**Influence of Energy Density on Metallurgy and Properties in Metal Additive Manufacturing**

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**Biographical note**

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# Abstract

Fabricating metallic components for highly specialised industries such as automotive and aerospace has become the main focus of Additive Manufacturing due to its many advantages over traditional processes. This review initially outlines current Additive Manufacturing techniques for processing metallic components, particularly on ‘Powder Bed Fusion’ and ‘Directed Energy Deposition’ categories. Various solidification and metallurgical aspects, microstructure and properties of fabricated parts are described in subsequent sections. In addition, the influence of energy density on metallurgy, microstructure and mechanical properties is addressed. The need to establish processing maps for various materials and techniques, and the challenges currently faced in metal Additive Manufacturing are then highlighted. The final section provides an outlook for the future of research in Additive Manufacturing of metals.

**Keywords:** Additive Manufacturing, 3D printing, energy density, metallurgy, mechanical properties

# Introduction

Additive manufacturing (AM), or 3D printing is defined as the production of three-dimensional (3D) objects via material addition process in layer by layer, point by point or line by line manner without deformation or material removal. 1,2 The capability of 3D printing to fabricate parts with complex geometries, and significant time, cost and waste material reduction makes this technology very competitive compared to conventional manufacturing techniques such as milling, welding, casting, forming, forging and turning.

During the development stage in the 1990s, 3D printing was mostly used for rapid prototyping (RP) and rapid tooling (RT) which involved only polymer and thermoplastic materials. However, advancement in laser and electron beam technologies has enabled to fabricate metallic components. As a result, 3D printing has now progressed to the stage of rapid manufacturing (RM) of end-use and functional metallic parts. The immense potential and benefits of 3D printing especially for low volume productions of complex geometries and mass customisation when compared to conventional manufacturing techniques have attracted widespread customers from various commercial sectors, particularly aerospace, automotive and biomedical industries.

A number of 3D printing techniques for processing metals are currently available including Selective Laser Sintering (SLS) commercialised by DTM, Selective Laser Melting (SLM) commercialised by SLM Solutions, Electron Beam Melting (EBM) commercialised by Arcam, molten droplet processes and Wire and Arc Additive Manufacturing (WAAM) which is based on welding technique. The application of these techniques has enabled the fabrication of near-net shape 3D objects. For example, SLM has been used to fabricate functional dental implants and aero engines in a small volume capacity without the need of expensive moulds, showing consistency in terms of mechanical properties. 3-5 This results in high material utilisation rate which implies minimal resource wastage and cost effectiveness.

The operating mechanisms of these processes differ according to the type of machine used, which typically involves one of the three; (i) intense heat source being emitted to the processed materials via laser or electron beam, (ii) metallurgical bonding of feedstock material and (iii) deposition of binder onto the feedstock material. They influence the way at which the raw materials solidify into a complete part i.e. the solidification mechanisms. In addition, such processes also exhibit complex metallurgical mechanisms depending on the processing parameters used (e.g. laser power and scan speed), material characteristics and chemical compositions. These inevitably lead to variations of microstructural and mechanical properties of the completed parts. Hence, the successful fabrication of commercial metallic parts in terms of desired solidification levels and mechanical properties are dependent on the type of machine, processing parameters used and the intended application

It is important to understand the relationship between processing parameters and the solidification and metallurgical mechanisms when processing a particular metallic material via 3D printing. This is because not all practical applications require 100% dense structures, as products such as medical implants, filters, heat exchangers and cooling machines require a certain level of porosity. 6,7 Thus, appropriate machine and AM technique as well as suitable processing parameters need to be selected to fabricate such artefacts. For example, Direct Metal Laser Sintering (DMLS) and SLM are used to fabricate porous objects from 316L stainless steel with controllable microcellular features which enables to achieve the intended porosity level. 8,9

Current research on metal AM processes reveal that various processing parameters have significant influence on the solidification, metallurgy and the resulting microstructures and mechanical properties of AM-fabricated metal parts. In particular, energy density, which is a collective term combining individual processing parameters of heat source power, scan speed, hatch/scan line spacing and layer thickness, has been widely studied. Accordingly, the aim of this paper is to provide an overview of the latest developments in metal AM processing research.

Firstly, 3D printing techniques for metallic components are classified and described based on the ISO/ASTM standard, primarily focusing on powder bed fusion (PBF) and Directed Energy Deposition (DED) categories. Various solidification and metallurgical aspects (mechanisms and defects) are then briefly outlined in the following section, followed by the typical microstructural evolution and mechanical properties of metal AM parts. The subsequent section summarises the influence of energy density parameter on the solidification and metallurgy during AM processing, and the resulting microstructures and mechanical properties of the fabricated parts. Current challenges faced in metal AM processes, especially to achieve 100% densification levels free from porosity and balling are then presented. The final section provides a summary of the review and an outlook on further advanced metal AM processing in the future.

# 3D printing techniques for metallic components

In general, every 3D printing process begins with digital 3D design in Computer Aided Design (CAD) software. This 3D CAD model can either be directly designed in CAD software or scanned to the computer via 3D scanners. 10 The model is then translated into a STL file format which stands for Standard Tesselation Language to be virtually ‘sliced’ before printing. 11 Any geometrical features requiring extra support could be determined automatically or adjusted manually. The model is then printed in a layer-wise manner and further post-processing may be required to enhance their mechanical properties and physical appearances. The generic process sequence of any 3D printing techniques are summarised in Table 1. 12 [Table 1 near here].

AM of metals can either be a single-step process or a multi-step process. 13 Single-step AM process imply that the intended geometry and properties required in the final parts could be acquired directly after complete solidification of individual layers. On the other hand, multi-step AM process means that the required geometry is obtained directly after multi-layer solidification, but further processing is required to improve the mechanical properties of the fabricated parts. However, most metal AM processes are multi-step because post-processing measures such as heat treatment and hot isostatic pressing (HIP) are carried out to improve mechanical properties such as yield and tensile strengths, and to achieve favourable metallurgical mechanisms. 14,15 The as-fabricated parts could also be in a near-net shape form which is close to the intended geometry but requires machining to attain the final geometry and surface finish. 16

There are five AM classifications pertaining to metal AM techniques from the seven described by the ISO/ASTM 52900:2015 standard: 13 (i) Powder Bed Fusion (PBF), (ii) Direct Energy Deposition (DED), (iii) material jetting, (iv) binder jetting, and (v) sheet lamination. Of these five categories, PBF and DED are the main focus of this review as they are more widely studied compared to the other two. The classification of these processes is shown in Fig. 1. (Ref. 13) [Figure 1 near here].

## *Powder Bed Fusion (PBF)*

In PBF processes, a layer of powder bed is selectively melted or sintered by using a focused thermal energy from either laser or electron beam. Both sintering (partially melting) and melting of raw feedstock materials are available for AM of metals, at which Selective Laser Sintering (SLS), Selective Laser Melting (SLM) and Electron Beam Melting (EBM) are common PBF techniques. In PBF systems, a powder bed is formed by spreading metallic powder layers from a powder dispensing mechanism onto a build area, also called substrate, which is typically made of metal materials. 17,18 The heat source then selectively sinters (partial melting) or melts (complete melting) the powder according to initial CAD data. 2,19,20 However, complete melting is now typically used instead of sintering in metal AM processing. 21,22 In addition, the reactive nature of metals also necessitates PBF processing to be conducted in an enclosed chamber, often either vacuum or containing inert gases such as Ar to prevent oxidation or other undesired reactions. 11,23-25

While SLS, SLM and EBM all follow the same powder bed principle to form 3D structures, the machine architecture of EBM is different from SLS and SLM in terms of dissemination of heat source and powder distribution method. EBM machines eject electron beam from an electron beam column which is firstly collimated and then spatially deflected using a filament and magnetic coils, whereas in SLS and SLM, the position of laser is controlled by lenses and scanning mirror or galvanometer. 22 In EBM, powder hoppers and a metal rake are used to distribute powder along the powder bed, while powder hoppers, feeding systems and recoater blades or piston and hopper systems are used in SLS and SLM. 22 In addition, EBM also requires pre-sintering of the powder bed to prevent particle repulsion during the melting process. 26 Nevertheless, PBF processes follow the same operational steps including setting up the machine, processing of raw materials, recovery of unused powders and finally removal of substrates.

On the other hand, the operating mechanism of PBF systems are governed by the processing parameters and scan strategy. Different raw materials usually require different set of processing parameters depending on the PBF technique used and the intended applications. As a result, the microstructure and mechanical properties of completed parts are influenced by these two factors, which will be detailed in later sections.

To date, PBF techniques have been used to process a wide range of materials including pure and pre-alloyed metals as well as metal-matrix composites (MMCs). 27-30 For example, SLS, which sinters (partially melts) the powder material before fusing them together to form a solid part is usually used to process pre-alloyed metals such as W-Cu and CuSnCuCuP. 31,32 In DMLS, SLM and EBM, the powders are completely melted before being fused into a complete part. Since complete melting PBF processes require higher beam power compared to partial melting techniques, high densification levels could be achieved. 33-38 In addition, some applications such as bone implants require certain levels of porosity with high interconnectivity levels. These requirements could be achieved due to the capability of PBF, specifically to fabricate intricate internal features and complex geometrical shapes with controllable porosity levels, e.g. bone implants and medical devices. 16,39,40

## *Directed Energy Deposition (DED)*

DED processes include all AM techniques where feedstock material is deposited onto a molten pool generated via focused energy, either laser or electron beam. The feedstock material may be in powder or wire form. DED is based on the conventional welding technology which enables material deposition without a build chamber by focusing shielded gas flow onto the molten pool. 22

Laser Engineered Net Shaping (LENS), developed by Sandia National Laboratories and commercialised by Optomec is one of the examples of DED-based techniques. In DED processes, a stream of feedstock powder is melted by laser beam, creating a molten pool which eventually solidifies into a solid object. 41-45 Other DED processes using powder as the raw material include Laser Metal Deposition (LMD), Direct Metal Deposition (DMD) and Laser Cladding (LC). In addition, another subset of DED involves feeding wire feedstock into a molten pool. This technology, termed Wire and Arc Additive Manufacturing (WAAM) is essentially an expansion of welding technology. 46-49

DED processes are primarily used to improve wear, impact and corrosion resistance of mechanical components as well as to repair worn-out or damaged high-value components. 50 Powder feed and wire feed DED processes have shown huge potential in building large objects. 51 In DED machines, the substrate is usually staged in either one of these two configurations: 22 (i) 3-axis system which means the substrate is in stationary position, or (ii) 5 or 6-axis system which allows the substrate to rotate while the feedstock is being deposited. The latter configuration enables fabrication of more complex geometries.

Similar to PBF, DED-based AM techniques also follow the same operational steps including setting up the machine, processing of raw materials, recovery of unused powders and finally removal of substrates. In addition, material processing in PBF and DED systems is conducted in an enclosed area, or build chamber to provide safety from the high energy beam and intense heat dissipation. In both cases, the build chamber may or may not be filled with inert gases, e.g. argon to prevent oxidation in the processed metals, depending on the reactivity of the metals. Finally, excess powder is usually vacuumed, or either recovered or disposed depending on the machine set-up. 22

## *Recent developments in 3D printing techniques*

Recent advancement in 3D printing technology has resulted in variations in 3D printing methods, most notably multiple-material additive manufacturing (MMAM) and 3D printing components with controlled porosity.

In the last ten years, efforts have been made to produce a single part from more than one material or different materials. For example, Multiple-Material Stereolithography (MMSLA) has been developed to incorporate multiple resin vats which contain different liquid polymers. 11 However, a scheduling process is required to cater for material change in this process. 52,53 Vaezi et al. 54 has comprehensively reviewed multiple-material additive manufacturing (MMAM) methods which include techniques for 3D printing non-metals derived from SLS, Fused Deposition Modelling (FDM), and Stereolighography (SLA) and also 3D printing metals derived from powder bed and laser cladding processes.

A number of studies on MMAM of metals have been conducted, but they are mostly based on DED processes 55-60 with limited research using PBF processes. 61-63 This is largely due to the better flexibility in control and architecture of DED machines compared to that of PBF. 62 In addition, research on MMAM also focuses on the creation of functionally graded materials (FGM). These FGM materials with alteration of material and microstructural compositions in different sections of a single component corresponds to variation in mechanical properties along that particular component. 64 These materials can be tailored for specialised applications such as biomedical, nuclear and optics. 59,65 One of the recent developments in this area is the introduction of a systematic approach for both linear and radial depositions of gradient metal alloy parts for aerospace applications. 66

Apart from fully dense parts (almost 100% of theoretical density) and near-net shape objects, 3D printing is also used to fabricate porous products for applications mainly in medical and automotive fields. Scaffolds for tissue growth, metallic bone implants and filter components with varying levels of porosity have all been successfully fabricated by various 3D printing methods. 6,8,9,67-69 The flexibility of 3D printing processes in controlling the processing parameters such as laser power and scan speed shows a huge potential in manufacturing porous components.

# Processing parameter variables in Additive Manufacturing of Metals

AM follows the progressive line-layer-bulk build philosophy. 70 This process begins with a single line scanning, in which heat source power, *P,* and scan speed, *v,* becomes the most important parameters. The quality of resulting scan tracks is assessed by a performance matrix which relates both quantities by . Once scanning of multiple lines is completed, a single layer is formed, which introduces another important parameter, hatch/scan line spacing, *h,* defined as the distance between two adjacent scan tracks. Finally, a suitable layer thickness, *d,* needs to be selected so that multiple successive layers could be consolidated to form the bulk component as required on the initial design. Here, all these terms are grouped into a single expression termed energy density, which is utilised to comprehensively evaluate the influence of each process parameters on the overall densification levels, surface roughness and defects such as porosity and balling in the final parts.

In laser based PBF processes such as SLM, this energy density, is represented as: 71

(1)

However, electron beam-based PBF (EBM) and DED-based techniques employ a slightly different energy density relationship. In EBM, the acceleration voltage, *U,* and current, *I* of the electron beam are taken into consideration for the *P* term, thus modifying the energy density, , as: 24,72  (2)

where *s* is the scan line spacing in this case. Meanwhile, the energy density, , in DED-based processes is related by material feed rate (powder or wire), *F*,heat source power, *P*, and cladding speed, *s,*: 41  (3)

where *s* is similar to scan speed in PBF processes. Nevertheless, in all cases, the unit used for energy density is typically either J mm-3 or J m-1.

In PBF and DED metal AM processes, homogeneous melting of each layer is usually desired to enable complete melting of feedstock materials. 73 However, the complex heat and mass transfer that occurs throughout processing could result in various defects such as balling, crack, delamination and porosity in the completed build structure when non-optimum processing parameters are used. Hence, the scan strategy used during processing is also another process variable, apart from energy density that affects the melting of feedstock materials and thus the resulting defects in the final part. Various scan strategies have been employed, and their influence on metallurgy and mechanical properties in 3D printed metallic structures have been well-described and reviewed. 22,73 Overall, these parameters are found to have significant influence on the solidification, metallurgy and the microstructural evolution in 3D printed metallic parts.

# Solidification and metallurgy in 3D printing metallic components

All metal AM processes involve the consolidation of raw feedstock material into solid 3D structure, either fully dense or with certain level of porosity depending on the desired applications. The solidification is achieved via sintering or complete melting of feedstock materials and then joining of the solidified layers. 13 Processing parameters and scan strategy play major roles in influencing solidification levels and the type of metallurgical mechanisms. These result in variation of mechanical properties for the fabricated parts as can be seen in the subsequent sections.

## *Solidification*

In processing metals, the sintering solidification mechanism is exclusively used in PBF processes such as SLS, while complete melting solidification mechanism is common to PBF and DED processes. There are three main sintering mechanisms, which are derived from powder metallurgy and extended to metal PBF AM processes: (i) solid state sintering (SSS), 74 (ii) liquid phase sintering (LPS), 75,76 and (iii) supersolidus liquid phase sintering (SLPS). 70,77 In SSS, solidification occurs via diffusion of sintering necks formed after heating near the melting point of materials. However, this is a slow process and unsuitable for metal AM processes that are characterised by the short interaction time between heat source and feedstock materials. On the other hand, LPS and SLPS enable rapid solidification of the sintered powder particles in SLS process. This because the presence of liquid phase flow driven by capillary forces promote faster particle rearrangement and solution re-precipitation, which are the main stages of LPS and SLPS. 21 Hence LPS and SLPS are more accurate descriptions of the sintering mechanisms for metallic parts fabricated by PBF processes, particularly SLS technique.

Overall, solidification of PBF and DED-processed materials is governed by the geometry of the molten pool generated, controlled mainly by the processing parameters. Apart from energy density, the *P*-*v* relationship is also an important factor which is often considered in order to achieve desired densification levels. In this relationship, values for *P* and *v* are the only adjustable parameters while others (*h* and *d*) are kept constant in order to determine densification levels in AM-built parts. 78-81 Densification levels are typically assessed by measuring the actual versus theoretical density and analysing porosity levels in the final parts. By adjusting *P,* *v*, and other processing parameters mentioned previously, metal AM processes are not only able to fabricate parts with near 100% theoretical density, but also lightweight parts, lattice and open-cell structures, as well as components with pre-designed porosity for specialised applications. 82-84

In sintering processes, e.g. SLS, the metal powders processed can be of pure, pre-alloyed, or some combinations between pure and pre-alloyed metals. However, early works on the SLS of pure metals such as Ni, Cu, Pb, Sn and Zn revealed that high densification levels are difficult to be achieved. 85-89 These studies showed that the limited liquid phase present when processing pure metals causes high viscosity and surface tension that leads to high porosity levels and the balling phenomenon. 90 On the other hand, high densification levels could be achieved when sintering pre-alloyed or pure-pre-alloyed metal combinations. 91-93 The influence of processing parameters, especially energy density on the densification levels of metal parts processed by partial melting AM have been investigated extensively. 94-97

For example, Olakanmi et al. 98 reported that 67 J mm-3 is the optimum energy density to obtain high densification and low porosity levels when processing Al-12Si by SLS. However, reducing this value results in insufficient liquid phase to promote sintering and enlarged pore sizes, while increasing this value above the optimum level causes balling to occur. Ghosh et al. 99 used the Taguchi method and determined layer thickness and hatch distance (scan line spacing) as important factors in preparing Al203, TiC and TiB2 composites by SLS. High densification levels with low porosity content were obtained when the layer thickness and hatch distance were reduced.

A different study by Manfredi et al. 96 revealed that in the SLS process of AlSiMg with 10 wt% SiC particles, high density levels (~ 95%) and low porosity content (~ 5%) could be obtained by using laser powers of 180 and 195 W and varying scan speed from 500 to 900 mm s-1. On the other hand, an experiment on indirect SLS process of porous 316L stainless steel samples observed that increase in laser power (10 W to 35 W) and decrease in scan line spacing (0.20 mm to 0.10 mm) result in the reduction of porosity levels from 73% to 55%. 100

Similar experiments were carried out to determine the influence of energy density on metal AM processes undergoing complete melting mechanism. For example, the analysis of variance (ANOVA) procedure revealed that porosity in SLM-processed ALSi10Mg parts is largely influenced by laser power, scan speed and scan line spacing. 78 Within the parameters studied (laser power: 100-200 W, scan speed: 700-2000 mms-1, laser track width: 150 µm), high scan speeds and low laser powers reduce the energy density input, leading to incomplete solidification and increased porosity levels.

In addition, a study on SLM of commercially pure Ti (CP-Ti) by Attar et al. 34 revealed two important observations. Firstly, although high densification levels (98.5-99%) were achieved when the energy density used was in the range of 95 J mm-1 and 125 mm-1, certain combinations of individual laser power and scan speed which correspond to this energy density range could only result in attained density levels <98%. This is caused by molten pool instabilities due to balling and powder overheating. Secondly, the maximum densification level could not exceed 99% even though the laser power is increased beyond 180 W. At this point, all the powders along the laser path have been completely melted, hence further increase in laser power could deter the quality of fabricated parts, following the occurrence of balling at very high laser powers.

## *Metallurgical mechanisms and metallurgical defects*

A number of important metallurgical phenomena occur in metal AM processes including molten pool behaviour, 101-103 Marangoni convection, 104-106 unmelt formation, 26,56 occurrence of spatter, 26 and formation of denudation zones. 107 The study on these phenomena, particularly spatter and denudation zones, has received significant interest in recent years because they are found to have important consequence on the defects, e.g. porosity and balling and thus the mechanical properties of the final parts. 108 The occurrence of these phenomena largely contribute to the microstructural evolution of the 3D printed metallic components. Nevertheless, these phenomena are largely determined by the thermal history experienced during processing, which are governed by the energy density as well as the chemistry of the feedstock materials. 22

### *Metallurgical mechanisms*

In PBF and DED processes, unmelts are the areas of incomplete melting of the raw materials. One of the reasons of unmelts formation in PBF processes is due to non-uniform powder spreading on the powder bed at the beginning of the process. 26 Material compositions have also been observed to cause the formation of unmelts in DED-processed parts. For example, too much vanadium content in the laser deposition of Ti6Al4V and 316L stainless steel gradient parts contributes to excess unmelts formation due to insufficient molten pool temperature, which is indicated by inhomogeneous powder mixing and grain morphology. 56 The presence of unmelts, together with other defects such as porosity are found to reduce the mechanical properties of the fabricated metal parts. 109-112 Post-processing measures e.g. heat treatment and Hot Isostatic Pressing (HIP) as well as optimisation of processing parameters are among the efforts that could be taken to minimise or remove the formation of unmelts. 113

Spatter ejection occurs due to the overheating of molten pool as a result of the high energy beam-feedstock material interaction during metal AM processing, which creates recoil pressure in the molten pool due to convective transport of fluid (Marangoni Convection). 22,107,114 The recoil pressure results in the spatter of metal droplets upwards into the surrounding in PBF and DED processes, 115-118 and an additional sideways spatter of raw metal powders on the powder bed in PBF processes. 119 Spatter ejection is found to contribute to the formation of oxide layers and also porosity, which could be detrimental to the mechanical properties of the fabricated samples. 117 However, some studies have proposed measures to reduce spatter and hence improve part quality in PBF processes. For example, Hopkinson et al. 120 initially sintered the powder bed with low energy density before re-melting each layer using high energy density (high *P*, low *v*).Mumtaz et al. 121 utilised pulsed laser instead of continuous wave emissions to improve energy density distribution and achieve better melting of feedstock materials. However, the influence of energy density on spatter in DED processes remain unclear.

On the other hand, denudation zone (DZ) is a region of displaced powders driven by the metal vapour ejection (flux) and entrainment of powder particles in a gas flow due to the high energy beam-material interaction, which clears the powders along the path of the laser towards the sides of the scan track (Fig. 2). 122 The earlier studies demonstrated that the formation of DZ only occurs in PBF processes 4,108,123,124, and only recently that researchers have found that the gaps on the sides of solidified tracks may contribute to porosity in the finished parts. This is because the available amount of powders along the laser scan route are reduced during subsequent scans. 107,122,125 [Figure 2 near here].

Up to now, only a limited research has been conducted to determine the influence of processing parameters on the formation of DZ. For example, Yadroitsev et al. 126 found that DZ reduces the height of scan tracks after each layer scan in SLM of stainless steel, but it could be controlled by varying scan line spacing, *h*. On the other hand, Cooper 127 observed increase in the height of scan tracks instead when processing Ni superalloys and MMCs. This is the result of the entrapped powder particles in the gas flow being absorbed into the molten pool, thus increasing height of the scan tracks instead of reducing it. These conflicting observations may be caused by a number of factors including energy density, gas pressure in the build chamber as well as feedstock material properties. 122,128 Further investigations need to be carried out to clarify the influence of various processing parameters on the denudation phenomenon, and hence on the quality of the fabricated metal parts.

### *Metallurgical defects*

Porosity and balling are two major metallurgical defects which occur due to non-optimum energy density values in metallic AM parts. These defects significantly reduce the resultant mechanical properties such as yield and tensile strengths, and microhardness required for various applications. Porosity in the final parts is either gas or porosity-induced as shown in Fig. 3. 129 Spherical gas-induced porosity is caused by internal porosity resulting from gas entrapment during metal powder production. On the other hand, non-optimum processing parameters used in AM contribute to process-induced porosity that is often characterised by irregular-shaped pores. Gas-induced pores are usually microscopic while the size of process-induced pores range from sub-microns to macroscopic. Process-induced porosity is divided into lack-of-fusion 130,131 and shrinkage porosities. 132,133 Insufficient power (Low *P*) applied results in unmelted material layers causing lack-of-fusion regions in the completed parts. On the other hand, spatter ejection caused by excessive heat power (very high *P*) contributes to shrinkage porosity as the evaporated materials leave regions of molten pool with incomplete material solidification causing formation of pores in the solidified structure. [Figure 3 near here].

Balling or liquid spheroidisation occurs as the result of limited liquid phase formation in the molten pool to initiate powder melting when the temperature becomes too high as the power of heat source increases. It is characterised by the development of rough, spherical structures on the surface of the 3D printed parts. 134 The occurrence of balling is governed by various metallurgical mechanisms including molten pool behaviour, Marangoni convection and wettability. 86,103,135 Altogether, these mechanisms are greatly dependent on the energy density and the characteristics of materials used during the 3D printing processes. Balling is an unfavourable phenomenon because it severely degrades the quality of the fabricated components. For example, the formation of discontinuous scan tracks during multiple-line scanning can result in poor inter-line and poor inter-layer bonding respectively. 136 These subsequently causes part delamination and porosity which lead to poor surface finish, low solidification levels, thus low attained density and low part strength. 76

To address the porosity and balling defects in metal AM processes, recent research efforts have focused on adjusting individual parameters to obtain optimum energy density values that could minimise the levels of both defects in the fabricated parts. Various studies on PBF and DED processes of Ti and Ti alloys, 34,104,137 Al and Al alloys, 138-140 Cu and Cu alloys, 25,141 stainless steel, 142-144 and FGM materials 57,64 all reveal that high energy density levels (high *P*,and low *v*, *h*, and *d*) are able to produce metallic parts with minimum balling and low porosity levels (< 1%).

## *Microstructural evolution and mechanical properties*

Mechanical properties of 3D printed metallic components are highly dependent on the microstructure obtained upon solidification. In general, the high energy-material interaction cycle typically associated with 3D printing of metals causes rapid heating and melting of the materials. This eventually leads to rapid solidification upon cooling. Generally, the high heating/cooling rates in the range of 103-108 K/s contribute to the formation of non-equilibrium phases. 102,145,146 Gu et al. 70 outlined four characteristics for 3D printed metallic components based on these observations:

Since there is insufficient time for grain development and grain growth in 3D printed materials, grain refinement is expected to occur.

The occurrence of Marangoni convection in the molten pool due to surface tension and temperature gradients causes the rapid solidification in materials processed by 3D printing to become a non-steady state process.

3D printed metals usually have anisotropic behaviour due to kinetic limitation of crystal growth that occurs due to the rapid solidification.

The variation of microstructures along the build direction (z-axis) is attributed to the change in heat flow conditions i.e. different thermal histories between different layers.

Additionally, different microstructures are obtained depending on the processing parameters and the properties of the materials used. For example, a study on the SLM of Inconel 718 (IN 718) showed that the columnar dendrites formed during this process evolve from a coarsened and cluttered structure at low energy density values to typical slender structure undergoing directional solidification at high energy density values. 147

Another study by Zhang et al. 148 observed the evolution of dendrite grains to equiaxed grains under high temperature gradients in processing Mg-9%Al by SLM. The equiaxed grains were further refined as scan speed, *v,* increases (Fig. 4). In addition, Gu et al. 91 found that decreasing scan line spacing, *h,* during the DMLS process of W-Cu could close the interconnected gaps caused by narrow columnar agglomerates and produce a completely dense structure with smooth surface (Fig. 5). These improvements are attributed to increased overlaps between scan lines at lower *h* values. [Figures 4 and 5 near here].

Furthermore, it is also established that the resulting mechanical properties of metal AM parts depend on the microstructures obtained after processing. Fine microstructures are typically obtained in metal AM parts as a result of rapid heating/cooling cycle, which often result in higher yield and tensile strengths 149-151 as well as higher microhardness (HV) values 142,152,153 compared to conventionally manufactured parts. The tensile and yield strengths of some metallic components fabricated by SLM in comparison with those produced by conventional manufacturing methods under optimum processing parameters are displayed in Table 2. [Table 2 near here].

However, the layer-wise build manner of AM processes also contributes to anisotropy in metal AM parts. 82,154-156 For example, various studies have shown that the ultimate tensile strength (UTS) values are lower in the build direction (z-axis) compared to that in the scan direction (x- or y-axis) due to the microstructural variation obtained at different orientations of the solidified parts. 153,157,158 Similarly, the differences in HV values at the horizontal (x-y plane) and vertical (x-z or y-z planes) surfaces of 3D printed parts are also attributed to the microstructural variation associated with anisotropy. 159-161  Fig. 6 shows an example of different HV values in SLM-fabricated NiCr parts at different scan speeds in different orientations due to anisotropy. 159 [Figure 6 near here].

Apart from anisotropy, the lower UTS values obtained in the build direction are also caused by a large amount of pore formation and high residual stress concentration in that orientation. 9,94,95,155,162 Post-processing such as heat treatment could improve the UTS values in these cases, but they are generally undesirable in SLM and EBM processes to avoid dimensional shrinkage in the final parts. 163 Apart from that, scan speed, *s,* is also found to influence the porosity and tensile strength of metal AM parts (Fig. 7). 81 Therefore, proper understanding in optimising processing parameters to control the microstructural evolution in metal AM processes is of utmost important so that high tensile and yield strength values could be achieved while maintaining dimensional integrity, especially for industrial applications. [Figure 7 near here].

# Processing maps for metal AM processes

Even though high energy density values are often desirable to attain favourable solidification and metallurgy as well as the desired microstructure and mechanical properties in the final parts, it is difficult to obtain a general set of optimum individual process parameters for all metal AM processes i.e. no ‘one size fits all’. This is because the desired metallurgy and mechanical properties for metal AM parts are dependent on the following factors: type of machine, and the type and chemistry of the materials being processed, as well as the desired application.

Hence, various experiments have been conducted to determine the set of optimum processing parameters (energy density and scan strategy) according to the intended end-application. Various methods are also used to assess the attained metallurgical mechanisms in these cases, which include classifications based on scan track formation, size and stability of molten pool, and microstructures obtained. Processing maps are then constructed based on the applied processing parameters for each feedstock materials to establish the relationship between energy density and the metallurgical mechanisms attained.

The establishment of different processing maps enables suitable processing parameters to be selected and favourable metallurgical mechanisms to be obtained according to the type of AM system used and the applications desired. An example of processing map showing optimum and non-optimum processing zones in DMLS-processed 316L stainless steel powder parts is shown in Fig. 8. 144 [Figure 8 near here].

Some examples of materials processed by PBF and DED techniques with the corresponding optimum processing parameters are listed in Table 3, which also includes the characteristics of favourable metallurgical mechanisms and the resulting mechanical properties of the 3D printed samples. [Table 3 near here].

# Challenges in AM of metallic components

While the presence of porosity may be advantageous for certain applications, achieving 100% densification level with structures free from porosity and balling is the major stumbling block in AM of metals at present. Throughout the studies reviewed, the maximum densification level that could be achieved in the fabricated part is 99%. Even though this might be adequate to provide the mechanical strength required under service conditions or for structural applications, improvements could always be made to further increase the properties and extend the service life of these metallic components. In addition, the occurrence of porosity and balling could also compromise the dimensional and geometrical integrity and the smoothness of the surface of metal AM parts. These limitations largely arise due to the complexity in controlling the non-equilibrium physical and metallurgical phenomena occurring during metal AM processing.

A significant amount of experimental work and simulation analysis have been explored, but it is often difficult to incorporate all possible variables within a single study. Hence, most studies usually focus on one or two variables and the outcome of these studies could provide some clues for how to control these physical and metallurgical phenomena, and thus improve the mechanical properties to fit the requirements of desired applications. Nevertheless, most studies have focused on optimising energy density to attain favourable metallurgical mechanisms in order to reduce porosity and balling as low as possible. These studies reveal that the stability of molten pool have a pronounced impact on the occurrence of balling and formation of pores. The stability of this molten pool is often affected by various metallurgical phenomena, which is governed primarily by the energy density.

One step that could be taken to address these issues is by establishing more processing maps, as one example shown in Fig. 8. Thus, it is a rational move to tabulate processing parameters with corresponding range of energy density values that results in the fabricated parts having minimum levels, or are free from porosity and balling However, the information listed by the authors of this review in Table 3 is only limited to a few materials because not many material combinations have been investigated yet to date. Nevertheless, the information in Table 3 would be useful for researchers and manufacturers to determine suitable processing parameters and produce high quality 3D printed metallic components.

# Summary and outlook

A variety of 3D printing techniques for processing metallic components have been developed to respond to the demands of highly specialised industries e.g. aerospace, automotive, biomedical for low production volumes of purpose-designed metallic components. Currently, only PBF and DED techniques are well-established while others 3D printing processes need to be further explored for commercial and industrial applications. The key to successful fabrication of 3D printing metallic components relies on fully understanding the underlying principles throughout the process.

This review presents the mechanism of PBF and DED metal AM categories, with particular focus on the influence of energy density parameter on the solidification and metallurgy during processing, as well as the resulting microstructures and mechanical properties of metal AM parts. PBF and DED processes undergo similar solidification mechanism, which is the complete melting of feedstock materials and then joining of the solidified material layers to form a complete structure. However, AM processes are subjected to high energy beam-material interaction which results in a number of complex physical and metallurgical phenomena during processing.

Understanding the mechanisms and factors which influence the complex metallurgical phenomena is of critical importance because they play a key role to ensure the characteristics and properties required for specific applications. Of the metallurgical mechanisms described, spatter ejection and denudation zone phenomena have become the current research focus in recent years. This is because even though the basic principles of these phenomena are known, it is only recently discovered that these two mechanisms are correlated with the occurrence of porosity and balling, which are the key metallurgical defects that weakens the mechanical properties of metal AM parts. Furthermore, their effects on part properties are not yet clearly understood.

In addition, it is well-established that the mechanical properties of metal AM parts are dependent on the microstructures obtained and are typically better than that of conventionally manufactured metallic parts. Metal AM processes are characterised by rapid heating/cooling cycle, which results in fine microstructures that improves yield and tensile strengths, and microhardness of the fabricated parts. However, microstructures of AM parts are also subjected to anisotropy due to the layer-wise build philosophy of AM processes, which usually results in slightly different properties in different orientations of the built parts.

Nevertheless, a large amount of literature reveal that the solidification process and various metallurgical phenomena are governed by the processing parameters used during processing, particularly energy density (heat source power, scan speed, scan line spacing and layer thickness) and to a lesser extent, the scan strategy employed. Altogether, these parameters are often optimised to: (i) attain favourable metallurgical mechanisms which could minimise porosity and balling, and (ii) control the microstructures in order to meet the characteristics and properties of various applications as required. In general, high energy density values are desirable as they often favour high densification values, good metallurgical characteristics, and better mechanical properties.

However, the difficulty to find a set of optimum individual processing parameters and hence the energy density which could be generalised for all metal AM processes have resulted in the establishment of processing maps for various materials processed by different metal AM techniques. The importance of establishing such processing maps would enable the visualisation of the borders between optimum and non-optimum processing conditions for producing high quality components, as shown in Table 3. In addition, regions at which unfavourable metallurgical phenomena e.g. porosity, balling and unstable molten pool are also able to be determined, making the avoidance of these disadvantageous phenomena possible.

Although internationally accepted standards, e.g. ISO/ASTM 52921:2013, can provide guidelines for test methodologies of AM parts, there is still no proper, centralised database which compiles the individual processing maps for all current materials that have been processed. Such standards, once validated, could be established and made available for scientists and manufacturers to facilitate a steady and accurate process control of 3D printing machines for research and industrial applications.

In the near future, research on metal AM will undoubtedly experience exponential growth and thus a variety of new materials will be developed and manufactured, e.g. new FGM material combinations, due to the development of advanced AM processing techniques, such as multiple-material metal AM. Hence, the establishment of the inter-dependence between processing parameters particularly energy density, solidification and metallurgy, and the resulting microstructures and mechanical properties in current metal AM research is extremely important as it becomes the basis for further advanced metal AM processing techniques which should be the focus of research in the future.

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# References

1. N. A. Waterman and P. Dickens: 'Rapid product development in the USA, Europe and Japan', *World Cl. Des. to Manuf.*, 1994, **1**, 27–36.

2. W. E. Frazier: 'Metal Additive Manufacturing: A Review', *J. Mater. Eng. Perform.*, 2014, **23**, 1917–1928.

3. S. Bremen, W. Meiners, and A. Diatlov: 'Selective Laser Melting. A manufacturing technology for the future?', *Laser Tech. J.*, 2012, **9**, 33–38.

4. L. Thijs, F. Verhaeghe, T. Craeghs, J. Van Humbeeck, and J.P. Kruth: 'A study of the microstructural evolution during selective laser melting of Ti–6Al–4V', *Acta Mater.*, 2010, **58**, 3303–3312.

5. B. Vandenbroucke and J.P. Kruth: 'Selective laser melting of biocompatible metals for rapid manufacturing of medical parts', *Rapid Prototyp. J.*, 2007, **13**, 196–203.

6. G. Gagg, E. Ghassemieh, and F. E. Wiria: 'Effects of sintering temperature on morphology and mechanical characteristics of 3D printed porous titanium used as dental implant', *Mater. Sci. Eng. C. Mater. Biol. Appl.*, 2013, **33**, 3858–3864.

7. J. Banhart: 'Manufacture, characterisation and application of cellular metals and metal foams', *Prog. Mater. Sci.*, 2001, **46**, 559–632.

8. Y. F. Shen, D. D. Gu, and P. Wu: 'Development of porous 316L stainless steel with controllable microcellular features using selective laser melting', *Mater. Sci. Technol.*, 2008, **24**, 1501–1505.

9. D. Gu and Y. Shen: 'Processing conditions and microstructural features of porous 316L stainless steel components by DMLS', *Appl. Surf. Sci.*, 2008, **255**, 1880–1887.

10. F. P. W. Melchels, J. Feijen, and D. W. Grijpma: 'A review on stereolithography and its applications in biomedical engineering', *Biomaterials*, 2010, **31**, 6121–6130.

11. K. V. Wong and A. Hernandez: 'A Review of Additive Manufacturing', *ISRN Mech. Eng.*, 2012, **2012**, 1–10.

12. N. Hopkinson: 'Additive Manufacturing : What’s happening and where are we going with printing in the third dimension?', *Becta*, 2010, 1–22.

13. 'Additive manufacturing - General principles - Terminology', 52900:2015(E), ASTM, Philadelphia, PA, USA, 2015.

14. C. Qiu, N. J. E. Adkins, and M. M. Attallah: 'Microstructure and tensile properties of selectively laser-melted and of HIPed laser-melted Ti-6Al-4V', *Mater. Sci. Eng. A*, 2013, **578**, 230–239.

15. N. Hrabe, T. Gnäupel-Herold, and T. Quinn: 'Fatigue properties of a titanium alloy (Ti-6Al-4V) fabricated via electron beam melting (EBM): Effects of internal defects and residual stress', *Int. J. Fatigue*, 2017, **94**, 202-210.

16. D. Manfredi, F. Calignano, M. Krishnan, R. Canali, E. P. Ambrosio, S. Biamino, D. Ugues, M. Pavese, and P. Fino: 'Additive Manufacturing of Al Alloys and Aluminium Matrix Composites ( AMCs )', in 'Light Metal Alloys Applications', (ed. W. A. Monteiro), 3–34; 2014, InTech Open.

17. I. Yadroitsev, P. Krakhmalev, I. Yadroitsava, S. Johansson, and I. Smurov: 'Energy input effect on morphology and microstructure of selective laser melting single track from metallic powder', *J. Mater. Process. Technol.*, 2013, **213**, 606-613.

18. I. Yadroitsev and I. Yadroitsava: 'Evaluation of residual stress in stainless steel 316L and Ti6Al4V samples produced by selective laser melting', *Virtual Phys. Prototyp.*, 2015, **10**, 67–76.

19. R. Bogue: '3D printing: the dawn of a new era in manufacturing?', *Assem. Autom.*, 2013, **33**, 307–311.

20. N. Guo and M. C. Leu: 'Additive manufacturing: technology, applications and research needs', *Front. Mech. Eng.*, 2013, **8**, 215–243.

21. J. P. Kruth, X. Wang, and T. Laoui: 'Lasers and materials in selective laser sintering', *Assem. Autom.*, 2003, **23**, 357–371.

22. W. J. Sames, F. A. List, S. Pannala, R. R. Dehoff, and S. S. Babu: 'The metallurgy and processing science of metal additive manufacturing', *Int. Mater. Rev.*, 2016, **61**, 1–46..

23. W. Cao and Y. Miyamoto: 'Freeform fabrication of aluminum parts by direct deposition of molten aluminum', *J. Mater. Process. Technol.*, 2006, **173**, 209–212.

24. C. Guo, W. Ge, and F. Lin: 'Effects of scanning parameters on material deposition during Electron Beam Selective Melting of Ti-6Al-4V powder', *J. Mater. Process. Technol.*, 2014, **217**, 148–157.

25. M. A. Lodes, R. Guschlbauer, and C. Körner: 'Process development for the manufacturing of 99.94% pure copper via selective electron beam melting', *Mater. Lett.*, 2014, **143**, 298–301.

26. R. J. Hebert: 'Viewpoint: metallurgical aspects of powder bed metal additive manufacturing', *J. Mater. Sci.*, 2016, **51**, 1165–1175.

27. E. Santos, K. Osakada, M. Shiomi, M. Morita, and F. Abe: 'Fabrication of Titanium Dental Implants by Selective Laser Melting', Proc. SPIE on 'Fifth International Symposium on Laser Precision Microfabrication', October 2004, Vol. 5662, 268–273.

28. E. C. Santos, K. Osakada, M. Shiomi, Y. Kitamura, and F. Abe: 'Microstructure and mechanical properties of pure titanium models fabricated by selective laser melting', *Proc. Inst. Mech. Eng. Part C J. Mech. Eng. Sci.*, 2004, **218**, 711–719.

29. S. Kumar and S. Pityana: 'Laser-Based Additive Manufacturing of Metals', *Adv. Mater. Res.*, 2011, **227**, 92–95.

30. J. J. Beaman, T. L. Bergman, C. Atwood, S. Hollister, A. Arbor, D. Rosen, and T. G. W. Woodruff: 'Additive/Subtractive Manufacturing Research and Development in Europe', WTEC panel, World Technology Evaluation Centre (WTEC), Boston, USA, 2004.

31. D. Gu, Y. Shen, S. Fang, and J. Xiao: 'Metallurgical mechanisms in direct laser sintering of Cu–CuSn–CuP mixed powder', *J. Alloys Compd.*, Jul. 2007, **438**, (1-2), 184–189

32. D. Gu and Y. Shen: 'Effects of dispersion technique and component ratio on densification and microstructure of multi-component Cu-based metal powder in direct laser sintering', *J. Mater. Process. Technol.*, Feb. 2007, **182**, (1-3), 564–573.

33. L. E. Murr, S. M. Gaytan, D. A. Ramirez, E. Martinez, J. Hernandez, K. N. Amato, P. W. Shindo, F. R. Medina, and R. B. Wicker: 'Metal Fabrication by Additive Manufacturing Using Laser and Electron Beam Melting Technologies', *J. Mater. Sci. Technol.*, Jan. 2012, **28**, (1), 1–14.

34. H. Attar, M. Calin, L. C. Zhang, S. Scudino, and J. Eckert: 'Manufacture by selective laser melting and mechanical behavior of commercially pure titanium', *Mater. Sci. Eng. A*, 2014, **593**, 170-177.

35. B. Vrancken, L. Thijs, J.P. Kruth, and J. Van Humbeeck: 'Heat treatment of Ti6Al4V produced by Selective Laser Melting: Microstructure and mechanical properties', *J. Alloys Compd.*, 2012, **541**, 177–185.

36. T. Vilaro, C. Colin, and J. D. Bartout: 'As-Fabricated and Heat-Treated Microstructures of the Ti-6Al-4V Alloy Processed by Selective Laser Melting', *Metall. Mater. Trans. A*, 2011, **42**, 3190–3199. *Metall. Mater. Trans. A*, vol. 42, no. 10, pp. 3190–3199, May 2011.

37. K. Osakada and M. Shiomi: 'Flexible manufacturing of metallic products by selective laser melting of powder', *Int. J. Mach. Tools Manuf.*, 2006, **46**, (11), 1188–1193.

38. E. C. Santos, M. Shiomi, K. Osakada, and T. Laoui: 'Rapid manufacturing of metal components by laser forming', *Int. J. Mach. Tools Manuf.*, 2006, **46**, 1459–1468.

39. A. Simchi, F. Petzoldt, and H. Pohl: 'On the development of direct metal laser sintering for rapid tooling', *J. Mater. Process. Technol.*, 2003, **141**, 319–328.

40. N. Harrison, J. R. Field, F. Quondamatteo, W. Curtin, P. E. McHugh, and P. McDonnell: 'Preclinical trial of a novel surface architecture for improved primary fixation of cementless orthopaedic implants', *Clin. Biomech.*, 2014, **29**, 861–868.

41. D. Novichenko, L. Thivillon, P. Bertrand, and I. Smurov: 'Carbide-reinforced metal matrix composite by direct metal deposition', *Phys. Procedia*, 2010, **5**, 369–377.

42. V. K. Balla, P. D. DeVasConCellos, W. Xue, S. Bose, and A. Bandyopadhyay: 'Fabrication of compositionally and structurally graded Ti-TiO2 structures using laser engineered net shaping (LENS)', *Acta Biomater.*, 2009, **5**, 1831–1837.

43. K. Zhang, W. Liu, and X. Shang: 'Research on the processing experiments of laser metal deposition shaping', *Opt. Laser Technol.*, 2007, **39**, 549–557.

44. L. Costa and R. Vilar: 'Laser powder deposition', *Rapid Prototyp. J.*, 2009, **15**, 264–279.

45. A. Gasser, G. Backes, I. Kelbassa, A. Weisheit, and K. Wissenbach: 'Laser Additive Manufacturing Laser Metal Deposition ( LMD ) and Selective Laser Melting (SLM) in Turbo-Engine applications', *Laser Tech. J.*, 2010, **7**, 58–63.

46. P. Kazanas, P. Deherkar, P. Almeida, H. Lockett, and S. Williams: 'Fabrication of geometrical features using wire and arc additive manufacture', *Proc. Inst. Mech. Eng. Part B J. Eng. Manuf.*, 2012, **226**, 1042–1051.

47. R. Martukanitz and J. Hollingsworth: 'Taking the Next Step in Additive Manufacturing', *Weld. J.*, 2014, **93**, 40..

48. F. Wang, S. Williams, P. Colegrove, and A. Antonysamy: 'Microstructure and Mechanical Properties of Wire and Arc Additive Manufactured Ti-6Al-4V', *Metall. Mater. Trans. A*, 2013, **44**, 968–977.

49. A. Uriondo, M. Esperon-Miguez, and S. Perinpanayagam: 'The present and future of additive manufacturing in the aerospace sector: A review of important aspects', *Proc. Inst. Mech. Eng. Part G J. Aerosp. Eng.*, 2015, **229**, 2132-2147.

50. P. L. Blackwell: 'The mechanical and microstructural characteristics of laser-deposited IN718', *J. Mater. Process. Technol.*, Dec. 2005, **170**, (1-2), 240–246.

51. S. W. Williams, F. Martina, A. C. Addison, J. Ding, G. Pardal, and P. Colegrove: 'Wire + arc additive manufacturing', *Mater. Sci. Technol.*, 2016, **32**, 641–647.

52. M. Szilvśi-Nagy and G. Mátyási: 'Analysis of STL files', *Math. Comput. Model.*, 2003, **38**, 945–960.

53. J. W. Choi, H. C. Kim, and R. Wicker: 'Multi-material stereolithography', *J. Mater. Process. Technol.*, 2011, **211**, 318–328.

54. M. Vaezi, S. Chianrabutra, B. Mellor, and S. Yang: 'Multiple material additive manufacturing – Part 1: a review', *Virtual Phys. Prototyp.*, Mar. 2013, **8**, (1), 19–50..

55. B. E. Carroll, R. A. Otis, J. P. Borgonia, J. Suh, R. P. Dillon, A. A. Shapiro, D. C. Hofmann, Z. K. Liu, and A. M. Beese: 'Functionally graded material of 304L stainless steel and inconel 625 fabricated by directed energy deposition: Characterization and thermodynamic modeling', *Acta Mater.*, 2016, **108**,46–54.

56. A. Reichardt, R. P. Dillon, J. P. Borgonia, A. A. Shapiro, B. W. McEnerney, T. Momose, and P. Hosemann: 'Development and characterization of Ti-6Al-4V to 304L stainless steel gradient components fabricated with laser deposition additive manufacturing', *Mater. Des.*, 2016, **104**, 404–413.

57. T. E. Abioye, P. K. Farayibi, P. Kinnel, and A. T. Clare: 'Functionally graded Ni-Ti microstructures synthesised in process by direct laser metal deposition', *Int. J. Adv. Manuf. Technol.*, 2015, **79**, 843–850.

58. R. M. Mahamood and E. T. Akinlabi: 'Laser metal deposition of functionally graded Ti6Al4V/TiC', *Mater. Des.*, 2015, **84**, 402–410.

59. A. Bandyopadhyay, B. V Krishna, W. Xue, and S. Bose: 'Application of laser engineered net shaping (LENS) to manufacture porous and functionally graded structures for load bearing implants', *J. Mater. Sci. Mater. Med.*, 2009, **20**, S29–S34.

60. B. V. Krishna, W. Xue, S. Bose, and A. Bandyopadhyay: 'Functionally graded Co-Cr-Mo coating on Ti-6Al-4V alloy structures.', *Acta Biomater.*, 2008, **4**, 697–706.

61. A. Hinojos*,* J. Mireles, A. Reichardt, P. Frigola, P. Hosemann, L. E. Murr, R. B. Wicker: 'Joining of Inconel 718 and 316 Stainless Steel using electron beam melting additive manufacturing technology', *Mater. Des.*, 2016, **94**, 17–27.

62. S. L. Sing, L. P. Lam, D. Q. Zhang, Z. H. Liu, and C. K. Chua: 'Interfacial characterization of SLM parts in multi-material processing: Intermetallic phase formation between AlSi10Mg and C18400 copper alloy', *Mater. Charact.*, 2015, **107**, 220–227.

63. Z. H. Liu, D. Q. Zhang, S. L. Sing, C. K. Chua, and L. E. Loh: 'Interfacial characterization of SLM parts in multi-material processing: Metallurgical diffusion between 316L stainless steel and C18400 copper alloy', *Mater. Charact.*, 2014, **94**, 116–125.

64. K. Shah, I. U. Haq, A. Khan, S. A. Shah, M. Khan, and A. J. Pinkerton: 'Parametric study of development of Inconel-steel functionally graded materials by laser direct metal deposition', *Mater. Des.*, 2014, **54**, 531–538.

65. S. Chianrabutra, B. G. Mellor, and S. Yang: 'A Dry Powder Material Delivery Device for Multiple Material Additive Manufacturing', Proc. Solid Free. Fabr. Symp., University of Texas, Austin, TX, USA, August 2014, 36–48.

66. D. C. Hofmann, S. Roberts, R. Otis, J. Kolodziejska, R. P. Dillon, J. Suh, A. Shapiro, Z.K. Liu, and J.P. Borgonia: 'Developing gradient metal alloys through radial deposition additive manufacturing', *Sci. Rep.*, 2014, **4**, 5357.

67. N. Harrison, J. R. Field, F. Quondamatteo, W. Curtin, P. E. McHugh, and P. McDonnell: 'Micromotion and friction evaluation of a novel surface architecture for improved primary fixation of cementless orthopaedic implants', *J. Mech. Behav. Biomed. Mater.*, 2013, **21**, 37–46.

68. M. Dewidar, A. Khalil, and J. K. Lim: 'Processing and mechanical properties of porous 316L stainless steel for biomedical applications', *Trans. Nonferrous Met. Soc. China*, 2007, **17**, 468–473.

69. F. Mangano, L. Chambrone, R. van Noort, C. Miller, P. Hatton, and C. Mangano: 'Direct metal laser sintering titanium dental implants: a review of the current literature', *Int. J. Biomater.*, 2014, DOI:10.1155/2014/461534.

70. D. D. Gu, W. Meiners, K. Wissenbach, and R. Poprawe: 'Laser additive manufacturing of metallic components: materials, processes and mechanisms', *Int. Mater. Rev.*, 2012, **57**, 133–164.

71. E. O. Olakanmi, K. W. Dalgarno, and R. F. Cochrane: 'Laser sintering of blended Al-Si powders,” *Rapid Prototyp. J.*, 2012, **18**,109–119.

72. W. Yan, W. Ge, J. Smith, S. Lin, O. L. Kafka, F. Lin, W. K. Liu: 'Multi-scale modeling of electron beam melting of functionally graded materials', *Acta Mater.*, 2016, **115**, 403–412.

73. J. Jhabvala, E. Boillat, T. Antignac, and R. Glardon: 'On the effect of scanning strategies in the selective laser melting process', *Virtual Phys. Prototyp.*, 2010, **5**, 99–109.

74. J. P. Kruth, P. Mercelis, L. Froyen, and M. Rombouts: 'Binding Mechanisms in Selective Laser Sintering and Selective Laser Melting', *Rapid Prototyp. J.*, 2005, **11**, 44–59.

75. J. P. Kruth, B. Van der Schueren, J.E. Bonsem and B. Morren: 'Basic Powder Metallurgical Aspects in Selective Metal Powder Sintering', *Ann. CIRP*, Jan. 1996, **45**, (1), 183–186.

76. J. P. Kruth, G. Levy, F. Klocke, and T. H. C. Childs: 'Consolidation phenomena in laser and powder-bed based layered manufacturing', *CIRP Ann. - Manuf. Technol.*, 2007, **56**, 730–759.

77. H. Momeni, H. Razavi, and S. G. Shabestari: 'Effect of Supersolidus Liquid Phase Sintering on the Microstructure and densification of the Al-Cu-Mg Pre-alloyed powder', *Iran. J. Mater. Sci. Eng.*, 2011, **8**, 10–17.

78. N. Read, W. Wang, K. Essa, and M. M. Attallah: 'Selective laser melting of AlSi10Mg alloy: Process optimisation and mechanical properties development', *Mater. Des.*, 2015, **65**, 417–424.

79. M. N. Ahsan, A. J. Pinkerton, R. J. Moat, and J. Shackleton: 'A comparative study of laser direct metal deposition characteristics using gas and plasma-atomized Ti–6Al–4V powders', *Mater. Sci. Eng. A*, 2011, **528**, 7648–7657.

80. L. Löber, F. P. Schimansky, U. Kühn, F. Pyczak, and J. Eckert: 'Selective laser melting of a beta-solidifying TNM-B1 titanium aluminide alloy', *J. Mater. Process. Technol.*, 2014, **214**, 1852–1860.

81. R. Li, J. Liu, Y. Shi, M. Du, and Z. Xie: '316L Stainless Steel with Gradient Porosity Fabricated by Selective Laser Melting', *J. Mater. Eng. Perform.*, 2010, **19**, 666–671.

82. T. Bormann, R. Schumacher, B. Müller, M. Mertmann, and M. Wild: 'Tailoring Selective Laser Melting Process Parameters for NiTi Implants', *J. Mater. Eng. Perform.*, 2012, **21**, 2519–2524.

83. O. A. M. Abdelaal and S. M. H. Darwish: 'Analysis, Fabrication and a Biomedical Application of Auxetic Cellular Structures', *Int. J. Eng. Innov. Technol.*, 2012, **2**, 218–223.

84. S. Zhang, Q. Wei, L. Cheng, S. Li, and Y. Shi: 'Effects of scan line spacing on pore characteristics and mechanical properties of porous Ti6Al4V implants fabricated by selective laser melting', *Mater. Des.*, 2014, **63**, 185–193.

85. N. Tolochko, S. Mozzharov, T. Laoui, and L. Froyen: 'Selective laser sintering of single- and two-component metal powders', *Rapid Prototyp. J.*, 2003, **9**, 68–78.

86. M. Agarwala, D. Bourell, J. Beaman, H. Marcus, and J. Barlow: 'Direct selective laser sintering of metals', *Rapid Prototyp. J.*, 1995, **1**, (1), 26–36.

87. Y. Song: 'Experimental study of the basic process mechanism for direct selective laser sintering of low-melting metallic powder', *Ann. CIRP 46*, Jan. 1997, **1**, 127–130.

88. F. Abe and K. Osakada: 'A study of laser prototyping for direct manufacturing of dies from metallic powders', Proc. Fifth ICTP on 'Advanced Technology of Plasticity', Columbus, Ohio USA, October 1996, Vol. II, 923–926.

89. B. Van der Schueren and J. P. Kruth: 'Powder deposition in selective metal powder sintering', *Rapid Prototyp. J.*, 1995, **1**, 23–31.

90. Y. P. Kathuria: 'Microstructuring by selective laser sintering of metallic powder', *J. Surf. Coatings Technol.*, 1999, **116-119**, 643–647.

91. D. Gu and Y. Shen: 'Effects of processing parameters on consolidation and microstructure of W–Cu components by DMLS', *J. Alloys Compd.*, Apr. 2009, **473**, (1-2), 107–115.

92. K. Schmidtke, F. Palm, A. Hawkins, and C. Emmelmann: 'Process and Mechanical Properties: Applicability of a Scandium modified Al-alloy for Laser Additive Manufacturing', *Phys. Procedia*, 2011, **12**, 369–374.

93. B. Verlee, T. Dormal, and J. Lecomte-Beckers: 'Density and porosity control of sintered 316L stainless steel parts produced by additive manufacturing', *Powder Metall.*, 2012, **5**, 260–267.

94. D. Manfredi, F. Calignano, E. P. Ambrosio, M. Krishnan, R. Canali, S. Biamino, M. Pavese, E. Atzeni, L. Iuliano, P. Fino, and C. Badini: 'Direct Metal Laser Sintering : an additive manufacturing technology ready to produce lightweight structural parts for robotic applications', *La Metall. Ital.*, 2014, **10**, 15–24.

95. D. Manfredi, F. Calignano, M. Krishnan, R. Canali, E. Ambrosio, and E. Atzeni: 'From Powders to Dense Metal Parts: Characterization of a Commercial AlSiMg Alloy Processed through Direct Metal Laser Sintering', *Materials (Basel).*, 2013, **6**, 856–869.

96. D. Manfredi, E. P. Ambrosio, F. Calignano, R. Canali, M. Krishnan, S. Biamino, M. Pavese, and P. Fino: 'Realization and Characterization of AlSiMg / SiC Composites by Direct Metal Laser Sintering', 15th Eur. Conf. on 'Composite Materials', Venice, Italy, June 2012, 1–6.

97. C. Yan, L. Hao, A. Hussein, S. L. Bubb, P. Young, and D. Raymont: 'Evaluation of light-weight AlSi10Mg periodic cellular lattice structures fabricated via direct metal laser sintering', *J. Mater. Process. Technol.*, 2014, **214**, 856–864.

98. E. O. Olakanmi, R. F. Cochrane, and K. W. Dalgarno: 'Densification mechanism and microstructural evolution in selective laser sintering of Al–12Si powders', *J. Mater. Process. Technol.*, Jan. 2011, **211**, (1), 113–121.

99. S. K. Ghosh, K. Bandyopadhyay, and P. Saha: 'Development of an in-situ multi-component reinforced Al-based metal matrix composite by direct metal laser sintering technique — Optimization of process parameters', 2014, **93**, 68–78.

100. F. Xie, X. He, S. Cao, and X. Qu: 'Structural and mechanical characteristics of porous 316L stainless steel fabricated by indirect selective laser sintering', *J. Mater. Process. Technol.*, 2013, **213**, 838–843.

101. S. Li, Q. Wei, Y. Shi, Z. Zhu, and D. Zhang: 'Microstructure Characteristics of Inconel 625 Superalloy Manufactured by Selective Laser Melting', *J. Mater. Sci. Technol.*, 2015, **31**, 946–952.

102. C. A. Brice and N. Dennis: 'Cooling Rate Determination in Additively Manufactured Aluminum Alloy 2219', *Metall. Mater. Trans. A Phys. Metall. Mater. Sci.*, 2015, **46**, 2304–2308.

103. W. Shifeng, L. Shuai, W. Qingsong, C. Yan, Z. Sheng, and S. Yusheng: 'Effect of molten pool boundaries on the mechanical properties of selective laser melting parts', *J. Mater. Process. Technol.*, 2014, **214**, 2660–2667.

104. D. Gu, Y.C. Hagedorn, W. Meiners, G. Meng, R. J. S. Batista, K. Wissenbach, and R. Poprawe: 'Densification behavior, microstructure evolution, and wear performance of selective laser melting processed commercially pure titanium', *Acta Mater.*, 2012, **60**, 3849–3860.

105. M. Rombouts, J.P. Kruth, L. Froyen, and P. Mercelis: 'Fundamentals of Selective Laser Melting of alloyed steel powders', *CIRP Ann. - Manuf. Technol.*, Jan. 2006, **55**, (1), 187–192

106. A. Simchi and H. Pohl: 'Effects of laser sintering processing parameters on the microstructure and densification of iron powder', *Mater. Sci. Eng. A*, Oct. 2003, **359**, (1-2), 119–128.

107. S. A. Khairallah, A. T. Anderson, A. Rubenchik, and W. E. King: 'Laser powder-bed fusion additive manufacturing: Physics of complex melt flow and formation mechanisms of pores, spatter, and denudation zones', *Acta Mater.*, 2016, **108**, 36–45.

108. I. Yadroitsev and I. Smurov: 'Surface morphology in selective laser melting of metal powders', *Phys. Procedia*, 2011, **12**, 264–270.

109. J. J. Lewandowski and M. Seifi: 'Metal Additive Manufacturing: A Review of Mechanical Properties', *Annu. Rev. Mater. Res*, 2016, **46**, 151–186.

110. M. Svensson and U. Ackelid: 'Titanium Alloys Manufactured with Electron Beam Melting Mechanical and Chemical Properties', Proc. Conf. on 'Materials and Processes for Medical Devices', Minneapolis, Minnesota USA, August 2009, 189–194.

111. U. Ackelid and M. Svensson: 'Additive Manufacturing of Dense Metal Parts by Electron Beam Melting', Proc. Conf. on 'Materials Science and Technology', Pittsburgh, PA USA, 2009, 2711–2719.

112. A. R. Vinod, C. K. Srinivasa, R. Keshavamurthy, and P. V. Shashikumar: 'A novel technique for reducing lead-time and energy consumption in fabrication of Inconel-625 parts by laser-based metal deposition process', *Rapid Prototyp. J.*, 2015, **22**, 269-280.

113. M. Seifi, D. Christiansen, J. Beuth, O. Harrysson, and J. J. Lewandowski: 'Process Mapping, Fracture and Fatigue Behaviour of Ti-6Al-4V Produced by EBM Additive Manufacturing', in 'Proceedings of the 13th World Conference on Titanium', (ed. A.L. Pilchak *et al.*), 1373–1378, 2016, San Diego, The Minerals, Metals and Materials Society.

114. H. Nakamura, Y. Kawahito, K. Nishimoto, and S. Katayama: 'Elucidation of Melt Flows and spatter formation mechanisms during high power laser welding of pure titanium', *J. Laser Appl.*, 2015, **27**, 032012-1 – 032012-10.

115. A. F. H. Kaplan and J. Powell: 'Spatter in laser welding', *J. Laser Appl.*, 2011, **23**, 32005-1 – 32005-7.

116. D. Bäuerle: 'Laser Processing and Chemistry', 2nd edn, 2000, Berlin, Springer Berlin Heidelberg.

117. M. Simonelli, C. Tuck, N. T. Aboulkhair, I. Maskery, I. Ashcroft, R. D. Wildman, and R. Hague: 'A Study on the Laser Spatter and the Oxidation Reactions During Selective Laser Melting of 316L Stainless Steel, Al-Si10-Mg, and Ti-6Al-4V', *Metall. Mater. Trans. A Phys. Metall. Mater. Sci.*, 2015, **46A**, 3842–3851.

118. B. H. Rabin, G. R. Smolik, and G. E. Korth: 'Characterization of entrapped gases in rapidly solidified powders', *Mater. Sci. Eng. A*, Apr. 1990, **124**, (1), 1–7.

119. Y. Liu, Y. Yang, S. Mai, D. Wang, and C. Song: 'Investigation into spatter behavior during selective laser melting of AISI 316L stainless steel powder', *Mater. Des.*, 2015, **87**, 797–806.

120. N. Hopkinson, R. J. M. Hague, and P. M. Dickens: 'Rapid Manufacturing: An Industrial Revolution for the Digital Age', 1st edn, 1–4; 2006, Chichester, John Wiley & Sons, Ltd.

121. K. A. Mumtaz and N. Hopkinson: 'Selective Laser Melting of thin wall parts using pulse shaping', *J. Mater. Process. Technol.*, 2010, **210**, 279–287.

122. M. J. Matthews, G. Guss, S. A. Khairallah, A. M. Rubenchik, P. J. Depond, and W. E. King: 'Denudation of metal powder layers in laser powder bed fusion processes', *Acta Mater.*, 2016, **114**, 33–42..

123. N. K. Tolochko*,* S. E. Mozzharov, I. A. Yadroitsev, T. Laoui, L. Froyen, V. I. Titov, and M. B. Ignaitev: 'Balling processes during selective laser treatment of powders', *Rapid Prototyp. J.*, 2004, **10**, 78–87.

124. I. Yadroitsev, A. Gusarov, I. Yadroitsava, and I. Smurov: 'Single track formation in selective laser melting of metal powders', *J. Mater. Process. Technol.*, 2010, **210**, 1624–1631.

125. J. A. Kanko, A. P. Sibley, and J. M. Fraser: 'In situ morphology-based defect detection of selective laser melting through inline coherent imaging', *J. Mater. Process. Technol.*, 2016, **231**, 488–500.

126. I. Yadroitsev and P. Bertrand: 'Use of track/layer morphology to develop functional parts by selective laser melting', *J. Laser Appl.*, 2013, **25**, 1–7.

127. D. E. Cooper: 'The High Deposition Rate Additive Manufacture of Nickel Superalloys and Metal Matrix Composites', University of Warwick, Warwick, UK, 2016, 208–220.

128. I. Yadroitsev and I. Smurov: 'Surface morphology in selective laser melting of metal powders', *Phys. Procedia*, 2011, **12**, 264–270.

129. W. J. Sames, F. Medina, W. H. Peter, S. S. Babu, and R. R. Dehoff: 'Effect of process control and powder quality on IN 718 produced using Electron Beam Melting', in 8th International Symposium on Superalloy 718 and Derivatives', (ed. E. Ott *et al.*), 409–423; 2014, Pittsburgh, The Minerals, Metals and Materials Society.

130. E. Yasa and J.P. Kruth: 'Microstructural investigation of Selective Laser Melting 316L stainless steel parts exposed to laser re-melting', *Procedia Eng.*, 2011, **19**, 389–395.

131. S. Dadbakhsh and L. Hao: 'Effect of Al alloys on selective laser melting behaviour and microstructure of in situ formed particle reinforced composites', *J. Alloys Compd.*, 2012, **541**, 328–334.

132. A. Simchi: 'Direct laser sintering of metal powders: Mechanism, kinetics and microstructural features', *Mater. Sci. Eng. A*, Jul. 2006, **428**, (1–2), 148–158.

133. D. Clark, M. R. Bache, and M. T. Whittaker: 'Shaped metal deposition of a nickel alloy for aero engine applications', *J. Mater. Process. Technol.*, Jul. 2008, **203**, (1–3), 439–448.

134. J. P. Kruth, L. Froyen, J. Van Vaerenbergh, P. Mercelis, M. Rombouts, and B. Lauwers: 'Selective laser melting of iron-based powder', *J. Mater. Process. Technol.*, Jun. 2004, **149**, (1–3), 616–622

135. N. K. Tolochko, T. Laoui, Y. V. Khlopokov, S. E. Mozzharov, V. I. Titov, and M. B. Ignatiev: 'Absorptance of powder materials suitable for laser sintering', *Rapid Prototyp. J.*, 2000, **6**, 155–161.

136. D. Gu and Y. Shen: 'Balling phenomena during direct laser sintering of multi-component Cu-based metal powder', *J. Alloys Compd.*, Apr. 2007, **432**, (1–2), 163–166.

137. C. Qiu, G. A. Ravi, C. Dance, A. Ranson, S. Dilworth, and M. M. Attallah: 'Fabrication of large Ti-6Al-4V structures by direct laser deposition', *J. Alloys Compd.*, 2015, **629**, 351–361.

138. N. T. Aboulkhair, N. M. Everitt, I. Ashcroft, and C. Tuck: 'Reducing porosity in AlSi10Mg parts processed by selective laser melting', *Addit. Manuf.*, 2014, **1-4**, 77–86.

139. E. Louvis, P. Fox, and C. J. Sutcliffe: 'Selective laser melting of aluminium components', *J. Mater. Process. Technol.*, 2011, **211**, 275–284.

140. E. O. Olakanmi, R. F. Cochrane, and K. W. Dalgarno: 'Spheroidisation and oxide disruption phenomena in direct Selective Laser Melting (SLM) of pre-alloyed Al-Mg and Al-Si powders', Proc. Conf. on 'TMS Annual Meetings and Exhibition', San Francisco, California USA, February 2009, Vol. 1, 371–380.

141. D. Dai and D. Gu: 'Thermal behavior and densification mechanism during selective laser melting of copper matrix composites: Simulation and experiments', *Mater. Des.*, 2014, **55**, 482–491.

142. K. Zhang, S. Wang, W. Liu, and X. Shang: 'Characterization of stainless steel parts by Laser Metal Deposition Shaping', *Mater. Des.*, 2014, **55**, 104–119.

143. J. A. Cherry, H. M. Davies, S. Mehmood, N. P. Lavery, S. G. R. Brown, and J. Sienz: 'Investigation into the effect of process parameters on microstructural and physical properties of 316L stainless steel parts by selective laser melting', *Int. J. Adv. Manuf. Technol.*, 2014, **76**, 869–879.

144. D. Gu and Y. Shen: 'Balling phenomena in direct laser sintering of stainless steel powder: Metallurgical mechanisms and control methods', *Mater. Des.*, 2009, **30**, 2903–2910.

145. Y. Li and D. Gu: 'Parametric analysis of thermal behavior during selective laser melting additive manufacturing of aluminum alloy powder', *Mater. Des.*, 2014, **63**, 856–867.

146. M. Das, V. K. Balla, D. Basu, S. Bose, and A. Bandyopadhyay: 'Laser processing of SiC-particle-reinforced coating on titanium', *Scr. Mater.*, 2010, **63**, 438–441.

147. Q. Jia and D. Gu: 'Selective laser melting additive manufactured Inconel 718 superalloy parts: High-temperature oxidation property and its mechanisms', *Opt. Laser Technol.*, 2014, **62**, 161–171.

148. B. Zhang, H. Liao, and C. Coddet: 'Effects of processing parameters on properties of selective laser melting Mg–9%Al powder mixture', *Mater. Des.*, 2012, **34**, 753–758..

149. B. Song, S. Dong, S. Deng, H. Liao, and C. Coddet: 'Microstructure and tensile properties of iron parts fabricated by selective laser melting', *Opt. Laser Technol.*, 2014, **56**, 451–460.

150. L. Facchini, E. Magalini, P. Robotti, A. Molinari, S. Höges, and K. Wissenbach: 'Ductility of a Ti-6Al-4V alloy produced by selective laser melting of prealloyed powders', *Rapid Prototyp. J.*, 2010, **16**, 450–459.

151. L. E. Murr, S. A. Quinones, S. M. Gaytan, M. I. Lopez, A. Rodela, E. Y. Martinez, D. H. Hernandez, E. Martinez, F. Medina, and R. B. Wicker', Microstructure and mechanical behavior of Ti-6Al-4V produced by rapid-layer manufacturing, for biomedical applications', *J. Mech. Behav. Biomed. Mater.*, Jan. 2009, **2**, (1), 20–32.

152. Z. Wang, K. Guan, M. Gao, X. Li, X. Chen, and X. Zeng: 'The microstructure and mechanical properties of deposited-IN718 by selective laser melting', *J. Alloys Compd.*, 2012, **513**, 518–523.

153. D. Wang, Y. Yang, X. Su, and Y. Chen: 'Study on energy input and its influences on single-track,multi-track, and multi-layer in SLM', *Int. J. Adv. Manuf. Technol.*, 2012, **58**, 1189–1199.

154. P. Kanagarajah, F. Brenne, T. Niendorf, and H. J. Maier: 'Inconel 939 processed by selective laser melting: Effect of microstructure and temperature on the mechanical properties under static and cyclic loading', *Mater. Sci. Eng. A*, 2013, **588**, 188–195.

155. E. Chlebus, B. Kuźnicka, T. Kurzynowski, and B. Dybała: 'Microstructure and mechanical behaviour of Ti―6Al―7Nb alloy produced by selective laser melting', *Mater. Charact.*, 2011, **62**, 488–495.

156. L. Thijs, M. L. Montero Sistiaga, R. Wauthle, Q. Xie, J. P. Kruth, and J. Van Humbeeck: 'Strong morphological and crystallographic texture and resulting yield strength anisotropy in selective laser melted tantalum', *Acta Mater.*, 2013, **61**, 4657–4668.

157. B. Baufeld, O. Van der Biest, and R. Gault: 'Additive manufacturing of Ti–6Al–4V components by shaped metal deposition: Microstructure and mechanical properties', *Mater. Des.*, 2010, **31**, S106–S111.

158. D. Dimitrov, K. Schreve, and N. De Beer: 'Advances in Three Dimensional Printing - State od the Art and Future Perspectives', *J. New Gener. Sci.*, 2006, **12**, 136-147.

159. B. Song, S. Dong, P. Coddet, H. Liao, and C. Coddet: 'Fabrication of NiCr alloy parts by selective laser melting: Columnar microstructure and anisotropic mechanical behavior', *Mater. Des.*, 2014, **53**, 1–7.

160. M. A. Pinto, N. Cheung, M. C. F. Ierardi, and A. Garcia: 'Microstructural and hardness investigation of an aluminum–copper alloy processed by laser surface melting', *Mater. Charact.*, 2003, **50**, 249–253.

161. T. Amine, J. W. Newkirk, and F. Liou: 'An investigation of the effect of direct metal deposition parameters on the characteristics of the deposited layers', *Case Stud. Therm. Eng.*, 2014, **3**, 21–34.

162. L. C. Zhang, D. Klemm, J. Eckert, Y. L. Hao, and T. B. Sercombe: 'Manufacture by selective laser melting and mechanical behavior of a biomedical Ti–24Nb–4Zr–8Sn alloy', *Scr. Mater.*, Jun. 2011, **65**, (1), 21–24.

163. T. Bormann, B. Müller, M. Schinhammer, A. Kessler, P. Thalmann, and M. de Wild: 'Microstructure of selective laser melted nickel–titanium', *Mater. Charact.*, 2014, **94**, 189–202.

164. K. Kempen, L. Thijs, J. Van Humbeeck, and J.P. Kruth: 'Mechanical Properties of AlSi10Mg Produced by Selective Laser Melting', *Phys. Procedia*, 2012, **39**, 439–446.

165. K. Guan, Z. Wang, M. Gao, X. Li, and X. Zeng: 'Effects of processing parameters on tensile properties of selective laser melted 304 stainless steel', *Mater. Des.*, 2013, **50**, 581–586.

166. D. D. Gu, Y. F. Shen, J. L. Yang, and Y. Wang: 'Effects of processing parameters on direct laser sintering of multicomponent Cu based metal powder', *Mater. Sci. Technol.*, 2006, **22** , 1449–1455.

167. Q. Jia and D. Gu: 'Selective laser melting additive manufacturing of Inconel 718 superalloy parts: Densification, microstructure and properties', *J. Alloys Compd.*, 2014, **585**, 713–721.

168. I. Shishkovsky, I. Yadroitsev, and I. Smurov: 'Direct Selective Laser Melting of Nitinol Powder', *Phys. Procedia*, 2012, **39**, 447–454.

169. T. Vilaro, C. Colin, J. D. Bartout, L. Nazé, and M. Sennour: 'Microstructural and mechanical approaches of the selective laser melting process applied to a nickel-base superalloy', *Mater. Sci. Eng. A*, 2012, **534**, 446–451.

170. A. J. Sterling, B. Torries, N. Shamsaei, S. M. Thompson, and D. W. Seely: 'Fatigue behavior and failure mechanisms of direct laser deposited Ti-6Al-4V', *Mater. Sci. Eng. A*, 2016, **655**, 100–112.

171. R. M. Mahamood, E. T. Akinlabi, M. Shukla, and S. Pityana: 'Material Efficiency of Laser Metal Deposited Ti6Al4V : Effect of Laser Power', *Eng. Lett.*, Feb. 2013, **21**, (1), 18–22.

172. P. K. Farayibi, J. A. Folkes, and A. T. Clare: 'Laser Deposition of Ti-6Al-4V Wire with WC Powder for Functionally Graded Components', *Mater. Manuf. Process.*, 2012, **28**, 514–518.

173. M. Rombouts, G. Maes, M. Mertens, and W. Hendrix: 'Laser metal deposition of Inconel 625: Microstructure and mechanical properties', *J. Laser Appl.*, 2012, **24**, 52007-1 – 52007-6.

174. X. Zhao, J. Chen, X. Lin, and W. Huang: 'Study on microstructure and mechanical properties of laser rapid forming Inconel 718', *Mater. Sci. Eng. A*, Apr. 2008, **478**, (1–2), 119–124.

**Figure captions**

Figure 1. Classification of 3D printing processes for metal components based on ISO/ASTM 52900 (Ref. 13).

Figure 2. Schematic showing the formation of denudation zones i.e. empty area in between powder bed and solidified track (rectangular shape irradiated by laser beam), caused by metal vapour flux and entrapment of powder particles in a shear gas flow. As a result, powder particles are either ejected away to the environment, or absorbed into the molten pool when the pressure is high (Kn<1). When the pressure is low (Kn>1), powder particles are cleared sideways of the molten pool. Kn is the Knudsen number. 122

Figure 3. Two types of porosity typically present in 3D printed metallic parts.129

Figure 4. Optical micrographs showing microstructural characteristics of etched Mg-9%Al samples processed by SLM with different parameters: (a) 10 W, 0.01 m s-1, (b)15 W, 0.02 m s-1,(c) 20 W, 0.04 m s-1, and(d) 15 W, 0.04 m s-1 (Ref. 148)

Figure 5. SEM images which show typical surface morphologies for DMLS-processed W-Cu samples with various scan line spacings: (a) 0.30 mm, (b) 0.25 mm, and (c) 0.15 mm. The fixed parameters are: laser power 850 W, scan speed 0.05 m s-1, powder layer thickness 0.15 mm (Ref. 91)

Figure 6. Different HV values for SLM-processed NiCr parts at different scan speeds in different orientations due to anisotropy. (TV = top view or x-y plane, SV = side plane or x-z/y-z plane) 159

Figure 7. Effects of scan speed on porosity and tensile strength of 316L stainless steel gradient porosity components processed by SLM 81

Figure 8. Processing map of DMLS-processed 316L stainless steel powder showing different solidification track regions: I unmelted tracks; II,III occurrence of balling; IV smooth, continuous tracks 144

Table 1. Generic process sequence for all 3D printing processes.12

|  |  |  |
| --- | --- | --- |
| No. | Step | Description |
| 1  2  3  4  5  6 | 3D CAD model design  Translation into STL file  CAD model sliced  Pre-processing  Printing  Post-processing | Model of the component to be printed is designed in CAD software.  CAD model converted into STL file format which is readable by 3D printing machines.  Continuous geometry of CAD model is sliced into 2D layers approximated by small triangles.  Model is oriented and support structure is planned for efficient printing  CAD model printed onto a build area (substrate) in layer by layer or point by point or line by line fashion to construct physical 3D components.  Finished part is further processed to improve surface finish quality, enhance mechanical properties or to remove support structures. |

Table 2. Mechanical properties of SLM-processed metal parts compared to that of those conventionally manufactured at optimum processing parameters.

|  |  |  |  |  |
| --- | --- | --- | --- | --- |
| SLM-processed material (SPM) | Conventionally processed material (CPM) | Ultimate Tensile Strength (UTS), MPa  SPM CPM | Yield strength, MPa  SPM CPM | Reference |
| Ti6Al7Nb  Ti-24Zr-4Nb-8Sn  AlSi10Mg  304 stainless steel | Wrought Ti6Al7Nb  Hot forged  Ti-24Zr-4Nb-8Sn  Cast and aged AlSi10Mg  ANSI stainless steel | 714-717 520  396 300  665 ± 18 755  1515 ± 60 1440 ± 59 | 566-570 205    563 ± 38 570  984-1024 933-952 | Chlebus *et al.* 155  Zhang *et al.* 162  Kempen *et al.* 164  Guan *et al.* 165 |

Table 3. Optimum processing parameters and characteristics of favourable metallurgical mechanisms for various metal AM processes and materials.

|  |  |  |  |  |  |
| --- | --- | --- | --- | --- | --- |
| Type and  composition  of metals | AM process category and machine used | Optimum processing parameters | Characteristics of favourable metallurgical mechanisms | Mechanical  properties | Reference |
| Al-12Si powder  CuCuSnCuP powder  CuSnCuCuP powder  CuSnCuCuP powder  316L stainless steel powder  316L stainless steel  (porous parts) | PBF; Synrad 240 W CO2 laser machine at University of Leeds  PBF; DMLS apparatus with continuous wave CO2 laser\*  PBF; DMLS system developed at CAEP, China\*  PBF; DMLS system developed at CAEP, China\*  PBF; DMLS system with Gaussian continuous CO2 wave laser\*  PBF; DMLS system, laser type Rofin-Sinar 2000 SM | VED 67 J mm-3, laser power 200 W, scan speed 120 mm s-1, scan spacing 0.1 mm, layer thickness 0.25 mm  Laser power 375 W, scan speed 0.04 m s-1, scan line spacing 0.15 mm, layer thickness 0.20 mm,  Laser power 350 W, scan speed 0.04 m s-1, scan line spacing 0.15 mm, laser spot size 0.30 mm, layer thickness 0.30 mm  Laser power 350 W, scan speed 0.04 m s-1, scan line spacing 0.15 mm, laser spot size 0.30 mm, layer thickness 0.10-0.50 mm  Laser power ≥ 400 W, scan speed ≤ 0.09 m s-1, spot size 0.30 mm, scan line spacing 0.15 mm, layer thickness 0.10-0.25 mm  LED 3400-6000 J m, scan speed ≤ 0.06 m s-1, layer thickness 0.10-0.25 mm | Fully dense dendrite grain agglomerates growing perpendicular to build direction, adequate liquid phase for solidification promotes high cooling rates  Dense sintered structure with dark grain boundary network, relatively homogeneous Sn distribution in Cu matrix due to high mutual solubility of both elements.  Good metallurgical bonding obtained by the formation of sintering nexk, sufficient liquid flow and particle re-arrangement due to capillary forces induced by Marangoni convection.  Formation of continuous network of small agglomerates via fusion of solid particles, formation of fully dense broad dendritic microstructures  Continuous and smooth track formation free from balling, sufficient liquid formation  Pore formation through liquid bridges between partially melted particles and due to growth of sintering necks | Relative density 84%  Porosity levels ~21 to ~55%, tensile strength 152 MPa  Relative density 94.6%, fracture strength 169.2 MPa, hardness 101.7 HV | Olakanmi *et al.*98  Gu *et al.* 31  Gu *et al.* 32  Gu *et al*. 166  Gu *et al.* 144  Gu *et al.* 9 |

\*Maximum power output 2000 W, laser wavelength 10.6 µm

Table 3. Continued.

|  |  |  |  |  |  |
| --- | --- | --- | --- | --- | --- |
| Type and  composition  of metals | AM process category and machine used | Optimum processing parameters | Characteristics of favourable metallurgical mechanisms | Mechanical  properties | Reference |
| AlSi10Mg powder  AlSi10Mg powder  6061 Al powder  ScalmalloyRP powder  WC-Cu powder | PBF; SLM apparatus with YLR-500-SM ytterbium  fiber laser output power ~500W, argon atmosphere  PBF; Modified Concept Laser M1 SLM Machine, 200 W fibre laser  PBF; MCP Realizer 100 (MTT Tooling Technologies, UK) SLM machine  PBF; EOS M 270 SLM  Machine, max. laser output 200 W  PBF; SLM apparatus, continuous wave Gaussian CO2 laser (max. power 2000W) | Laser power 250 W, scan speed 200 mm s-1, powder layer thickness 50 µm, hatch spacing 50 µm, laser spot size 70 µm  Laser power 200 W, scan speed 1400 mm s-1, scan line spacing 105 µm  Laser powers 50 W and 100 W, scan speeds 100-200 mm s s-1, hatch distance 0.15 mm  Laser power 195 W, layer thickness 20 µm, contour scan strategy for cross-section outline, fill scan strategy for the remaining cross-sections  LED 17.5 kJ m-1, layer thickness 250 µm | High molten pool depth to powder layer thickness ratio results in sound metallurgical bonding between adjacent layers, formation of dense cross-sectional morphology free from any pores  Very fine microstructure and fine Si phase distribution due to rapid cooling and solidification, some anisotropy in the elongation at break  Narrow process window at which minimum balling and smooth hatch line formation occur, narrow process window for optimum overlapping area between adjacent melt tracks in order to obtain minimum porosity  Hyper-eutectic Al-Scandium composition obtained  Improvement in gas pore removal efficiency, reduction in mean porosity level, decreased melt viscosity, reduction of pore size from 200 to 50 µm | Relative density at least 98.5%, 99.8% when every layer re-melted with alternating scan directions of 90°, UTS 391 ± 6 MPa at XY direction and 396 ± 8 MPa at Z direction, Microhardness 127 HV  Relative density 83.7-89.5%  Yield strengths >500 MPa, Tensile strengths >520 MPa, Microhardness: as built, ̴100 HV0.3, after ageing, ̴100 HV0.3  Maximum relative density of 96% | Li *et al.* 145  Kempen *et al.* 164  Louvis *et al.* 139  Schmidtke *et al.* 92  Dai *et al.* 141 |

Table 3. Continued.

|  |  |  |  |  |  |
| --- | --- | --- | --- | --- | --- |
| Type and  composition  of metals | AM process category and machine used | Optimum processing parameters | Characteristics of favourable metallurgical mechanisms | Mechanical  properties | Reference |
| Commercially pure (CP) Ti powder  Commercially pure (CP) Ti powder  β-solidifying Titanium aluminide (TiAlNbMoB) powder  316L stainless steel powders  (porous parts)  Mg-9%Al powder | PBF; Fraunhofer ILT SLM system with YLR-200 ytterbium fiber laser (~200 W)\*  PBF; MTT SLM 250 HL machine with 400 W Yb:YAG fiber laser\*  PBF; SLM 250 hl machine (SLM Solutions GmbH) with 400 W Nd-YAG continuous wave fiber laser  PBF; HRPM-II SLM machine developed at HUST, China with 100 W fiber laser\*  PBF; MCP Realize 250  II SLM machine with Nd:YAG laser\* | Laser power 90 W, scan speeds 200 and 300 mm s-1, LED values 450 and 300 J m-1  VED 120 J mm-3, laser powers 85-180 W, scan speeds 70-150 mm s-1, layer thickness 100 µm, hatch distance 100 µm,  Volume contour: Laser power 100 W, scan speed 50 mm s-1  Outer contour: Laser power 175 W, scan speed 1000 mm s-1  In general: hatch distance 0.3 mm, layer thickness 75 µm, stripe hatching strategy  Laser power 100 W, scan speeds 90-180 mm s-1, scan interval 0.1 mm, layer thickness 0.06 mm  Laser power 15 W, scan speed 0.02 m s-, layer thickness  50 µm, hatch spacing 80 µm | Clear and regular solidification front free from balling, stable and continuous scan tracks, dense metallurgically bonded layers without interlayer pores or cracks, clear configuration of molten pool  Formation of refined α’ phase from β, balanced viscosity obtained leading towards homogeneous deposition of fresh powder layers  Fine grained nearly lamellar β microstructure, formation of stable melt tracks without balling or crack, increased melt track width enables good interaction between molten pool and substrate which results in good fusion  Gradient porosity structure formed via layer-wise scan speed variation along certain gradient directions, formation of true gradient microstructure with constant porosity changes, porosity levels 5-35%  Fine equiaxed grain, complete melting of Mg and Al, low porosity (18%), good liquid-solid wetting characteristics | Densification up to 99.5%, Hardness 3.89 GPa  Relative density 99%, UTS 757 ± 12.5, Vickers hardness 261 Hv  Relative density 99%,  Without heat treatment: Ultimate Tensile Strength (UTS) 18160-1903 MPa, Yield strength 1620-1651 MPa  With heat treatment: UTS 1428-1671 MPa, yield strength 886-1071 MPa  Maximum relative density 82% | Gu *et al.* 104  Attar *et al.* 34  Löber *et al.* 80  Li *et al.* 81  Zhang *et al.* 148 |

\* Laser beam in continuous wave mode and all experiments were conducted in Ar atmosphere.

Table 3. Continued.

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| Type and  composition  of metals | AM process category and machine used | Optimum processing parameters | Characteristics of favourable metallurgical mechanisms | Mechanical  properties | Reference |
| Inconel 718 powder  Inconel 718 powder  Inconel 718 powder  NiCr powder  NiTi (Nitinaol) powder  Nimonic 263 (Ni superalloy) powder | PBF; IPG Photonics Ytterbium YLR-200-SM fiber laser\*  PBF; IPG Photonics Ytterbium YLR-200-SM fiber laser\*  PBF; Self-developed SLM machine (LSNF-I) IPG YLR-200 fiber laser\*  PBF; Realizer 250 (MCP Ltd) with YLR-100-SM Ytterbium fiber laser\*  PBF; PM 100 (Phenix Systems)., IGP Photonics YLR-50 Ytterbium fiber laser\*  PBF; Trumpf SLM machine, laser output 500 W | LED 330 J m-1  VED 130 J mm-3, powder layer thickness 50 µm, hatch distance 50 µm, laser beam spot size 50 µm, scan vector length 4 mm  Laser power 170 W, scan speed 25 m min-1, overlap rate 30.5, powder layer thickness 20 µm  Laser power 100 W, scan speeds 0.1-0.3 m s-1, diameter of laserbeam 34 µm, scan line spacing 40 µm, powder layer thickness 50 µm  Laser power 50 W, scan speeds 100 and 160 mm s-1, hatch distance 100 µm, laser diameter 60 µm  Laser power 200 W, scan speed 100 mm s-1, laser beam diameter 250 µm, layer thickness 30 µm | Uniformly distributed columnar dendrites  Sound cross-section morphologies free from any significant pores, dispersion of primary dendrites, formation of typical directional solidified slender columnar dendrites  Closed stacked ends of melted tracks to form good metallurgical bonding between two adjacent layers, grain refinement due to rapid heating and cooling  Stable molten pool as a result of sufficiently high LED, dense structure with columnar grains in the build direction  Absence of free Ni and Ni-Ti intermetallic phase transformation which are free from impurities and suitable for medical applications  Porosity < 0.3%, good layer bonding due to re-melting of previously solidified layers | Near full density 98.4%, microhardness 395.8 HV0.2  Dense part of near full density (98.9%)  Relative density nearly 100%, UTS before and after heat treatment; 1137-1148 MPa and 1280-1358 MPa respectively, average microhardness before and after heat treatment; 365 and 470 Hv respectively  98-99% of theoretical density  Relative bulk density 97%, microhardness ~540-735 Hv  UTS for as-processed samples; 1085 ±11 MPa | Jia *et al.* 167  Jia *et al.* 147  Wang *et al.* 152  Song *et al.* 159  Shishkovsky *et al*. 168  Vilaro *et al.* 169 |

\* All machines are based on SLM technique and utilised continuous wave laser beam.

Table 3. Continued.

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| Type and  composition  of metals | AM process category and machine used | Optimum processing parameters | Characteristics of favourable metallurgical mechanisms | Mechanical  properties | Reference |
| Ti6Al4V  Ti6Al4V powder  Ti6Al4V powder  Ti6Al4V wire with WC powder  \*FGM parts | DED; 6.5-axis TRUMPF DLD (blown powder) system fitted with a 4 kW disc laser and an automatic spot change collimator\*  DED; OPTOMEC LENS 750 machine, 1 kW Nd:YAG laser  DED; KUKA robot equipped with high power Nd:YAG laser and coaxial nozzles  DED; 2 kW Ytterbium doped, continuous wave fiber laser (IPG Photonics) with beam delivery system and Precitec YC 50 cladding head, all mounted on a 4-axis CNC table for substrate movement | Set 1: Laser powers 1215-1315 W, scan speeds 750-850 mm min-1, powder feed rate 6.5-7.5 g min-1, Z-step 0.76 mm  Set 2: Laser powers 136-1460 W, scan speeds 750-850 mm min-1, powder feed rate 5.5-6.5 g min-1, Z-step 0.84 mm  Laser power 350 W, hatch spacing 0.508 mm, laser traverse speed 16.93 mm s-1, powder flow rate 0.16 g min-1, alternate scan strategy  Laser power 1.5 kW, laser travel speed 0.005 m s-1, powder flow rate 1.44 g min-1 , laser spot size 2 mm, laser focal distance 195 mm  Laser power 1800 W, traverse speed 172 mm min- , wire feed rate 800 mm min- 1 powder feed rate 10-40 g min- 1, carrier gas flow rate 10 l min-1, shielding gas flow rate 30 l min-1 | Minimum porosity levels, columnar grain and martensitic needle structure formed in as-fabricated samples, sufficient melting of Ti6Al4V which avoids lack-of-fusion pore formation  Predominantly columnar, epitaxial grain growth of the deposited layers  Max. powder efficiency 97.22%, fine acicular α microstructure, min. dilution between substrate and powder for max. deposited height, fully dense deposits free from porosity  Fully dense beads deposited with no porosity, min. melt depth (dilution) between substrate and feedstock indicating strong metallurgical bonding at clad/substrate interface, no crack and delamination for all deposits, clad height increases with powder flow rate, clad width decreases as powder flow rate increases | Yield strength after HIP processing; 850 ± 2 MPa, UTS; 920 ± 1 MPa  Yield stress (0.2%): as built 908 MPa; annealed 959 MPa; heat treated 957 MPa, UTS: as built 1038 MPa; annealed 1049 MPa; heat treated 1097 MPa  Microhardness increases from 600 HV to 1030 HV when powder flow rate increases from 10-40 g min- 1 | Qiu *et al.* 137  Sterling *et al.* 170  Mahamood *et al.* 171  Farayibi *et al*. 172 |

Table 3. Continued.

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| Type and  composition  of metals | AM process category and machine used | Optimum processing parameters | Characteristics of favourable metallurgical mechanisms | Mechanical  properties | Reference |
| 316L stainless steel powder  316L stainless steel powder  Inconel 625 powder  Inconel 718 powder | DED; LMD based, 6-axis NC machine (Shenyang Machine Tools Co., Ltd.) with 3kW CO2 laser (Dalu Laser Group)  DED; DLD system, 5-axis FADAL CNC (VMC 3016) with 1 kW coherent diode laser  DED; LMD system with 7 kW IPG fiber laser  DED; Laser Rapid Forming (LRF) system consisting of 5 kW ROFIN SINAR continuous wave CO2 laserm LPM-408 CNC working table, DPSF-1 powder feed system (powder feed rate error ± 2%), off-axial nozzle | Laser power 1000 W, laser travel speed 8 mm s-1, powder feed rate 4 g min-1  Laser power 600 W, laser travel speed 450 mm min-1 , powder feed rate 12 g min-1 , beam diameter 5 mm  Laser power 500 W, scan speed 750 mm min-1 , scan spacing 0.6 mm, layer thickness 0.4 mm, powder flow rate 2.46 g min- 1, transport gas 4 l min-1  Laser power 2350 W, scan speed 8 mm s-1 , spot diameter 3 mm, powder feed rate 6 g min- 1, shield gas flow rate 7.5 l min-1 | Continuous and uniform cladding track, smooth cladding surface, clad height and width 0.46 mm and 2 mm respectively, clad layers free from crack and porosity, strong metallurgical bonding between clad and substrate indicated by bright white band  Homogeneous cellular microstructure at the cross-section, grain sizes ranging from 15 to 25 µm, deposited layers free from cracks, epitaxial grain growth  Dense parts without defects except for a few spherical gas inclusions, anisotropic fine cellular/dendritic microstructure  Superior mechanical properties related to thin and dense columnar dendrites with 5 µm primary arm spacing, parts produced free from defects due to the optimum processing parameters and usage of powder manufactured using PREP technique | Microhardness 331-427 HV, yield strength for as-deposited parts; a) parallel direction: 558 MPa, b) vertical direction: 352 MPa  Microhardness 185-280 HV  Relative density ~99.8%, yield strength; a) horizontal orientation: 631 ± 7 MPa, b) vertical orientation: 641 ± 12 MPa  After heat treatment: UTS 1360 MPa (1340 MPa), yield strength, YS0.2 1170 MPa (1100 MPa), elongation 18% (12%)  \*values in brackets indicated that of wrought manufactured Inconel 718 | Zhang *et al.* 142  Amine *et al.* 161  Rombouts *et al*. 173  Zhao *et al*. 174 |