Recent progress and scientific challenges in multi-material additive manufacturing via laser-based powder bed fusion

Chao Wei and Lin Li

Laser Processing Research Centre, Department of Mechanical, Aerospace and Civil Engineering, The University of Manchester, Manchester, United Kingdom

ABSTRACT
Multi-material additive manufacturing provides a new route for fabricating components with tailored physical properties. Laser-based powder bed fusion (L-PBF), also known as selective laser melting, is a powder bed-based additive manufacturing technology. This technology affords the advantage of manufacturing metallic and non-metallic materials with high geometrical resolution. An emerging field relevant to the foregoing is multi-material L-PBF. This paper reviews the latest progress in this field including multiple material powder deposition mechanisms, molten pool behaviour, process characteristics of printing metal–metal, metal–ceramic, and metal–polymer multiple material components, and potential applications. Finally, scientific and technological challenges are presented.

1. Introduction
Over the last two decades, additive manufacturing (AM) has been extensively studied and applied to manufacturing technology owing to its inherent flexibility and efficiency in producing highly complex components (Huang et al. 2013). Conventional AM approaches are limited to the manufacturing of single material components; thus, they cannot fabricate products with distributed functional properties, such as high wear resistance, high-temperature resistance, and corrosion resistance, in target regions while maintaining strength and low cost in other parts (Hall and Hall 2018). The emerging multi-material AM (MMAM) technologies not only overcome the above problems but also afford more complexity and functionality to new applications, such as the embedding of anti-counterfeiting features in AM parts (Pa et al. 2015). MMAM is defined as an AM process in which no less than two types of materials are physically deposited onto any spatial location according to pre-programmed codes (Vaezi et al. 2013).

Laser-based powder bed fusion (L-PBF), also known as selective laser melting, is a powder bed-based AM method. Thin layers of dry powders are spread and levelled on the powder bed layer by layer using a powder recoating device (e.g. blade) (Khairallah et al. 2016). A galvo scanner directs a focused laser beam and selectively melts powders in pre-designed zones according to sliced three-dimensional (3D) model data (Goodridge and Ziegelmeier 2017).

Conventionally, manufacturing methods, such as dissimilar material welding and explosive welding, can manufacture multi-material parts with simple geometrical structures. Even functionally graded material (FGM) parts can be manufactured by centrifugal casting. However, the above-mentioned conventional manufacturing methods are incapable of producing parts with complex geometries and multi-function characteristics (Li et al. 2020). The AM methods provide designers and manufacturing engineers with a new processing route to overcome the foregoing deficiencies and achieve spatial gradient changes in material composition and functions. At present, the commonly used metal AM methods, such as wire arc AM (WAAM), L-PBF, and laser-based directed energy deposition (L-DED), have been used to manufacture multi-metal material parts. The surface roughness and dimensional accuracy of parts processed by WAAM are considerably lower than those of L-DED and L-PBF using powder materials. This is because the heat input from the arc in the WAAM process is considerably greater than that from the small laser beam spot in the L-DED and L-PBF processes (McAndrew et al. 2018). The processing accuracy of L-PBF is better than that of L-DED because the powder size, laser spot, and layer thickness used in the former are smaller than those in the latter (in the L-PBF and L-DED, powder: 10–50 and 100–150 μm (Singh, Kapil, and Das 2020); laser beam spot diameter: 50–80 and 1.0–4.0 μm (Luo et al. 2020); layer thickness: <100 μm and 0.25–2 mm (Gibson,
Rosen, and Stucker 2015, respectively). Consequently, the molten pool and heat-affected zones are smaller in the L-BPF. Compared with the above-mentioned methods, the advantages of L-PBF are attributed to its high processing accuracy and high degree of freedom in printing complex parts of different materials (e.g., metals, ceramics, and polymers) if the laser wavelength is appropriate (Pereira and Bártolo 2014; Rafiee, Farahani, and Therriault 2020).

This paper presents an overview of recent progress in MMAM via L-PBF, including dissimilar powder deposition mechanisms, molten pool behaviour, and materials used in multi-material L-PBF, applications, and scientific/technical challenges.

2. Material-spreading mechanisms in multi-material L-PBF

In the L-PBF process, unmelted powder is retained in the powder bed. Therefore, the deposition of at least two types of dissimilar powders across different powder layers or to the same powder layer is a technical challenge for realising multi-material L-PBF (Stichel et al. 2016). To date, several material-spreading methods have been proposed: blade-based, ultrasonic-based, electrophotographic-based, and ‘blade + ultrasonic’ hybrid powder spreading. The schematics of these methods are illustrated in Figure 1-a to Figure 1-d.

2.1. Blade-based dissimilar powder spreading

Researchers in Singapore (Pedersen 2013; Andriani 2014) modified the commonly used blade-based powder recoating device in the commercial L-PBF system. They used two powder containers to store and spread different powders (Figure 1a) to achieve bimetallic copper–stainless steel (SS) L-PBF processing with material changes in the vertical direction only, as shown in Figure 2-a. The two metals have sharp material interfaces. The abrupt transition between two materials is prone to high stress concentrations at the interfaces and can even lead to delamination under complex loading conditions (Mahamood and Akinlabi 2015). Such a problem is resolved using a gradient interface with a mixture of both materials known as FGM (Zhang et al. 2018). Scaramuccia et al. (2020) improved the blade-based dual-powder recoater design by adding a function of in situ powder mixing and realising L-PBF of a vertical Ti6Al4 V/In718 FGM structure. However, the above-mentioned blade-based powder spreading solution cannot achieve the objective of depositing dissimilar powders in the same build layer. To overcome this problem, the unmelted powder in the build layer should be cleaned up before spreading the second powder (Lappo, Jackson, and Wood 2003). Lappo, Jackson, and Wood (2003) introduced vacuum cleaning to remove all unmelted powder after powder laser melting and then utilised the powder recoater to spread the second type of powder. The disadvantage of this approach is that it may cause the cross-contamination of dissimilar powders in the same build layer. Wu et al. (2019) conducted a similar study in which a soft blade was used to clear away the unmelted powder. However, the repeated scraping of the sample surface may damage the soft blade tip (Guo, Ge, and Lin 2015), which can subsequently roughen the surface of
processed parts, leading to an uneven interface of dissimilar materials (Figure 2-b).

2.2. Ultrasonic-based dissimilar powder spreading

Ultrasound is a mechanical wave with a frequency exceeding 20 kHz that can be generated through a piezoelectric or magnetostriction oscillator (Vock et al. 2019). Ultrasonic waves can efficiently propagate in liquids and powders, and their application to the selective delivery of dry powders has been extensively researched (Kumar et al. 2004; Yang and Li 2003; Chian-rabutra, Mellor, and Yang). A schematic of a typical ultrasonic-based powder-dispensing device for dual-powder deposition is illustrated in Figure 1-b. The foregoing indicates that this method can feed powder at a uniform feed rate with a precision that can reach the micron-scale level. In 2008, researchers from the University of Manchester (Al-Jamal, Hinduja, and Li 2008) demonstrated the use of ultrasonic wave vibration to dispense multi-material dry powder particles to a powder bed following different geometrical patterns. They also laser printed two-dimensional (2D) multiple metallic material components, as shown in Figure 2-c. The factor restricting the use of this method for 3D printing is that the processing efficiency of point-by-point ultrasonic powder feeding is extremely low (Seidel and Anstätt 2016). Moreover, the uniformity of the powder layer thickness produced by this method requires further improvement (Wei, Gu, et al. 2019).

2.3. Electrophotographic-based dissimilar powder spreading

The working principle of electrophotographic powder deposition is similar to that of office laser printers. The latter uses static electricity to print toner particles onto paper following six steps: charging, exposing, developing, transferring, fusing, and cleaning (Schein 1988). Benning and Dalgarno (2018) described a belt-based electrophotographic system for the L-PBF of a single material. Eijk, Mugaas, and Karlsen (2014) reported a bimetallic copper–iron pattern-processed by the electrostatic powder deposition method. Aerosint SA developed a dual-toner cartridge-selective powder-spreading device (Figure 1-c) (Rafiee, Farahani, and Therriault 2020) and successfully employed it for depositing complex patterns comprising SS–copper alloy, as shown in Figure 2-d. The working principle of the device is based on using micro-airflow that can be controlled point by point to attract the powder particles onto a cylindrical mesh. Then they are blown off the mesh and deposited on the build platform to form a design pattern. Inappropriate parameters can cause the powder to fall unintentionally in certain areas, resulting in the contamination of the powder layer (Stichel et al. 2018; Stichel et al. 2020).

2.4. ‘Blade + ultrasonic’ hybrid method for dissimilar powder spreading

Researchers from the University of Manchester demonstrated a new multi-material L-PBF strategy that integrated a powder blade-assisted L-PBF system and ultrasonic powder dispensers (Figure 1-d). The powder blade spreads the powder constituting the main part of the component. Ultrasonic-assisted powder dispensers are used to deposit other types of powders that occupy a small volume of the component. Between these two steps, a micro-vacuum powder suction device was used to remove excess unmelted single-layer powders according to the working principle described by Glasschroeder, Prager, and Zäh (2015). This hybrid material deposition has been demonstrated to improve the powder deposition efficiency of ultrasonic-assisted L-PBF. A set of 3D bimetallic 316L–Cu10Sn samples, including a sphinx statue (Figure 2-i), was produced using the above experimental setup and verified the feasibility of this new processing strategy (Wei, Li, et al. 2018). To improve the stability of the powder flow rate of ultrasonic powder dispensing, Wei et al. (2020) added a miniature vibration motor to an ultrasonic powder dispenser that used the radial high-frequency vibration of the motor to loosen the tightly packed powder near the powder feeding nozzle. To print FGM components, Wei et al. (2019) integrated six ultrasonic powder dispensers to form a powder feed array. These were used to fabricate a 316L–Cu10Sn FGM turbine disk component with a spatial FGM material distribution (Figure 2-f). The researchers from the University of Manchester employed the same experimental setup to produce metal–glass (Figure 2-g) and metal–polymer (Figure 2-h) multi-material samples (Zhang et al. 2019; Chueh, Zhang, et al. 2020).

2.5. Combination of L-PBF with other manufacturing methods for printing multiple material components and comparison of different material deposition methods

The combination of L-PBF with other AM methods, such as L-DED (Gradl et al. 2019), fused deposition modelling (FDM), laser foil printing (Rittinghaus and Wilms 2020), stereolithography (SLA) (Silva, Felismina, et al. 2017), and cold spraying (CS) (Yin et al. 2018), can also
produce multi-material components. However, the integration of different AM methods can lead to a prolonged production cycle and severely restrict the design freedom of multi-material parts, consequently reducing the advantages of AM technology. In addition, by melting powders on substrates of different materials through L-PBF, bimetallic samples (consisting of solidified powder and substrate) can also be manufactured (Nguyen, Park, and Lee 2019; Tan, Zhou, and Kuang 2019; Wang et al. 2020).

The summary in Table 1 compares various powder deposition methods considering powder types, distribution dimensions, spreading efficiency, patterning resolution, and powder cross-contamination. A conventional powder recoater is used as a reference for comparison. The L-PBF can print complex...
geometrical structures because the unmelted powder material in the powder bed is used as a support material. Powder deposition methods, including ultrasonic-based and electrostatic-based methods, are fundamentally powder bed-based techniques. Therefore, theoretically, these powder-feeding methods can be applied to fabricate complex geometric structures composed of dissimilar materials. However, if the materials are only selectively deposited in regions to be melted, then, as the height of the printed part increases, gravity can cause the unmelted support powder in the proximity of the melting area contour to collapse. This is a severe problem when printing thin-walled features and of the melting area contour to collapse. This is a cause the unmelted support powder in the proximity the height of the printed part increases, gravity can selectively deposited in regions to be melted, then, as also fabricate 3D parts with randomly spaced materials.

3. Understanding molten pool behaviour in multi-material L-PBF by modelling and simulation

The modelling and simulations performed at Lawrence Livermore National Laboratory underpin the physics of complex melt flow and defect formation mechanisms in single-material L-PBF (Khairallah et al. 2016; Khairallah and Anderson 2014; Khairallah et al. 2020; King et al. 2015a). This section mainly reports the latest research status of the modelling and simulation of the molten pool behaviour of multi-material L-PBF. The simulations of L-PBF processes can be categorised into three: macroscopic, mesoscopic, and microscopic (Kyogoku and Ike-shoji 2020; Tan, Sing, and Yeong 2020). A limited number of studies are based on the macroscopic (Ban-dyopadhyay and Heer 2018) and microscopic methods (Mohanty and Hattel 2017) for multi-material L-PBF modelling. Most investigations in this field have been conducted on the mesoscopic scale. These simulations generally include two steps: discrete element modelling (DEM) and computational fluid dynamics (CFD) modelling. A typical integrated DEM–CFD flow for multi-material L-PBF is shown in Figure 3. In contrast to single-material L-PBF modelling, multi-material L-PBF modelling involves two or more materials, and different material physical parameters must be assigned to the corresponding powder particles on the same powder layer (Gu et al. 2020).

Table 1. Summary of multiple material deposition mechanisms in multi-material L-PBF.

| Powder-spreading mechanism | Powder types | Powder Distribution dimensions | Spreading efficiency | Patterning resolution | Powder cross-contamination | Refs. |
|-----------------------------|--------------|---------------------------------|----------------------|-----------------------|-----------------------------|-------|
| Conventional recoaters      | One material | 1D                              | Second-level for spreading one powder layer | Milli-scale            | No powder cross-contamination | Vock et al. 2019 |
| Blade-based                 | Two materials/compositions | 2D/3D                           | Second-level for spreading one powder layer | Milli-scale            | All unfused powders are cross-contaminated | Lappo, Jackson, and Wood 2003 |
| Ultrasonic-based            | Two materials/compositions | 2D/3D                           | Sub-minute level depending on dimensions of sliced cross-section | Micro-scale            | No cross-contamination if only one material is used as support | Chianrabutra 2015 |
| Electrostatic-based         | Two materials/compositions | 2D/3D                           | Sub-minute level depending on dimensions of sliced cross-section | Micro-scale            | No cross-contamination if only one material is used as support | Eijk, Mugaas, and Karlsen 2014 |
| Ultrasonic-blade hybrid     | Seven materials/compositions | 3D                              | Sub-minute level depending on dimensions of sliced cross-section | Micro-scale            | No cross-contamination if only one material is used as support | Wei et al. 2019 |

3D parts is also interesting. Furthermore, in hybrid AM processes, the material distribution is strictly limited by stepwise processing methods. Hence, it is challenging to fabricate 3D parts with randomly spaced materials.
The thermodynamic behaviour of the molten pool of L-PBF is considerably complex. As shown in Figure 4-a, the Marangoni convection and recoil force are the main driving forces for liquid flow (Khairallah et al. 2016). In 1982, Heiple and Roper proposed the Marangoni convection theory for describing the molten pool behaviour. They found that the temperature difference along the molten pool and its surface tension produced a driving force that stirred the pool causing the liquid to circulate (Dass and Moridi 2019). In multi-material L-PBF processes, the mixed flow of dissimilar elements in the molten pool can occur at the sharp material interface or in the FGM structure. Figure 4-b shows the microstructure of sharp material interfaces of bimetals with similar physical properties processed by L-PBF (Hengsbach et al. 2018). Figure 4-c shows the microstructure of the FGM of bimetals with significantly different physical properties (Wei, Gu, Li, et al. 2020). All of these two studies indicate that the Marangoni convection stirs the molten pool. The circular flow generated by the convection improved the distribution of elements in the solidification zone and modified the material properties. The elements inside the molten pool were first stirred and then redistributed along the boundary (Shakerin et al. 2019). The Marangoni convection-induced element rearrangement is also observed in the mesoscopic simulation of the L-PBF of Cu10Sn–Inconel718 (Sun, Chueh, and Li 2020), as shown in Figure 4-d.

The thermophysical properties and laser absorptivity of powder materials considerably affect the size of the molten pool during L-PBF (Guo et al. 2019). The material combination of high-thermal-conductivity copper alloy and high-melting-point iron–nickel alloys, has been well studied in multi-material L-PBF. This material combination can achieve high thermal conductivity, high operating temperature, and high corrosion resistance; hence, it has potential applications in the aerospace industry (Onuike, Heer, and Bandyopadhyay 2018). Mesoscopic numerical simulations (Wei, Gu, Li, et al. 2020; Sun, Chueh, and Li 2020; Gu et al. 2020) of Cu–Ni/SS FGM based on the coupled DEM–CFD modelling framework indicate that the diameter, depth, and temperature of the molten pool in multi-material L-PBF are considerably reduced as the copper alloy content in FGM increases, as shown in Figure 4-e. An experimental study of the L-PBF of Cu10Sn–Invar36 verified this phenomenon (Wei, Gu, Li, et al. 2020). Copper alloys have low laser absorption and high thermal conductivity; thus, increasing the copper alloy content in the FGM reduces the powder layer’s laser radiation absorption (Sun, Chueh, and Li 2020).

4. Materials and process characteristics in multi-material L-PBF investigations

The metal powder materials used in L-PBF include SS, iron-based alloys excluding SS, aluminium alloys, titanium alloys, cobalt-based alloys, copper alloys, nickel-based alloys, and other materials (Vock et al. 2019). The L-PBF of ceramics, glass, and polymers has also been investigated. As summarised in Table 2, the researchers combined Al, Cu, SS, Ti, Fe alloys with other materials to study the microstructure and mechanical properties of the multi-material processed by L-PBF.
Compared with the studies on multi-material L-DED that usually fabricate FGMs, the investigations on multi-material L-PBF typically bond two dissimilar materials directly rather than adopt an FGM transition structure. This is due to the technical difficulty of spreading multi-material powders in L-PBF. A sharp material interface may cause problems, such as cracks or embrittlement, which usually result from the lack of solubility of the elements in the two materials, mismatched lattice structure, changes in thermal expansion, and formation of thermodynamically stable brittle intermetallic phases (Sun and Karppi 1996). In contrast, the material composition of the FGM structure gradually changes, eliminating the distinct material boundary. This reduces the residual thermal stress concentration in the crack-sensitive area, thus avoiding crack propagation and material delamination. Hence, the FGM structure is more robust and has a longer fatigue life than sharp material interfaces (Li et al. 2020). Researchers from the University of Manchester demonstrated an ultrasonic-assisted L-PBF processing strategy that can fabricate FGMs through the L-PBF process, as elaborated below.

### 4.1. L-PBF of Al-involved bimetal

Wang et al. (2020) employed L-PBF to produce bimetallic samples with satisfactory metallurgical bonding composed of two aluminium alloys, Al–12Si and Al–Cu–Mg–Si. The Al2Cu phase appearing in the interface area interrupts the microstructure of eutectic Al–Si with its low microhardness. The research on L-PBF of AlSi10Mg–C18400 (Sing et al. 2015) found that Al2Cu is generated by the diffusion of Al and Cu elements. Owing to the refined grains caused by L-PBF, the tensile strength and elongation of the bimetallic samples in these two studies were higher than those of certain base metals. The fracture occurred at the base material and not at the material interface. Nguyen, Park, and Lee (2019) reported that the intermetallic layer thickness affects the bonding strength of Fe and Al, that is, the thicker the film, the lower the bonding strength. The thickness of the intermetallic layer increases with the laser energy density.

In multi-material L-PBF, the materials are directly melted, inevitably producing intermetallic compounds.
On the other hand, CS can deposit metals, metal matrix composite materials, and ceramics on various substrates without melting, thus avoiding residual stress, phase change, cracks, and thermal effects in the underlying substrate caused by material melting (Yin et al. 2017). In the study of Yin et al. (2018), Al and Al + Al2O3 were deposited on the L-PBF-processed Ti6Al4V parts through CS. As shown in Figure 5-a, no defects have been observed at the material interface.

4.2. L-PBF of Cu-involved bimetal (Cu–SS)

Wei et al. (2018) designed a micro-transverse finger-cross structure to increase the material contact area of the 316L–Cu10Sn sample (Figure 5-b), thereby increasing the interface bonding strength. The authors found pores and cracks in the Cu10Sn zone (Figure 5-b). This was because the powder discharged out of the nozzle of the ultrasonic powder dispenser freely fell on the powder bed with no external compacting force, rendering the powder layer loose and highly porous (Wei, Gu, et al. 2019). Applying an external force to compress the ultrasonic-deposited powder layer effectively reduced porosity (Figure 5-c). Using an ultrasonic powder dispensing method, Wei et al. (2019) fabricated a horizontal 316L–Cu10Sn FGM sample (Figure 5-d), which was difficult to manufacture using conventional L-DED and L-PBF. The metallurgical bonding between the two powders was found satisfactory despite their different composition ratios, (Figure 5-d); however, it was reported that the 316L powder, which has a high melting point, possibly lacked fusion. The reasons for this phenomenon, which are the same as those that caused the lack of fusion in the L-PBF of Cu–Fe, are discussed in later sections.

Some researchers (Liu et al. 2014; Chen et al. 2020; J. Chen et al. 2019) studied the microstructure and mechanical properties of L-PBF-processed 316L–C18400–Cu10Sn samples. The results showed that the bimetal tensile strength and elongation were between those of the two base metals. The microstructure of the cross-section of the bimetallic samples had element diffusion zones at the material interface, which aided in improving the bonding strength. It is noteworthy that these investigations found no brittle intermetallic phases formed in the parts manufactured by L-PBF. By contrast, microcracks were observed on the 316L side at the material interface (Figure 5-e and -f) but not on the copper alloy side. This is considered as a typical liquid metal embrittlement (LME) defect resulting from the loss of ductility and subsequent embrittlement of solid metals after coming into contact with a specific liquid metal. This phenomenon
Figure 5. a) Al–Ti6Al4V bimetal sample manufactured by L-PBF + CS and the microstructure of its interface (Yin et al. 2018), b) 316L-Cu10Sn bimetallic sample with a finger-cross interlock structure and its microstructure at the material interface (Wei, Li, et al. 2018), c) comparison of fused powder layers with and without powder compression (Wei, Gu, et al. 2019), d) horizontal 316L–Cu10Sn FGM sample and the microstructure at different composition zones (Wei, Sun, et al. 2019), e) microcracks at 316L–C18400 interface (Liu et al. 2014), f) microcracks at 316L–Cu10Sn interface (Chen et al. 2020), g) delamination of fused Invar36–Cu10Sn powder layer (Wei, Gu, Li, Sun, Chueh, et al. 2020), h) interface microstructure of W–Cu (Tan, Zhou, and Kuang 2019), i) micro-interface structure at SS–PET interface (Chueh, Wei, et al. 2020), j) microstructure of cross-section of Cu10Sn–PA11 FGM (Chueh et al. 2020), k) microstructure of cross-section of Cu10Sn–glass FGM (Zhang et al. 2020).
has been investigated in-depth in the study of welding dissimilar metals (Chen et al. 2013). The formation mechanism of the LME defect is elaborated in the section ‘Discussions and challenges in multi-material L-PBF’.

4.3. L-PBF of Cu-involved bimetal (Cu–Fe)

A number of researchers (Al-Jamal, Hinduja, and Li 2008; Beal et al. 2006; Beal et al. 2004) have investigated the L-PBF of Cu–H13. The satisfactory mutual diffusion of Cu and Fe mutually has been reported. The tensile strength of the Cu-H13 sample was observed to be between those of Cu and H13. Cracks were observed owing to thermal stress and rapid cooling during solidification. Anstaett et al. (2017) reported that the deposition sequence of copper alloy and tool steel affected the crack formation caused by the thermal expansion mismatch between the two.

The numerical simulation of the L-PBF of Ni–Cu performed by Sun, Chueh, and Li (2020) showed that the high-melting-point nickel alloy powder could not reach its melting temperature; unmelted nickel alloy powders were blended with liquid copper alloy with a low melting point; however, they did not provide further theoretical analysis to verify this work. In a study of the L-PBF of Cu–W, a similar phenomenon was reported (Tan, Zhou, and Kuang 2019): high-melting-point W particles fell into the pool molten Cu without melting. During L-PBF, most of the energy of incident laser beams was absorbed by the powder on the layer surface, whereas the powder at the bottom layer only absorbed an extremely small amount of energy (Boley, Khairallah, and Rubenchik 2014). This phenomenon was verified by the investigation of the L-PBF of bimetallic Invar36–Cu10Sn (Wei, Gu, Li, et al. 2020).

As observed, the energy density that could initially melt the high-melting-point pure metal (Invar36) was unable to melt the Invar36–Cu10Sn mixture. The solidified powder layer exhibited distinct delamination (Figure 5-g). The copper alloy was completely melted, and the high-melting-point Invar36 particles were melted when they were on top of the powder layer. Moreover, the unmelted Invar36 shielded by the Cu10Sn powder was embedded in melted Cu10Sn. The authors reported that this was due to the higher thermal conductivity as well as the lower laser absorptivity and melting point of copper alloy. When the copper powder partially covered the upper surface of the powder layer, it considerably reflected the laser beam. However, because copper had high thermal conductivity, the absorbed energy quickly dissipated. Consequently, the Invar36 powder located under the copper powder cannot obtain sufficient energy to melt. This problem can be resolved by increasing the energy density.

4.4. L-PBF of SS-involved bimetals

Mohd Yusuf et al. (2021) used L-PBF to manufacture 316 L–In718 bimetallic samples. The samples had low porosity and no cracks, and the material interface exhibited satisfactory metallurgical bonding. The two types of alloy crystal structures in the solid and molten states have the same single-sided centre cubic crystal structures; and there is no allotropic phase transition (Hinojos et al. 2016). The thermal expansion coefficients of the two materials are also similar (Hinojos et al. 2016). The main components of the two alloys, including Fe, Cr, and Ni, have satisfactory solubilities (Croll and Wallwork 1969). These three factors make L-PBF suitable for processing Ni–SS FGMs (Carroll et al. 2016). Hengsbach et al. (2018) reported that the Marangoni convection effect determined the solidification of the fused 316L–H13 mixture microstructure during the L-PBF of the 316L–H13 bimetal.

The Ti and Fe elements in Ti6Al4 V and 316 L, respectively, synthesise harmful brittle intermetallic phases, such as TiFe and TiFe2, during laser melting, thereby decreasing the bonding strength and cracks (Yin et al. 2018; Bobbio, Otis, Borgonia, et al. 2017). Accordingly, Tey et al. (2020) presented an L-PBF-manufactured Ti6Al4 V/316L bimetallic sample with a 0.5 mm CuA copper alloy interlayer. They found that the Ti6Al4V–CuA interface containing three detrimental phases (i.e. L21 ordered phase, amorphous phase, and Ti3Cu) was the critical interface that affected the mechanical strength of the entire component. The non-homogeneity of the melt pool results in tough reinforcement phases within a relatively brittle matrix, thus controlling the laser energy input. This aids in increasing the proportion of the relatively tougher interfacial α-Ti phase and decreasing other brittle phases.

4.5. L-PBF of Ti alloy-involved bimetals

Both Ti5Al2.5Sn and Ti6Al4 V exhibit satisfactory oxidation resistance as well as excellent metallurgical compatibility and weldability; accordingly, they can be easily connected by welding methods, such as laser welding (Wang, Liu, and Chen 2013). Wei et al. (2020) proved that L-PBF was also suitable for processing complex parts composed of these two materials. A narrow defect-free metallurgical bonding interface between the Ti5Al2.5Sn and Ti6Al4 V layers where the elements diffused each other was observed. The interface bonding strength exceeded that of the Ti5Al2.5Sn layers.
The Ti and Ni elements in Ti6Al4 V and In718, respectively, are metallurgically incompatible and produce brittle intermetallic phases, such as TiNi3 and Ti2Ni (Chatrerjee, Abinandanan, and Chattopadhyay 2006). Scaramuccia et al. (2020) reported that cracks caused by Ti2Ni were widely observed in the Ti6Al4V–In718 FGM manufactured by L-PBF when the In718 content exceeded 20 wt%.

### 4.6. L-PBF of metal–polymer components

Chueh et al. (2020) combined two AM methods, L-PPF and FDM, for processing metal–polymer (SS–PET) parts. The specially designed interlock structure (Figure 5-i) allowed PET to form interlocking structures with SS. The SS–PET joint exhibited satisfactory shear and tensile strengths. In another study, Chueh et al. (2020) designed a unique dual-vibration powder deposition device to dispense light and low-fluidity polymer powder (PA11). Using this device they successfully manufactured a Cu10Sn–PA11 FGM sample (Figure 5-j) and further studied the interaction between metal and polymer during laser melting (Chueh, Zhang, et al. 2020). They indicated that because of the significant difference between the melting points of the two materials, direct contact of the two must be avoided in the design of metal–polymer parts to prevent the polymer from evaporating due to the high melting temperature of metal. They also found that adding a small amount of PA11 (5 vol%) to Cu10Sn powder significantly improved the surface quality of L-PBF-processed Cu10Sn.

### 4.7. L-PBF of metal-ceramic/glass components

Metal–ceramic composite materials can considerably improve the surface hardness, wear resistance, chemical inertness, and temperature stability of parts (Hu and Cong 2018). Trenke, Müller, and Rolshofen (2006) proposed three types of possible metal–ceramic powder combinations during L-PBF: layered metal–ceramic, metal–ceramic powder composite, and ceramic coating on fused metal layers, which could all be produced via L-PBF. Koopmann, Voigt, and Niendorf (2019) found that molten ceramics could not completely bond to the steel surface. However, the bonding strength improved after applying a re-melting strategy, which resulted in a jagged interlocking microstructure at the metal–ceramic interface.

Boride-reinforced Ti matrix composites are promising materials for aerospace applications. For instance, the AM of composition-graded turbine blades employs such materials. Wang et al. (2019) studied L-PBF-processed TiB2–Ti6Al4V-graded materials and observed the nano-hardness gradient of TiB2–TiB–Ti6Al4 V at the TiB2–Ti6Al4 V interface produced by L-PBF. The TiB phase formation results from the in situ reaction between Ti and TiB2 triggered by high-power laser irradiation (Attar et al. 2014). This phenomenon was also reported in a similar study by Shishkovsky, Kakovkina, and Sherbakov (2017). They observed two types of microstructure heterogeneity, including the unmelted TiB2 particles at the interlayer interfaces and element chemical segregation at the track boundaries. Wang et al. (2020) investigated the formation mechanism of TiB2 tracks on Ti6Al4 V alloy during L-PBF. They reported that the optimum processing parameter window for creating a sound track was in the laser power range of 400–450 W and laser energy density of 0.67–1.50 J/mm.

Zhang et al. (2019) presented L-PBF-manufactured metal–soda–lime glass pendants and achieved satisfactory mechanical bonding at the material interfaces. The authors also reported defects, including insufficient melting of glass particles in heat-affected zones (HAZs) and cracks in fused glass layers. In another investigation (Zhang et al. 2020), metal–glass FGM samples were presented. The cross-sections of these samples are shown in Figure 5k. No element diffusion is observed at the material interface, and the fracture occurred at the interface close to the ceramic matrix composite zone.

All processing conditions of the above investigations on multi-material L-PBF are presented in Table 3.

Regardless of whether the material properties are compatible, the interface between the two materials processed in L-PBF has been observed to usually exhibit a satisfactory metallurgical bond. It is worth noting that for bimetallic L-PBF, defects (including intermetallic phases, LME, microcracks, and lack of fusion) can easily occur if the lattice parameters of the two metals do not match, the elements are not compatible, or the melting points considerably differ. The defects identified thus far are summarised in Table 4. These L-PBF-specific defects can further lead to other typical L-PBF defects, such as porosity, balling, residual stress-induced cracks, and delamination (Grasso and Colosimo 2017), eventually degrading the mechanical properties of the material. The mechanisms that cause these defects and their corresponding solutions are discussed in the next section.

### 5. Discussions and challenges in multi-material L-PBF

#### 5.1. Material science challenges

Although L-PBF can be employed to manufacture bi-materials and FGM structures, it remains confronted with challenges in materials science, including
understanding the phenomena of brittle intermetallic phase formations, LME, lack of fusion, and element segregation. These defects are likely to occur if the crystal lattices of the two materials do not match, the elements lack solubility, or the melting points and densities of the two materials considerably differ. The formation

Table 3. Summary of processing conditions in multi-material L-PBF investigations.

| Combination | Materials | Laser power (W) | Hatch distance (um) | Scanning speed (mm/s) | Layer thickness (um) | Refs |
|-------------|-----------|----------------|--------------------|----------------------|---------------------|------|
| Al–Al       | Al12Si powder, Al3.5Cu1.5Mg15Si, Al12Si powder | 320 | 110 | 1455 | 50 | Wang et al. 2020 |
| Al–Cu       | Al12Si10Mg powder, Cu1240 powder | 350 | 170 | 1140 | 50 | Sing et al. 2015 |
| Al–Ti       | Ti6Al4 V powder, cold-sprayed aluminium | 260 | 30 | 1400 | 50 | Yin et al. 2018 |
| Al–Fe       | Pure Al powder, SS 316L plate, Cu10Sn powder, | / | / | / | / | Nguyen, Park, and Lee 2019 |
| Cu–SS       | Cu10Sn powder | 125 | 140 | 150 | 50 | Wei, Li, et al. 2018 |
| Cu–Fe       | Cu powder, H13 powder, CuSn powder, Cu10Sn powder, In718 powder | 200 | 250 | 130 | 100 | Al-Jamal, Hinduja, and Li 2008 |
| Cu–W        | T2 copper plate, Pure W powder | / | / | / | / | Tan, Zhou, and Kuang 2019 |
| SS–Ti       | 316L powder, K220 Cu powder, Ti6Al4 V powder, In718 powder | 380 | 40 | 1750 | 50 | Tey et al. 2020 |
| SS–Ni       | 316L powder, Ti6Al4 V powder, In718 powder | 370 | 90 | 400 | 530 | 30 | Mohd Yusuf et al. 2021 |
| SS–Fe       | 316L powder, H13 powder, Cu10Sn powder, Invar36 powder | 400 | 75 | 300 | 500 | 50 | Beal et al. 2005 |
| Ti–Ni       | Ti6Al4V powder, Cu10Sn powder, Ti6Al4 V powder, Ti6Al4 V powder | 200 | 80 | 1000 | 20 | Wei et al. 2020 |
| Ti–Fe       | Ti6Al4V powder, In718 powder, MS1 powder, H13 plate | 250–400 | 50–100 | 500–1900 | 50 | Shakerin, Hadadzadeh, Amirkhiz, et al. 2019 |
| Fe–Fe       | MS1 powder, H13 plate, Cu10Sn powder, PET, PLA, Steel, Cr3C2, WC | 285 | 110 | 960 | 40 | Koopmann, Voigt, and Niendorf 2006 |
| Metal–Polymer | Cu10Sn powder, PET, PLA, Cu10Sn powder, PA11 powder, PA11 powder, Cu10Sn powder | 125 | 140 | 800 | 50 | Chueh, Wei, et al. 2020 |
| Metal–SS-glass | 316L powder, Steel, Cr3C2 | 100, 125 | 100 | 150, 200 | 50 | Chueh, Zhang, et al. 2020 |
| Metal–Ti-ceramic | Ti6Al4 V powder, TiB2 powder | 170 | 35 | 800 | / | Zhang et al. 2019 |
| Metal–Fe–ceramic | Ti6Al4 V powder, TiB2 powder, Ti6Al4 V powder | 300–450 | 50 | 400–1000 | 50 | Chen et al. 2019 |

- LME: Liquefied Metal Ejection
- MS1: Materials Science 1
- X38CrMoV5-3: X38CrMoV5-3 steel
mechanisms and solutions for these defects are discussed below.

5.1.1. Brittle intermetallic phases

In the melting process of dissimilar metals in L-PBF, the elements in the metals diffuse and combine according to a certain number of atoms. This may result in new phases (i.e. intermetallic phases) whose lattice type is different from that of the base material. Such new phases can easily form if the electronic layer structure, atomic radius, and crystal type of the elements comprising the alloy relatively differ. These intermetallic phases are usually considerably brittle as well as have high hardness values and melting points, which induce defects, such as cracks, in the solidified components (Stoloff and Sikka 1996). Three techniques may be applied to overcome the problems caused by intermetallic phases.

(1) Calculation of phase diagram (CALPHAD)

An important feature that has a considerable effect on the physical properties of inorganic material is phase. The phase diagram, consisting of experimental measurements and statistical thermodynamic analysis, provides essential information for understanding the properties of a material at various temperatures and compositions. The phase diagram calculation based on experiments and thermodynamic analysis is usually called CALPHAD (Ohtani 2006), which was first introduced by Kaufman and Bernstein (Xu et al. 2016).

The CALPHAD method establishes a thermodynamic model based on the crystal structure of each component phase (i.e. gas phase, liquid phase, solid solution, and compound). The Gibbs free energy of each phase in a material system can be determined by evaluating and screening the experimental and theoretical calculation data (from first-principles calculations, statistical methods and experience, and semi-empirical formulas) of the multi-material system under certain temperature and pressure. Fitting and optimising the model parameters are essential during this step. Finally, a thermodynamic database of multi-component material systems is established using CALPHAD (Ohtani 2006). Figure 6 shows the flow of the CALPHAD method. CALPHAD is a useful thermodynamic calculation method that can be employed to determine the thermodynamic properties of multi-component systems. Moreover, it is the thermodynamic basis of material dynamics and microstructure evolution simulation. Accordingly, the CALPHAD method is widely used in the research and development of new materials and processes (Hofmann et al. 2014).

The CALPHAD method is a useful technique for showing the driving forces of intermetallic phases, kinetics of precipitation nucleation, and solute segregation/partitioning during the solidification stage of L-DED and L-PBF manufacturing of FGM. The correct selection of phase equilibrium calculation (Bobbio, Otis, Paul, et al. 2017) or non-equilibrium thermodynamic calculation (Liu et al. 2020) can provide critical information for accurately predicting the phase formation of FGMs in L-DED. This method, which has been widely used in the study of L-DED of FGM, is worthy of consideration.

Table 4. Summary of defects in multi-material L-PBF investigations.

| Material combination | Intermetallic phases | LME | Lack of fusion | Refs |
|----------------------|----------------------|-----|----------------|------|
| Al-Al                | $Al_2Cu$             |     |                | Wang et al. 2020 |
| Al-Cu                | $Al_2Cu$             |     |                | Sing et al. 2015 |
| Al-Ti                |                      |     |                | Yin et al. 2018 |
| (CS + L-PBF)         |                      |     |                |                  |
| Al-Fe                |                      |     |                | Nguyen, Park, and Lee 2019 |
| Cu-Si                | $Fe_3Al_5$           | ●   | ●              | Wei, Li et al. 2018; Wei, Sun et al. 2019; Liu et al. 2014; Chen et al. 2020; Chen et al. 2019; Gu et al. 2020; Bai et al. 2020 |
| Cu-Ni                |                      |     |                | Sun, Cheueh, and Li 2020 |
| Cu-Fe                |                      |     |                | Al-Jamal, Hinduja, and Li 2008; Anstaett et al. 2017; Zhang et al. 2019; Beal et al. 2006; Beal et al. 2004; Wei, Gu, Li, et al. 2020 |
| Cu-W                 |                      |     |                | Tan, Zhou, and Kuang 2019 |
| SS-Fe                | Ti$_3$Cu             |     |                | Tey et al. 2020 |
| SS-Ni                |                      |     |                | Mohd Yusuf et al. 2021 |
| SS-Fe                |                      |     |                | Hengsbach et al. 2018 |
| Ti-Cr                |                      |     |                | Wei et al. 2020 |
| Ti-Ni                |                      |     |                | Scaramuccia et al. 2020 |
| Fe-Fe                |                      |     |                | Shakerin, Hadadzadeh, Amirkhiz, et al. 2019 |
| Metal                |                      |     |                | Chueh, Wei, et al. 2020; Chueh, Zhang, et al. 2020 |
| Metal                |                      |     |                |                  |
| SS-glass             |                      |     |                | Wei, Gu, Zhang, et al. 2020; Zhang et al. 2019 |
| Ti-ceramic           |                      |     |                | Wang et al. 2019; Shishkovsky, Kakovkina, and Sherbakov 2017; Bai et al. 2020; Chen et al. 2019 |
| Fe-ceramic           |                      |     |                | Trenke, Müller, and Rolshofen 2006; Koopmann, Voigt, and Niendorf 2019 |
for application in the study of multi-material L-PBF. The formation of intermetallic phases is directly related to the composition ratio of materials. Therefore, in FGMs, the transition path from material A to B can be artificially designed based on the results of thermodynamic calculations. This makes it possible to skip the material ratio range that produces intermetallic phases, thereby fundamentally avoiding the generation of these harmful phases (Reichardt 2017). The cooling rate has an essential influence on the phase transition and grain growth during the AM of FGM (Bobbio, Otis, Paul, et al. 2017). Hence, the CALPHAD method can be used to predict the phase formation during processing based on the phase diagram, and the processing temperature can be controlled in the ideal range to avoid the appearance of secondary phases (Bobbio et al. 2018).

(2) Addition of transition layer between two base materials
Another practical solution to brittle intermetallic phase problems is the addition of a transition layer composed of elements with satisfactory metallurgical compatibility related to the two base materials (Atasoy and Kahraman 2008; Wang et al. 2010). The transition layer prevents the direct contact between the two base materials, thereby ultimately precluding the formation of intermetallic compounds.

(3) Third metal element addition to increase material ductility
The addition of a small amount of a third metal element to the multi-material L-PBF process is a potential solution for increasing the ductility of intermetallic phases. For example, by adding 0.02–0.05 wt% of B to brittle polycrystalline Ni3Al, the room temperature tensile elongation of Ni3Al can be increased from approximately 0–40%–50% (Aoki and Izumi 1977). By partially replacing Co in the Co3V intermetallic phase with Ni and Fe, the crystal lattice of Co3V can change from the non-plastic hexagonal structure (D019) to the plastic face-centred cubic structure (L12) (Chen, Matsuo, and Tatara 2000).

5.1.2. LME
The LME defect is essentially a type of stress corrosion cracking, which can be described by Galvele’s atomic surface mobility (ASM) model, as shown in Figure 7. According to this model, the crack propagates through the diffusion of metal and ions from the crack tip to the crack wall surface (Popov 2015). Fredriksson, Hansson, and Olsson (2001) proposed the formation mechanism of LME-induced cracks in an Fe–Cu system. During the laser melting process, the liquefied Cu diffuses into the iron grains, inducing the Kirkendall effect. This causes the vacancies to diffuse to their

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Figure 6. Flowchart of CALPHAD (Xu et al. 2016).

Figure 7. ASM model (Galvele 1987).
condensed grain boundaries, forming grain boundary cracks that are filled with liquefied Cu. The condensation of vacancies and the change in the free energy of the surface/interface provide the driving force of the liquid Cu element to penetrate the iron grain boundaries. The two prerequisites for the appearance of LME are as follows: the miscibility between the solid and liquid metals is low, and there are no intermetallic phases between the solid and liquid couple (OLD 1980). Different from ferritic SSs and nickel alloys, austenitic SSs, such as 316L, are particularly prone to this type of cracking when they encounter liquefied copper alloys (Fredriksson, Hansson, and Olsson 2001). Huang et al. (2019) reported that as long as the base metal temperature is higher than the melting point of copper, LME-induced cracks would continue to propagate.

5.1.3. Lack of fusion and element segregation
To form a 3D component, L-PBF uses the heat provided by the laser beam to melt powder particles point by point, line by line, and layer by layer (Zhang, Li, and Bai 2017). The laser energy density ($E_d$) in L-PBF is expressed by the following.

$$E_d = \frac{\text{Laser power}}{\text{hatch distance} \times \text{layer thickness} \times \text{scanning speed}}$$

If the laser energy density is insufficient, then the width or depth of the adjacent molten pool may not be adequate such that the overlap of molten pools becomes insufficient. Moreover, defects due to the lack of fusion, including pores and unmelted particles, can appear at the junction of solid/liquid interface in the powder bed, as illustrated in Figure 8-a (Tang, Pistorius, and Beuth 2017). The powder particle diameter range for L-PBF is usually 15–45 μm (Gaussian distribution with an average diameter of approximately 20 μm), and the powder bed layer thickness typically exceeds 30 μm. Thus, the powder layer can be stacked with numerous powder particles whose diameter is less than the powder layer thickness, as shown in Figure 8-b1. The multiple scattering among powder particles (Figure 8-b1 and b2) can cause the laser absorptivity of the powder to exceed that of the flat solid bulk metal. However, most of the laser energy is absorbed by the top-layer powder followed by the subsurface-layer powder. Only approximately 1% of the energy is absorbed by the substrate/pre-melted powder layer particles (King et al. 2015b). In the multi-material L-PBF process, the surface and subsurface powder in the same layer may be composed of different materials. The difference in their melting point and laser absorptivity can easily lead to defects caused by the lack of fusion.

The considerable difference between the melting points of the two materials causes the evaporation of the material with a low melting point during the melting process. In contrast, the material with a high melting point insufficiently melts. This can cause defects, such as element segregation (Zhang et al. 2018), pores (Liu et al. 2015), and powder infusibility (Brueckner et al. 2018), which can eventually reduce the mechanical properties of processed parts (Taheri Andani et al. 2017). Element segregation can be reduced by homogenisation heat treatment (Simonelli et al. 2018). Moreover, the simultaneous use of laser power in the medium range and high scanning speed can suppress the evaporation of metal elements with low melting point and reduce segregation (Vora et al. 2015). Researchers attempted to reduce the defects caused by the lack of fusion by optimising laser

![Figure 8.](image)
processing parameters (e.g. scanning speed (Kuo et al. 2020; Huang et al. 2021), hatch distance (Kuo et al. 2020), and laser power (Huang et al. 2021), optimising scanning strategies (Nadammal et al. 2021; Sing and Yeong 2020), and using a top-hat laser profile (Huang et al. 2021)). If the melting points of the two materials are similar, the above measures are suitable for suppressing the occurrence of defects due to the lack of fusion. However, if the difference between their melting points is extremely large, the effectiveness of the above strategy may be limited (Wei, Gu, Li, Sun, Liu, et al. 2020).

The surface roughness of L-PBF-printed parts is usually worse than that produced by traditional machining (Snyder and Thole 2020). Accordingly, researchers have attempted to improve the surface quality of L-PBF parts by the optimisation of processing parameters (Yu et al. 2019) and simulations (Khorasani et al. 2020). For multi-materials, poor surface quality at the material interface may be an advantage because it can increase the contact area between the two materials, thereby increasing their metallurgical bonding strength. For instance, a high laser power remelting strategy was implemented to improve the bonding strength between a ceramic powder layer and steel substrate surface (Koopmann, Voigt, and Niendorf 2019). The circular flow caused by the Marangoni convection in the molten pool thoroughly mixed the two materials, resulting in a jagged interlocking microstructure in the ceramic-steel. Researchers from the University of Manchester designed a macro-mechanical interlocking structure at the metal–polymer interface of multi-material components produced by L-PBF. Such a structure is favourable for the penetration of melted polymer into the metallic part (Chueh et al. 2020).

Considerable differences in the density of various powder mixture elements and viscosity of liquids can also lead to element segregation (including the separation of element-rich areas, and insufficient diffusion) (Yadroitsev, Krakhmalev, and Yadroitseva 2017)). Elements with a higher material mass density sink to the bottom of the molten pool and gather near the molten pool boundary. In the FGM design, high-density and low-density materials should be placed at the bottom and top, respectively. Component segregation is also suppressed by the smooth material gradient transition (Loh et al. 2018). Heat treatment can fine-tune the phase distribution and transition as well as release the internal stress, thereby improving the mechanical properties (Aboulkhair et al. 2016). To avoid elemental segregation, in situ alloying using L-PBF may be an alternative solution. This method uses a laser beam to fuse and form an alloy from a mixture of elemental powders directly (Brodie et al. 2020). It affords flexibility and yields high-throughput productivity for the study of new materials, such as particle-reinforced metal matrix composites (Khmyrov, Safronov, and Gusarov 2017), high-entropy alloys (Sarswat et al. 2019), and intermetallic compounds (Grigoriev et al. 2017). Compared with pre-alloyed powder, the powder mixing method affords many advantages, such as flexibility in elemental composition, high controllability of powder particle size, low cost, and time efficiency. Despite these apparent advantages, the L-PBF in situ alloying method also has drawbacks, such as poor repeatability, unmelted powders, and highly heterogeneous regions in specimens (Krakhmalev et al. 2017). The homogeneity of the microstructure processed by L-PBF mainly depends on the particle size of the mixed powders and the mixing methods (Qiu et al. 2016).

A low laser absorptivity in one of the metals in the metal powder mixture can result in the lack of fusion in the multi-material L-PBF process. The formation mechanism of this defect is elaborated in Section 4.3. Oliveira, Lalonde, and Ma (2020) proposed the reduction in the hatch distance or layer thickness to increase the laser beam penetration depth and suppress the defects caused by the lack of fusion.

5.1.4. Challenges in L-PBF of metal–ceramic components
Ceramic is a solid compound composed of metallic and non-metallic elements. It is similar to metals with crystal grain aggregates, crystal grains, and grain boundaries. However, they are fundamentally different from metals: they do not contain a large number of free electrons and are merged through ionic bonding, covalent bonding, or a highly stable combination of these two. Hence, the processing of metal–ceramic composite powders through L-PBF is challenging because crystal structures of metal and ceramic are different, resulting in considerably different melting points. The strong bonds of ceramic crystals can make elemental diffusion extremely difficult. Their thermal expansion coefficients considerably differ, resulting in significant thermal stresses at the joints and cracks on the ceramic side. The brittle and glass phases on the bonding surface weaken the ceramic performance (Bengisu 2001).

Researchers have successfully used ultrafast lasers in welding ceramics (Penilla et al. 2019) and glass (Carter et al. 2017). Therefore, the use of an ultrafast laser as the energy input of L-PBF may be useful to realise the direct printing of metal–ceramic parts. In addition, combining the AM method with other processing methods is also a solution for processing metal–ceramic materials. For example, the combination of L-PBF and CS is
applied to bond Ti alloy and Al₂O₃ ceramic (Yin et al. 2018).

5.1.5. Challenges in L-PBF of metal–polymer components
The challenge for L-PBF of hybrid metal/polymer mixture is how to avoid decomposition and gasification of low melting point polymer powders. The resulting fumes adversely affect the molten pool of the metal. Among the strategies for resolving this is the use of metal and polymer powders with melting points that are as similar as possible (Amoabeng and Velankar 2018). Moreover, the metal and polymer parts are set at a certain distance in the component design to avoid the thermal decomposition of the polymer due to the high temperature in the HAZ of the molten metal (Chueh, Zhang, et al. 2020). Integrating L-PBF with other AM methods is also an alternative solution. For instance, L-PBF and SLA may be combined for SS–polymer binding (Silva, Felismina, et al. 2017). This hybrid AM processing can be implemented step-by-step. First, L-PBF is used to print the metal part with a high-melting-point. Then, the semi-finished product is placed in the SLA equipment for printing the polymers with a low-melting-point. The disadvantage of this approach is that the processing freedom of SLA may be limited by the pre-formed metal parts.

5.2. Other technical challenges
5.2.2. Modelling and simulation challenges
Based on the foregoing review, the present modelling of multi-material L-PBF processes have been observed to usually involve the thermodynamic simulation at the mesoscopic scale, revealing the melt pool development in the multi-material L-PBF process. Microscopic scale modelling and simulation are practical tools for understanding the elements involved in grain growth process, such as element diffusion, phase change, grain density, and dendrite size, orientation, and morphology. However, at present, research works on the microscopic modelling of multi-material L-PBF with a complex heat history are considerably limited. Most of the current microscopic modelling methods (e.g. phase-field modelling, cellular automata modelling, and kinetic Monte Carlo modelling) are mainly developed for binary alloys. These methods are difficult to apply for predicting the growth of dendrites and grain structures with a mixture of different powders/compositions inside the molten pool.

All microscopic modelling methods require thermal history information (including temperature, cooling rate, and temperature gradient) obtained via macroscopic or mesoscopic modelling. Therefore, the accurate prediction of the thermal history of multi-material L-PBF is essential. A significant uncertainty in obtaining an accurate thermal history is associated with the material properties in the high-temperature range. These properties include material density, specific heat, surface tension coefficient, viscosity, laser absorptivity, thermal conductivity, and latent heat for material melting and vaporisation. Currently, experimental data on the universal physical properties of mixed materials are insufficient. This hinders calibration modelling for accurately predicting the results of multi-material L-PBF at both the mesoscopic and microscopic scales.
5.2.3. Challenges in experimental method
Defects easily form in laser-based AM processes; for multi-material L-PBFs, defect control is even more challenging. Each material composition may require an optimal process parameter in a multi-material part particularly if such a part is of the FGM structure. Therefore, the experimental workload based on conventional trial-and-error testing methods and simple orthogonal experiments may substantially increase, undoubtedly substantially increasing the time devoted to research and economic costs. The statistical design of experiments and artificial intelligence prediction methods can be useful in deriving the best process parameters and reducing the volume of actual physical experiments (Goh, Sing, and Yeong 2021; Johnson et al. 2020; Meng et al. 2020). Rankouhi et al. reported the use of machine learning to optimise the processing parameters of L-PBF of the 316L–Cu composite (Rankouhi et al. 2021). Furthermore, many researchers (Okaro et al. 2019; Gobert et al. 2018) have implemented machine learning algorithms to analyse the molten pool information collected by co-axis/off-axis sensing technology (e.g. pyrometers, high-speed cameras, and infrared cameras) for defect identification and classification. This type of technology also aids in improving printing quality and research efficiency.

5.2.4. Production efficiency
Compared with AM methods, such as L-DED, where the material is deposited and melted simultaneously, L-PBF requires material deposition before melting, leading to inherently low deposition efficiency. Although multi-material spreading methods solve the problem of multi-material spatial distribution, they render the powder spreading process in multi-material L-PBF to be more time-consuming, consequently lowering the production efficiency. Developing a highly efficient and high-quality dissimilar powder spreading device based on the new powder spreading mechanisms reported in this paper is a prerequisite for the industrial application and commercialisation of multi-material L-PBF.

5.2.5. Challenges in multi-material component design software
Conventional single-material parts can be modelled with the usual computer-aided design (CAD) software, which only requires geometry as input. For multi-material parts, their different material properties, spatial distributions, and geometric shapes should be defined. The 3D CAD software based on voxel modelling can realise this function (Doubrovski et al. 2015); several related software tools such as ParaMatters and Monolith are available on the market. However, these tools are typically used for the MMAM of polymer materials; whether they are suitable for the MMAM of other materials in L-PBF processes is not certain.

6. Potential applications of multi-material L-PBF
Multi-material L-PBF applications usually combine the advantages of the physical characteristics of different materials into one part to derive a special function that is difficult to achieve using traditional processing methods. Previous investigations on multi-material L-PBF applications are summarised in Table 5.

7. Conclusion
In this paper, a review of the latest research progress on multi-material L-PBF is presented. Potential challenges and applications are also discussed.

Multi-material L-PBF has been enabled by the development of selective powder deposition technology, and researchers have demonstrated a series of multi-material samples processed using this method. Usually, the multi-metallic samples processed by L-PBF exhibit satisfactory metallurgical bonding at the material

| Table 5. Investigations on applications of multi-materials L-PBF. |
|------------------|-----------------|---------------------------|-------------------|
| Material combinations | Applications | Physical characteristics used | Refs |
| Cu–Ni | Liquid rocket combustor comprised of copper alloy liner and nickel alloy jacket | Thermal conductivity, mechanical strength | Gradl et al. 2019 |
| SS–polymer | Bio-inspired metal–polymer arm orthosis | Mechanical strength, material density | Silva et al. 2017 |
| Cu–SS | Bimetal porous energy-absorbing structure | Elongation, mechanical strength | Zhang et al. 2019 |
| Cu–SS | Anti-counterfeiting of L-PBF-processed components | Thermal conductivity, material density | Wei, Sun, et al. 2018 |
| Cu–Steel | Cu conformal cooling channel in mould | Thermal conductivity, wear resistance | Aerosint |
| SS–SiC | Easy-to-remove support structure of L-PBF processed components | Melting point, hardness | Wei, Chueh, et al. 2019 |
| Cu–Ceramic | Cu circuits deposited on ceramic substrates | Electrical conductivity | Syed-Khaja and Franke 2016 |
| Cu–polymer | Cu circuits deposited on polymer substrates | Electrical conductivity | Hou et al. 2018 |
interface. However, the variations in the physical properties of materials and insufficient compatibility can easily lead to defects, such as unmelted powder with a high melting point, cracks, brittle intermetallic compounds, and LME at the interface. Although producing hybrid metal–ceramic and metal–polymer compositions with L-PBF is technically feasible, further in-depth material science research is necessary.

Multi-material L-PBF is practicable and has broad application prospects. However, the current technological maturity remains considerably inadequate for direct industrial applications; accordingly, the technology must be investigated further. In particular, theoretical research must be conducted on phase transformations, thermodynamic calculations, modelling, and numerical simulations. These investigations are essential to increase process efficiency as well as reduce processing defects and cost.

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Notes on contributors
Dr. Chao Wei is a Research Associate in the Laser Processing Research Centre at the University of Manchester, UK. He obtained PhD degree and MSc degree from The University of Manchester in 2019 and 2015 respectively. His research focuses on laser-based multi-material additive manufacturing.

Professor Lin Li is Director of Laser Processing Research Centre at The University of Manchester, UK. He obtained PhD in laser engineering from Imperial College London in 1989, followed by 6 years of postdoctoral research at Liverpool University in laser processing. His academic career started at University of Manchester Institute of Science and Technology as a lecturer in 1994 and established the first laser-processing laboratory. In 2000 he was promoted to a full professor. His research is in laser-based manufacturing including additive manufacturing and has received a number of awards including Schawlow award from Laser Institute of America and Sir Frank Whittle medal from Royal Academy of Engineering, UK. He is an elected fellow of Royal Academy of Engineering.

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