared to pure ferritic stainless steels and can exhibit higher strength characteristics than pure austenitic steels [167, 168]. Dual-phase stainless steels share the advantages of both austenitic and ferritic stainless steels due that enable its widespread applications in various engineering fields. By modifying the elemental compositions of the feed material powder, it is possible to promote the growth of either austenite or ferrite concentrations. Davidson et al. found that the growth of Ni elements led to an increase of austenite content improved the material plasticity and reduced the residual stress on clad materials [169].

An SEM image was taken near the deposited track patterns seen from the microscopy images. In the EDX analysis, Oxygen element was detected at the fusion lines area that is not seen in the deposition body area. This is due to the natural formation of a protective oxide layer after each track of AISI Stainless Steel 316L is deposited and exposed to the atmosphere. An SEM image was also taken within a deposited deposition body area. In the EDX mapping analysis for this image, columnar microstructures were observed in the Molybdenum, Chromium, and Iron element maps. This observation is the result of an AF solidification mode that leads to the formation of columnar δ -ferrites parallel to the solidification direction.

Chapter 4: Build Strategy Effect on Materials-Process-Performance Properties (Study 2)

4.1 Introduction

Anisotropy is defined as a property of a part that exhibits variations and differences in physical properties along different axes. The anisotropic properties in AM parts as a result of build strategy have been widely investigated by various researchers in the past few years. In general, the anisotropy is a result of the build's effects on the microstructure that grow from the substrate, arising from the layer by layer deposition process. Zhang et al. [170] demonstrated that the layer-based principle of AM techniques hold restrictions like the anisotropy of mechanical and structural properties and showed that tensile tests and SEM morphologies of specimen fractures built in differing orientations exhibit anisotropic properties. Guo et al. [111] reported an anisotropic trend for different build orientations during tensile testing, whereby the specimens with a horizontal build orientation exhibited a significantly higher tensile strength, yield strength, and elongation, as opposed to the specimens with a vertical build orientation due to the higher anisotropy of the dendritic microstructure. Kono et al. [171] showed distinct anisotropic properties in laser metal deposited Stainless Steel 316L parts, where the ductility is lower in the transverse direction as compared to the longitudinal direction. Differences in build orientations of AM components have a significant influence on the anisotropy of mechanical properties due to the differences in thermal gradients and cooling rates. For instance, Yu et al. [172] demonstrated by finite element modelling simulation that various deposition patterns like raster, offset out, offset in, and fractal generates different temperature gradients. Amine et al. [173] found that the deposition scan speed and laser power have a significant effect on the microstructural growth, and that the thermal history of LMD can be described as a series of reheating cycles, or discrete pulses of heat. Yadollahi et al. [174] explains that for LMD processes, one can control the degree of anisotropy by manipulating the scanning strategy and the build orientation to make the thermal history more uniform.

The motivation of this research is to characterize the effects of build strategy on LMD built parts and quantify its influence on the anisotropy in mechanical properties that arises from its microstructure. Numerous existing studies

have looked into anisotropic behaviour that arises from varying the build orientation of additively manufactured parts [110, 175], however little work has been done in investigating other sources of anisotropy like build strategy. Build orientation has been demonstrated to be a critical source of anisotropy in AM and much effort has been made in this area to minimise its effects. This study explores the extent of an additional source of anisotropy and measures its impact on the mechanical behaviour of the LMD build. An illustration of the difference in LMD build methodology between existing studies on build orientation and this study on build strategy is illustrated in Figure 4.1. Tensile specimens were extracted from build-up specimens in order to evaluate the build strategy's effect on its mechanical anisotropy. Microstructural studies were conducted in order to observe the grain growth differences between each build strategy and correlate the grain characteristics with the anisotropy from the tensile tests.

Figure 4.1: Illustration of LMD build orientation types from existing studies on anisotropy (a) base build orientation, (b) longitudinal build orientation, (c) transverse build orientation, and illustration of LMD build strategy types from this study on alternative sources of anisotropy (d) long unidirectional build strategy, (e) bidirectional build strategy, (f) short unidirectional build strategy

Tensile specimens were extracted from these LMD blocks in order to evaluate the build strategy's effect on the tensile properties' anisotropy. Process data like melt pool size and temperature and laser power regulation in response to a closed-loop control implementation were recorded and analysed alongside the build strategy

implementation by constructing a MatLab process plot for each characteristic. Melt pool heat maps were also constructed in ImageJ using the process data. The LMD specimens' volumes were measured by Laser coordinate measuring machine (CMM) in order to derive the powder efficiency per build strategy, and conclusions were made about the build strategy's effect on powder efficiency. Porosities, density, and defects were characterised for the LMD specimens to determine the quality of the LMD builds. Microscopy depth of focus reconstruction was used to analyse the fracture surface topologies from the tensile test fractures to postulate the possible source of the anisotropic tensile behaviour. Lastly, micro-indentation techniques were used to verify the fracture surface topology analyses.

4.2 Methodology

4.2.1 Laser Metal Deposition Process

A DMG MORI LaserTec 65 *3D* machine was used to fabricate AM built LMD specimens from Stainless Steel 316L powder feed using three different build strategies: long unidirectional, bidirectional, and short unidirectional raster scan deposition process. The LMD process used in this experiment is illustrated in a schematic diagram as shown in Figure 4.2, whereby a diode laser is used as a laser source, and argon as the shielding gas. The metal powder was fed coaxially and melted by a diode laser to form a melt pool at the substrate. The deposition head moves with accordance to the defined build strategies' scan paths and builds the part layer by layer.

Figure 4.2: LMD schematic diagram

The LaserTec 65 3D has an integrated closed-loop control system to monitor the build process in real time. The process monitoring system uses a charge-coupled device (CCD) camera to monitor the melt pool during the build process. The melt pool characteristics, like melt pool size and melt pool temperature, are measured in-situ and fed to the APC to adjust the laser power according to user-defined closed-loop control parameters. The build parameters and closed-loop control parameters are specified in Table 4.1

Table 4.1: Build parameters and closed-loop control parameters employed in LMD process

Build Parameters	1.	Laser Spot Diameter	3 mm
	3.	Deposition Rate	14 g/min
	4.	Scanning velocity	1000 mm/min
	5.	Track Overlap	60 %
	6.	Temperature Melt Threshold	1560 °C
Closed-loop control parameters	7.	Target Melt Pool Area	8000 px
	8.	Laser Power Boundary	Min = 700 W, Max = 1800 W
	9.	Maximum Laser Power Ramp	50.0 W/sec
	10.	Melt Pool Size Boundary	Upper = 12000 px, Lower = 6000 px

4.2.2 Powder Material

The powder material used in the LMD process was Stainless Steel 316L. The SS316L powder feed is gas atomized spherical metallic granules prepared by Carpenter. A Scanning Electron Microscope (SEM) image of the powder material with a particle size range of 45 μm – 105 μm was taken using a Zeiss EVO HD 25 and is shown in Figure 4.3. The LMD process was fabricated on a mild steel base plate substrate of dimensions 250 mm $\times$ 250 mm $\times$ 9 mm. The chemical composition of the SS316L powder feed is shown in Table 4.2.

Figure 4.3: SEM image of powder feed particles

Table 4.2: Nominal chemical composition of SS316L powder feed (wt.%)

Material	C	Mn	P	S	Si	Cr	Ni	Mo	Fe
SS316L	0.03	2.00	0.045	0.03	1.00	16.00-18.00	10.00-14.00	2.00-3.00	Bal.

4.2.3 LMD Build Specimens

Three specimens were fabricated for each build strategy type: long unidirectional, bidirectional, and short unidirectional raster scan. The long unidirectional raster scan build strategy employs a raster scan pattern that runs longitudinally, parallel to the long edge of the specimen. For unidirectional raster scan patterns, the deposition track overlap was set at 60%. The short unidirectional raster scan build strategy's deposition method is similar to the long unidirectional raster scan build strategy, with the exception that the raster scan runs transversely, parallel to the short edge of the specimen. The bidirectional raster scan build strategy alternates between a longitudinal and transverse raster scan pattern between the deposited layers. For all build strategies, the scan path initiation point per layer is situated diagonally across from the previous layer's initiation point. The build strategy scan patterns as described are illustrated in Figure 4.4. The dimensions of the block specimens produced by LMD are 206 mm long, 26 mm wide, and 9 mm thick. These dimensions were chosen based on an inclusion of a 3 mm overbuild at all sides to allow for tensile specimens to be extracted from the LMD block specimens. Tensile specimens dimensions of 60 mm long, 12.5 mm wide, 4 mm thick were selected with accordance to ASTM E8/E8M [176]. To ensure that the LMD samples were built to specification and has the build volume required to extract tensile specimens which were verified using an articulated laser coordinate-measuring machine (CMM).

Figure 4.4: Laser CMM scans of LMD specimens built using (a) long unidirectional raster scan, (b) bidirectional raster scan, and (c) short raster scan build strategies

Tensile specimens were subsequently extracted from each of the nine LMD built blocks using milling and grinding techniques. Stainless Steel 316L specimens were also extracted from stock bar material from extrusion followed by cold rolling process, with accordance to ASTM A276/A276M-17 [177], and was used as a point of reference for comparison with the specimens produced by LMD. A total of twelve tensile specimens were fabricated, with three repetitions for the four experimental test specimen conditions as shown in Table 4.3. The over-build is removed later by wire-cutting, milling, and grinding, to achieve the desired surface finish for testing.

Figure 4.5: Tensile specimens for four experimental test specimen conditions
(a) long unidirectional raster scan, (b) bidirectional raster scan, (c) short unidirectional raster scan and (d) stock

Table 4.3: LMD Specimen dimensions

	ASTM-E8M-16a: Sheet type (mm)	LMD specimen dimensions: inclusive of over-build (mm)
T – Thickness	3	$3 + 3 + 3 = 9$
L – Length	200	$200 + 3 + 3 = 206$
C – Width of grip	20	$20 + 3 + 3 = 26$

4.2.4 Experimental procedure

Tensile tests were conducted on the specimens using a Shimadzu Tensile Tester equipped with a 50 kN load cell with a strain rate of 0.05 /min, and a digital video extensometer was used for strain measurement. After the tensile tests, the specimens' fracture surface topologies were scanned using the Zeiss Smart Zoom 5, via a 3D depth-of-focus reconstruction method using 34 times magnification with a 30 μm resolution. The specimens were sectioned at the grip length area, 30 mm away from the edge, and the sectioned surface was etched in a solution of 5% nitric acid, 5% hydrochloric acid, and 90% ethanol. The section location and measurement surface are illustrated in Figure 4.6.

Figure 4.6: Specimen section location on tensile specimen for micrograph and micro-indentation measurements

Micrographs of the microstructure were captured using a Zeiss Light Microscope and characterised using Fiji software. To measure the grain size, the ASTM E112-13's [178] Abrams Three-Circle procedure was used due to its application method for non-equiaxed grain structures, typically observed for LMD build materials. The results were then plotted against the yield strength from the tensile tests and correlated via the Hall-Petch relationship. The Hall-Petch relationship describes the correlation between microstructural grain size and yield stresses via the following equation [179, 180]:

$$\sigma_y = \sigma_0 + \frac{k}{\sqrt{d}} \quad (4.1)$$

Where σ_y is yield strength, σ_0 is a material constant that governs the stresses for grain dislocation movement, k is the Hall-Petch slope coefficient, and d is the grain width. The Hall-Petch relationship generally describes how the yield strength of the material increases with decreasing microstructural grain width.

SEM images were captured using Zeiss EVO SEM equipment and analysed with Inca software. Aztec software was further used to conduct Energy-dispersive X-ray spectroscopy (EDX) analysis. SEM was set at Backscatter

Electron (BSE) mode, 300 times magnification, 9 mm working distance and an Extra High Tension (EHT) voltage level of 15kV.

Subsequently, micrographs of the deposition track body and fusion zones were captured using the Zeiss Light Microscope. Vickers micro hardness tests were conducted with accordance to ASTM E384-17 [181] using an Innova Test Falcon 500 micro hardness tester. A micro-indentation load of 500 g and a 15 second dwell time were used at fusion zones and deposition track body locations.

4.3 Results and Discussion

4.3.1 LMD Process Evaluation

During the LMD process, the melt pool temperature, melt pool size, and laser power process data were captured and monitored using an in-situ system. A CCD camera positioned along the main optical axis captures the melt pool's luminance and the process monitoring system converts the luminance data to a temperature reading per pixel and produces a melt pool temperature and melt pool size reading at each capture instance. The sampling rate used in the recording of the process data was 10 samples per second. This is shown in Figure 4.7.

Figure 4.7: Empirical melt pool heat map model recorded at capture time instances (a) 0 s; (b) 0.5 s; (c) 1.0 s

From each sample's heat mapping, the melt pool size, melt pool temperature characteristics were measured and recorded. The process monitoring data (mean melt pool temperature, melt pool size, and laser power input per deposition layer) was subsequently modelled respective to their XYZ coordinates and the laser power input as

shown Figure 4.8, Figure 4.9 and Figure 4.10. A high degree of melt pool homogeneity was achieved during LMD process as a result of the in-situ adjustment of the laser power input as seen in Figure 4.8 and Figure 4.9. The mean laser power input across the LMD build for each condition was 1440 ± 45 W for the long unidirectional raster scan condition, 1304 ± 55 W for the bidirectional raster scan condition, and 1330 ± 77 W for the short unidirectional raster scan condition. To maintain a high degree of melt pool homogeneity, the laser power regulation for each condition varies widely, with the long unidirectional raster scan condition requiring approximately 8-10% more energy input than the other two conditions.

The first deposition layer exhibited a lower melt pool temperature and smaller melt pool size due to the initial thermal conditions of the LMD process. The substrate's temperature at the point of the LMD initiation is cooler and hence requires a higher laser power input to achieve the necessary energy required to melt the deposited powder particles in the first deposition layer as seen in Figure 4.11 (c). Beyond the first deposition layer, the heat energy and dissipation reaches a quasi-equilibrium state that results in a consistent mean melt pool temperature and melt pool size as seen in Figure 4.11 (a) and (b). Each new deposition layer's initiation point is located diagonally across the previous layer's initiation point. During the deposition head's translation to the new initiation point for the next layer, the previously deposited layer undergoes a significant cooling effect that necessitates a higher laser power input to achieve the target energy density. Furthermore, as more layers are deposited, lower laser power input is required for the LMD process due to the thermal energy build-up within the component body, that acts as a pre-heating element for each new layer that is deposited as seen in the decreasing laser power input reading from layer to layer Figure 4.11 (c). The long unidirectional raster scan's pattern that runs longitudinally across the long edge of the LMD block that produces a longer interval from one raster deposition path to the next, compared to the short unidirectional raster scan. The longer interval allows the area adjacent to the deposition area to cool down significantly before the subsequent deposition path in the raster pattern is produced. Hence, a higher energy density in the laser input is required to compensate for this cooling effect that results in a higher laser input for the long unidirectional raster scan build strategy than the other build strategies as seen in Figure 4.11 (c).

Figure 4.8: LMD empirical melt pool temperature process data plot for (a) short unidirectional, (b) long unidirectional, (c) bidirectional raster scan

Figure 4.9: LMD empirical melt pool size process data plot for (a) short unidirectional, (b) long unidirectional, (c) bidirectional raster scan

Figure 4.10: LMD empirical melt pool size process data plot for (a) short unidirectional, (b) long unidirectional, (c) bidirectional raster scan

Figure 4.11: LMD process monitoring mean data per deposition layer for (a) melt pool size, (b) melt pool temperature and (c) laser power input

The empirical melt pool temperature and size process data plots exhibited a high degree of melt pool homogeneity throughout the process, and little difference when comparing between build strategies, as seen in the melt pool temperature and melt pool size process plots in Figure 4.8 and Figure 4.9. In order to attain process consistency, the laser power was regulated differently between build strategies, as seen in Figure 4.10. The temperature of the substrate prior to the LMD process is cooler and hence requires a higher laser power regulation to compensate for this initial condition for the first deposition layer. For the subsequent deposition layers, the method in which the raster scan is implemented for the LMD process is such that each new deposition layer starts diagonally across from the previous layer's initiation coordinates. At the beginning of each new deposition layer, the LMD build would have cooled down significantly. Hence, it is necessary for the laser power to be regulated to a higher value to compensate for this cooling effect. As the layer is deposited, the component rises in temperature and a lower laser power is required to maintain the melt pool temperature and size within the user-defined boundaries, and

hence the laser power is regulated to a lower value. This effect occurs cyclically as the LMD block is built layer by layer.

4.3.2 Powder Efficiency

The volume of the LMD builds are measured by laser CMM using a Hexagon HP-L-20.8 Laser Scan mounted on a ROMER Absolute Arm. The stereolithography (STL) scans are subsequently processed in PolyWorks, RDS Software, and Autodesk MeshMixer. The processed STL scans are seen in Figure 4.12. Using the volume measured, the powder efficiency for each LMD block specimen is calculated using equation (4.1) :

$$k_d = V_m \div (G_p \times t \div \rho_p) \quad (4.1)$$

Where k_d is powder efficiency, V_m is the measured volume, G_p is the powder deposition rate, t is the process time, and ρ_p is the density of the deposited powder material. The results are shown in Table 4.4.

Table 4.4: Powder efficiency results for LMD specimens per build strategy

Specimen No.	Short Unidirectional Raster Scan	Long Unidirectional Raster Scan	Bidirectional Raster Scan
1	55.8%	58.8%	54.9%
2	55.0%	63.4%	57.6%
3	55.9%	61.5%	60.4%
Mean	55.6%	61.2%	57.6%

From Table 4.4, the long unidirectional raster scan build strategy exhibited the highest powder efficiency, followed by bidirectional, and short unidirectional raster scan respectively. This difference in powder efficiencies between build strategies is a result of the manner in which the raster scan was implemented. During the movement of the deposition head from one deposition track to the next, the laser is momentarily turned off, and is turned on again at the beginning of the subsequent deposition track. This is done to improve the deposited volume consistency within the LMD build. During the laser off instances, the deposition powder flow is maintained due to the amount of time it takes to reinitiate a stable powder flow. For an LMD block with rectangular dimensions, as used in this experiment, the different build strategies would result in different number deposition tracks, and

hence a different number of laser off instances as illustrated in Figure 4.12. The short unidirectional raster scan would have the highest number of laser off instances, followed by the bidirectional, and long unidirectional raster scan respectively. Since less powder bonds during the laser off instances, the powder efficiency would hence adhere to the same trend per build strategy.

Figure 4.12: Laser ON/OFF implementation for each build strategy type
(a) short unidirectional raster scan, (b) long unidirectional raster scan, and
(c) bidirectional raster scan

4.3.3 Porosity Analysis

The LMD SS316L were sectioned using a bench saw, extracting three cross-section specimens. These specimens were subsequently hot mounted in resin under heat and high pressure, and polished for density and porosity analysis by microscopy. Microscope images were captured, stitched, and analysed using a Zeiss Light Microscope as seen in Figure 4.13. Specimen density was measured using Zeiss software and porosity characteristics were analysed using ImageJ software.

Figure 4.13: LMD cross-section microscopy images for each build strategy type
(a) short unidirectional raster scan, (b) long unidirectional raster scan, and (c) bidirectional raster scan

The LMD Stainless Steel AISI 316L's object count distribution per porosity size class for each build strategy are compiled in Figure 4.14, and overall density and object count per target measurement area is consolidated in Table 4.5. The majority of the objects detected concentrated within the Class 1 ($0.0\ \mu\text{m} - 15.0\ \mu\text{m}$). All build strategies exhibited a high degree of part density of more than 99.97%. The small sized pores detected suggests minor issues with miniscule air entrapments during the LMD process, with no severe build defects. This indicates a good quality LMD build process, and that the user-defined LMD process parameters are satisfactory.

Figure 4.14: Object count distribution per pore size class for build strategy

Table 4.5: Density and object count detection per build strategy

Build Strategy	Part Density	Object Count
Bidirectional	99.97%	371
Long Unidirectional	99.98%	1328
Short Unidirectional	99.98%	1000

4.3.4 Tensile Test

A series of tensile tests were conducted on the tensile specimens, as seen from Figure 4.5, using a Shimadzu Tensile Tester equipped with a 50 kN load cell. The tensile test was conducted with accordance to ASTM standards [176]. A strain rate of 0.05 /min was used, and strain was measured using a digital video extensometer. Results for each experiment condition's ultimate tensile strength (UTS), Young's modulus, 0.2% yield strength, and strain taken using the digital video extensometer, were measured and compiled in Figure 4.15. The stress-strain relations and the reduction of area (ROA) are given as:

$$\sigma = \frac{F}{A_0} \quad (4.2)$$

$$\varepsilon = \frac{l - l_0}{l_0} \quad (4.3)$$

$$ROA\% = \frac{A - A_0}{A_0} \times 100\% \quad (4.4)$$

Where σ is engineering stress, ε is the fracture strain, l is the length of the specimens after elongation, l_0 is the original length of the specimens, ROA% is the reduction of area %, A is the cross-sectional fracture surface area at fracture, and A_0 is the original cross-sectional area.

The stress parameters were established using the original cross-sectional area of the tensile specimen, and the strain measurements were established using the original gauge length. These parameters were used in establishing the engineering stress-strain results are relations that reflect the uniform strain regime. As the load increases beyond the uniform strain regime, the cross-sectional area at the failure region undergoes a rapid reduction also known as necking. During this stage, the distribution of stresses and strain across the necking region become complex and quantifying the magnitudes of the stress-strain quantities become challenging to determine. The necking phase results in large quantities of axial deformation and is also known as the non-uniform stress-strain regime. The rapid changes in the dimensions of the tensile specimens at this deformation stage up to the point of fracture need to be considered on top of evaluating the engineering stress and strain. Using true stress instead of engineering stress can give a more direct measure of the material's response in the plastic flow range. True strain is expressed as:

$$d\varepsilon_t = \frac{dL}{L} \rightarrow \varepsilon_t = \int_{L_0}^L \frac{1}{L} dL = \ln \frac{L}{L_0} \quad (4.5)$$

Where ε_t is the true strain, L is the current specimen length, an L_0 , is the original specimen length. During the yield and plastic-flow regime that follows after the yield point, the material undergoes negligible changes in volume, and any increase in specimen length is offset by a decrease in cross-sectional area. Before the necking phase, the strain is considered uniform along the specimen length. The volume constraint can hence be expressed as:

$$dV = 0 \rightarrow AL = A_0L_0 \rightarrow \frac{L}{L_0} = \frac{A_0}{A_f} \quad (4.6)$$

Where V is the specimen volume, A is the current cross-sectional area, and A_0 is the original cross-sectional area.

From here, the relationship between true and engineering measures of stress and strain can be expressed as:

$$\sigma_t = \sigma_e(1 + \varepsilon_e) = \sigma_e \left(\frac{L}{L_0} \right), \quad \varepsilon_t = \ln(1 + \varepsilon_e) = \ln \left(\frac{L}{L_0} \right) \quad (4.7)$$

Where σ_t is true stress, σ_e is engineering stress, and ε_e is engineering strain. These equations can be used to plot a stress-strain curve from the engineering curve, up to the strain at which necking begins. After the necking phase initiates, the strain is non-uniform at the gage length and computing the true stress-strain curve beyond this point is not meaningful. However, a complete true stress-strain curve could be drawn if the necking area is monitored in real-time throughout the tensile test. The evaluation of stress-strain parameters at fracture is also known as the true stress and true strain where:

$$\frac{L}{L_0} = \frac{A_0}{A} \rightarrow \varepsilon_t = \ln \frac{L}{L_0} = \ln \frac{A_0}{A} \quad (4.8)$$

Figure 4.15: Stress-strain data for stock, and three different LMD built strategy samples

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Tensile test results for each experimental test specimen condition's UTS, Young's modulus, yield strength, fracture strain, and ROA are shown from Table 4.6 to Table 4.12

Table 4.6: UTS results

Build Strategy	N	UTS Results (MPa)			Mean	Std Dev
Long Unidirectional	3	634	652	658	648	10
Bidirectional	3	622	629	632	628	4
Short Unidirectional	3	618	617	617	617	1
Stock	3	604	602	607	605	2

Table 4.7: Young's Modulus results

Build Strategy	N	Young's Modulus Results (GPa)			Mean	Std Dev
Long Unidirectional	3	170	182	164	172	8
Bidirectional	3	267	279	211	252	30
Short Unidirectional	3	283	296	288	289	5
Stock	3	183	207	189	193	10

Table 4.8: Yield strength results

Build Strategy	N	Yield Strength Results (MPa)			Mean	Std Dev
Long Unidirectional	3	436	443	437	439	3
Bidirectional	3	439	406	436	427	15
Short Unidirectional	3	348	400	407	385	27
Stock	3	322	350	329	334	12

Table 4.9: Fracture strain results

Build Strategy	N	Fracture Strain Results			Mean	Std Dev
Long Unidirectional	3	0.60	0.62	0.58	0.60	0.02
Bidirectional	3	0.52	0.56	0.54	0.54	0.02
Short Unidirectional	3	0.52	0.50	0.54	0.52	0.02
Stock	3	0.82	0.78	0.82	0.81	0.02

Table 4.10: Reduction of area results

Build Strategy	N	Reduction of Area Results (%)			Mean	Std Dev
Long Unidirectional	3	12.65	4.70	12.78	10.04	3.78
Bidirectional	3	14.93	14.17	24.26	17.79	4.59
Short Unidirectional	3	26.60	33.04	26.74	28.79	3.01
Stock	3	62.12	62.29	62.05	62.16	0.10

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Table 4.11: True stress at fracture results

Build Strategy	N	True Stress Results			Mean	Std Dev
Long Unidirectional	3	864.06	927.36	898.18	896.53	25.87
Bidirectional	3	731.60	733.46	834.92	766.66	48.27
Short Unidirectional	3	707.67	647.34	707.65	687.55	28.43
Stock	3	1594.47	1597.42	1600.79	1597.56	2.58

Table 4.12: True strain at fracture results

Build Strategy	N	True Strain Results			Mean	Std Dev
Long Unidirectional	3	0.31	0.40	0.31	0.34	0.043
Bidirectional	3	0.16	0.15	0.28	0.20	0.057
Short Unidirectional	3	0.14	0.05	0.14	0.11	0.041
Stock	3	0.97	0.98	0.97	0.97	0.003

In order to test if LMD build strategy has an effect on mechanical properties, an analysis of variance (ANOVA) was conducted with a confidence level of 95%, $\alpha = 0.05$, and the stock condition was excluded. The exclusion of the stock condition was done because the stock specimens were not produced by LMD techniques and would confound the statistical analysis. The ANOVA results for the UTS, Young's modulus, yield strength, fracture strain, reduction of area, true stress at fracture, and true strain at fracture are shown from Table 4.13 to Table 4.19.

Table 4.13: Analysis of variance for UTS

Source	DF	Adj SS	Adj MS	F-Value	P-Value
Build Strategy	2	1453.5	726.74	12.14	0.008
Error	6	359.2	59.87	-	-
Total	8	1812.7	-	-	-
S = 7.74 $R^2 = 80.18\%$					

Table 4.14: Analysis of variance for Young's Modulus

Source	DF	Adj SS	Adj MS	F-Value	P-Value
Build Strategy	2	21512	10756.2	22.20	0.002
Error	6	2907	484.5	-	-
Total	8	24420	-	-	-
S = 22.01 $R^2 = 88.09\%$					

Table 4.15: Analysis of variance for yield strength

Source	DF	Adj SS	Adj MS	F-Value	P-Value
Build Strategy	2	4882	2441.2	5.28	0.048
Error	6	2774	462.3	-	-
Total	8	7656	-	-	-
S = 21.50 $R^2 = 63.77\%$					

Table 4.16: Analysis of variance for fracture strain

Source	DF	Adj SS	Adj MS	F-Value	P-Value
Build Strategy	2	0.0104	0.0052	13.00	0.007
Error	6	0.0024	0.0004	-	-
Total	8	0.0128	-	-	-
S = 0.02 $R^2 = 81.25\%$					

Table 4.17: Analysis of variance for reduction of area

Source	DF	Adj SS	Adj MS	F-Value	P-Value
Build Strategy	2	532.6	266.32	12.01	0.008
Error	6	133.1	22.18	-	-
Total	8	665.7	-	-	-
S = 4.70936 $R^2 = 80.01\%$					

Table 4.18: Analysis of variance for true stress at fracture

Source	DF	Adj SS	Adj MS	F-Value	P-Value
Build Strategy	2	66798	33399	17.54	0.003
Error	6	11423	1904	-	-
Total	8	78221	-	-	-
S = 43.6334 $R^2 = 85.40\%$					

Table 4.19: Analysis of variance for true strain at fracture

Source	DF	Adj SS	Adj MS	F-Value	P-Value
Build Strategy	2	0.08334	0.041672	12.26	0.008
Error	6	0.02040	0.003400	-	-
Total	8	0.10374	-	-	-
S = 0.0583102 $R^2 = 80.34\%$					

Compiled results for each experimental test specimen condition's ultimate tensile strength (UTS), Young's modulus, yield strength, and fracture strain are shown in Figure 4.16. The compiled bar chart also illustrates the main effects for each condition.

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Figure 4.16: Tensile test results for long unidirectional, bidirectional, short unidirectional raster scan build strategies and stock experiment conditions' (a) UTS, (b) Young's modulus, (c) yield strength, (d) fracture strain (e) Reduction of area %, (f) true stress at fracture, (g) true strain at fracture

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From the ANOVA tables, the results for UTS, Young's modulus, yield strength, fracture strain, ROA, true stress at fracture, and true strain at fracture all have P-values < 0.05 , indicating that build strategy does has a significant effect on the tensile properties. The R^2 value is coefficient of determination and is the proportion of the variation in the dependent variable that is predictable from the independent variable. The general application of the R^2 value is its measure of well the observable outcome is replicated by the model, which is a function of the proportion of total variation of outcomes explained by the model. The R^2 values for each mechanical characteristic ranges between 60% and 90% approximately, which indicates that the regression model is a good fit the observed data. The stock material specimens exhibit a lower UTS and yield strength, and a higher fracture strain than the LMD specimens. The long unidirectional raster scan build strategy exhibited the highest UTS, yield strength, and fracture strain among the LMD specimens, followed by the bidirectional and short unidirectional raster scan build strategies respectively, and the reverse trend is true for the Young's modulus result. The stock specimen exhibited the highest ROA % followed by the long unidirectional, bidirectional, short unidirectional raster scan build strategies and stock respectively. The tensile test results indicate that there is a significant anisotropic tensile effect between the different build strategies.

4.3.5 Microstructure analysis

Microstructure images for the long unidirectional, bidirectional, short unidirectional raster scan build strategies and stock bar specimens are shown in Figure 4.17 and Figure 4.18.

Figure 4.17: Microstructure images for four experimental test specimen conditions (a) long unidirectional raster scan, (b) bidirectional raster scan, (c) short unidirectional raster scan build strategies and (d) stock bar material

Figure 4.18: SEM images for four experimental test specimen conditions (a) long unidirectional raster scan, (b) bidirectional raster scan, (c) short unidirectional raster scan build strategies and (d) stock bar material

In general, the key to understanding the mechanical behaviour of AM parts is in its microstructural properties. The solidification behaviour that is a result of the high-powered laser's interaction with the powder feed material results in highly rapid heating and melting of the material, forming a melt pool. This is very quickly followed by a rapid cooling and solidification, forming a deposition track. This process of melting and solidification result in a very specific type of microstructure pattern that could be characterised as typical for LMD-built parts.

The mean grain size for each specimen type condition was taken from the micrograph images as seen in Figure 4.17 and compiled in Table 4.20. To measure these grain characteristics, the ASTM Abrams three-circle procedure [182] was used to derive the mean grain size. The mean grain size for each condition was subsequently plotted against the experimental yield strength results in a Hall-Petch plot as shown in Figure 4.19. From the Hall-Petch

plot, a best fit linear trendline was applied to derive the σ_0 constant and k coefficients and were found to be 231 MPa and 647 MPa respectively.

The stock specimens exhibited the largest mean grain size, indicating the presence of large austenitic grain structures. These grain characteristics are consistent with the mechanical test data where the stock specimens exhibited low UTS values, yield strength values, and Young's modulus values, and the highest strain measurement values. The LMD specimens' microstructure observed from Figure 4.17 (a), (b), and (c) show needle-like, acicular crystallite phase. The acicular crystals are observably preferential in their orientation in the vertically oriented direction from the substrate. This type of grain growth nucleation is a result of the thermal history of the LMD process, resulting from heat flows through previously deposited layers and the substrate. The LMD process induced thermal gradients produces a non-steady-state solidification process with re-melting and preferential metallurgical coarsening. In a previous study [2], it was found that the rapid cooling rates during the LMD process result in the retention of δ -ferritic phases that strengthens the austenitic matrix. Similar observations were reported for LMD Stainless Steel 316L specimens by Akbari et al. [183] using x-ray diffraction techniques.

The rapid solidification that occurs during LMD results in a type of grain refinement that can be deemed as characteristic of AM produced parts. This is largely due to a rapid quenching effect that happens via conduction through the previously deposited layers and the substrate that results in a lack of grain development and growth. Hence, conventional full dendritic solidification of Fe-based materials in AM is typically not well developed in AM processes. Secondary dendrite arms do not have time to develop in such a rapid solidification sequence, hence forming a highly directional cellular microstructure [184, 185]. The crystal structure growth is kinetically limited by the direction of the maximum heat flow, driving the growth to occur in a single direction. This phenomenon is in direct competition with a non-steady-state solidification process that occur from the surface tension gradient and Marangoni convection at the melt pool [74, 186]. The resultant crystal structure growth result in a range of crystal orientations with localized regularity. This is the main driver for the anisotropic characteristics that occur in the LMD produced SS316L part.

Table 4.20: Grain size data used in Hall-Petch relationship and plot

	Grain size measurement	Hall-Petch model parameters	
	Mean Grain Size, d (μm)	$d^{-0.5}$ ($\mu\text{m}^{-0.5}$)	Yield Strength, σ_y (MPa)
1. Long Unidirectional	9.7 ± 0.9	0.32	439 ± 3
2. Bidirectional	13.0 ± 0.8	0.28	427 ± 15
3. Short Unidirectional	15.7 ± 1.2	0.25	385 ± 27

Figure 4.19: Hall-Petch plot correlation between $d^{-0.5}$, where d is the measured grain size, and experimental yield strength results for four experimental test specimen conditions (a) long unidirectional raster scan, (b) bidirectional raster scan, (c) short unidirectional raster scan build strategies

Wang et al. [110] reported the effects of LMD process parameters on microstructure and tensile properties whereby the microstructural grain growth at the heat affected zone (HAZ) was presented as a function of the initial temperature conditions and heat input conditions. In these models, the microstructural grain growth is diffusion controlled, driven by surface energy, and requires no nucleation. It is shown that a higher initial temperature leads to a larger microstructural grain size growth. The initial temperature can be taken as either a pre-heat temperature or thermal cycles that are a consequence of LMD's multi-pass depositions in raster scan build patterns. The differences in inter-track intervals per build strategy are attributed to the different thermal histories that arise from their respective deposition track lengths and number of deposition tracks per layer. During the LMD process, the

temperature at the initiation point of each consecutive short unidirectional raster scan path is higher due to the shorter length of each raster scan path and the heat generated from the previous deposition path, resulting in lower thermal gradient. The temperature at the initiation point of each long unidirectional raster scan path is lower due to its comparatively longer length of each raster scan path, resulting in a steeper thermal gradient.

Based on their respective thermal histories, the short unidirectional raster scan build strategy employed in this study would result in a larger grain size than the long unidirectional raster scan build strategy. This is verified from the grain size measurements as seen in Table 4.20, where the mean grain size is $15.7 \pm 1.2 \mu\text{m}$ for the short unidirectional raster scan, $13.0 \pm 0.8 \mu\text{m}$ for the bidirectional raster scan, and $9.7 \pm 0.9 \mu\text{m}$ for the long unidirectional raster scan build strategies. The grain size measurement results are consistent with the Hall-Petch relationship, whereby the short unidirectional raster scan samples, which has the largest mean grain size, has resulted in the lowest yield strength. The inverse is true for the long unidirectional raster scan samples. Furthermore, the Stainless Steel 316L samples produced by LMD have significantly higher yield strength than the stock samples due to the smaller grain sizes. Bahl et al., similarly, used a Hall-Petch analysis on additively manufactured Stainless Steel 316L and attributed the resultant high yield strength to a high dislocation density of the grains [187].

During the solidification process, ferrite becomes unstable compared to austenite with decreasing temperature. While the austenitic dendrites grow, some of the heavier elements begin to segregate, resulting in differences in the elemental composition at the interdendritic zones. As the solidification process continues, the remaining molten material retains ferrite stabilizing elements that favours the formation of δ -ferrite at the interdendritic areas [188]. The higher the Cr_{eq}/Ni_{eq} ratio, the lower the starting temperature of the transformation in contrast to the liquidus temperature. A high Cr_{eq}/Ni_{eq} ratio produces a rapid cooling effect resulting in a Widmanstätten effect, where the growth of new phases occurs within the grain boundaries of the parent metal. This solidification effect produces a non-equiaxed grain morphology that increases the hardness and results in embrittlement of the material [189].

The ferrite concentration within the LMD-built part can be estimated using Schaeffler's diagram [190] as seen in Figure 4.20. The x-axis and the y-axis are the Cr equivalent and Ni equivalent respectively. The Cr equivalent is

estimated from the wt. % of the ferrite stabilizing elements measured, and the Ni equivalent is estimated from the wt. % of the austenite stabilizing elements measured. The Cr equivalent and Ni equivalent equations are seen in Eq. (2) and (3). SEM images of the microstructure of each build strategy and the respective EDX area analysis locations are shown in Figure 4.21. The measured elemental composition (wt. %) for each build strategy and the calculated Cr_{eq} and Ni_{eq} using Eq. (4.9) and (4.10) are compiled in Table 4.21.

$$Ni_{eq} = Ni + (30 \times C) + (0.5 \times Mn) \quad (4.9)$$

$$Cr_{eq} = Cr + Mo + (1.5 \times Si) + (0.5 \times Nb) \quad (4.10)$$

Figure 4.20: Schaeffler diagram

Figure 4.21: SEM images and EDX area analysis locations for (a) long unidirectional raster scan, (b) bidirectional raster scan, and (c) short unidirectional raster scan

Table 4.21: EDS measured composition (wt. %) of long unidirectional raster scan, bidirectional raster scan, and short unidirectional raster scan measurement areas indicated in boxes 1-3 respectively from Figure 4.21, and Ni and Cr Equivalent

	C	Si	Cr	Mn	Ni	Mo	Fe	Cr _{eq}	Ni _{eq}	Cr _{eq} /Ni _{eq}
1	0.04	0.42	18.81	1.49	10.57	1.48	67.19	12.52	20.92	1.67
2	0.04	0.45	18.21	1.61	10.58	1.56	67.55	12.59	20.45	1.62
3	0.04	0.41	17.84	1.42	10.89	1.47	67.93	12.80	19.93	1.56

The Cr_{eq}/Ni_{eq} ratio values of all three build strategies were found to be within the range of 1.56 to 1.67. This places all three build strategy specimens to be on the lower end of the ferritic-austenitic solidification mode region, where a duplex austenite and δ -ferrite microstructural composition is expected. This is in agreement with the results of the Schaeffler's diagram for the LMD-built SS316L samples where a volumetric fraction of approximately 5% δ -ferrite was measured. The Cr_{eq}/Ni_{eq} ratio values were measured to be in a descending order from long unidirectional, to bidirectional, to short unidirectional raster scan build strategies. Since a higher

Cr_{eq}/Ni_{eq} ratio produces an embrittlement and hardening effect due to its differences in solidification, the trend in the Cr_{eq}/Ni_{eq} ratios between each build strategy suggests that the long unidirectional raster scan has a higher hardness property, followed by bidirectional and short unidirectional raster scan build strategies respectively. By applying Tabor's model for hardness of austenitic steels [191], the Cr_{eq}/Ni_{eq} trend between each build strategy is consistent with the measured yield strength from Figure 4.16 (b), where long unidirectional raster scan build strategy exhibited the highest yield strength, followed by the bidirectional and short unidirectional raster scan build strategies respectively.

The dynamic thermal history and interactions that occur during the deposition process of LMD results in a variable of metallic powder melt and solidification states across the build. The rapid cooling associated with LMD leads to a non-equilibrium phase transformation, which results in the transformation of austenite phase into martensite in the matrix. The formation of austenite from δ -ferrite is a diffusion-controlled phase transformation and its completion is strongly time-dependent, however, the LMD process provides a narrow window for this to occur due to the rapid solidification that occurs in the fusion zone.

Murterlle et al. investigated the influence of δ -ferrite on mechanical properties of AISI Stainless Steel 316L and found that δ -ferrite increases hardness and both yield strength and ultimate tensile strength correspondingly, while decreasing ductility [192]. It was noted in the same study that although δ -ferrite can have an effect on these properties, it is possible for these changes in mechanical properties to be overshadowed by other properties like defects and density as well as microstructural texture. Although it was recorded that it is possible for the content of δ -ferrite to have an impact on the mechanical properties of the part, the amount of δ -ferrite % that is needed in variation to produce a significant enough difference that is comparable to the present study is a very large. In order to produce a difference of 15 HV hardness, the difference in δ -ferrite content had to be of a difference between <1% and 9% δ -ferrite content.

Additionally, although the loss of elements are possible in LMD, the loss rate of elements have been reported to be low or negligible [193]. Although generally the application of Schaeffler diagram and calculations into Ni/Cr ratios can hold true to a significant degree, the considerations of any loss of elements during the deposition process means that the computed results made in this section can hold some degree of error. LMD AISI Stainless Steel

316L In order to interpret the results observed in the differences in mechanical properties between each build strategy, a combination of an analyses into both the solidification mode and the grain sizes have to be utilized.

4.3.6 Fracture surface topology

The 3D depth-of-focus reconstruction and 2D extended depth-of-field images of the fracture surface topologies for each experimental test specimen conditions are shown in Figure 4.22 and Figure 4.23 respectively.

Figure 4.22: Tensile specimen fracture surface 3D reconstruction topologies for the four experimental test specimen conditions (a) long unidirectional raster scan, (b) bidirectional raster scan, (c) short unidirectional raster scan, and (d) stock

Figure 4.23: Micrograph of fracture surface for the four experimental test specimen conditions (a) long unidirectional raster scan, (b) bidirectional raster scan, (c) short unidirectional raster scan, and (d) stock

The LMD specimens from Figure 4.22 and Figure 4.23 (a, b, c) exhibited little cross-sectional area reduction and distinct brittle fracture cleavage lines. The acicular grain crystal morphologies in the LMD specimens induce a strengthening effect that is evident in the brittle mechanical property observed from the tensile tests and are consistent with the fracture surface topologies. The brittle fracture surface topology observed in the LMD Stainless Steel 316L samples can be attributed to its martensitic structure [194]. The long unidirectional raster scan from Figure 4.22 (a), exhibited an irregular, 45° fracture surface topology resembling a longitudinal anisotropic fracture regular 45° fracture surface topology resembling a transverse anisotropic fracture effect. The bidirectional raster scan condition from Figure 4.22 and Figure 4.23 (b), the fracture surface topology exhibited features that are a combination of that observed from the long and short unidirectional raster scan condition. The fracture surface topologies for the stock specimen condition shown in Figure 4.22 and Figure 4.23 (d) exhibited significant necking, with a typical ductile cup-and-cone topology, and multiple identifiable ductile dimples marked on the 2D extended depth-of-field images. The brittle fracture topology observed from the LMD specimens were observed to have an inclination of approximately 45° to the principal stress direction which indicates that shear stress is the crack initiation mechanism. The LMD specimens were measured to have a shear crack at the gage length location with a stable crack propagation with cleavage and tearing morphology typical of brittle fracture. These observations are consistent with the high degree of plastic strain measured from the instrumented tensile experiments, and the large grain crystal structures where the LMD specimens have a lower degree of ductility and ROA than the stock specimens.

Chen et al. [195] reported that laser additively manufactured parts are susceptible to liquation cracking at the fusion zone localities that propagate along the interdendritic region in the Heat Affected Zone (HAZ). Specific types of defects have the propensity to cluster between deposition layers due to the non-steady-state deposition process. Different arrays of deposition tracks within LMD components built with different build strategies play a role in contributing to the degree of anisotropy. To verify the role of the array of deposition track patterns on the mechanical anisotropy, a series of micro-indentation measurements was conducted at the fusion zone and deposition body locations.

4.3.7 Micro-indentation

The samples from Figure 4.13 were etched in 5% nital acid etchant in order to reveal the fusion lines. The fusion lines are distinguished from the deposition body zones by a lighter colour. Images of the specimen's cross-sectioned etched surface were captured using a Zeiss Smart Zoom 5, and the fusion lines are shown in Figure 4.24.

Figure 4.24: Microscope images of etched cross-sectioned LMD specimens for each build strategy (a) bidirectional raster scan, (b) short unidirectional raster scan, and (c) long unidirectional raster scan

Next, using the Agilent Technologies Nano Indenter G200, a comparison of the micro-indentation properties at the fusion lines and deposition bodies' locations were conducted. In order to define the indentation depth value, the dimensions of the projected indentation area and the intended measurement target have to be considered. The Agilent Technologies Nano Indenter G200 uses a Berkovich tip type as seen in Figure 4.25.

Figure 4.25: Berkovich Tip (a) dimensions and, (b) projected indentation area

Where, h is the indentation depth, a is the base length, b is the side length, l is the height, and z is the edge-to-tip distance. The indentation depth and projected indentation area are correlated by the equation (4.2):

$$A_{proj} = 3\sqrt{3} h^2 \tan^2 65.3^\circ \quad (4.2)$$

Where A_{proj} is the projected indentation area, and h is the indentation depth. From equation (4.2), an indentation depth of $5 \mu\text{m}$ produces a projected indentation area of $614 \mu\text{m}^2$. The projected indentation dimensions' base length a , and height l are hence derived to be $a = 32.6 \mu\text{m}$, $l = 37.6 \mu\text{m}$ and was intentionally selected to be dimensionally suitable to fit and detect the fusion lines. The micro-indentation settings are compiled in Table 4.22.

Table 4.22: Micro-indentation Settings

	Parameter	Value
1.	Allowable Drift Rate	0.10 mm/s
2.	Indentation Depth	$5 \mu\text{m}$
3.	Frequency Target	45.0 Hz
4.	Harmonic Displacement Target	2.0 nm
5.	Percent To Unload	90 %
6.	Strain Rate Target	0.05 /min

Eight micro-indentation measurement locations were selected for each fusion line and deposition body and is repeated for each build strategy type as seen in Figure 4.26. 24 micro-indentation measurements were taken for fusion lines and deposition bodies locations each.

Figure 4.26: Micro-indentation at fusion line and deposition body locations for each build strategy (a) bidirectional raster scan, (b) short unidirectional raster scan, and (c) long unidirectional raster scan

The elastic modulus and hardness characteristics are evaluated and calculated using equations (4.3) and (4.4) from the micro-indentation tests respectively. The micro-indentation results are compiled in Figure 4.27, Figure 4.29 and Figure 4.28.

$$\frac{1}{E_r} = \frac{1 - \nu^2}{E} + \frac{1 - \nu_i^2}{E_i} \quad (4.3)$$

Where E_r is the reduced elastic modulus, E is the elastic modulus of the sample, E_i is the elastic modulus of the indenter tip, ν is the Poisson's ratio of the sample, and ν_i is the Poisson's ratio of the indenter.

$$H = \frac{P}{A} \quad (4.4)$$

Where H is hardness, P is the load applied to the sample, and A is the contact area.

Figure 4.27: Micro-indentation results for (a) hardness at deposition body location, (b) hardness at fusion line's location

Figure 4.28 (a) Young's modulus at deposition body location, and
(b) Young's modulus at fusion line's location

Figure 4.29: Micro-indentation results for deposition body and fusion lines
(a) hardness, and (b) Young's modulus

From Figure 4.29, the fusion lines exhibited higher hardness and Young's modulus characteristics than deposition body. The fusion lines are harder and stiffer than the deposition body locations. Since the array of fusion lines within the LMD blocks for each build strategy are distributed differently for each build strategy, this would in turn affect each build strategy component's mechanical properties. This difference is responsible for the fracture propagation differences seen in the fracture surface topologies from Figure 4.22, as well as the anisotropic tensile properties in different build strategies observed from Figure 4.16. An earlier EDX analysis from Chapter 3.3.3 indicated traces of chromium oxides at the fusion line's locations. This is a result of a protective layer of oxide that forms at the deposition track surface during the deposition process. This difference in elemental composition between the fusion lines and deposition body, when taken into consideration with the micro-indentation results,

suggest that the fusion lines have a significant role to play in the anisotropic tensile behaviour of the LMD specimens.

The section surface location for the micro-indentation measurements was illustrated earlier in Figure 5. Images of the specimen's prepared etched surface were captured using a Zeiss Light Microscope and shown in Figure 4.30. Micro-indentation tests were conducted to characterise the micro-hardness differences between the fusion zone region and deposition track body locations. Eight micro-indentation measurements were conducted in a line at the fusion zone and deposition track body locations each per specimen as shown in Figure 4.31. Measurements were repeated on three LMD specimens, one per build strategy, and the result is summarized in Table 4.23.

Figure 4.30: Microscope images of etched cross-sections of LMD specimens for each build strategy (a) long unidirectional raster scan, (b) bidirectional raster scan, and (c) short unidirectional raster scan

Figure 4.31: Microscope images of micro-indentation sites at (a) fusion zone and (b) deposition track body locations

Table 4.23: Micro-indentation measurements at deposition track body and fusion zone locations

Measurement Location	Hardness (HV)
Deposition Track Body	221.4 ± 6.8
Fusion Line	255.4 ± 5.5

From Table 4.23, the fusion zones exhibited higher hardness values than the deposition track body. The brittle characteristics of the fusion zones serve as fracture initiation sites during the tensile tests that result in the observed fracture surface topologies in Figure 4.22, whereby a transverse and longitudinal anisotropic fracture effect were observed for the short and long unidirectional raster scan build strategies respectively. These findings are also consistent with the anisotropic behaviour in mechanical properties, like a lower UTS and yield strength for the short unidirectional than the long unidirectional raster scan build strategies seen from Figure 4.16.

Yu et al. [196] found that crack propagation tend to occur at the interdendritic regions of LMD Stainless Steel 316L due to the differences in the material constitutions. Since, the array of fusion zones within the LMD builds are distributed differently for each build strategy, the mechanical properties would hence reflect this difference due to the fracture propagation that occurs across the different arrays of fusion zones as illustrated in Figure 4.32.

Figure 4.32: Schematic of LMD tensile specimens under tension for (a) long unidirectional raster scan specimen undergoing longitudinal fracture effect and, (b) short unidirectional raster scan specimen undergoing transverse fracture effect

4.4 Conclusion

In this study, the effects of Additive Manufacturing build strategies using LMD fabrication of Stainless Steel 316L specimens provided a framework for characterizing the anisotropic properties and behaviour arising from build strategy pattern effects. A series of tensile tests were used to characterise the mechanical anisotropy of LMD specimens that were produced using long unidirectional, bidirectional, and short unidirectional raster scan build strategies. Mechanical anisotropy was established by correlating the LMD build process induced grain size effects to yield strength using the Hall-Petch relations. Micro-indentation hardness measurements show differences

between the fusion zones and deposition track body locations showing the effects of build pattern on localized mechanical properties. The δ -ferrite constituent of each sample was determined using Schaeffler diagram and Cr_{eq}/Ni_{eq} measurement methods. Micro-indentation hardness measurements show differences between the fusion zones and deposition track body locations showing the effects of build pattern on localized mechanical properties. The following conclusions can be made from this study:

Mechanical anisotropy was observed between different build strategies: short unidirectional, long unidirectional and bidirectional raster scans in the tensile tests, whereby the long unidirectional raster scan build strategy exhibited the highest UTS, yield strength, and fracture strain, followed by the bidirectional and short unidirectional raster scan build strategies. The reverse trend was observed for the Young's modulus. LMD specimens exhibit higher UTS and yield strength, and lower fracture strain than conventionally manufactured stock bar specimens. This is due to the δ -ferritic phase, acicular crystal grain structures that are vertically oriented from the substrate in the LMD fabricated part. The stock bar specimens showed the typical austenitic grain structures giving ductile mechanical properties from the tensile test result. The differences in thermal histories between the three build strategy raster scan patterns result in differences in the microstructural grain sizes, whereby the short unidirectional raster scan build strategy exhibited the largest grain size, followed by the bidirectional and long unidirectional raster scan build strategies respectively. The Hall-Petch model fitted in this study show anisotropy in yield strength with respect to the grain size arising from the different build strategy pattern effect.

The Cr_{eq}/Ni_{eq} ratio and the estimated δ -ferrite constituent for each build strategy was observed to be in a descending order from long unidirectional, bidirectional, and short unidirectional raster scan build strategies respectively, indicating an embrittlement and hardening effect for each build strategy in the same descending order. The fracture surface topologies exhibited a transverse and longitudinal fracture effect in the short unidirectional and long unidirectional raster scans respectively. These fracture topology patterns and the anisotropic observations from the tensile tests show that the arrays of deposition track patterns play a role in the eventual built specimen anisotropic properties and fracture mode. Microindentation at the fusion zone and deposition track body locations show localized strengthening at the fusion zones. The arrays of fusion zones

arising from the build strategy effects leads to differences in the distribution of localized strengthening across each sample.

Chapter 5: Interfacial Structural Design and Fabrication Methods for Additive Manufacturing Applications (Study 3)

5.1 Introduction

Structural optimisation of LMD components is crucial area of interest for the AM and remanufacture of functional parts. Numerous engineering applications require the secure joining of two metallic bodies. Examples include applications from the aerospace industry like rotor blades, and air foils, as well as applications from the automotive industry like exhaust manifolds. These applications require a high quality joining technique that is able to withstand the stresses in their respective environments. Other AM techniques like SLM and EBM are powder bed processes and have limited capabilities to add/modify features on existing components. These techniques may have hybrid manufacturing (subtractive and additive) solutions in the future, but do not currently exist. Attempts to solve this issue include the use of some implementation of fasteners that can secure two bodies together. An example of a fastener implementation is in the form of bolts. Bolts secure two bodies by passing through a hole on one metallic body, crossing the interface and through another hole at the adjacent metallic body. The use of bolts as a fastening agent poses several disadvantages that may weaken the structure by introducing stress concentrations at the bolting points. The use of bolts also increases the complexities of the assembly process and typically creates protrusions at the bolt heads that may not be wanted in a component application.

LMD applications require building metal for cladding, filling gouged defects in repair of components and 3D printing of metal parts with complex geometry transitions and interfaces. The aerospace and automotive industries have potential applications that utilize LMD to build 3D printed metal engine structures like air-foils and exhaust manifolds.

Combining subtractive and additive manufacturing capabilities provide the ability to design 3D AM built parts with complex transition shapes with metallurgical bond and mechanical interlocking interfacial structure design demonstrated in this novel work. This invention proposes a novel subtractive and additive manufacturing method

to fabricate interfacial interlocking metallurgical bond structures in LMD built parts with complex transition in geometries at interface structures.

5.2 Provisional Patent

To address the issues with the pre-existing joining solutions, a superior novel joining method is offered in the form of a provisional patent filing. The provisional patent, entitled “Interfacial Structural Design And Fabrication Methods For Additive Manufacturing Applications”, Singapore provisional patent application number 10201801121S was filed on 9 February 2018. The method covers the fabrication of interfacial joint step features between a metallic base and an LMD structure build-up.

This invention pertains to a novel combined subtractive and additive manufacturing method to fabricate stronger interface interlocking design features in AM parts. The technique employs subtractive manufacturing techniques by metal machining to prefabricate contoured trenches to be filled by additive manufacturing with laser metal deposition to strengthen interface interlocking with metal fusion built parts formed by additive manufacturing techniques. The interfacial joint is created by forming a recess having one or more steps in the substrate by a material removal process (metal machining, mechanical forming, laser treatment, hybrid machining or the like); and depositing metal material on the recess layer by layer, using laser or direct energy metallic deposition to form the projection, such that a metallic bond is formed at the interface between the substrate and the projection. This produces AM Parts with improved interfacial interlocking fusion bond when building AM parts on new built or existing substrate parts that involves complex transition in geometry and interfacial designs.

A schematic example of this invention is shown in Figure 5.1 where a stepped joint interfacial projection is additively deposited over a substrate body. In this scenario, the substrate interfacial structures can be manufactured subtractively via CNC milling, and the interfacial projection is manufactured additively via LMD techniques.

Figure 5.1: Stepped joint interfacial projection for solid bodies

5.3 Methodology

5.3.1 Interfacial Joint Design Trial Study

A study was conducted to investigate the quality of parts produced using various types of interfacial joint designs. Three design parameters were varied in the interfacial joint design specimens: step height, step angle, and geometry feature. The objective of the study is to see what effect the design parameters have on the part quality. Part quality was examined using both non-destructive and destructive techniques, to detect any form of defects and porosities within the part. The non-destructive techniques chosen for this study are ultrasound measurement and x-ray tomography, and the destructive technique chosen for this study is characterization via optical microscopy.

Optical microscopy was used as the primary method to quantify the amount of defects in the specimens. Due to limited access to the ultrasound and x-ray tomography measurement resources, two samples were selected to be examined in further detail to characterise the defects. In this case, the geometry feature design parameters were selected because among the three design parameters, the geometry feature parameter is the only categorical type, where the concave and convex design features are binary states. The other two design parameters depict a range of numerical values that are non-binary scalar states, where it is possible to extract a proportional trendline between the upper and lower boundaries of the design parameter, and the use of the optical microscopy method for quantifying defects in these cases is sufficient. On the other hand, it is not possible to plot a trendline between the geometry design features of concave and convex due to their categorical states. Although quantifying the

defects for geometry designs parameter via optical microscopy is useful, the ultrasound and x-ray tomography techniques can provide further insight into the qualitative aspects of the formed defects that arise from the categorical states of concave and convex geometry features, like any clustering of defects or a deeper analysis into non-circular and non-round defects.

5.3.2 Hybrid Manufacturing Process

SS316L material was deposited via LMD techniques over SS316L substrates with stepped interfacial joint designs of varying design parameters. In this study, a fundamental investigation was conducted into the effects of design parameters like step height, step angle, and geometry feature have on part quality. In order to investigate the effects of the selected design parameters on the amount of defects, a 2^2 full factorial design of experiment (DOE) was used for this study. The experiment design and each parameter's abbreviation are shown in Table 5.1. The range of values selected for the Step Height and Step Angle design parameters were selected based on a few criteria. From previous experiments completed in Chapter 3 and Chapter 4, select process parameters were able to produce a deposition layer height of approximately 1 mm. Hence, when selecting the design parameter range for step height, the values were chosen in intervals of 1 mm. When selecting the lower and upper boundary ranges of the design parameters, the effects of the different parameters on the quality of the build are complex. A conservative approach to the selection of the step height parameters were used in this scenario because any minor variations in the deposition layer height would compound in error toward the higher layers. It is important that the deposited layer height is closely matched to the height of the step features on the substrate to a high degree in order to achieve a good quality bond. To reduce the risk of any compounding error due to the deposited layer height variation, a range of 1 mm and 2 mm were used for each step, for three steps across the samples. The step angle values were selected based on the drive to observe how an approximate diagonal feature of 60° would compare to a more extreme right angled step feature. A more gradual angle was not selected because the flat interfacial bond is already well understood by existing literature and past experiments. The design parameter range was selected in order to evaluate how the quality of bond is affected by the angle of the substrate. The concave and convex geometry features were selected based on the motivation to understand how any adjacent walls to the entrant deposition material would affect the quality of the build due to the powder flow dynamics relation to the

geometry. For instance, the concave geometry may have an effect on how the unfused powders interact with the additive process due to a “basin-effect” of the powder particles that get entrapped in the sunken concave feature. Meanwhile the convex geometry may not have this issue due to the lack of any obstructive adjacent features that may prevent the powder particles from being scattered.

A DMG MORI LaserTec 65 3D machine was used to fabricate LMD interfacial joint specimens. The hybrid manufacturing capabilities of the LMD machine was utilised to create the interfacial features at the substrate via subtractive manufacturing, and to produce the LMD portion via additive manufacturing. The fabrication of the test specimens and experimental procedure for this test is adapted from ASTM standards [197]. This two-step process is illustrated in Figure 5.2 whereby the interfacial features were created using CNC coded milling process and LMD was used to deposit SS316L material over the interfacial features additively. For the LMD process, a diode laser was used as the laser source, and argon was used as the shielding gas. A unidirectional raster scan was used, parallel to the long edge of the substrate. The LMD process parameters selected for this study were chosen based on the stability of the process parameters used in Chapter 3 and Chapter 4 and are shown in Table 5.2.

Table 5.1: Design of experiment

Design Parameter	Levels (Abbreviation)	
Step Height	1 mm (H1)	2 mm (H2)
Step Angle	60° (A60)	90° (A90)
Geometry Feature	Concave (FCC)	Convex (FCV)

Figure 5.2: Hybrid manufacturing two-step process
(a) subtractive and (b) additive

Table 5.2: LMD process parameters

	Process Parameter	Value
1.	Laser Power	1400 W
3.	Laser Spot Diameter	3 mm
4.	Deposition Rate	14 g/min
5.	Scanning velocity	1000 mm/min
6.	Track Overlap	50 %
7.	Deposition Track Height	1 mm

5.3.3 Powder Material

The powder feed material used in the LMD process is SS316L. The SS316L powder feed is gas atomized spherical metallic granules prepared by Carpenter. An SEM image of the powder material with a particle size range of 45 μm – 105 μm was taken using a Zeiss EVO HD 25 with a 300 times magnification, Secondary Electron mode, with EHT of 15.00 kV and working distance of 24.0 mm. The SEM image taken of the SS316L powder feed particles is shown in Figure 5.3. The chemical composition of the SS316L powder feed material is shown in Table 6.5.

Figure 5.3: SEM image of SS316L powder feed particles

Table 5.3: Nominal chemical composition of SS316L powder feed (wt.%)

Material	C	Mn	P	S	Si	Cr	Ni	Mo	Fe
SS316L	0.03	2.00	0.045	0.03	1.00	16.00-18.00	10.00-14.00	2.00-3.00	Bal.

5.3.4 Interfacial Joint Design Specimens

The interfacial joint design was produced by CNC milling techniques and its dimensions are shown in Figure 5.4. The eight interfacial joint designs' geometries and dimensions at the SS316 substrate are based on the experiment design listed in Figure 5.4. Three steps were employed for each design and SS316L metal powder was deposited at the interfacial joint location along the step features. For the concave geometry features, a 6 mm valley was used at the lowest step and each subsequent step width was 3 mm. With a 3 mm laser spot diameter and a 50% track overlap, a total of three raster tracks was deposited at the bottom most valley, seven raster tracks were deposited at the second step, eleven raster tracks were deposited at the topmost step. For the convex geometry substrates, one track was deposited at the lowest most step, three raster scan tracks were deposited at the second step, and five raster scan tracks were deposited at the topmost step. The same deposition sequence was conducted on each side of the convex geometry feature. The LMD over the interfacial joint design at the substrate is illustrated in Figure 5.5. An image of the test samples after the LMD process is shown in Figure 5.6.

Figure 5.4: Dimensions for each interfacial joint design (a) A60-H1-FCC, (b) A90-H1-FCC, (c) A60-H2-FCC, (d) A90-H2-FCC, (e) A60-H1-FCV, (f) A90-H1-FCV, (g) A60-H2-FCV, and (h) A90-H2-FCV

Figure 5.5: LMD dimensions (a) A60-H1-FCC, (b) A90-H1-FCC, (c) A60-H2-FCC, (d) A90-H2-FCC, (e) A60-H1-FCV, (f) A90-H1-FCV, (g) A60-H2-FCV, and (h) A90-H2-FCV

Figure 5.6: Test specimens (a) A60-H1-FCC, (b) A90-H1-FCC, (c) A60-H2-FCC, (d) A90-H2-FCC, (e) A60-H1-FCV, (f) A90-H1-FCV, (g) A60-H2-FCV, and (h) A90-H2-FCV

5.3.5 Ultrasound Analysis

A preliminary defect examination was conducted on two samples using ultrasound inspection techniques. A60-H2-FCC and A60-H2-FCV samples were used to investigate if the feature geometry has a significant effect on the formation of defects at the interfacial joint.

In order to detect the defects non-destructively, ultrasound inspection was conducted using Scanmaster's LS-25 Ultrasonic Inspection System equipped with Olympus Ultrasonic Transducer probes as seen in Figure 5.7.

Figure 5.7: LS-25 Ultrasonic Inspection System

In order for the ultrasound system to measure defects within the LMD part, the probe must be able to be run across a smooth even surface. Because of this limitation, it is unsuitable to run the probe across the upper surface of the specimen because of the step features. Although it is possible to run the probe on the underside of the specimen, a section cut was made using an abrasive saw 5 mm away from the upper surface. Furthermore, the substrate's surface toward the LMD part of the largest volume was used as the level surface, hence, only the ultrasound results toward that LMD location will be taken into consideration for the ultrasound analysis. The section cut and the level surface location is illustrated in Figure 5.8. Images of the two samples after the section cut and the LMD area that will be focused for the ultrasound measurement is shown in Figure 5.9.

Figure 5.8: Specimen section location for
(a) A60-H2-FCV and (b) A60-H2-FCC

Figure 5.9: Specimens for ultrasound measurement and measurement location for (a) A60-H2-FCV and (b) A60-H2-FCC

5.3.6 X-Ray Tomography Analysis

A supplementary defect examination was conducted on A60-H2-FCC and A60-H2-FCV samples using x-ray tomography techniques. The results from this non-destructive examination will be used to have an approximation of the size and location of the defects that exist in three-dimensions of the specimen. The Nikon XT H 225 x-ray system as seen in Figure 5.10 was used to capture the Raw CT file and slicing. Fiji software was used to conduct the image processing with custom written macros that use a series of processing steps: 8-bit conversion, thresholding, find maxima, inversion, and 3D viewer. The settings used for the X-ray tomography is shown in Table 5.4 below:

Table 5.4: X-ray tomography settings

	Process Parameter	Value
1.	Voxel Size	14
3.	Voltage	210 kV
4.	Filter	1 mm
5.	Num. of Projections	3141
6.	Resolution	12 μ m

Figure 5.10:Nikon XT H 225 x-ray tomography system

5.3.7 Microscopy Analysis

In order to conduct the porosity measurement, the SS316L have to be cross-sectioned and polished first. The specimens were sectioned across the middle of the largest volume LMD using an AbrasiMet 250 abrasive saw. Subsequently, they were hot mounted in resin using a SimpliMet XPS1 and polished using a MetaServe 250. Microscopy images were then captured, stitched, and analysed using a Zeiss Light Microscope and the porosity characteristics were analysed using Fiji software.

5.3.8 Dilution and Heat Affected Zone Analysis

To verify the results from the defects analysis, the size of the dilution and the heat affected zone (HAZ) for all eight specimens were measured using a Vickers hardness tester. The same cross-sectioned specimens from the defect analysis were used for this measurement. Ten indentations in a line were made perpendicular to and across the joint interface as illustrated in Figure 5.11. Vickers micro hardness tests were conducted with accordance to ASTM E384-17 [181] using an Innova Test Falcon 500 micro hardness tester. A micro-indentation load of 500 g and a 15 second dwell time were used. The width of each indentation was measured to be approximately 0.045 mm, hence, to maintain a minimum of 2.5 times distance between each indentation as per standard, an indentation spacing of 0.15 mm was used. A sample microscope image of the indentation measurement is shown in Figure 5.12.

Figure 5.11: Vickers hardness indentation location for samples with
(a) 90° step angle and (b) 60° step angle interfacial joint designs

Figure 5.12: Microscope image of one specimen after conducting Vickers hardness measurement

5.4 Results and Discussion

5.4.1 Ultrasound Measurement

Ultrasound measurement was conducted on specimens A60-H2-FCV and A60-H2-FCC. The ultrasound imaging results are shown in Figure 5.13 and Figure 5.14. From this imaging result, the spectrum-type image labelling assist in indicating where anomalies or defects are located throughout the specimen from a bottom-up perspective due to the ultrasound probe's scan across the underside of the specimen.

Figure 5.13: Results for A60-H2-FCC

(a) ultrasound measurement imaging focal area (b) measurement location at step and (c) measurement location at LMD material

Figure 5.14: Results for A60-H2-FCV

(a) ultrasound measurement imaging focal area (b) measurement location at step and (c) measurement location at LMD material

The ultrasound imaging from Figure 5.13 and Figure 5.14 show that defects tend to form along the corners of the interfacial steps, parallel to the longitudinal edge of the specimen. The location and depth of the defects were verified by examining points of interest across the ultrasound image in Figure 5.13 (b), (c) and Figure 5.14 (b), (c). The relationship between the depth of the defect detected and the time of the reflected pulse is described in the equation:

$$L = \frac{ts}{2} \quad (5.1)$$

Where L is the depth of the defect, t is the duration of time between the initiation and the ultrasound pulse, s is the ultrasonic velocity in the material, which in the case of SS316L, is ~5740 m/s.

Two locations of interest from A60-H2-FCC, Figure 5.13 (c) and A60-H2-FCV, Figure 5.14 (c) were examined using Equation (5.1) and the calculated depth of the defect detected was calculated and compiled in Table 5.5.

Table 5.5: Calculated depth of defect at two locations

Specimen	Ultrasound pulse from initiation	Depth of defect
A60-H2-FCC, Figure 5.13 (c)	2.80 μ s	8.03 mm
A60-H2-FCV, Figure 5.14 (c)	2.38 μ s	6.83 mm

From Table 5.5, by examining the depth of defects and corresponding the depth to the location of the defects from the ultrasound images in Figure 5.13 (a) and Figure 5.14 (a), the location of the defects were found to be located along the edges of the joint interface. The size of the defects was also observed to be large, which further indicates that incomplete fusion may have occurred during the LMD process. The source of these defects is further discussed in detail in Chapter 5.4.3 Microscopy Analysis.

5.4.2 X-Ray Tomography Analysis

The x-ray tomography individual frames for specimens A60-H2-FCV and A60-H2-FCC were imported and sequenced in an image stack in Fiji. A limitation with using x-ray tomographic techniques for dense materials like SS316L is that the small defects become challenging to distinguish from the grainy noise. The noise within the bounds of each specimen appears as dark artefacts in each frame that need to be removed in order to examine the defects. However, in doing so, the small pores are also removed in the denoising process. Hence, in this examination, the x-ray tomographic technique is only highly useful for the detection of larger defects that exist in 3D. These limitations can be seen in a grainy and noisy raw pre-processed x-ray tomography frame is shown in Figure 5.15.

Figure 5.15: Pre-processed frame taken from x-ray tomography image stack for (a) A60-H2-FCV (b) A60-H2-FCC

From Figure 5.16, it can be observed that defects tend to form toward the corners or the interface between the LMD build-up and the substrate. Using the image stack, the specimens' outer scan was constructed in 3D as seen in Figure 5.16. The two specimens' image stacks were processed to isolate the internal defects and is constructed in 3D as seen in Figure 5.17. From the 3D construction of the defects, the volume of the defects was measured against the bounds of the LMD build-up. The results for the measurement are compiled in Table 5.6 and sorted in a histogram in Figure 5.18.

Figure 5.16: X-ray tomography outer scan
(a) A60-H2-FCV (b) A60-H2-FCC

Figure 5.17: X-ray tomography of internal defects
(a) A60-H2-FCV (b) A60-H2-FCC

Table 5.6: X-ray image processing bounds, and defects detected

	Specimen	
	A60-H2-FCV	A60-H2-FCC
Total defects volume	1.300 mm ³	1.228 mm ³
Average defect volume	0.052 mm ³	0.053 mm ³

Figure 5.18: Histogram of defect sizes detected for each specimen

The bulk of the defects were observed to form along the corners of the interfacial joint design, parallel to the orientation of the longitudinal edge of the raster scan. The preliminary defect examination for A60-H2-FCV and A60-H2-FCC for both the ultrasound taken in Chapter 5.4.1 and x-ray tomography show that defects tend to form at the corners and longitudinally, parallel to the raster scan direction. Majority of the defects detected were also observed to be in the 0.025 mm³ - 0.050 mm³ range as seen in Figure 5.18. Although the abnormally large defects of >0.200 mm³ detected in both specimens make up <10% of total defects detected, these large defects pose as high risk for part failure due to their tendency to be crack initiation sites. The source of these defects is discussed in detail in Chapter 5.4.3.

5.4.3 Microscopy Analysis

The cross-sectional microscope images of the defects for each specimen condition are shown in Figure 5.19. The substrate and LMD areas are designated on each image. The results for the defect size analysis were compiled into size bins of $20 \mu\text{m}^2$ as seen in Figure 5.20 and Table 5.7.

Figure 5.19: Threshold microscopy images of specimens for porosity analysis (a) A60-H1-FCC, (b) A90-H1-FCC, (c) A60-H2-FCC, (d) A90-H2-FCC, (e) A60-H1-FCV, (f) A90-H1-FCV, (g) A60-H2-FCV, and (h) A90-H2-FCV

Figure 5.20: Histogram of defects sizes detected for each specimen

Table 5.7: Percentage of total defects detected for each defect size range

Specimen	Defect Size (μm ²)					
	0-20	21-40	41-60	61-80	81-100	>100
A60-H1-FCC	62.16%	10.81%	8.11%	8.11%	5.41%	5.41%
A90-H1-FCC	68.89%	17.78%	0.00%	4.44%	0.00%	8.89%
A60-H2-FCC	59.26%	25.93%	3.70%	0.00%	1.85%	9.26%
A90-H2-FCC	61.76%	14.71%	2.94%	0.00%	0.00%	20.59%
A60-H1-FCV	61.29%	19.35%	3.23%	6.45%	0.00%	9.68%
A60-H1-FCV	64.71%	23.53%	0.00%	2.94%	0.00%	8.82%
A60-H2-FCV	51.02%	16.33%	12.24%	6.12%	4.08%	10.20%
A90-H2-FCV	70.27%	5.41%	0.00%	0.00%	5.41%	18.92%

From the defect measurement results, it can be seen clearly that the majority of defects detected for all specimens fall within the range of 0-20 μm^2 . However, a significant number of large defects of $>100 \mu\text{m}^2$ was detected as well. These large defects are problematic because defects of this size risk being crack initiation sites when subjected to critical stresses that may result in catastrophic failure. This is especially the case for a specimen like A90-H2-FCC as seen in Figure 5.19 (d), where extremely large pores of sizes $>4000 \mu\text{m}^2$. In cases where large defects are detected, it indicates that there were significant issues during the LMD process that has a lot of room for process optimization. From the microscopy images, it is evident that the larger pores detected in the specimens mostly formed at the first LMD layer or near the substrate surface.

There are numerous reasons for this observation which have discussed in detail in Chapter 2.3.1. It is generally well understood that interrune pores are a common occurrence at the first layer of LMD, at the substrate surface, and can be prevented by increasing the laser power or lowering the scanning speed. In many cases of AM-built parts, the additively built part is typically sectioned away from the substrate, which in doing so, removes the interrune defects in the process. This applies to cases where solely the AM-built part is required for its application, and the substrate is discarded. Increasing the laser power or lowering the laser scan speed at the first few layers are easily adjusted in laser cladding applications in order to prevent these interrune porosities, because only one to three layers are required for this form of LMD application, and the thermal history is not complex. However, in this study, both the substrate and the AM-build up needed to be kept intact as well as, an LMD 3D build-up of multiple layers are required. This complicates the process optimization due to the complex thermal history involved in the 3D build-up from the substrate.

In order to determine the effect interfacial design parameter has on detrimental defects, an ANOVA was conducted for the percentage of defects detected that are $>100 \mu\text{m}^2$ in size. In this ANOVA, a confidence level of 95%, $\alpha = 0.05$ was used. The interfacial design parameters from Table 5.1 and the percentage of defects detected of $>100 \mu\text{m}^2$ in size from Table 5.7 were used in the analysis. The results for the ANOVA are compiled in Table 5.8. The main effects plot and the interaction plot are shown in Figure 5.21 and Figure 5.22.

Table 5.8: ANOVA results for percentage of $>100 \mu\text{m}^2$ sized defects

Source	DF	Adj SS	Adj MS	F-Value	P-Value
Step Angle	1	64.241	64.2411	171.72	0.048
Step Height	1	85.609	85.6086	228.83	0.042
Geometry Feature	1	1.505	1.5051	4.02	0.294
Step Angle * Step Height	1	37.976	37.97561	101.51	0.063
Step Angle * Geometry Feature	1	6.038	6.0378	16.14	0.155
Step Height * Geometry Feature	1	3.038	3.0381	8.12	0.215
S = 0.611647 $R^2 = 99.81\%$ $R^2(\text{adj}) = 98.68\%$					

Figure 5.21: Main effects plot for percentage of large defects detected

Figure 5.22: Interaction plot for percentage of large defects detected

From the 2^2 full factorial DOE analysis, the regression equation was calculated to be:

$$D_{large} = 20.17 - 0.2468 A - 15.24 H + 6.63 G + 0.2905 AH - 0.0579 AG - 1.232 HG \quad (5.2)$$

Where D_{large} is percentage of large defects, A is step angle, H is step height, and G is feature geometry. From the ANOVA in Table 5.8, the results for the step angle and step height were shown to have P-values <0.05 , indicating that they have a significant effect on the amount of large defects formed. On the other hand, the geometry feature and the interactions have a P-values of >0.05 , indicating that these factors do not have a significant effect on the number of large defects formed. This result is reflected in the main effects plot in Figure 5.21 where step height was observed to have the largest effect, followed by step angle, while geometry feature was observed to have a minor difference between concave and convex geometries. This is also consistent with the interaction plot Figure 5.22 in where no clear interaction between the factors was observed.

The ANOVA results as well as the microscopy images from Figure 5.19 show that there is an issue with fusion at the interfacial side walls that is exacerbated by the step angle and step height. This is evidenced by observing the poor fusion at interfacial joint designs with 90° step angles, where large detrimental defects at the interfacial side walls were formed and the formation of defects are aggravated with a higher step height of 2 mm. This is generally due to problems with the powder flow dynamics and the melt pool that forms along the interfacial side walls. The coaxial powder feed and the orientation of the deposition nozzle in LMD processes are generally optimised for fusion at surfaces perpendicular to the deposition head. The powder flow blockage at step corners and the partial melt pool formation that results in poor fusion at the interfacial joint design's side walls are illustrated in Figure 5.23 (a) and (b) respectively.

Figure 5.23: Schematic diagram of LMD process for A90-H2-FCV indicating problematic areas (a) powder flow blockage at step corners and (b) partial melt pool formation at side walls

From the results of the two-level DOE study, in order to have an interfacial joint design that minimises the amount of large detrimental defects, a smaller step angle and a shorter step height is necessary. This is consistent with the results from the main effects plot in Figure 5.21 and Equation (5.2) where if the smallest possible percentage of

large defects is set as the optimisation goal, the A60-H1-FCC would be the optimal design, followed closely by A60-H1-FCV.

5.4.4 Dilution and Heat Affected Zone Analysis

The Vickers hardness measurement results are shown in Figure 5.24. The substrate and LMD build-up areas, as well as the interface are indicated on the graph. The hardness in the LMD-build up zone was measured to have an average hardness value of 166 ± 4 HV while the substrate zone was measured to have an average hardness value of 188 ± 9 HV. The transition area between the LMD-build up zone and the substrate zone can be divided into two phases: The dilution zone and the HAZ. This transition area is characterised a gradual change in hardness value from the substrate to the LMD-build up, which in the case of this study, is an increasing hardness value.

From the hardness plot, the length of the dilution zone and HAZ for each specimen was measured by summing the length of the indentation gaps that cover the transition area. The Vickers hardness test results for each condition are shown in Figure 5.24.

Figure 5.24: Vickers hardness results measuring the dilution zone at the interfacial joint

The higher hardness value measured at the LMD build-up area compared to the substrate area is a consequence of the rapid solidification rates in LMD processes that result in a grain refinement of the SS316L material. The rapid quenching effect that occurs via a conductive action through previously deposited layers restricts the granular growth of the deposited SS316L material, producing grain sizes that are significantly smaller. This also results in a lack of development for the secondary dendrite arms and leaving behind the formation of highly directional cellular structures. The details of the grain refinement and its correlation with mechanical properties have been explored in earlier studies in Chapter 4.3.5 as well as a literature review in Chapter 2.3.2.

The correlation between the grain refinement and the yield strength and hardness can be deduced from the Hall-Petch equations [179, 180] and the general relationship between hardness and yield strength [198]:

$$\sigma_y = \sigma_0 + \frac{k}{\sqrt{d}} \quad (5.3)$$

$$\Delta\sigma_y = K \Delta H_V \quad (5.4)$$

Where σ_y is yield strength, σ_0 is a material constant that governs the stresses for grain dislocation movement, k is the Hall-Petch slope coefficient, d is the grain width, $\Delta\sigma_y$ is the change in yield strength, K is a material constant, and ΔH_V is the change in Vickers hardness. Substituting Equation (5.3) into Equation (5.4) gives:

$$H_V = \frac{1}{K} \left(\sigma_0 + \frac{k}{\sqrt{d}} \right) \quad (5.5)$$

The main effects for each factor (step angle, step height, and feature geometry) was subsequently plotted against the size of dilution zone and HAZ as seen in Figure 5.25.

Figure 5.25: Main effects for step angle, step height, and feature geometry on the size of dilution zone and HAZ

The results from the DOE's main effects plot in Figure 5.25 show that the 60° step angle, 1 mm step height, and the concave feature geometry produces a larger dilution zone and HAZ. The larger dilution zone and HAZ indicate that there is a comparatively better fusion than the factors of the opposing levels. An interfacial joint design with a smaller dilution zone and HAZ suggests that there is a lower degree of fusion at these sites which if severe enough, can lead to the formation of defects that arise from incomplete melting or lack of fusion. This observation is consistent with the main effects from the defects analysis in Figure 5.21 where the 60° step angle, 1 mm step height, and the concave feature geometry produces a lower number of defects detected. The combination of interfacial design parameters that produces the largest dilution zone and HAZ coincides with the same parameters the least number of large defects and vice versa. This observation affirms the hypothesis that the large defects observed at the side walls are driven by the lack of fusion.

5.5 Conclusion

In this preliminary study, a fundamental review of interfacial joint designs for select design parameters was conducted. Two levels for three design parameters: Step angle, step height, and feature geometry were selected to understand their effects on the build quality.

The ultrasound, x-ray tomography, and microscopy techniques were used to characterise defects that form within the interfacial joint design. Measurements for A60-H2-FCV and A60-H2-FCC specimens show that large-sized defects were inclined to form at the sharp corners of the interfacial joint design. These defects were also observed to form in a line parallel to the longitudinal edge of the specimen. This is a result of a lack of powder flow access to these corners due to issues from the powder flow dynamics that arise from interference with the step design geometry.

The defect measurement by microscopy showed that the Step Height and Step Angle factors have a significant effect on the formation of large defects, while the Geometry Feature factor has an insignificant effect. The main effects trend from the microscopy measurement and the dilution and heat affected zone analysis are in agreement with defect study, for which factor had the most effect on the formation of large defects.

From this preliminary study, several lessons have been learned to improve the quality of the interfacial joint build. By setting an optimisation goal of lowest degree of large defects formed, the most effective combination of interfacial joint design would be to lower the Step Height and Step Angle. On top of this, tighter tolerances can be made for the step joint features in order to ensure that more precise initiation point can be made for the LMD process that produces a more consistent process and a higher quality build.

Chapter 6: Interfacial Joint Design Comparative Study (Study 4)

6.1 Introduction

A study was conducted to investigate the mechanical performance of three interfacial joint design types: Flat interfacial joint, V-Shaped interfacial joint, and stepped interfacial joint. The flat interfacial joint design is the conventional interfacial design for additively manufactured joints. The V-shaped interfacial joint design and the stepped interfacial joint design are two variants of the invention whose mechanical performance are compared to the conventional flat interfacial joint design in this study. Lessons that were gleaned from the preliminary study in Chapter 3 were applied in this comparative study.

6.2 Methodology

6.2.1 Hybrid Manufacturing Process

An SS316L LMD cuboid of 170 mm × 15 mm × 37 mm was built over an SS316L substrate designed with each interfacial joint design type. The LMD cuboid dimensions were selected based on the necessary build volume required to extract six Charpy samples, where the notch is located at the middle of the joint interface. This results in a Charpy sample where half of its volume is in the LMD region, and the other half is in the substrate region. Two variants for the Charpy sample for each interfacial joint design was used. The two variants differ in where the notch is located for each Charpy sample type. Three Charpy samples were extracted for each notch variant type. The objective of using two notch variants is to investigate whether directionality of the impact affects the mechanical performance of the interface.

The DMG MORI LaserTec 65 3D machine was used to fabricate LMD interfacial joint specimens. The hybrid manufacturing capabilities of the LMD machine was utilised to create the interfacial features at the substrate via subtractive manufacturing and produce the LMD portion via additive manufacturing. The fabrication of the test specimens and experimental procedure for this test is adapted from ASTM standards [197]. SS316L powder feed

was fed coaxially via the deposition head and melted by a diode laser to form a melt pool at the substrate. The factors used in the experiment design for this study was chosen based on the conclusions made from the preliminary study in Chapter 3. In this study, two design parameters were chosen: Interfacial joint design and notch variant. For the interfacial joint design factor, two types of angled features, V-shaped and stepped, were chosen as a comparison against a flat interfacial joint. The dimensions for the stepped interfacial joint design used in this study were selected based on the optimised design from the preliminary study, where the A60-H2-FCC specimen produced the least number of large defects. The V-shaped interfacial design was added as an alternative angled design as an attempt to eliminate all corners in the design. This is due to the tendency for pores to form at the corners of the stepped design from the preliminary study, and to evaluate if removing corners within the interfacial joint design would improve the build quality. The experiment design used in this study is shown in Table 6.1. The LMD build parameters used in this process are specified in Table 6.2. Images of the LMD process and LMD-built specimen are shown in Figure 6.1.

Table 6.1: Experiment design

Design Parameter	N	Levels (Abbreviation)		
Interfacial joint design	3	Flat (F)	V-shaped (V)	Stepped (S)
Notch variant	3	0°		90°

Table 6.2: Build parameters employed in LMD process

	Process Parameter	Value
1.	Laser Power	1200 W
3.	Laser Spot Diameter	3 mm
4.	Deposition Rate	14 g/min
5.	Scanning velocity	1000 mm/min
6.	Track Overlap	50 %

Figure 6.1: (a) LMD process and (b) LMD-built specimen

6.2.2 Powder Material

The powder feed material used in the LMD process is SS316L. The SS316L powder feed is gas atomized spherical metallic granules prepared by Carpenter. The SS316L powder feed is gas atomized spherical metallic granules prepared by Carpenter. An SEM image of the powder material with a particle size range of $45\text{ }\mu\text{m}$ – $105\text{ }\mu\text{m}$ was taken using a Zeiss EVO HD 25 with a 300 times magnification, Secondary Electron mode, with EHT of 15.00 kV and working distance of 24.0 mm. The SEM image taken of the SS316L powder feed particles is shown in Figure 6.2. The particle size range used in the LMD process is $45\text{ }\mu\text{m}$ – $105\text{ }\mu\text{m}$. The chemical composition of the SS316L powder feed material is shown in Table 6.5.

Figure 6.2: SEM image of SS316L powder feed particles

Table 6.3: Nominal chemical composition of SS316L powder feed (wt.%)

Material	C	Mn	P	S	Si	Cr	Ni	Mo	Fe
SS316L	0.03	2.00	0.045	0.03	1.00	16.00-18.00	10.00-14.00	2.00-3.00	Bal.

6.2.3 Interfacial Joint Design Specimens

The fabrication sequence for the LMD Charpy impact specimens is illustrated in Figure 6.1, Figure 6.4, and Figure 6.3. SS316L material within the stepped or V-shaped slot at the interface was deposited using a unidirectional build strategy across the long edge of the substrate, and the raster scan build strategy above the slot was deposited using a bidirectional build strategy. This is done due to the space limitation within the slot and the disproportionately large thermal build-up if a bidirectional or a unidirectional raster scan build strategy that runs across the short edge of the substrate were used. In Figure 6.3 (a), a slot is first machined at the substrate by milling with accordance to the step joint interface design parameters. Next, as in Figure 6.3 (b), metal is deposited by LMD over the slot to form the interfacial joint projection, creating a step joint interface. The LMD block is subsequently machined down to size as in Figure 6.3 (c). Lastly, multiple Charpy impact specimens as seen in Figure 6.3 (d) can be extracted from the LMD block. A similar fabrication sequence can be used to produce flat

interface Charpy impact specimens, as in Figure 6.3 (e), by excluding the milling slot at the first step. Figure 6.4 depicts the step-by-step deposition sequence that is used in the manufacture of the LMD section of the Charpy impact samples. A unidirectional raster scan build strategy parallel to the long edge of the substrate was used between Step 1 and Step 7 to fill the recessed section at the substrate section. A bidirectional raster scan build strategy is not recommended for Step 1 to Step 7 because conducting a raster scan deposition parallel to the short edge of the substrate would result in a steep thermal gradient and poor deposition quality. A unidirectional raster scan build strategy parallel to the long edge of the substrate results in a more gradual thermal gradient and improves the bond between the LMD section and substrate. A bidirectional raster scan build strategy was subsequently used from Step 8 to Step 10 where more space is available on each side of the substrate which provides more time during the deposition for the heat to dissipate resulting in a more gradual thermal gradient.

Figure 6.3: Fabrication sequence for Charpy impact specimens with step joint interface (a) substrate with machined slot, (b) LMD over substrate, (c) machine to shape, (d) stepped interface Charpy impact specimen (e) flat interface Charpy impact specimen

Figure 6.4: LMD build strategy and deposition sequence

The substrate design and dimensions for each interfacial joint design type are detailed Figure 6.5, Figure 6.6, and Figure 6.7. The dimensions used in the substrate design was determined based on the lessons learned from the preliminary study. From the preliminary study, defects were observed toward the edges of the deposition tracks partly due to poor tolerancing and difficulties for the deposition to fill the slot volume. To compensate for these errors, the maximum tolerance used was set at 0 mm, the minimum tolerance used was set at 0.05 mm, and slot opening's width was set at 5.70 mm for the V-shaped and stepped joint interface substrates.

The dimensions for the substrate and the interfacial joints to be produced by subtractive manufacturing are shown in Figure 6.5, Figure 6.6, and Figure 6.7. The LMD build up over the substrate for each interfacial joint design type is illustrated in Figure 6.8, Figure 6.9, and Figure 6.10. An illustration of the Charpy sample extraction location from the substrate and LMD build up is illustrated in Figure 6.12. The Charpy sample for each interfacial joint design type and the location of the notch for each Charpy notch variant is illustrated in Figure 6.13.

Figure 6.5: Substrate dimensions for Flat Interfacial Joint Design
(a) isometric view, (b) front view (c) side view

Figure 6.6: Substrate dimensions for V-Shaped Interfacial Joint Design
(a) isometric view, (b) front view (c) side view

Figure 6.7: Substrate dimensions for Stepped Interfacial Joint Design
(a) isometric view, (b) front view (c) side view

Figure 6.8: LMD build-up over substrate for Flat Interfacial Joint Design
(a) front view and (b) isometric view

Figure 6.9: LMD build-up over substrate for V-Shaped Interfacial Joint Design (a) front view and (b) isometric view

Figure 6.10: LMD build-up over substrate for Stepped Interfacial Joint Design (a) front view and (b) isometric view

Figure 6.13: Illustration of Charpy samples extracted from each interfacial joint design type (a) F-0°, (b) V-0°, (c) S-0°, (d) F-90°, (e) V-90°, and (f) S-90°

6.2.4 Defects Analysis

Specimens used for defects analysis were extracted from the LMD build-up and substrate for each condition, with the measurement surface positioned at the centre of the part as illustrated in Figure 6.14. The specimens were extracted using an AbrasiMet 250 abrasive saw and hot mounted in resin using a SimpliMet XPS1. Microscopy images of the measurement surface were then captured, stitched, and analysed using a Zeiss Light Microscope and the defects characteristics were analysed using Fiji software.

Figure 6.14: Specimen for defects analysis extraction location and microscopy measurement surface

6.2.5 Charpy Impact Test

Charpy tests were conducted using a Zwick Roell, Amsler RKP 450 equipped with a 300 J pendulum head. Images of the Charpy tester and the Charpy sample mounting is shown in Figure 6.15.

Figure 6.15: (a) Zwick Roell, Amsler RKP 450 equipped with a 300 J pendulum head and (b) mounted Charpy sample

After the Charpy impact tests, the specimens' fracture surface topologies were scanned using the Zeiss Smart Zoom 5, via a 3D depth-of-focus reconstruction method using 34 times magnification with a 30 μm resolution.

6.3 Results and Discussion

6.3.1 Process Monitoring Data Analysis

During the LMD process, the melt pool temperature, melt pool size, and laser power process data were captured and monitored using an in-situ process monitoring system. A CCD camera positioned along the main optical axis captures the melt pool's luminance and the process monitoring system converts the luminance data to a temperature reading per pixel and produces a melt pool temperature and melt pool size reading at each capture

instance. The sampling rate used in the recording of the process data was 10 samples per second. The melt pool temperature and size process data plots for each specimen are shown in Figure 6.16 and Figure 6.17.

Figure 6.16: LMD empirical melt pool temperature process data plot for (a) Flat, (b) V-shaped, and (c) Stepped interfacial joint designs

Figure 6.17: LMD empirical melt pool size process data plot for (a) Flat, (b) V-shaped, and (c) Stepped interfacial joint designs

The melt pool temperature and size, as seen in Figure 6.16 and Figure 6.17 respectively were observed to be non-homogeneous and the variation in the readings at each point were very much location dependent. For instance, the measured melt pool temperature and size were observed to be far lower and smaller at the first few layers. This is because the substrate's initial thermal condition is that of room temperature, far cooler than the melt pool temperature. The difference in temperature means that the energy required to form a melt pool within the target

size range is insufficient. The disparity in the energy input and the required energy for full melting results in a lower temperature and smaller melt pool size result at the first few layers. This is observed within the slots of the V-shaped and Stepped joint interface design as well as their adjacent flat interface and the Flat joint interface design. These issues with the measured melt pool temperature and size at the first few layers may result in issues with fusion between the deposited metal and the substrate. Any poor quality fusion at the interface location may lead to the formation of detrimental defects that have a significant impact on the performance of the LMD-built part. The defect measurement and source of defects are discussed in Chapter 6.3.2.

6.3.2 Defects Analysis

Based on the preliminary study, the bulk of the large detrimental defects were observed to form close to the interface between the LMD build-up and the substrate. Hence, the bounding perimeter of the defect's analysis within the area of the specimen cross-section was set 1 mm (one deposition layer height) above the interface and bounding along the interface geometry.

The cross-sectional microscope images of the defects for each specimen condition and the defect measurement area are shown in Figure 6.18. The substrate and LMD areas are designated on each image. The results for the defect size analysis were compiled into size bins of $20 \mu\text{m}^2$ as seen in Figure 6.19 and Table 6.4. The overall part density for each specimen is shown in Figure 6.20.

Figure 6.18: Microscopy image of specimens and demarcated defect measurement area for (a) Flat, (b) V-shaped, and (c) Stepped specimens.

Figure 6.19: Histogram of defect sizes detected for each specimen

Table 6.4: Percentage of total defects detected for each defect size range

Specimen	Defect Size (μm)					
	0-20	21-40	41-60	61-80	81-100	>100
Flat	24%	20%	11%	9%	13%	22%
Stepped	23%	12%	12%	14%	5%	33%
V-Shaped	46%	11%	10%	6%	3%	24%

Figure 6.20: Percent part density for each specimen

From Figure 6.19 and Table 6.4, the majority of the porosities detected were found to be within the 0 – 20 μm and the >100 μm size range. Although the Stepped joint design specimen exhibited the most number of >100 μm sized pores, it also had the highest part density among the three samples as seen in Figure 6.20. On the other hand, the Flat interfacial joint design specimen was observed to have the least number of >100 μm sized pores but the lowest part density. This difference is attributed to the different types of pores formed in both cases. The Flat interfacial joint design's defects were observed to be solely consisting of interrump porosities that were few in count but large in size. These interrump porosities form between deposition tracks and are hence limited to the number of deposition tracks that exist within the first deposition layer of the specimen. The V-shaped and Stepped interfacial joint design types were observed to have disjointed elongated pores that are parallel to the sloped edges of the interfacial joint due to partial fusion at the side walls. This type of defect is discussed at length in the preliminary study in Chapter 5.4.3.

6.3.3 Charpy Impact Test

Images of the post-test Charpy samples are shown in Figure 6.21. All specimens were observed to fracture at the interfacial joint. The measured absorbed energy from the Charpy impact tests for each interfacial joint type is compiled in Table 6.5 and Figure 6.22.

Figure 6.21: Post-test Charpy samples for each interfacial joint design type
(a) F-0°, (b) V-0°, (c) S-0°, (d) F-90°, (e) V-90°, and (f) S-90

Table 6.5: Charpy impact test results

Interfacial Joint	Replicates	Absorbed Energy (J)			Mean (J)
F-0°	3	16	16	16	16.0 ± 0.0
V-0°	3	17	18	17	17.5 ± 0.5
S-0°	3	20	20	20	20.0 ± 0.0
F-90°	3	12	14	12	13.0 ± 0.9
V-90°	3	20	23	20	21.5 ± 1.4
S-90°	3	31	26	26	28.5 ± 2.4

Figure 6.22: Charpy test results for each interfacial joint design type

In order to test if the interfacial joint type has an effect on mechanical properties, an ANOVA was conducted with a confidence level of 95%, $\alpha = 0.05$. The ANOVA results for the absorbed energy from the Charpy impact test is compiled in Table 6.6. The interaction plot and main effects plot for absorbed energy is shown in Figure 6.23 and Figure 6.24 respectively.

Table 6.6: Analysis of variance for Charpy impact test results

Source	DF	Adj SS	Adj MS	F-Value	P-Value
Interfacial Joint Design	2	270.78	735.389	62.49	< 0.001
Notch Variant	1	32.00	32.000	14.77	0.002
Interfacial Joint Design * Notch Variant	2	93.00	46.500	21.46	< 0.001
Error	6	2907	484.5	-	-
Total	8	24420	-	-	-
S = 1.47196 R ² = 93.84% R ² (adj) = 91.27%					

Figure 6.23: Toughness interaction plot for each interfacial joint design type and notch variant

Figure 6.24: Toughness main effects plots for each interfacial joint design type and notch variant

From the factorial design analysis, the regression equation was calculated to be:

$$P = 19.11 - 4.78 I_F + 0.06 I_V + 4.72 I_S - 1.33 N_0 + 1.33 N_{90} + 3.00 I_F N_0 - 3.00 I_F N_{90} - 0.50 I_V N_0 + 0.50 I_V N_{90} - 2.50 I_S N_0 + 2.50 I_S N_{90} \quad (6.1)$$

Where, P is the impact toughness, I_F is the flat interfacial joint design, I_V is the V-shaped interfacial joint design, I_S is the stepped interfacial joint design, N_0 is the 0° notch variant, and N_{90} is the 90° notch variant.

From the ANOVA in Table 6.6, the results for interfacial joint design, notch variant, and their interaction were all shown to have P -values < 0.05 , indicating that they have significant effects on the absorbed energy. The significance of the interaction can be traced to the inverse trend seen in the flat interfacial joint design in Figure 6.23. To understand this phenomenon, the fracture surface topology from Figure 6.25 and the macroscopic fracture trajectory from Figure 6.26 has to be examined. The fracture surface topologies show that the crack propagation occurs along the interface for all the interfacial joint design and notch variant types. Although this is generally the case for all the samples, one key difference between the fracture topology between the notch variant

0° and 90° is the crack propagation path as seen in Figure 6.26. Samples V-0° and S-0° as seen in Figure 6.26 (b) and (c) exhibited a crack propagation path that fractures at the notch, through the substrate material, and once the path intersects with the interfacial line, the crack begins to propagate in alignment with the V-shaped and stepped interfacial designs respectively. On the other hands samples V-90° and S-90° as seen in Figure 6.26 (e) and (f) exhibited a crack propagation path that fractures away from the notch, at the interfacial joint location and immediately begins to propagate in alignment with the V-shaped and stepped interfacial joint designs respectively.

The images also show the existence of intertrack defects that occur along the direction of the raster scan strategy across the surface of the first layer and the substrate. This suggests that in the case of the V-shaped and stepped joint interface, the mechanics of the interfacial joint design plays a strong role in strengthening the interface and the results of the absorbed energy can be deemed as driven predominantly by the interfacial features. On the other hand, in a flat joint interface, in the absence of interfacial features, the results of the absorbed energy are hence driven predominantly by the directionality of the intertrack defects.

Overall, the V-shaped and stepped interfacial joint designs produced a 9% to 119% improvement in toughness compared to the conventional flat interfacial joint design. The stepped joint interface with a rotated 90° notch produced the most improvement in toughness, showing that the invention has a stronger mechanical performance in one direction over the other. The main effects plot from Figure 6.24Figure 6.25 show that although the interfacial joint design type and the directionality of the impact (via the notch variants) play an important role in affecting the mechanical performance, the interfacial joint design factor has a more significant effect than the directionality of the impact. The fracture surface topology of each specimen after conducting the Charpy tests are compiled in Figure 6.25.

Figure 6.25: Fracture surface topology for each interfacial joint design type (a) F-0°, (b) V-0°, (c) S-0°, (d) F-90°, (e) V-90°, and (f) Stepped-90°

Figure 6.26: Macroscopic photos showing crack trajectory for each interfacial joint design type (a) F-0°, (b) V-0°, (c) S-0°, (d) F-90°, (e) V-90°, and (f) Stepped-90°

6.4 Summary

The fabrication of six types of interfacial joint designs (F-0°, F-90°, V-0°, S-0°, V-90°, and S-90°) was done by depositing SS316L material using LMD techniques over three SS316L substrates, dimensionally designed to extract a series of Charpy impact test samples such that the notch is located at the interfacial joint. An instrumented Charpy impact test was done to evaluate the impact strength of the interfacial joint designs.

The results from the Charpy impact test show that the S-90° design produced the highest impact strength, followed by V-90°, S-0°, V-0°, F-0° and F-90° respectively. The Charpy impact test results show that the V-shaped and stepped interfacial joint designs produce a significant improvement in the mechanical properties of the interfacial joints. This improvement offers a validation to the proof-of-concept to the proposed design in the published patent.

One challenge posed in this study was the formation of defects at the interface that reduced the impact strength of the samples. Reducing the defects at the interface is a significant challenge due to the complexities that arise from the initial thermal conditions at the substrate. The steep thermal gradient between the substrate and the first deposition layer result in the formation of various defects at the interface. Future work following this study should include attempts at pre-heating the substrate or applying a higher laser power input at the first few layers and adjusting down the energy density as more layers are deposited. One measure of the effectiveness of pre-heating the substrate is the evaluation of the dilution zone between the additively deposited material and the substrate. Dilution is the phenomenon of mixing the substrate metallic material with the additively deposited material by the means of melting at the fusion zone. The degree of dilution is highly dependent on the processing parameters of the LMD process. An increase in thermal energy input during the pre-heating of the substrate results in a larger dilution zone. In cladding processes, a general rule of thumb of acceptable dilution is between 10-15%, dependent on the component's environment, conditions, and application. To achieve an acceptable dilution zone that enhances the strength of the interface requires a pre-heating temperature of the substrate of between 250°C to 300°C [199, 200]

The future study should also include full parametric study to achieve a more consistent thermal history and reduce the number and size of defects that occur at the interface. Further work can be done to continue to examine the effects of the interfacial design on the other forms of mechanical performance. For instance, tensile test coupons

could be fabricated to evaluate the strength of the interfacial joint designs when tested in tension. Other forms of examination could also be used to compliment the analyses of the interfacial joint design specimens like SEM and EDX techniques, focusing on the nature of the defects that arise from one instrumented mechanical test coupons to another, and how defects play a role in affect the various forms of tests.

Chapter 7: Finite Element Modelling of Novel Interfacial Joint Designs (Study 5)

7.1 Introduction

In this section, an analytical modelling methodology for the prediction of mechanical testing of various types of interfacial joint designs presented. Three interfacial joint designs, as described from Chapter 6 as Flat, V-shaped, and Stepped interfacial joint designs, undergoing a Charpy impact test were modelled in this study. Significant research has been devoted in the recent years to investigating the LMD processing parameters effects on the mechanical and material characteristics via finite element modelling techniques. Despite the amount of progress, little to no research has been done on the modelling LMD built parts with the focus on optimising and improving the strength of the interfacial joint. The present simulation work presents a model based on the Charpy impact tests conducted in the previous study, comparing the energy absorbed across the three interfacial joint designs.

7.2 Methodology

FE models of the Charpy impact test specimens were developed in ABAQUS Explicit software incorporating the Johnson-Cook plasticity and damage models. The Johnson-Cook model is a Mises plasticity model with analytical forms of hardening law and rate dependence and is suitable for cases where the material undergoes a high strain-rate deformation. The Johnson-Cook plasticity model was used in the finite element modelling of the process in this study due to its model accuracy for high strain rate deformation cases and its suitability for FEM implementation and computational use. Since the Johnson-Cook model is an empirical model, this method works best for the purposes of finite element modelling when empirical data can be fed to the model for a more accurate result. In the case of this experiment, the instrumented test was conducted at room temperature for AISI Stainless Steel 316L material samples, hence modified versions of the Johnson-Cook model are not necessary to compensate for material fracture behaviour at elevated temperatures. The same material and process parameters are used in Chapter 3 and Chapter 4, and therefore, the previous studies form a substantive foundation for the

Johnson-Cook finite element model used in this study. The Johnson-Cook plasticity model can be described as a function of three analytical forms: the hardening law, strain rate, and temperature effect as seen in Equation (7.1).

$$\sigma_{eq} = [A + B\varepsilon_p^n] \left[1 + C \ln \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right] \left[1 - \left(\frac{T - T_{room}}{T_{melt} - T_{room}} \right)^m \right] \quad (7.1)$$

Where A is the yield strength, B is the strain hardening parameter, ε_p is the effective plastic strain, n is the strain hardening exponent, C is the strain rate sensitivity parameter, $\dot{\varepsilon}$ is the effective plastic strain rate, $\dot{\varepsilon}_0$ is the reference strain rate under quasi-static test conditions, and m is the temperature exponent. From Equation (7.1), it can be seen that the Johnson-Cook model is a multiplicative decomposition of the current yield stress in three terms. The first term describes the model's strain effects expresses stress as a function of strain, also known as the hardening law. The second set of brackets describes the effects of the strain rate on the material's yield strength. The third set of brackets describes the thermal softening effect on yield strength [201, 202]. It should be noted that in the temperature term, at room temperature, the term reduces to unity.

Damage is defined as the process by which the load carrying capacity of the material gradually decreases until the complete failure of the material occurs. In this finite element model, the Johnson-Cook damage model was utilised alongside the Johnson-Cook plasticity model. The Johnson-Cook damage model is based on the equivalent strain at the element integration points where the cumulative strain term is updated at every analysis increment. The Johnson-Cook damage model is described in a similar analytical form as seen in Equation (7.2)

$$\varepsilon_f = \left[D_1 + D_2 \exp \left(D_3 \frac{P}{\bar{\sigma}} \right) \right] \left[1 + D_4 \ln \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right] \left[1 - D_5 \left(\frac{T - T_{room}}{T_{melt} - T_{room}} \right)^m \right] \quad (7.2)$$

Where D_1 is the initial failure strain, D_2 is the exponential factor, P is the pressure stress, $\bar{\sigma}$ is the Von-Mises stress, D_3 is the triaxiality factor, D_4 is the strain rate factor, and D_5 is the temperature factor.

The modelling parameters used in this study were taken from Chapter 4.3.4 tensile test results. The stock condition's specimen material and the substrate material were procured similarly and has the same elemental constituency, hence the mechanical test results for the Stock condition from Chapter 4.3.4 were taken to be the

substrate SS316L material in this study. On the other hand, the Short Unidirectional condition's results were taken to be the LMD SS316L material in the modelling due to the similarities between the two study's deposition raster scan patterns with respect to the fracture propagations.

In order to determine the parameters for the Johnson-Cook plasticity model, mechanical properties obtained from previous tensile test results in Chapter 4.3.4 were used as parametric input into the model. The Stock condition was used as the substrate's base material properties, and the Short Unidirectional condition was used as the LMD's base material properties. The build strategy used in the interfacial joint design, within the stepped or V-shaped slot joint is a unidirectional raster scan oriented perpendicular to the striking edge of the Charpy impact hammer as seen in Figure 6.4. The Short Unidirectional raster scan build strategy condition's results were used as the LMD condition in this FEM because of its resemblance to the model in terms of its deposition track's orientation to the fracture propagation. From the stress-strain graphs of these two conditions, the parameters for the hardening law term were obtained by applying a power-law fit to the stress-strain curves as seen in Figure 7.1, Figure 7.2 and Table 7.1. Since the plasticity model parameters are taken from the LMD empirical studies from the previous studies, the present model does not require the simulation of the thermal history of the LMD process.

Figure 7.1: Stress-strain plots and power-law trendline fit for Stock SS316L samples (a) a, (b) b, and (c) c

Figure 7.2: Stress-strain plots and power-law trendline fit for LMD SS316L samples (a) a, (b) b, and (c) c

Table 7.1: Power-law fit equations for Stock and LMD SS316L specimens

Specimen	Trendline Fit	Fitted Equation
Stock SS316L	Power-Law Fit - a	$y = 707.91x^{0.17}$
	Power-Law Fit - b	$y = 682.71x^{0.15}$
	Power-Law Fit - c	$y = 702.24x^{0.16}$
LMD SS316L	Power-Law Fit - a	$y = 725.23x^{0.12}$
	Power-Law Fit - b	$y = 722.05x^{0.12}$
	Power-Law Fit - c	$y = 730.47x^{0.12}$

To obtain the Johnson-Cook plasticity model's parameters, input values were taken from a combination of past work as well as existing publications. For the hardening law term, the yield stress results from Chapter 4.3.4 were used as input for the yield stress A term and the power-law fit equations from Table 7.1 were used as input for the strain hardening parameter B and the strain hardening exponent n terms. The test was conducted at room temperature; hence as observed from Equation (7.1), the temperature exponent constant becomes null, and the temperature term reduces to unity. The strain rate sensitivity parameter C was obtained from existing literature on the material plastic behaviour of SS316L in stock [203] and in LMD forms [204] and the strain rate value was obtained from the study in Chapter 4.3.4. The values for the Johnson-Cook plasticity model for the Stock SS316L and the LMD SS316L conditions are compiled in Table 7.2 and Table 7.3. The Johnson-Cook damage model's parameters were obtained from existing literature on the material damage behaviour of SS316L [205]

Table 7.2: Johnson-Cook plasticity constitutive model parameters for Stock SS316L condition

Terms	Units	a	b	c	Mean
A	MPa	322	350	329	334 ± 12
B	MPa	708	683	702	698 ± 11
n	-	0.17	0.15	0.16	0.16 ± 0.01
C	MPa	-	-	-	130
$\dot{\epsilon}/\dot{\epsilon}_0$	-	-	-	-	0.05
T	°C	-	-	-	T_{room}
m	-	-	-	-	N.A.

Table 7.3: Johnson-Cook plasticity constitutive model parameters for LMD SS316L condition

Terms	Units	a	b	c	Mean
A	MPa	348	400	407	385 ± 26
B	MPa	725	730	722	726 ± 3
n	-	0.12	0.12	0.12	0.12 ± 0
C	MPa	-	-	-	176
$\dot{\epsilon}/\dot{\epsilon}_0$	-	-	-	-	0.05
T	°C	-	-	-	T_{room}
m	-	-	-	-	N.A.

Table 7.4: Johnson-Cook plasticity constitutive model parameters for LMD SS316L condition

Terms	Units	Value
D ₁	-	0.05
D ₂	-	3.44
D ₃	-	2.12
D ₄	-	0.002
D ₅	-	0
T	°C	0
m	-	NA

In ABAQUS, the Charpy impact test specimen and the Charpy machine's anvil hammer were modelled with accordance to the empirical test dimensions. The anvil was modelled as an elastic body and the Charpy specimen as an elasto-plastic material. The section geometry modelled toward the centre of the simulated Charpy sample was designed with accordance to the interfacial joint design used in the instrumented Charpy test from Chapter 6: F-0°/F-90°, V-0°, S-0°, V-90°, and S-90°. The Flat interfacial design for both notch orientations (F-0° and F-90°) are virtually indistinguishable in the implementation of the FEM, hence the computational results from the FEM were used for both conditions. The FEM in this study does not include any form of defects in its model and will be a simulated representation of an ideal defect-free part.

The meshing of the part in FEM plays a crucial role in the formulations and calculations involved in the simulation of the crack propagation and initiation. The mesh size used at different sections and regions were optimally determined after a few runs. In FEM, using uniformly refined meshes across the model is considered a highly impractical from a computational viewpoint regardless of the capability of high performance computers. It is necessary to identify a region of interest within the model where high-accuracy computation is required and consolidate most of the computation effort to that zone as opposed to a uniform approach to the entire system. By experimenting with mesh refinement or mesh coarsening at different areas of the Charpy specimen, it was

concluded that introducing mesh coarsening at a significant distance away from the impact zone can keep the number of elements low without losing much accuracy in the solution. A few preliminary runs were partially run to test the validity of this and a balance between the computational time and model accuracy was eventually found and presented in this study. Using a uniform mesh across the model was found to use extensive computational power and have insignificant difference to the accuracy between the model and the instrumented Charpy test.

The C3D8R 8-node linear brick meshing was applied at all sections of the Charpy specimen where possible, and locations unable to be meshed with this technique were partitioned off and meshed using C3D4 4-node linear tetrahedron. A fine mesh was applied at sections where crucial loading is applied and where the crack initiation and propagation is known to occur, and a coarse mesh was applied at all remaining non-crucial sections. The element size used in the fine mesh is 0.1, while the element size used in the coarse mesh is 0.8. The mesh seeding for regions where the fine and coarse mesh is applied to is illustrated in Figure 7.3 (a). Boundary conditions were added to apply a vertical load from the hammer at 5 m/s, and simply supported Charpy sample located 7.5 mm away from each end with accordance to a 40 mm span length as per standard as seen in Figure 7.3 (b). The Johnson-Cook failure model was used as the criterion for deletion of elements when the damage parameter reaches its pre-defined value. The model uses an explicit time integration and was set to run for a total of 8 ms. The partitioning and the mesh used in the FEM for each condition is shown in Figure 7.5 to Figure 7.9. The number of meshed elements for the five modelled Charpy samples are $>1.4 \times 10^6$.

In order to simulate the interfacial defects that were observed in Chapter 6.3.2, defects of 0.01 width in the form of ellipses of varying major lengths were added at the interface of each finite element model. The size of each interfacial defect used are approximated based on the defect measurements done from the previous study. Due to the complexity of distribution of the smaller porosities and the dominating effect of the larger interfacial porosities, only the large interfacial porosities were considered in the finite element model. The dimensions and locations of each defect are shown in Figure 7.4.

Figure 7.3: (a) Mesh seed for fine and coarse mesh size and (b) Boundary conditions for hammer and Charpy sample

Figure 7.4: Defect dimensions for each interfacial joint design
(a) F, (b) S-0°/90°, and (c) V-0°/90°

Figure 7.5: F-0°/F-90° interfacial joint design partitioning (a) isometric view, (b) front view, and mesh's (c) isometric view, (d) front view

Figure 7.6: V-0° interfacial joint design partitioning (a) isometric view, (b) front view, and mesh's (c) isometric view, (d) front view

Figure 7.7: S-0° interfacial joint design partitioning (a) isometric view, (b) front view, and mesh's (c) isometric view, (d) front view

Figure 7.8: V-90° interfacial joint design partitioning (a) isometric view, (b) front view, and mesh's (c) isometric view, (d) front view

Figure 7.9: S-90° interfacial joint design partitioning (a) isometric view, (b) front view, and mesh's (c) isometric view, (d) front view

7.3 Results and Discussion

The deformation energy absorbed in an instrumented Charpy test is typically recorded as a single value at the end of the experiment. However, in an FEM, the deformation energy absorbed can be measured as a function of time. The results for the FEM simulations at crack initiation and post-fracture are shown in Figure 7.10 and Figure 7.11, and the impact energy data recorded for the simulation are graphed in Figure 7.12 and Figure 7.13. From the results, it is evidently clear that the impact energy absorbed is significantly different for each interfacial joint design. The S-90° condition was observed to have the highest absorbed energy recorded among the specimens, followed by V-90°, S-0°, V-0°, and F-0°/F-90° conditions. This trend in the toughness properties between each condition are similar to that measured empirically from the instrumented Charpy test results obtained from the previous study in Chapter 6.3.3 in terms of the ranking of impact toughness. The toughness properties in the finite element model are in good agreement to the instrumented Charpy test results.

From previous studies in Chapter 3 and Chapter 4, it has been established that the directionality of the build orientation has a direct effect on the microstructural grain structures and growth. In the case of this additively manufactured Charpy test samples, the substrate acts as heat sink where the thermal energy applied at the melt pool dissipates downwards into the previously deposited layers and into the substrate. This directionality of the heat transfer produces a microstructure that is needle-like, acicular crystallite phase that is oriented vertically from the substrate. This preferential growth is further explored in Chapter 4.3.5 where the LMD samples in that chapter used the same process parameters used in Chapter 6 and this study. The anisotropy reported in earlier chapters is similarly applicable in this study, where the mechanical strength of the additively manufactured part is stronger in the direction parallel to the acicular grain growth as opposed to perpendicular. Karnati et al. investigated the effects of the build orientation in the 0°, 45°, and 90° angles with respect to the substrate plane of additively manufactured stainless steel Charpy samples on their impact strength and found that the samples oriented 90° to the substrate plane exhibited the lowest impact strength followed by 45° and 0° respectively [206]. P. Moore et al. conducted a similar study where stainless steel Charpy samples were additively manufactured in 0° and 90° orientations relative to the substrate and underwent an instrumented Charpy test to compare the differences in impact strength, and found that the samples oriented 90° also exhibited a lower impact strength than samples

oriented 0° [207]. Numerous reasons were cited as the main contributors to the anisotropic properties arising from the differently oriented samples. The directional crystallite phase that forms as a result of the directional thermal energy flow generally results in more plastic deformation before failure in vertically oriented samples. The melt pools produced by laser metal deposition typically undergo high cooling rates and have steep thermal gradients that exacerbate the growth of the columnar crystallite texture.

Despite the similarities in the trend, there are very slight differences between the results of the empirical and the simulated Charpy tests. The simulated Charpy tests were observed to be slightly higher in magnitude in its impact energy compared to the empirical tests. During the LMD process, the melt pool size and temperature for the first few deposition layers were observed to be smaller and lower, especially within the V-shaped and Stepped interfacial joint design slots, via the process monitoring data as seen from Chapter 6.3.1. This observation is the case due to a combination of the substrate's cooler initial temperature as well as an increase in substrate surface area contact in the interfacial design geometry. The larger surface area contact within the interfacial slot exacerbates the thermal diffusion effects via conduction through the substrate that results in difficulties in the fusion between the interface and the LMD build. Defects like interrune pores and incomplete fusion that occur uniquely at the interface arise due to these thermal issues as seen in Chapter 6.3.2. These defects prove to be critical due to their role in directing the crack initiation to propagate along the defect locations as seen their fracture surface topology from Figure 6.25. This is evidenced in the differences in the fracture propagation between the empirical study and the model, where the fracture occurs along the interfacial joint design's interface as opposed to the fracture occurring across the interfacial joint design's slot. Although the interfacial defects were attempted to be modelled in the simulation model, the effects of the smaller porosities were not considered and an exact match of the geometry of every defect is not possible at this juncture. These exceptions account for the slight differences in the model and the instrumented test results.

In the FE model, the fracture propagation that occurs through the Charpy impact specimen and its interaction with the sectioned partitions and mesh play a crucial role in the computed impact energy. The FE model, for the most part, mimics the setup of a simple two layer metallic composite. The Flat interfacial joint design's section geometry is essentially a flat plane that favours the fracture propagation across the flat interface, as seen in Figure

7.11 (a). On the other hand, the V-shaped and Stepped interfacial joint designs have angled section geometries that do not favour the direction of fracture propagation, resulting in a coarse and uneven form of fracture as seen in Figure 7.11 (b) – (e). This difference in the fractures is reflected in the impact energy results seen in Figure 7.12 and Figure 7.13.

Figure 7.10: Von-Mises stress contours at point of crack initiation for (a) F-0°/F-90°, (b) V-0°, (c) S-0°, (d) V-90°, (e) S-90° interfacial joint designs

Figure 7.11: Von-Mises stress contours after fracture for
(a) F-0°/F-90°, (b) V-0°, (c) S-0°, (d) V-90°, (e) S-90°
interfacial joint designs

Figure 7.12: FEA results of impact energy as a function of time

Figure 7.13: FEA cumulative impact energy

7.4 Summary and Recommendations

An analytical FEM was done for the prediction of Charpy impact testing of specimens designed to improve the interfacial strength between a substrate and an LMD built part. The modelling results confirm the trend observed empirically in an instrumented Charpy impact study for interfacial joint designs: F-0°/F-90°, V-0°, S-0°, V-90°, and S-90°. The results for the FEM show a similar trend of impact strength per interfacial joint design to the instrumented empirical Charpy test results. The S-90° design exhibited the highest impact strength, followed by V-90°, S-0°, V-0°, F-0°/F-90°.

Future work to improve the accuracy of the finite element model should include the geometrical and locational mapping of each micro and macro defect and applying these mapped defects into the model. Without a one-to-one mapping, the model merely serves as an approximation of the broad trends of LMD. Although the current finite element model is sufficient for the purposes of the scope of the study, any further forms of precision measurement would require a more complex model that includes the effects of the many types of defects. The common forms of defects come in the form of porosities and incomplete melting that have distinct geometrical forms that serve as stress concentration sites in fracture behaviour.

Optimising the LMD build to minimise the defects at the interface to an amount where the instrumented test can be comparable to the model poses a significant challenge. The novelty of the patented interfacial design compounded with the unique ability for LMD to add features to existing components means that publications that touch on the subject even adjacently are virtually non-existent. In order to realise the full extent of the design's potential, an in-depth optimisation study to minimise the detrimental effects of the interfacial defects is required. Future work following this study should include a parametric study on the number, size, and type of defects that occur at the interface. An extensive study that establishes a process window for the design features of the interfacial joint design can be explored empirically and verified by finite element modelling.

The scope of this study covers the application of Charpy impact at room temperature for stainless steel and hence the standard Johnson-Cook model is suitable in this form. Future iterations of this experiment can include instrumented tests done for different materials at elevated temperatures. For metallic alloys like tungsten alloys, WC-Co alloys, Fe-Al alloys, NiAl-based alloys have ductile-brittle transition phases that occur at elevated

temperatures. In future case studies of additively manufactured materials that have this transition phase, other forms of the Johnson-Cook model have to be considered where modifications to the model take into account the relationships of the flow stress, strain rate, and temperature of these materials at elevated temperatures by compensation of strain and strain rates. Classical stress-dislocation relations and kinematics of dynamic recrystallization, flow stress constitutive equations for work hardening-dynamical recovery periods and dynamical recrystallization periods can be established.

Chapter 8: Conclusions and Future Work

8.1 Summary

The research findings in the present work can be divided into five contributions, Study 1 to 5. In Study 1, defects, and microstructure characterization, as well as the study of hardness properties of LMD Stainless Steel AISI 316L was explored. In Study 2, defects, process mapping, material characterization and mechanical testing of LMD Stainless Steel AISI 316L was conducted. In Study 3, a methodology for a novel interfacial joining technique was developed for LMD that yields an improvement to mechanical strength at the interface, and an evaluation of various interfacial designs were explored. In Study 4, as a follow-up to the previous study, a more in-depth review of different interfacial joint designs was conducted by analysing the process characteristics, defects, and conducting instrumented Charpy impacted tests to evaluate the performance characteristics. In Study 5, process and material parameters were taken from all four previous studies to simulate the performance characteristics of the interfacial joint designs, which provides a prediction tool for future iterations and modifications to the design.

8.2 Conclusion

8.2.1 Defect and Microstructure Characterization (Study 1)

In Study 1, a systematic materials-process-performance approach to characterising laser metal deposition in the form of two studies that encompass defect and microstructure characterizations was presented. The build strategy's effect on Materials-Process-Performance tensile anisotropy that serve as parametric input for a subsequent study of the development of a novel interfacial joint design.

The study on “Defect and Microstructure Characterization”, investigated the material characterization of LMD Stainless Steel AISI 316L's porosity, density, and defects at four localities within an LMD specimen by microscopy. The majority of pores detected fell within the $0.0\ \mu\text{m}$ – $20.9\ \mu\text{m}$ range, and part densities were measured to be 99.8% and above. A larger population of small pore sizes, and slightly higher part densities were measured in the top zone area than the bottom zone area. Further optimisation can be employed in the dynamic

control of laser power input, or a pre-heating to the substrate prior to the build can be applied to achieve a more consistent build.

In this study, conclusions about the sources of defects, fusion line and track body compositions, and austenitic and ferritic characterizations are offered. The primary source of defects in this build is air entrapment from moisture evaporation, and local voids after powder layer deposition. Oxygen element was detected at the fusion lines that is not seen in the deposition body areas. This is due to the natural formation of a protective oxide layer after each track of Stainless Steel AISI 316L is deposited and exposed to the atmosphere. In the EDX mapping analysis columnar microstructures were observed in various element maps. This observation is the result of an AF solidification mode that leads to the formation of columnar δ -ferrites parallel to the solidification direction.

8.2.2 Build Strategy Effect on Materials-Process-Performance Properties (Study 2)

In Study 2, “Build Strategy Effect on Materials-Process-Performance Anisotropy”, looks into the build strategy’s effect on LMD Stainless Steel AISI 316L materials-process-performance characteristics. In this study, the effects of Additive Manufacturing build strategies using LMD fabrication of Stainless Steel 316L specimens provided a framework for characterizing the anisotropic properties and behaviour arising from build strategy pattern effects. A series of tensile tests were used to characterise the mechanical anisotropy of LMD specimens that were produced using long unidirectional, bidirectional, and short unidirectional raster scan build strategies. Mechanical anisotropy was established by correlating the LMD build process induced grain size effects to yield strength using the Hall-Petch relations. Micro-indentation hardness measurements show differences between the fusion zones and deposition track body locations showing the effects of build pattern on localized mechanical properties. Mechanical anisotropy was observed between different build strategies: short unidirectional, long unidirectional and bidirectional raster scans in the tensile tests, whereby the long unidirectional raster scan build strategy exhibited the highest UTS, yield strength, and fracture strain, followed by the bidirectional and short unidirectional raster scan build strategies. The reverse trend was observed for the Young’s modulus. LMD specimens exhibit higher UTS and yield strength, and lower fracture strain than conventionally manufactured stock bar specimens.

This is due to the δ -ferritic phase, acicular crystal grain structures that are vertically oriented from the substrate in the LMD fabricated part. The stock bar specimens showed the typical austenitic grain structures giving ductile mechanical properties from the tensile test result. The differences in thermal histories between the three build strategy raster scan patterns result in differences in the microstructural grain sizes, whereby the short unidirectional raster scan build strategy exhibited the largest grain size, followed by the bidirectional and long unidirectional raster scan build strategies respectively. The Hall-Petch model fitted in this study show anisotropy in yield strength with respect to the grain size arising from the different build strategy pattern effect. The fracture surface topologies exhibited a transverse and longitudinal fracture effect in the short unidirectional and long unidirectional raster scans respectively. These fracture topology patterns and the anisotropic observations from the tensile tests show that the arrays of deposition track patterns play a role in the eventual built specimen anisotropic properties and fracture mode.

Microindentation at the fusion zone and deposition track body locations confirm that another source of anisotropy is due to localized strengthening at the fusion zones. The different arrays of fusion zones arising from the build strategy effects contribute to the anisotropy in the built specimen's tensile properties. In order to achieve melt pool size and temperature process consistency for different build strategies, the laser power regulation has to be dynamically controlled based on in-situ measurements. This is due to the differences in the state of each deposition layer has cooled with respect to the raster scan implementation when moving from one layer to the next. Different build strategies cause different temperature readings at the beginning of each new deposition layer, hence affecting the manner in which the power regulation has to be applied.

The effects of Additive Manufacturing build strategies using LMD fabrication of Stainless Steel 316L specimens provided a framework for characterizing the anisotropic properties and behaviour arising from build strategy pattern effects. In-situ process control was used to investigate the build strategy effects on the laser power control required to achieve consistent melt pool characteristics. A series of tensile tests were used to characterise the mechanical anisotropy of LMD specimens that were produced using long unidirectional, bidirectional, and short unidirectional raster scan build strategies. Mechanical anisotropy was established by correlating the LMD build process induced grain size effects to yield strength using the Hall-Petch relations. The δ -ferrite constituent of each

sample was determined using Schaeffler diagram Cr_{eq}/Ni_{eq} measurement methods. Micro-indentation hardness measurements show differences between the fusion zones and deposition track body locations showing the effects of build pattern on localized mechanical properties.

8.2.3 Interfacial Structural Design and Fabrication Methods for Additive Manufacturing Applications (Study 3)

In Study 3, a preliminary study of a set of pre-defined novel interfacial joint designs based on a published patent for select design parameters was conducted. A DOE analysis with two levels for three design parameters: Step angle, step height, and feature geometry were selected to determine their effects on the build quality.

Two levels for three design parameters: Step angle, step height, and feature geometry were selected to understand their effects on the build quality. The ultrasound, x-ray tomography, and microscopy techniques were used to characterise defects that form within the interfacial joint design. Measurements for A60-H2-FCV and A60-H2-FCC specimens show that large-sized defects were inclined to form at the sharp corners of the interfacial joint design. These defects were also observed to form in a line parallel to the longitudinal edge of the specimen. This is a result of a lack of powder flow access to these corners due to issues from the powder flow dynamics that arise from interference with the step design geometry. The defect measurement by microscopy showed that the Step Height and Step Angle factors have a significant effect on the formation of large defects, while the Geometry Feature factor has an insignificant effect. The main effects trend from the microscopy measurement and the dilution and heat affected zone analysis are in agreement with defect study, for which factor had the most effect on the formation of large defects.

From this preliminary study, several lessons have been learned to improve the quality of the interfacial joint build. By setting an optimisation goal of lowest degree of large defects formed, the most effective combination of interfacial joint design would be to lower the Step Height and Step Angle. On top of this, tighter tolerances can

be made for the step joint features in order to ensure that more precise initiation point can be made for the LMD process that produces a more consistent process and a higher quality build.

8.2.4 Interfacial Joint Design Comparative Study (Study 4)

In Study 4, the mechanical strength of several types of interfacial joint designs were studied. The fabrication of six types of interfacial joint designs (F-0°, F-90°, V-0°, S-0°, V-90°, and S-90°) was done by depositing SS316L material using LMD techniques over three SS316L substrates, dimensionally designed to extract a series of Charpy impact test samples such that the notch is located at the interfacial joint. An instrumented Charpy impact test was done to evaluate the impact strength of the interfacial joint designs. The results from the Charpy impact test show that the S-90° design produced the highest impact strength, followed by V-90°, S-0°, V-0°, F-0° and F-90° respectively. The Charpy impact test results show that the V-shaped and stepped interfacial joint designs produce a significant improvement in the mechanical properties of the interfacial joints. This improvement offers a validation to the proof-of-concept to the proposed design in the published patent.

8.2.5 Finite Element Modelling of Novel Interfacial Joint Designs (Study 5)

In study 5, An analytical FEM was done for the prediction of Charpy impact testing of specimens designed to improve the interfacial strength between a substrate and an LMD built part. The modelling results confirm the trend observed empirically in an instrumented Charpy impact study for interfacial joint designs: F-0°/F-90°, V-0°, S-0°, V-90°, and S-90°. The results for the FEM show a similar trend in ranking of impact strength per interfacial joint design to the instrumented empirical Charpy test results. The S-90° design exhibited the highest impact strength, followed by V-90°, S-0°, V-0°, F-0°/F-90. Optimising the LMD build to minimise the defects at the interface to an amount where the instrumented test can be comparable to the model poses a significant challenge. The novelty of the patented interfacial design compounded with the unique ability for LMD to add features to existing components means that publications that touch on the subject even adjacently are virtually non-existent.

In order to realise the full extent of the design's potential, an in-depth optimisation study to minimise the detrimental effects of the interfacial defects is required.

8.3 Research Contribution

AM is a fast-growing emergent technology that is proving to be highly disruptive to the manufacturing market. Gives way to novel manufacturing steps and techniques that has opportunities for optimization in designs and methods. LMD allows for fully dense metallic components to be built. LMD can build features or modify existing components for coatings and remanufacturing purposes. The interfacial strength between the added feature and the existing surface can be optimised by introducing the interfacial features detailed in this invention. Essentially, by employing an interfacial projection design in the form of a stepped joint interface, an improved interfacial bond can be achieved. A stepped interface spreads an acting load over a larger area at the joint, hence strengthening it. The interfacial projection recipient is typically the substrate material. The projection recipient's stepped features can be prepared by, but not limited to, applying subtractive manufacturing methods to the substrate prior to depositing material to form the joint body.

In this research, an investigation into the MPP characterization into LMD AISI Stainless Steel 316L where early into the study, a fundamental examination of LMD build strategy's effects on MPP properties that serve as crucial parametric input for the development of a novel interfacial joint technique for DED applications. An overview of the research work is shown in Figure 1.1. The novelty of the research presented in this thesis is the work done on a novel interfacial joint design. In this study, several forms of novel interfacial designs were compared against the conventional flat interfacial joint design using MPP techniques. Parametric input was taken from preliminary studies documented in Chapter 3 and Chapter 4 and fed to the development of the novel interfacial joint design. Empirically, the novel interfacial design produced significant improvements in mechanical strength over the conventional interfacial joint design. This finding was subsequently modelled and verified via FEM techniques.

The novel interfacial joint design unlocks three key competitive advantages that can propel LMD and hybrid technologies forward in a broader and more rapid industry adoption. The first competitive advantage is the enablement of superior interfacial strength. The additive manufacturing process allows for unique structures and

geometries – enabling the development of products that were once too expensive to make, or not feasible at all. Before industrial-scale additive manufacturing production, engineers needed to consider how manufacturing processes worked. As products are now built up layer-by-layer more complex elements, such as overhanging features, can be introduced. These complex structures can be strengthened and optimized by using the proposed invention's interfacial structure design method.

The second competitive advantage is the ability to capitalize on hybrid manufacturing technologies. Hybrid manufacturing is an emergent technology within the additive manufacturing sphere that attempts to streamline and simplify the additive manufacturing process into conventional manufacturing lines. By combining additive and subtractive in a single machine, the incorporation of additive manufacturing into a manufacturing line becomes much more simplified. Hybrid manufacturing can be used to create the invention's interfacial features. Subtractive manufacturing can first be used to create the interfacial steps prior to using additive manufacturing to build up the intended feature.

The third competitive advantage is its ease of incorporation with existing manufacturing processes. Conventional manufacturing process lines typically include subtractive manufacturing methods like computer numeric control (CNC) milling and turning machines. Companies that incorporate additive manufacturing into their conventional process lines can utilise the invention easily. Apart from an additional process step, no other significant cost will be incurred when using the invention.

8.4 Limitations and Recommendations for Future Work

The research work done in thesis encompasses several fundamental LMD studies that lead up to the development of a novel interfacial joint method. Although strides have been made in this research that amounts to a proof of concept for the proposed interfacial joining technique, additional work can be done in the form of future studies that follow this research.

In this thesis, the effects of LMD build strategies on material and mechanical properties provided framework for characterizing the anisotropic properties and behaviour arising from raster scan patterns. Although the relationship between the process parameters and the performance properties were established in this thesis, a more in-depth

look into a quantified relationship between the material properties and the process-performance properties could be explored. In future work, the formation of an empirical model that establishes a regression model correlating the material properties to the process properties can be done via DOE techniques like the response surface method and statistical analysis using ANOVA. In doing so, the regression model can form an approximate model between the input and output parameters of process parameters and material properties respectively. Process parameters like laser power, scanning speed, and powder flow rate could be evaluated as a function of material properties like grain sizes, grain orientation, or austenitic/ferritic concentrations. For instance, two levels of process parameters can be used as the input for the DOE parametric setup. Grain size measurements, EBSD grain orientation mapping, or δ -ferrite concentration contents can be added as the DOE output. From here a regression model can be computed, serving as an approximation of the relationship between the process parameters and material properties domains can be made.

Another potential area for future work following this thesis is the establishment of a type of dynamic control that utilises LMD. As the technology matures and finds its place within the manufacturing industry, the demand for a reliable control system that utilize a closed-loop feedback framework becomes a growing priority. Although a lot of work has been done in literature relating to process monitoring of LMD processes, a gap remains between the use of process monitoring techniques and the development of closed-loop controls. One possible area for future research is the exploration into both intrusive and non-intrusive closed-loop systems to achieve and maintain a homogeneous LMD build with a consistent thermal history across different forms of build geometries. For instance, although many process monitoring systems use sensors like pyrometers and infrared sensors to observe the melt pool characteristics and for the purposes of closed-loop control, little effort is made into more intrusive versions of controls that monitors parameters like the consistency of the powder flow rate. Future work can look into the integration of sensors and the development of closed-loop controls for the key process variables of LMD.

One challenge posed in this study was the formation of defects at the interface that reduced the impact strength of the samples. Minimising the defects at the interface is a significant challenge due to the complexities that arise from the initial thermal conditions at the substrate. The steep thermal gradient between the substrate and the first deposition layer result in the formation of various defects at the interface. Optimising the LMD build to minimise

the defects at the interface to an amount where the instrumented test can be comparable to the model poses a significant challenge. Future work following this study should include attempts at pre-heating the substrate or applying a higher laser power input at the first few layers and adjusting down the energy density as more layers are deposited. The future study should also include full parametric study to achieve a more consistent thermal history and reduce the number and size of defects that occur at the interface. In order to realise the full extent of the design's potential, an in-depth optimisation study to minimise the detrimental effects of the interfacial defects is required. Future work should also include a parametric study on the number, size, and type of defects that occur at the interface. Farshidianfar et al. applied a pre-solidification pre-heating to define the primary solidification heat cycle's initial temperature before deposition that achieves a more consistent microstructural grain growth and minimise defect formation [208]. Apart from applying a pre-solidification pre-heating, additional localised control was used to regulate the cooling rate to a uniform rate at all deposited layers. This is achieved by using a closed-loop DED system to eliminate cooling rate variations associated with open-loop depositions. Nassar et al. utilized an optical emission spectroscopy coupled with a custom-built data acquisition and control infrastructure to detect and minimise defects in real-time pushing the way forward for sensing strategies for defect detection and in-situ repair [209]. A range between 450°C to 600°C can be used as a pre-heating temperature in order to ensure a metallurgical bond between the deposited material and the substrate without the substrate becoming over-melted [210].

In future iterations of this study, a form of closed-loop control could be implemented to achieve a more consistent build, minimise the amount of defects, and better isolate the effects of the interfacial design parameters on the build quality. Kistler et al. examined the effects of processing parameters on the mechanical properties and repair quality of DED produced samples where variables including interlayer dwell time, initial substrate temperature, substrate thickness, hatch pattern deposition strategy, and number of layers deposited were studied [211]. Porosities were found to concentrate near the substrate surface resulting in a lower overall density for samples composed of fewer layers compared to samples with more layers. It was concluded that no single individual factor can be considered a main driving parameter in its impact on part porosity, but further investigation is warranted in better understanding the effects of the variables on the part quality. In future iterations of the interfacial joint

design study, an optimization study could be explored to minimize the amount of defects at the interface by varying a similar set of processing parameters.

To recreate a one-to-one finite element model that is much closer to the instrumented test studies, mapping of every micro and macro defects in the specimen onto the model is necessary before running the simulation. Including the mapping of the defects would produce a far better approximation of the mechanical properties of LMD specimens. Jiang et al. incorporated various forms of defects into the finite element modelling of AM fabricated lattice structures to predict the stress-strain and stiffness as a function of the introduced defects [212]. Surface roughness was introduced by including elliptical porosities, and surface roughness was introduced by discretizing a sinusoidal function and fitting experimental roughness data by Fourier Transformations. These methods produce a broad approximation to actual defects that serve as an elementary approach to computationally quantifying the effects of common defects in AM parts. Biswal et al. explored ideal spherical and oblate spherical pore geometries in FE predictions of fatigue life and maximum stress in strain-life relationship via a Smith-Watson-Topper model [213]. The effects of isolated porosity defects on the fatigue strength of AM materials were evaluated based on the gas pore morphologies found empirically. Although this FE analyses has its many uses in the evaluation of AM parts, the computational power required is considered unscalable for larger sized components.

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Appendix A : Other Additive Manufacturing Technologies

A.1 Selective Laser Melting

Selective Laser Melting (SLM) is an AM technique whereby a laser beam is used to scan over a bed of powder material controlled by a piston mechanism. After a layer is fused by the laser scan, the piston lowers the bed by a layer's thickness and a roller spreads a new layer of powder material over the previous layer. The process repeats until the entire component is manufactured. The SLM process makes use of a laser beam that is controlled and directed via a series of optic lenses to scan the laser across the layer of powder to create each layer forming a near 100% dense component. A schematic of the SLM process can be found in Figure A-1 below.

One of the limitations of SLM is the need for the laser beam to be positioned perpendicularly to the powder bed. This disallows angled deposition or over any non-flat surfaces. SLM deposits layers of powder from a bed of compact powder which act as support structures for any overhangs that may be necessary, hence eliminating the need for conventional support structures. Warping can still be an issue for SLM components, and may require an anchor base-plate for the first layer deposition. [214] An issue with SLM is the liquid metal flowing into the pores of the powder bed via balling effects or capillary action, resulting in formations of unconsolidated agglomerations of powder. [215]. To control this effect, the SLM process parameters can be optimised and controlled to ensure that the melt pool solidifies rapidly. Powders used in the SLM process are designed specially to control this effect. This is done by blending various types of materials in order to reduce the melting point, reduce the surface tension of the melt pool, and reduce the oxidation effects [214].

Figure A-1: Schematic of Selective Laser Melting [216]

The advantages and disadvantages of SLM can be summarised in Table A-1 below.

Table A-1: Advantages and disadvantages of SLM

Advantages	Disadvantages
• High quality metallic components • Large range of metal materials • High accuracy • Complex geometries possible	• High power consumption • Relatively slow build process • Size of components limited by chamber • May require machining to improve surface finish and geometrical finish.

A.2 Electron Beam Melting

Electron Beam Melting (EBM) is another AM technique whereby a computer-controlled electron beam melt metallic powders layer by layer. During the EBM process, a thin layer of metallic powder is spread onto a vertically controllable surface. The layer is fused with accordance to the CAD file with a computer-controlled electron beam in a vacuum chamber. The surface is subsequently lowered, and the next layer of powder is spread

over the previous layer. The process repeats until the entire component is built layer by layer. A schematic of the EBM process is shown in Figure A-2 below. The advantages and disadvantages of the EBM process is summarised in Table A-2.

Figure A-2: Schematic of Electron Beam Melting [217]

Table A-2: Advantages and disadvantages of EBM

Advantages	Disadvantages
• Superior material properties • Good geometrical accuracy • Excellent finishing • No support structures required	• Need to maintain vacuum chamber • High power consumption • Gamma rays

A.3 Summary

This chapter provides a summary of the three competitive metallic AM technologies, SLM, EBM, and LMD. Information covered includes their respective working principles, applications, advantages, and disadvantages. Amongst the metallic AM technologies, LMD holds the greatest flexibility in its configuration and can be

deployed in a free-format for large scale AM fabrication applications. The implementation of its coordinates system control allows it to be compatible with conventional 5-axis CNC systems. Hybrid manufacturing machines like DMG MORI's LaserTec 65 *3D* and LaserTec 4300 *3D*, as well as Mazak's INTEGREX i-400AM machines leverage on this compatibility by combining additive and subtractive processes in a single machine, hence streamlining the manufacturing process. The combination of additive and subtractive processes opens new opportunities in strategic build applications where structural features can be added to optimise the strength in AM built parts.

Despite the advantages in LMD's capabilities in large scale AM applications, a crucial drawback is its restriction in part complexity. SLM and EBM on the other hand have restrictions in its build size but have superior part complexity capabilities. A summary of the capabilities of each technology is compiled in Table A-3 and a comparison in the advantages and disadvantages of the three metallic AM technologies is compiled in Table A-3.

Table A-3: Summary of SLM, EBM, and LMD capabilities

	Selective Laser Melting (SLM)	Electron Beam Melting (EBM)	Laser Metal Deposition (LMD)
Energy Source	Laser	Electron Beam	Laser
Multi-material processing	Yes	No	Yes
Productivity vs Cost	Poor	Medium	Good
Part complexity	High	Medium	Low
Common Applications	Tooling, implants, metal components	Implants, near-net-shape manufacturing, turbine blades, prototypes	Near net-shape manufacturing, repair, coating, vanes, shafts, and ducts
Geometry Limitations and support structure requirements	Supports needed everywhere (thermal, anchorage)	Less supports needed, but powder is sintered	No supports needed. No powder bed. Similar to 5 axes milling.
Build job change	Fast	Medium	Fast
Materials	SS, tool steel, titanium, aluminum, ceramics, etc.	Only conductive materials (Ti6Al4V, CrCo, TiAl, Tool Steel, Cu alloy, etc.)	Steels, Ti, Ni-based alloys, composites, ceramics, etc.
Build Size (mm)	800 × 400 × 500 (Large) 250 × 250 × 280 (Mid-Range)	250 × 350 × 380 (Large) 200 × 200 × 180 (Mid-Range)	4000 × 2000 × 1000 (Large) 800 × 600 × 400 (Mid-Range)

Appendix B : Lasing Fundamentals

B.1 Lasing/ Laser Phenomenon

An atomic system can interact with electromagnetic radiation in three distinct ways. Energy band theory for different energy states of absorption, spontaneous emission, and stimulated emission are shown in Figure B-1. Absorption occurs when a system receives energy by interaction with other electrons, photons, or atoms, and transfers into an excited state. Spontaneous emission occurs when excited atoms decays spontaneously to a lower energy state by emitting a photon.

Lasers operate on the principle of stimulated emission where photons interact with a physical system in an excited state E_2 when the frequency matches the excitation energy of the system, hence stimulating the emission process. The emitted photon moves in the same direction and is in phase with the initial photon, hence enabling light amplification [218].

Figure B-1: Absorption, spontaneous emission, and stimulated emission [218]

When an atom, ion, or molecule interacts with an electromagnetic wave, an excitation proportional to the difference between the number of atoms at the upper and lower energy levels for the given transition is known as

the population difference. The population found between the two energy levels is governed by the probabilistic equation, Boltzmann ratio [219]

$$\frac{N_m}{N_n} = e^{\frac{-h^*f_{mm}}{k^*T}} \quad (\text{B.1})$$

Where “ N_m ” and “ N_n ” are populations at the excited level “ E_m ” and the ground level “ E_n ” respectively, “ T ” is temperature in Kelvin, and “ k^* ” is the Boltzmann constant (1.380622×10^{-23} J/K).

The temperature from equation (B.2) is in Kelvin, hence the equation will always be “ $N_m/N_n < 1$ ”. This indicates that at thermal equilibrium, there will always be fewer atoms at the higher energy level than the lower energy level. When in this state, the resonant nature of the “ m ” to “ n ” is absorptive. In order to achieve emission, a state must be reached such that there are more atoms at the higher energy level than in the ground level such that [219]:

$$N_m > N_n, \quad E_m > E_n, \quad \Delta N_{nm} < 0 \quad (\text{B.2})$$

A laser beam is light amplified by stimulated emission of radiation. An atom inherently has an upper energy state E_2 , and a lower energy state E_1 . Atoms interact with photons by an external pump source and are excited to a higher energy level. The atom restores thermal equilibrium by decaying to a lower energy level and releasing a photon with a frequency. This phenomenon is known as spontaneous emission. If a photon interacts with another atom in an excited state, another photon is emitted and travels in the same direction and phase as the incoming photon, resulting in a highly coherent photon emission. This phenomena is called stimulated emission, and is governed by [220-222]:

$$E' = E_2 - E_1 = \frac{hc}{\lambda} = h\nu \quad (\text{B.3})$$

Where “ h ” is Planck’s constant ($4.13566733 \times 10^{-15}$ eVs), “ c ” is the velocity of light (3×10^8 m/s), “ λ ” is the wavelength of electromagnetic radiation, “ ν ” is the frequency of electromagnetic radiation, and “ E ” is the energy of electromagnetic radiation.

A laser works based on the principle of light amplification by stimulated emission and can be generated in an optical laser cavity. The active medium serves to amplify light, the pumping source serves to excite the active medium to the amplified state, and the optical resonator serves to provide an optical feedback for amplification. A schematic of a resonator cavity is illustrated in Figure B-2 [218].

Figure B-2: Laser resonator cavity schematic [218]

To achieve a coherent laser beam, an optical cavity with highly reflective mirrors are positioned with the medium such that light can be reflected back and forth through the medium. One of the mirrors is 100% reflective and the other is a partially transmitting mirror. Light in the excited resonator is amplified several times, and is emitted through the partially transmitting window as a monochromatic and unidirectional laser beam [222-224].

B.2 Laser Types

Depending on the lasers' pumping method or mode of operation, the laser can be categorised into different types of classifications. The classification of lasers is generally dependent on the active medium used. Laser types include solid-state lasers, gas lasers, liquid lasers, and diode lasers [225-227].

Gas lasers use a gas as a gain medium and commonly employ an electrical discharge as the excitation pumping source. The container for the cavity is sealed in order to hold the gas medium. Some advantages of gas lasers are that it is comparatively inexpensive and is amorphous. This negates any concerns regarding damage to the medium due to thermal deformations. By incorporating a gas circulation system, heat can be regulated easily. Common gas lasers include helium-neon, also known as He-Ne lasers ($\lambda = 632.8 \text{ nm}$), argon lasers ($\lambda = 334 - 514 \text{ nm}$), CO_2 lasers ($\lambda = 10.6 \text{ }\mu\text{m}$), and excimer lasers like KrF ($\lambda = 249 \text{ nm}$), and ArF ($\lambda = 191 \text{ nm}$) [22, 225, 228, 229]

Solid state lasers use a solid as a gain medium. The solid is typically a nonconductive crystal doped with ions compatible with lasing. The crystals are typically manufactured into a cylindrical shape for lasing purposes. A high intensity lamp is typically used as an optical pump source. To avoid thermal distortion or cracking at the crystal, power limits are typically assigned, or the use of pulsed lasing can be used. Common solid state lasers include yttrium-aluminium-garnet doped with neodymium ions Nd:YAG ($\lambda = 1.06 \text{ }\mu\text{m}$), Ti:sapphire ($\lambda = 0.7 - 1.1 \text{ }\mu\text{m}$), Yb:YAG ($\lambda = 1.03 - 1.05 \text{ }\mu\text{m}$), Er:YAG ($\lambda = 2.94 \text{ }\mu\text{m}$) [225, 228, 230].

Liquid state lasers use a liquid as a gain medium. An example of a liquid laser is a dye laser ($\lambda = 585 - 595 \text{ nm}$). In a dye laser, the solvent containing the dye molecules is continuously pressed through a nozzle jet. The laser radiation passes the jet under a Brewster angle. The laser can be optically pumped with a separate laser or with a lamp [225, 228, 229].

A diode laser operates fundamentally differently from a solid, liquid, or gas state lasers. Instead of emitting photons by the process of energy decay between energy levels, photons are emitted by a change of energy between electrons in a semiconductor material. The gain medium in this case is a small semiconductor material that acts

as a diode. Unlike the other classifications of lasers, a pump source is not necessary for excitation. An advantage of diode lasers is it ease of delivering lasers at very high-power outputs at high energy efficiencies [228, 231, 232].

Appendix C : Laser Metal Deposition Fundamentals

C.1 Materials Used in Laser Metal Deposition

Table C-1: Comparison of findings from LMD materials

Material	Powder Feed	Laser Source	Findings	Ref
Ti-6Al-4V	Particle size: –100/+325 mesh	CO ₂ laser 6 kW	As deposited Yield Strength: 1105 ± 19 MPa UTS: 1163 ± 22 MPa Elongation: 4 ± 1 %	[120]
Ti-6Al-4V	Particle size: 25 – 45 µm	Nd : YAG laser (pulsed)	Yield Strength: 1100 ± 12 MPa UTS: 1211 ± 31 MPa Elongation: 6.5 ± 0.6 %	[121]
Ti-4Al-1.5Mn	Particle size: 45 – 420 µm	CO ₂ laser 5 kW	As deposited Toughness 599 ± 57 kJ m ⁻²	[122]
Inconel 625	Particle size: 45 – 135 µm	CO ₂ laser 6 kW	Defect free Columnar dendrites Hardness 254 ± 6 HV	[123]
Inconel 718	Particle size: 44 – 150 µm	CO ₂ laser 5 kW	As deposited Yield Strength: 590 MPa UTS: 845 MPa Elongation: 11 %	[123]
Rene 88DT	Particle size: 44 – 150 µm	CO ₂ laser 5 kW	HIP + heat treated Yield Strength: 1010 – 1030 MPa UTS: 1400 – 1440 MPa Elongation 16.5 % - 17.5 %	[124]
Tool Steel H13	Particle size: – 70 mesh	CO ₂ laser 4.5 kW	Yield Strength: 1505 MPa UTS: 1820 MPa Strain: 6% Hardness: 690 HK	[15]
Tool Steel H13	Particle size: – 70 mesh	CO ₂ laser 4.5 kW	Yield Strength: 1505 MPa UTS: 1820 MPa Strain: 6% Hardness: 690 HK	[15]
AISI 4340 Steel	Particle size: – 140/+325 mesh	Diode laser 1 kW	Microhardness 681 – 480 HV Porosity 4.13%	[125]
Stainless Steel 316L	-	CO ₂ laser 5 kW	1000 mm/min scanning velocity Yield Strength: ~400 MPa UTS: ~450 MPa Elongation: ~20%	[126]
CoCrMo	Particle size: – 100/+325 mesh	Nd : YAG laser 500 W	Hardness 40 HRC	[127]
Cu-30Ni alloy	Particle size: – 100/+325 mesh	CO ₂ laser 5 kW	Yield Strength: 316.16 MPa UTS: 240.49 MPa Elongation: 13.9%	[128]

C.2 In-situ Controls

Table C-2: Compilation of studies on in-situ controls for LMD

Sensor	Detection	Controller	Findings	Ref
2 CCD matrix cameras	CCD 1: Particle speed CCD 2: surface temperature distribution, height, and width of deposition track	-	Developed a low cost closed loop control system. Interpolated a linear relationship between grey level pixel and temperature Schematic:	[233]
CCD camera	Clad height Light intensity from melt pool	PID	Limited to simple straight deposition tracks, not to complex curvature tracks. Produced a thin wall structure. Schematic: Continued →	[234]

Sensor	Detection	Controller	Findings	Ref
Trinocular optical CCD-based detector	Clad height	Fuzzy thresholding technique	Feature tracking and trained network showed promising performance in detection of clad height in various trajectories angle in real time with 10 Hz and ± 0.15 mm precision. Schematic:	[235]
Trinocular optical CCD-based detector	Clad height Width Rate of solidification	Fuzzy adaptive threshold technique	Developed an algorithm that provides a good estimate of the solidification rates for four regions. Schematic: Continued →	[236]

Sensor	Detection	Controller	Findings	Ref
Trinocular CCD	Track dimensions Surface finish	-	Developed an in-time motion adjustment strategy based on the relationship between laser head motion relative to the growing part and other parameters in effects on clad profile. Remedy and eliminate defects, improve dimensional accuracy, and surface finishing. Schematic:	[237]
CCD camera Line laser	CCD: Clad height Line laser: Clad profile	-	A complete non-contact clad height measurement approach. Line detection algorithms developed to locate laser line despite laser line diffraction and variation in the laser line brightness profile. Able to reliably produce surface measurements within an error of 50 μm Schematic: Continued →	[238]

Sensor	Detection	Controller	Findings	Ref
Photodiode-based radiant thermometer	Melt pool infrared temperature	-	Thermal cycle during deposition of thin wall structure recorded. Abnormalities in process (melt pool temperature, cooling rate) can be recognised by IR-temperature signal. Schematic:	[239]
3 contact thermometers 2 Nd:YAG photodiodes Germanium photodiode CCD camera	Contact thermometer: Temperature of dichroitic beam splitter, protective glass, coaxial nozzle Nd:YAG photodiode: Measure laser power outpower, scattered laser light at protective glass Germanium photodiode: IR temperature signal emitted from melt pool CCD camera: monitoring cladding process coaxially		Developed a laser-cladding head with integrated sensors and tested successfully. Operating status of components can be successfully monitored in real time. CCD camera can monitor the melt pool in-situ, useful for distance control Uses integrated Ge-photodiode to realize a process monitoring system. Schematic: Continued →	[240]

Sensor	Detection	Controller	Findings	Ref
3 Photodetector	Deposition height	-	Improvements in microstructure consistence, surface roughness. Schematic:	[241]
Optoelectronic sensor (laser diode, photo diode)	Metal powder delivery rate	-	Sensing signal has good linearity with average powder delivery rate. Schematic:	[242]
-	DC motor (that drives the powder flow rate) angular speed with a Kalman filter	-	Uses a controller capable of tracking general reference powder flow rate trajectories, a measurement-model-based powder flow rate estimator, a disturbance estimator, and a powder flow rate reference trajectory generation scheme. Schematic: Continued →	[243]

Sensor	Detection	Controller	Findings	Ref
CCD camera	Clad height (SMC) Nozzle standoff distance control (PID)	SMC PID	Closed-loop nonplanar laser cladding process was developed. Schematic 	[244]
CCD	Clad height	PID	Feedforward PID controller was developed for the process. Controller designed based on semi-empirical nonlinear model. Relates scanning velocity to clad height. Schematic: 	[245]
CCD	Clad height	-	Developed an adaptive sliding mode control method with uncertainty estimator for the laser cladding process. Elimination of chattering, lumped uncertainty. Schematic: 	[246]

Appendix D : Surface Roughness

The surface finish of LMD produced parts has been known to be a problematic issue. There are numerous factors that come into play when considering the effects of process parameters on the geometrical accuracy or the quality of the surface finish on the part.

Generally, an optimal surface finishing in AM-built parts are achieved by applying thin deposition layers with a highly stable deposition process [247]. Achieving this, however, is not simple. Pinkerton et al. [185] demonstrated that the stability of the melt pool are driven by a number of internal and external factors. The Marangoni flow, shielding gas flow rate, recoil forces at the vaporization point, and the dynamic forces applied by the powder flow rate are all contributing factors. In Pinkerton's study, it was demonstrated that using different types of laser beams, continuous and pulsed, produced different degrees of surface roughness. Adjusting the pulse length: cycle ratio was demonstrated to increase both the mean layer height, as well as the surface roughness, while reducing the microhardness of the deposited material. Furthermore, Mahamood et al. [248] showed that increasing the laser power within a range of 0.8 kW and 3.2 kW resulted in an overall improvement on surface finishing on the LMD produced part. The explanation offered for this observation is that a higher laser power results in a complete melt of the deposited powder, hence producing a smoother surface. At a lower laser power, portions of the deposited powder were observed to be left partially melted, resulting in a rougher surface.

Resch et al. [249] showed that surface roughness can also be improved by controlling the powder flow rate and scanning velocity. By reducing the powder flow rate and increasing the scanning velocity in LMD, a significant improvement in surface roughness was observed. A lower feed rate produces a smaller layer thickness with a smoother surface roughness. At higher feed rates, the remelting of the previously deposited layer in the LMD process is shallow resulting in a higher surface roughness. Mazumder et al. [241] also showed that the same concept is applied on the deposition track level where a larger deposition track overlap would likewise produce a smoother surface roughness. The Marangoni forces that arise from the surface tension gradient at the melt pool

and vapour pressure causes ripples to occur at the melt pool's surface, which, when combined with the rapid solidification rate of the molten metal, the ripples get locked in position creating a large amount of waviness in the surface profile.

Appendix E : Residual Stress and Distortion

Parts produced by DED techniques are widely known to have complex residual stresses that are difficult to control. In most cases, tensile residual stresses in DED parts are undesirable because they reduce the overall effective fatigue and tensile properties. Klingbeil et al. [250] showed that plate-shaped specimens created by DMD techniques exhibited significant warping and that modifying the deposition scan strategy have a significant effect on the residual stress. Two types of scan strategies were used to produce DMD samples, a raster scan strategy, and an inside-out scan strategy. It was demonstrated that the raster scan strategy resulted in significantly amounts of deflections as well as different degrees of uniformity. The measured deflections for two deposition scan strategies are shown in Figure E-1 and Figure E-2.

Dai and Shaw [251] showed similar results, explaining that a raster scan pattern, which has a long scanning direction parallel to one axis, leads to a saddle-shaped distortion, whereas an inside-out scan pattern can bring about the cancellation of concave upward and downward distortions, reducing the out-of-plane distortions significantly.

Figure E-1: Measured deflections for raster scan strategy [250]

Figure E-2: DMD measured deflections for inside-out scan strategy [250]

Rangaswamy et al. [252] investigated the differences in residual stresses for LENS produced thin wall and square pillar structures using neutron diffraction and contour techniques. It was shown that for thin wall structures, compressive stresses along the deposition direction dominated near the deposition initiation point. Beyond this point, tensile stresses along the longitudinal direction would dominate close the surface at the top.

Wang et al. [253] demonstrated that residual stresses are significantly influenced by the laser mode, either pulse or continuous wave laser. Parts produced by continuous wave lasers were observed to have tensile stresses at the edge region and compressive stresses toward the central region. On the other hand, parts produced by pulse laser were observed to have an opposite trend, where compressive stresses were observed toward the edge region and tensile stresses were observed toward

Appendix F : Hybrid Manufacturing Machines

Within the last few years, companies like DMG MORI, Mazak, and Optomec have launched their own versions of hybrid manufacturing machines in attempt to streamline and speed up the manufacturing process chain. The effort to offer a complete end-to-end process chain solution serves as a key incentive and significant advancement toward further industry adoption. Information of some hybrid manufacturing machines is summarised in Table F-1.

Table F-1: Hybrid manufacturing machines

Company	Hybrid Machine	Description	Ref.
DMG MORI	LaserTec 65 3D 	5-axis Work area (machining): 735×650×560 mm Work area (additive): Ø500×400 mm Laser type: Fiber coupled diode laser, 2500 W, 1.6mm/3mm laser spot size Multi-material possible	[254]
DMG MORI	LaserTec 4300 3D 	Work area (machining): 660×400×1500 mm Work area (additive): Ø660×660 mm in Z axial or Ø545×1500 mm in Z radial Laser: Fiber coupled diode laser, 3000 W Multi-material possible	[255]
Mazak	INTEGREX i-400AM 	Work area (Head 1): Ø300×1011 mm Work area (Head 2): Ø250×160 mm Laser: Multi-laser “High Speed” and “Fine” deposition heads available	[256]
Optomec	LENS 3D Hybrid 20 	3-axis Work area (machining): 500×300×500 mm Work area (additive): 350×300×500 mm Fiber laser, 2000 W	[257]