

Multimaterial powder bed fusion techniques: laser-based powder deposition approaches and interface properties

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Abstract

Purpose – This study aims to provide the latest powder deposition approaches in multimaterial laser powder bed fusion (LPBF) and to discuss the produced interface properties based on effects caused by the intrinsic material compatibility, process parameters, powder characteristics, melt pool dynamics, and thermal models. It reviews the current metallurgical challenges at the diffusion zone of the pairs, including the formation of brittle intermetallic phases and other microstructural and thermodynamical challenges to provide better solutions to tackle these problems.

Design/methodology/approach – This review paper used a systematic approach to search for and investigate notable works and peer-reviewed publications concerning multimaterial LPBF.

Findings – The key results explain how the interface type and dynamic process parameter optimization for different layers at the interface, near the interface, and within single materials affect the avoidance of energy-related defects. Understanding the differences in intrinsic material properties such as thermal conductivity, thermal expansion, and melting point, allows us to initially select the most appropriate interface type. The effects of process parameters and powder characteristics are discussed. Common challenges, including microstructural defects, melt-pool-induced defects, and unwanted brittle intermetallic phases, can be solved by nonequilibrium phase diagram simulations, proper interface-type selection, better alloy design, and the use of interlayers at the interface.

Originality/value – By outlining the powder deposition approaches and their interface properties and challenges in multimaterial LPBF, this study provides deeper metallurgical aspects shared among material pairs. This helps researchers in troubleshooting common interface challenges based on the guidelines suggested in this paper.

Keywords Multimaterial additive manufacturing, Multimaterial laser powder based fusion, Bimetallic laser powder bed fusion, Multimaterial 3D printing

Paper type Literature review

1. Introduction

Metal additive manufacturing (AM) enables the manufacturing of three-dimensional (3D) objects from a computer-aided design (CAD) model by feeding metals in various forms layer-by-layer to create a product (Zhang and Liou, 2021). This has emerged as a competitor to traditional metal production technologies such as casting, rolling, extrusion, and forging, aiming to serve industry needs better and offer customized solutions (Attaran, 2017; Pereira *et al.*, 2019). Metal AM has subsequently opened up new solutions to existing challenges in industries such as aerospace, energy, and healthcare (Tao and Leu, 2016; Chen *et al.*, 2021a).

One limitation to designing a new metallic product using conventional manufacturing is the material selection. Engineers are generally prevented from designing products that can meet more than one functional requirement owing to technological limitations. As an alternative, engineers can find solutions in structural and architectural designs. For example, to address stress shielding in orthopedic hip implants, porous or cellular structures are used in the affected regions to dissipate energy to the adjacent bone (He *et al.*, 2018; Cortis *et al.*, 2022). This is where integrating a local secondary

material with less stiffness close to the affected region can be a complementary solution to the current architectural design. With recent technological advancements in multimaterial metal AM and their feeding systems, this limitation can be overcome.

However, even if better feeding systems for multimaterial metal AM are developed, bonding both similar and dissimilar alloys remains metallurgically challenging. Researchers have bonded dissimilar metals using welding technologies such as arc, high-energy beam, friction-stir, resistance-spot, and ultrasonic (Sun and Ion, 1995; Murr, 2010; Wang *et al.*, 2012; Kicukov and Gursel, 2015; Murphy, 2015; Razzaq *et al.*, 2024). Although the literature on these welding techniques can help us understand the general bonding mechanisms on multimaterial metal interfaces, they allow less control over the local energy input. This can cause the formation of harmful intermetallic compounds (IMCs) and large heat-affected zones (HAZs) at and near the interface, leading to poor mechanical properties. Multimaterial AM overcomes this issue by adjusting process parameters to control

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heat input at the multimaterial interface, thus preventing the formation of detrimental IMCs and other thermally induced stress issues (Razzaq *et al.*, 2024).

Multimaterial metal additive manufacturing

The term “multimaterial metal additive manufacturing” was searched in the Web of Science database, and a general yearly upward trend was visible for both publication and citation, as shown in Figure 1. The continuation of this upward trend is anticipated in future years, thus advancing multimaterial metal AM development.

Some AM technologies have shown significant promise for multimaterial purposes. Figure 2 illustrates these technologies, including liquid metal deposition [Figure 2(a)], material extrusion [Figure 2(b)], sheet lamination [Figure 2(c)], binder jetting [Figure 2(d)], direct energy deposition [Figure 2(e)], and powder bed fusion [Figure 2(f)]. Table 1 compares them considering feeding material type, energy source, resolution, geometrical distortion, and general benefits and limitations. Among the most developed technologies, binder jetting is a post sinter-based process. For multimaterial sinter-based processes, different alloys require different sintering parameters. Therefore, there is an additional complexity for sintering green multimaterial structures. They are also more prone to lower mechanical properties and higher porosity than in-process melting technologies, owing to partial local surface melting of the particles during sintering. Considering the benefits and limitations of all multimaterial metal AM processes, PBF is the best candidate for complex and fine-featured designs (Tang *et al.*, 2020b; Mussatto, 2022; Suwanprecha and Manonukul, 2022; Villa *et al.*, 2024).

2. Multimaterial laser powder bed fusion powder deposition techniques

Initially, the only way to produce multimaterial laser powder bed fusion (LPBF) parts was to print a second material on top of

another material that had already been printed. New multimaterial powder-conveying add-on modules allow the creation of fine, complex, and relatively dense multi-metallic parts (Mussatto, 2022; Wang *et al.*, 2022). These multimaterial metal LPBF systems can be classified into three categories based on technological limitations: 2D multimaterial (vertical), 3D multimaterial, and gradient structure (vertical and horizontal) presented schematically in Figure 3.

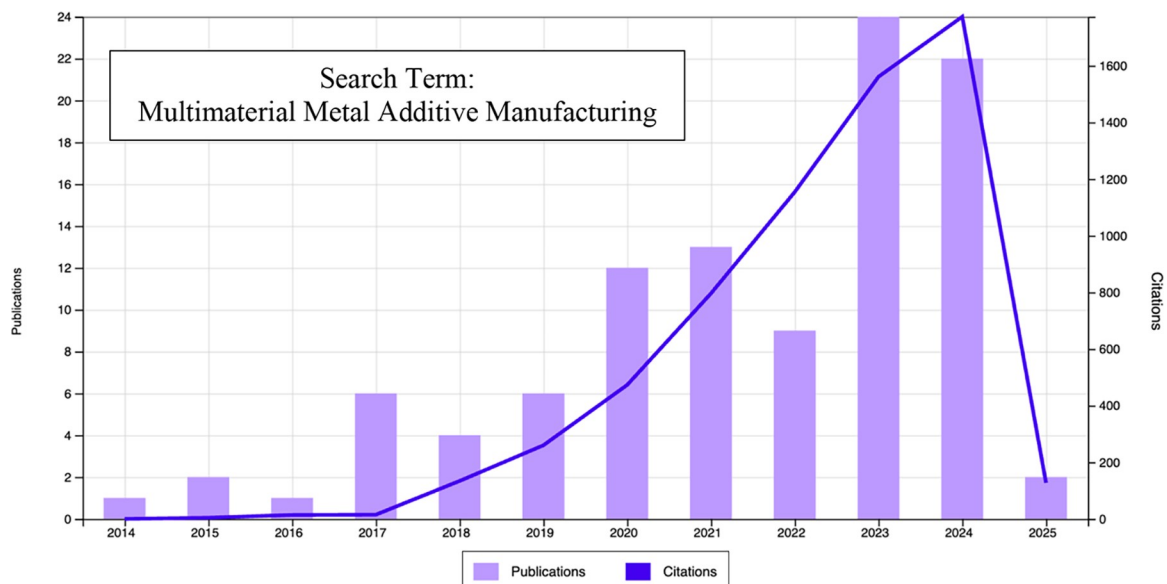
These enable the production of 2D multimaterial sandwich structures [Figure 3(a)]. Traditionally, this involves stopping the machine after completing the first build and restarting it with another powder, known as “layer-on-layer” (Sahu *et al.*, 2024). Using single-material equipment, this method is time-consuming and not industrially feasible. There is also a risk of contamination at the interface resulting from changing the powder, leading to debonding of the interface layers. Various pairs have been processed using this approach; 316 L stainless steel has been processed with In718 (Duval-Chaneac *et al.*, 2021), HX (Rankouhi *et al.*, 2022), SS 15-5PH (Liang *et al.*, 2023), HuvadorK220 (Tey *et al.*, 2016), and NiTi (Ekoi *et al.*, 2022). Ti6Al4V–AlSi10Mg (Müller and Woizeschke, 2021), SS174PH–CoCrMo (Steponavičiūtė *et al.*, 2022), SS420–MS300 (Tan *et al.*, 2020), SS304–MS300, and CS45–MS300 (Tan *et al.*, 2021) are other pairs processed using layer-on-layer.

Alternating powder deposition is a possible solution for contamination, using two powder recoaters, parallel or perpendicular to each other. This is economical, simple, and ideal for micro multimaterial LPBF, enabling faster production of vertical 2D multimaterial sandwich structures. 316 L was processed with bronze via alternating powder deposition. However, powder contamination and mixing limits this configuration and size (Schanz *et al.*, 2022).

2.1 3D multimaterial LPBF

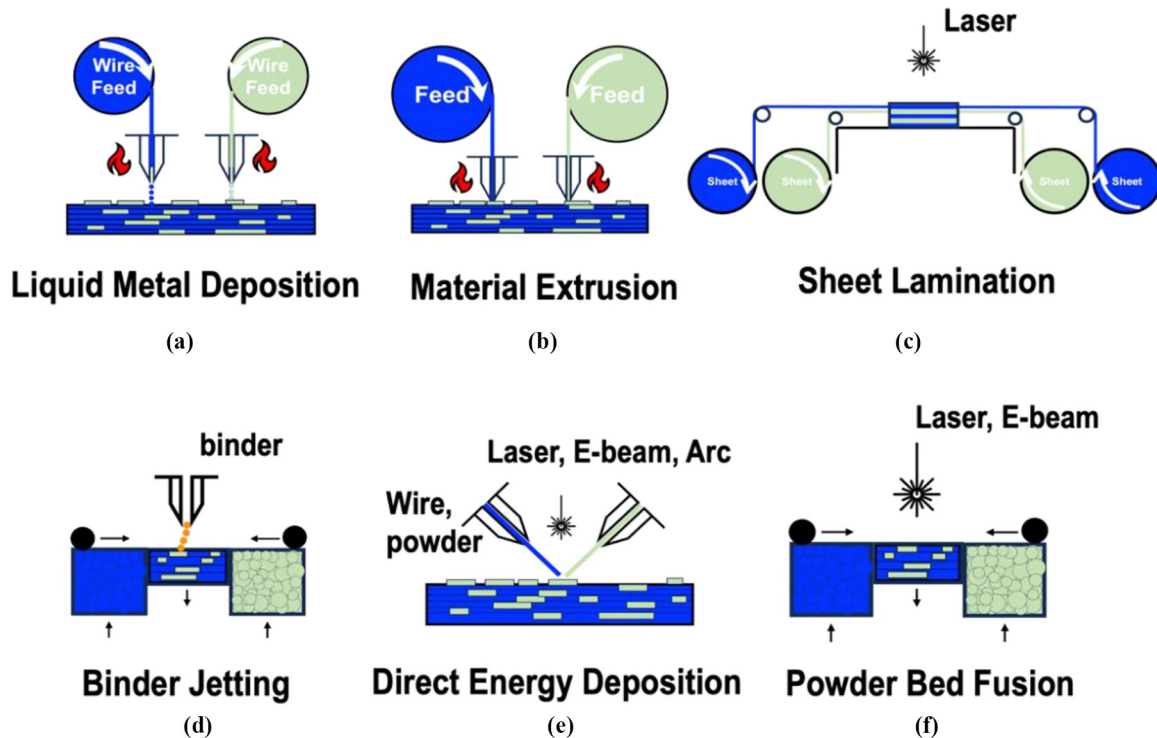
To obtain 3D multimaterial metallic components, new powder deposition systems are required to selectively deposit two or more

Figure 1 Yearly publication and citation numbers for the search term “multimaterial metal additive manufacturing”



Source: Figure by authors, Web of Science

Figure 2 Schematic overview of the metal additive manufacturing technologies. (a)–(c) Not fully developed yet to produce solid multimaterial AM components due to technical, economical, and/or design limitations, (d)–(f) main developing additive manufacturing techniques for multimaterial metal components



Source: Figure by authors

powders with reasonable resolution within a layer. Figure 3(b) illustrates different approaches used for this purpose. The first utilizes a deposition-suction mechanism. After the initial powder layer is deposited and selectively melted, the unmelted powder is removed using a suction device. This creates a void for another material, which is deposited using a second powder-conveying device. This approach enables powder reusability with less powder waste. However, it is relatively slow owing to the repeated deposition and suction within each layer. The powders in each layer may be slightly contaminated during each suction and deposition cycle. Commercially developed machines include SLM Solutions (models 280HL and 250HL, Germany), University of Minho, (3DMMLPBF process, Portugal), and Aconity (model AconityONE, Germany) (Pires *et al.*, 2022). Pairs of 316L–C18400 (Liu *et al.*, 2014), TS1.2709–C2.1293 (Anstaett, 2017; Bareth *et al.*, 2022), TS1.2709–CW106C (Schneck *et al.*, 2021), SS420–Cu (Cunha *et al.*, 2022), 316L–In718 (Wits and Amsterdam, 2021), In718–Cu (Marques *et al.*, 2022, p. 718), Ti6Al4V–Ti (Borisov *et al.*, 2021), and CoCrMo–Ti6Al4V (Bartolomeu *et al.*, 2023) were processed using this approach.

Another approach for selectively depositing metal powders in a layer involves patterning drums. These drums use electromagnetic micro-airflow technology to selectively blow powder particles adhering to the drums onto the deposition layer. Schaeffler Aerosint SA, a Belgian company, patented this technology (Neirinck *et al.*, 2021). Being module-based, this technology can be integrated into different LPBF machines. Currently, drums can selectively deposit up to three powders over

a build plate, with a resolution of $300\mu\text{m}$. This relatively high resolution enables the production of 3D designs with high multimaterial complexities. However, it is challenging to print functionally graded materials (FGMs) because of the abrupt interface between two materials. Another challenge is powder recyclability because the two alloys are mixed at the working stage and near the recoaters. It is possible to use a third drum to fill the powder bed with a cheaper, separable material, to aid recycling. The technique is complex and currently heavily dependent on technical skills. Currently, MS300–CuCrZr pairs are printed using the patterning drum technique (Li *et al.*, 2024).

In another approach, similar to ink printing using piezoelectric actuators (Jabari *et al.*, 2019), powder particles can be deposited selectively using nozzles with small orifice diameters. This was applied at the University of Manchester, UK, USING up to six nozzles containing different powders to be deposited in a layer. This allows the combination of metal powders with various percentages to be deposited, thus creating possible FGMs. This approach has two significant challenges. First, the fluidization of the powders with pulsed voltages should be consistent. To ensure this, ultrasonic assistance is necessary to control the potential and kinetic energy of powder particles during deposition. Second, the resolution of the print depends on the nozzle head diameter; therefore, small nozzle orifice diameters of a few tens of millimeters are necessary. Pairs of 316L–Cu10Sn (Wei *et al.*, 2018, 2019), 316L–In718 (Wei *et al.*, 2018), Invar36–Cu10Sn (Wei *et al.*, 2021), and Ti6Al4V–Cu10Sn (Wei *et al.*, 2022) have already been printed using this technique.

Table 1 Summarizing and comparison of current multimaterial additive manufacturing technologies

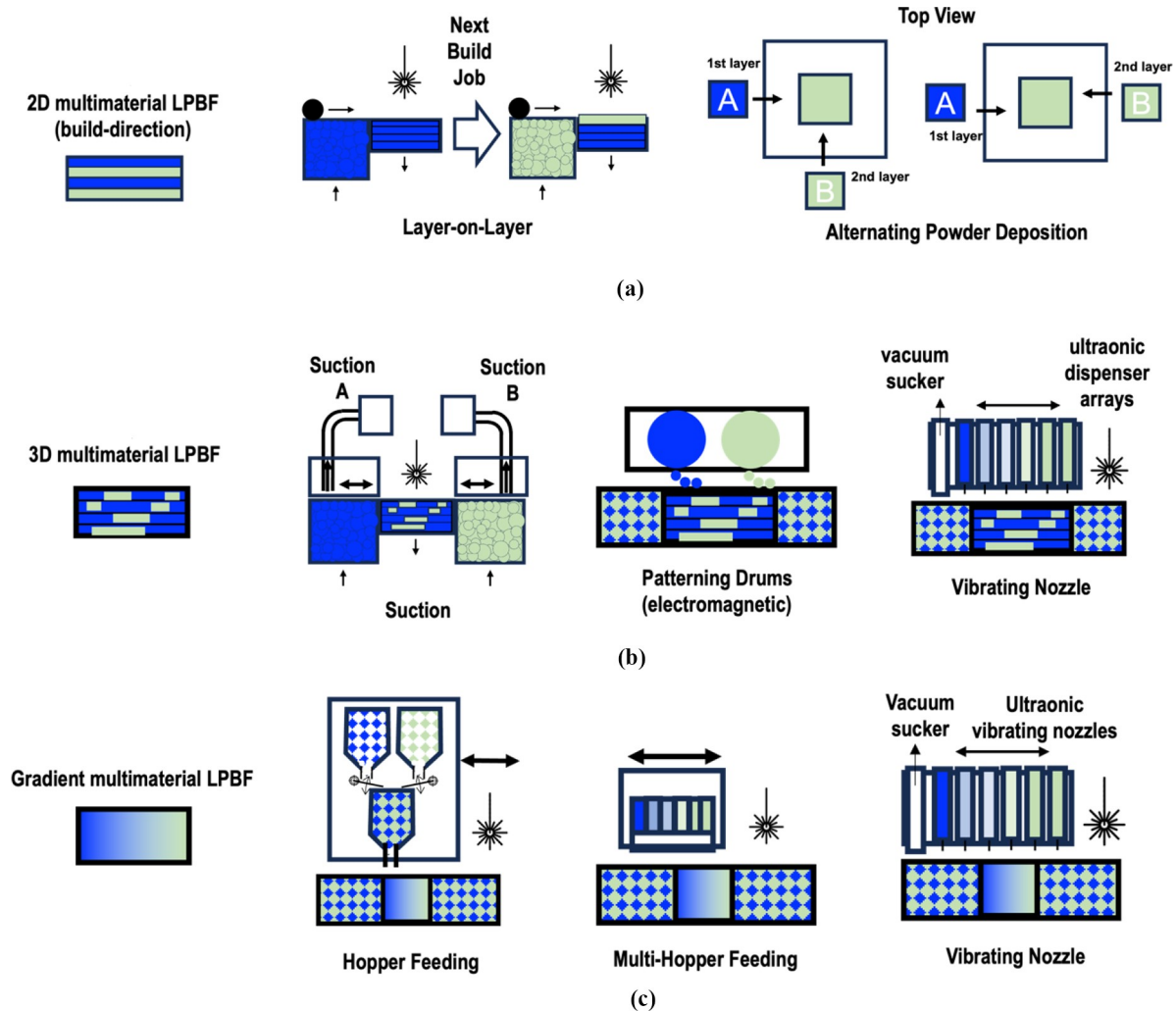
Techniques	Feeding material	Energy source	Resolution dependence	Geometrical deviation and distortion	Benefits and limitations	Ref
Material extrusion	Metal-filled Thermoplastic granulated feedstock, bars, and filaments	Heating elements to heat the extrusion nozzles to melt the binder, and sintering in the furnace	On nozzle diameter nozzle (diameters of 0.25–0.8 mm) and filler metal powder size	Deviation: Relatively high due to binder decomposition and powder particle rearrangement, and integration Distortion: Relatively low, such as curling and wrapping due to being high-power-source-free Very high	Benefits: Relatively quick (printing speed of 5–80 mm/s), low cost, simple, no direct metal powder involved, a wide range of materials available, no high-power energy source Limitations: No developed systems for multimaterial metal parts yet, low resolutions, lower relative density, sintering required, difficult to apply for complex parts, binder pollution, poor surface quality Benefits: Low cost, relatively quick, cheap feeding materials Limitations: Delamination, low density, low resolution, very limited geometrical designs, thermally-induced stresses, not well developed for metals yet	(Watschke <i>et al.</i> , 2018; Putra <i>et al.</i> , 2020; Suwanprecha and Manonukul, 2022)
Sheet lamination	Metallic sheets	Laser	Minimum sheets/foils thickness		Benefits: Low cost, relatively quick, cheap feeding materials Limitations: Delamination, low density, low resolution, very limited geometrical designs, thermally-induced stresses, not well developed for metals yet	(Wimpenny, Bryden and Pashby, 2003; Guo <i>et al.</i> , 2019; Salmi, 2021)
Liquid metal deposition	Wires of low temperature alloys (e.g. Aluminium)	Heating elements to melt the wire inside the crucible before deposition	On deposition nozzle	Relatively high	Benefits: Suitable for low temperature alloys like aluminium alloys, low-cost feeding materials, Limitations: Limited technological advancements, not yet developed for multimaterial purposes Benefits: Relatively quick, clean, low-cost equipment, no support structures required in general, low distortion due to missing thermally induced stresses induced by high power energy sources, high relative densities (higher than 99%) Limitations: Sintering challenge for green multimaterial components (shrinkage and possible gravity-based distortion during sintering), requiring a multimaterial powder handling system similar to PBF, relatively weaker material properties than other high-power techniques, sintering dependent	(Sukhotskiy <i>et al.</i> , 2017; Ruggles, 2018; Kapil <i>et al.</i> , 2023)
Binder jetting	Metal powders of 25–53 μm , adhesives for binding	Sintering furnaces	On metal powder particle sizes and printhead resolution	Deviation: Relatively high deviation due to binder decomposition and powder particle rearrangement Distortion: Relatively lower distortions	Benefits: Relatively quick, clean, low-cost equipment, no support structures required in general, low distortion due to missing thermally induced stresses induced by high power energy sources, high relative densities (higher than 99%) Limitations: Sintering challenge for green multimaterial components (shrinkage and possible gravity-based distortion during sintering), requiring a multimaterial powder handling system similar to PBF, relatively weaker material properties than other high-power techniques, sintering dependent	(Chou <i>et al.</i> , 2013; Hong <i>et al.</i> , 2016; Bai and Williams, 2018; Mostafaei <i>et al.</i> , 2018; Toursangsarak, 2018; Ziaee and Crane, 2019; Li <i>et al.</i> , 2020; Putra <i>et al.</i> , 2020)

(continued)

Table 1

Techniques	Feeding material	Energy source	Resolution dependence	Geometrical deviation and distortion	Benefits and limitations	Ref
Directed energy deposition	Powder (between 20–30 μm and 200 μm), wire	Laser, electron beam, plasma arc	The energy source spot size (few tenths to hundreds of μm), feeding rate, powder size, nozzle diameter (minimum feature size of about 250 μm , minimum reported layer thickness of about 200 μm)	Relatively high	Benefits: Relatively higher deposition rates, low oxidation, ideal for large scale fabrications and <i>in situ</i> repairing, good development for multimaterial printings, suitable for FGMs Limitations: Relatively large resolution/features, limited designs for smaller features such as for fine lattices, relatively large geometrical deviation and distortions, risk of inhomogeneity, and relatively lower densities than LPBF	(Gibson, Rosen and Stucker, 2015; Thompson <i>et al.</i> , 2015; Bobbio <i>et al.</i> , 2017; Woo <i>et al.</i> , 2019; Ahn, 2021; Feenstra <i>et al.</i> , 2021)
Powder bed fusion	Powder	Laser, electron beam	Laser or electron beam spot size, multimaterial powder deposition system limitations, powder size	Relatively high	Benefits: Suitable for complex geometries with fine features such as lattices, porous materials, topology-optimized complex designs, etc., relatively low geometrical deviation than Binder Jetting due to absence of a binder Cons: Relatively expensive and complex technology, Higher cooling rates than DED prone to distortions, hot cracks, distortions through thermally induced stresses, supporting required, challenging for re-cyclability and reusability of the remaining powder	(Mussatto, 2022; Wang <i>et al.</i> , 2022; Sahu <i>et al.</i> , 2024)
Source(s): Table by authors						

Figure 3 Schematic overview of multimaterial metal LPBF techniques. (a) 2D vertical multimaterial LPBF systems: layer-on-layer, alternating powder deposition, (b) 3D multimaterial LPBF: suction, patterning drums, vibrating nozzle, and (c) gradient multimaterial LPBF: hopper feeding, multi-hopper feeding, vibrating nozzle



Source: Figure by authors

2.2 Gradient multimaterial LPBF

Producing 3D LPBF gradients is more complex than creating 3D multimaterial structures with abrupt bonding, owing to limitations in powder deposition systems. Currently, powders for multimaterial LPBF gradients are deposited using a hopper powder feeding or vibrating nozzle [see Figure 3(c)]. In hopper systems, a mixture of two alloys at specific percentages are prepared on top of the deposition chamber, either using a third hopper or a multi-hopper system (Wen *et al.*, 2021; Demir *et al.*, 2022). In vibrating systems, ultrasonic vibration is used to control the mixture of the powder (Wei *et al.*, 2018, 2019). These technologies allow the production of both horizontal and vertical gradient structures. However, these technologies yet to be commercialized.

The current printed pairs using each technique with a summary of their benefits and limitations is provided in Table 2.

3. Multimaterial LPBF interfaces

The interface at multimaterial LPBF samples can be assessed by techniques such as EBSD mapping and XRD for phase detection, and hardness gradients from material A to B. In detail, the simplest method to achieve a multimaterial interface is through direct melting. Although there appears to be a distinct interface boundary between the two alloys, intermixing always occurs around the boundary for a few hundred micrometers. Therefore, alloys must have similar material, chemical, and thermophysical properties for successful bonding. If the material and thermophysical properties (e.g. thermal conductivity, thermal expansion coefficient [CTE], and/or melting points) do not match, the chemical properties (e.g. solubility or low risk of intermetallic formation) and functionally graded joints (FGMs) can be designed to suppress the effects of these materials and their thermophysical properties. If the material and chemical

Table 2 Summary of multimaterial 2D, 3D, and gradient structures reported in the literature

Techniques	Current pairs	Benefits and limitations
Layer-on-layer	316-In718 (Duval-Chaneac <i>et al.</i> , 2021) 316-HX (Rankouhi <i>et al.</i> , 2022) 316-SS 15-5PH (Liang <i>et al.</i> , 2023) 316-HuvadorK220 (Tey, Yeong and Chen, 2016) 316-NiTi (Ekoi <i>et al.</i> , 2022) Ti6Al4V-ALSi10Mg (Müller and Woizeschke, 2021) SS 17-4PH-CoCrMo (Steponavičiūtė <i>et al.</i> , 2022) SS420-MS300 (Tan <i>et al.</i> , 2020) SS304-MS300 (Tan <i>et al.</i> , 2021) CS45-MS300 (Tan <i>et al.</i> , 2021)	Benefits: Possibility of powder recyclability, using single-material equipment Limitations: Very time-energy consuming, not efficient, only 2D multimaterial, contaminated interfaces, interface layers debonding
Alternating powder deposition	316L – Bronze (Schanz <i>et al.</i> , 2022)	Benefits: Small system, simple, economic, ideal for micro multimaterial LPBF Limitations: 2D multimaterial, limitations in configuration and size, powder contamination, and mixing
Suction technique	316L – C18400 (Liu <i>et al.</i> , 2014) TS1.2709-C2.1293 (Anstaett, 2017; Bareth <i>et al.</i> , 2022) TS1.2709 – CW106C (Schneck <i>et al.</i> , 2021) SS420 – Pure Cu (Cunha <i>et al.</i> , 2022) 316L – In718 (Wits and Amsterdam, 2021) In718 – pure Cu (Marques <i>et al.</i> , 2022, p. 718) Ti6Al4V – Ti (Borisov <i>et al.</i> , 2021) CoCrMo – Ti6Al4V (Bartolomeu <i>et al.</i> , 2023)	Benefits: Powder reusability, 3D multimaterial Limitations: Relatively a slow process, time-energy consuming, prone to powder bed contamination during removal steps
Patterning drums (aerosint technique)	MS300 – CuCrZr (Li, Sukhomlinov and Que, 2024)	Benefits: 3D multimaterial of up to three alloys, fine feature/resolution (300 μ m), module-based (enabling integration into different single-material AM machines), possible to have low mixed powder waste, easy to be industrialized, relatively quicker than other 3D multimaterial LPBF techniques Limitations: Challenging for FGMs, challenging with mixed powder recyclability, high dependency on skilled technicians
Hopper feeding	316L – Fe35Mn (Demir <i>et al.</i> , 2022a) 316L – Cu10Sn (Chen <i>et al.</i> , 2022) In718 – GRCo42 (Walker <i>et al.</i> , 2022) Pure Fe – ALSi12 (Demir and Previtali, 2017) Pure Ti – Pure Ta (Lesko <i>et al.</i> , 2021; Walker <i>et al.</i> , 2022) 316L – MS1 (Nadimpalli <i>et al.</i> , 2019) CoCrMo – In718 (Wen <i>et al.</i> , 2021)	Benefits: Ideal for FGMs, 3D multimaterial, printing up to six alloys within a part Limitations: Prone to powder contamination
Vibrating nozzle	316L – Cu10Sn (Wei <i>et al.</i> , 2018, 2019) 316L – In718 (Wei <i>et al.</i> , 2018) Invar36 – Cu10Sn (Wei <i>et al.</i> , 2021) Ti6Al4V – Cu10Sn (Wei <i>et al.</i> , 2022)	Benefits: Enabling FGMs, enables 3D multimaterial printing with up to six alloys within a single part Limitations: Challenging fluidization of the powders (ultrasonic assistance is necessary for controlling the potential and kinetic energy), resolution dependence on nozzle head diameter

Source(s): Table by authors

properties do not match, alloys with an interlayer compatible with both may be used (Kavousi Sisi *et al.*, 2024). This basic guideline, however, considers only the intrinsic material properties, excluding the process parameters and feeding powder effects of multimaterial LPBF processes.

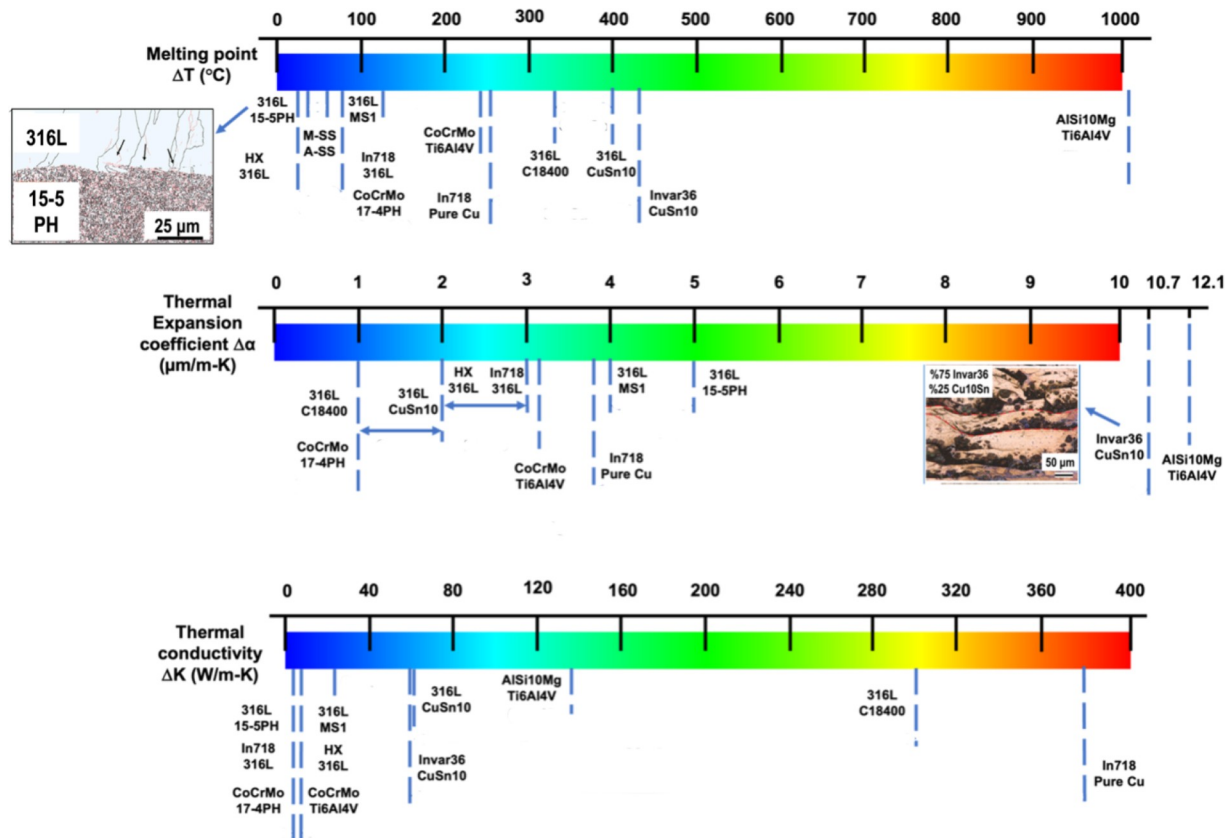
3.1 Influencing parameters at multimaterial LPBF interfaces

3.1.1 Intrinsic material properties

The most influential thermophysical properties reported for multimaterial DED and dissimilar weld joint approaches are differences in the melting points, CTEs, and thermal conductivities (Reichardt *et al.*, 2021). A significant mismatch

in melting points leads to nonuniform heat flow and dilution, causing excessive residual stresses and cracks, particularly in alloys with lower melting points during solidification. In 316 L and 15-5PH steels produced by multimaterial LPBF (Liang *et al.*, 2023), lower melting point differences show interfaces with no critical cracks (Figure 4). Similarly, a mismatch in thermal conductivity can lead to heat dissipating quickly from the interface to the base material, leading to uneven heat flow at and near the interface. This results in distortion and lack of fusion in the lower thermally conductive material owing to insufficient heat (Reichardt *et al.*, 2021). In Invar36 and Cu10Sn produced by multimaterial LPBF (Wei *et al.*, 2021), high thermal conductivity differences result in a lot of lack of

Figure 4 Qualitative differences in melting point (ΔT), thermal conductivity (ΔK), and thermal expansion coefficient ($\Delta \alpha$) of alloys paired with multimaterial LPBF in the current literature



Source(s): makeitfrom.com database, Figure courtesy of microstructures by Liang *et al.* (2023) and Wei *et al.* (2021)

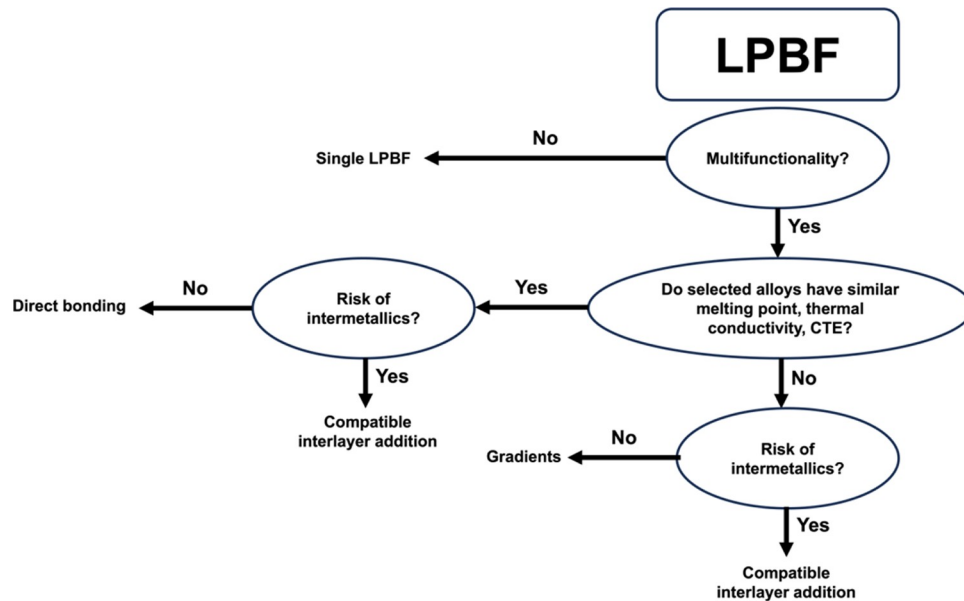
fusion and pores (Figure 4). To overcome this mismatch in fusion welding, heat is directed towards a higher thermally conductive material, either as a preheating or weld heating source. In multimaterial LPBF, this can be achieved with more controllability over the effective laser process parameters (adaptive volumetric energy densities for both alloys) and scanning strategies.

Another significant mismatch in thermophysical properties is the CTE. Differences in the CTE in multimaterial DED processes result in unequal thermal contraction, causing stress concentrations at the interface. This causes residual tensile and compressive stresses in metals with higher CTE and lower CTE, respectively (Reichardt *et al.*, 2021; Feenstra *et al.*, 2021). The gradient CTE was reported to solve these excessive residual stresses (Du Pont, 2010; Brentrup, 2011; Brentrup *et al.*, 2012; Hofmann *et al.*, 2014). The same situation applies to multimaterial LPBF. Figure 4 qualitatively illustrates the differences in the melting points, thermal conductivities, and CTE of multimaterial LPBF pairs in the current literature (Makeitfrom.com engineering material database, 2024, www.makeitfrom.com/). When we see pairs with higher relative differences (parts more towards the red region), we can expect microstructural defects and stress concentrations caused by these differences. Although other effective parameters, such as process and powder parameters, play

a role in intermetallic formation, this graph can be an initial approach for understanding the relative range of thermophysical property differences at the interface of multimaterial LPBF pairs. In the future, a more advanced graph with diverse data can be constructed using new research data.

Figure 5 can initially help multimaterial LPBF designers to select the right interface type based on the intrinsic material properties. This selection can vary in the following steps, considering other influencing factors, such as process parameters, input energy densities, scanning strategies, and the constituent nonequilibrium phases at the interface.

Elemental intermixing can occur at the multimaterial LPBF interface, leading to the local creation of new phases. Although these new intermetallic phases can be beneficial, such as nickel-iron aluminides (Ni_3Al , Fe_3Al , FeAl , etc.) with high strength and good high-temperature properties (Palm *et al.*, 2020), they are often brittle because of their ionic and covalent bonding nature and long-range ordering (Reichardt *et al.*, 2021). This brittleness can cause failure in the presence of thermal and residual stresses in AM processes, including the LPBF technology (Chen *et al.*, 2020; Ghasemi, Yildiz and Malekan, 2024). Therefore, awareness of the risks of intermetallic formation for each selected pair is a necessary step for multimaterial LPBF.

Figure 5 An initial complimentary flowchart is used to decide the type of multimaterial LPBF interface

Source: Figure by authors

Density is another important parameter of multimaterial LPBF. In vertical interfaces, a significant difference in density can cause melt-pool segregation, which can result in alternating regions at the interface owing to the higher buoyancy of the lighter material (Pulugurtha *et al.*, 2009; Reichardt *et al.*, 2021). The effect of density on vertical multimaterial LPBF interfaces requires further studies to provide a solid statement on the placement of the two materials, considering their thermal and physical properties.

3.1.2 Process parameters

Process parameters directly affect the size, morphology, stability, penetration depth, and microstructural heterogeneity of the melt pool in LPBF processes. For example, suboptimal laser power and scanning speed can cause distortion, part shrinkage, delamination, porosity, surface roughness, cracks, and premature failure (Oliveira *et al.*, 2020; Ahmed *et al.*, 2022; Liu *et al.*, 2023). In multimaterial LPBF, optimization of the interface process parameters for each material is necessary. In addition, scanning strategy and beam shaping can be designed at the direct interface to increase intermixing and further facilitate diffusion (Shi *et al.*, 2020; Zhang *et al.*, 2020; Nadammal *et al.*, 2021; Bakhtari *et al.*, 2024). Self-grading can also occur when the interface is in the vertical direction (Pulugurtha *et al.*, 2009; Liang *et al.*, 2014; Reichardt *et al.*, 2021). Therefore, multimaterial LPBF has more influencing parameters than single-material LPBF (Wei *et al.*, 2020). Depending on the type of bonding (direct printing, printing in a gradient, or printing with an interlayer), parameter sets for each material need to be optimized according to trial-and-error, Design of Experiments (DoE), or machine learning (ML) (Durão *et al.*, 2019; Wang *et al.*, 2020; Guan and Wang, 2023).

3.1.3 Powder characteristics

The morphology, size, and distribution of powder particles directly affect the powder packing density, continuity, and

homogeneity, affecting the internal stresses, part distortion, porosity, and surface roughness in the final build part of the LPBF process (Sun *et al.*, 2017). Achieving a higher, more consistent packing density is possible using a broader range of particle sizes, where smaller particles fill the gaps between larger particles (Kumar, 2014; Sun *et al.*, 2017). In the case of 316L stainless steel and In718 multimaterial LPBF, the particle size distributions (PSDs) of 28.2 ± 12 and $22.5 \pm 7 \mu\text{m}$ demonstrate this diversity as the broader range creates better powder stacking at the interface, reducing the risk of *in situ* porosity (Duval-Chaneac *et al.*, 2021). Similarly, a wider particle size range in a gradient LPBF structure from 316L to Fe35Mn improves the powder bed homogeneity (Demir *et al.*, 2022).

Finer powders with a narrow distribution may lead to powder agglomeration, whereas coarser particles in a narrow distribution can cause segregation (Simchi, 2004; Manfredi *et al.*, 2013; Kumar, 2014). Spherical powder morphology enhances the powder flowability and increases packing density (Niu and Chang, 1999; Sun *et al.*, 2017). Nonspherical powders lead to mechanical interlocking and powder entanglement, which results in nonhomogeneous and noncontinuous powder layers with variable densities. This can cause porosity and a lack of fusion (Attar *et al.*, 2015). Therefore, the metal powders in multimaterial LPBF must be as spherical as possible. Smaller particles have a higher surface area, which increases the absorption of laser energy and increases the melt pool temperature. This can result in a more significant densification and higher porosity in the final microstructure and interface (Simchi, 2004).

The thermal conductivity of a powder bed largely depends on the packing density of the powder rather than the intrinsic thermal conductivity of the metal (Gusarov *et al.*, 2003; Sun *et al.*, 2017). Because of their larger surface area, powder

layers absorb more laser energy than solid metal layers with the same composition (Stacy *et al.*, 2014; King *et al.*, 2015). However, with highly reflective metals, such as copper or aluminum, the laser energy absorption is low; therefore, the absorptivity of both alloys must be considered in multimaterial LPBF. For example, in 316-Cu10Sn or 316 L-C18400 pairs, the higher thermal conductivity, heat dissipation, and reflectivity of copper can lead to unmelted copper particles. In these cases, the steel component absorbed more energy and melted more completely. Simultaneously, the copper remains partially melted, leading to a dense microstructure on the steel side and a more porous structure on the copper side (Liu *et al.*, 2014; Chen *et al.*, 2022). This issue also appears in the AlSi10Mg-pureCu combination, where the lower thermal conductivity of the substrate helps heat accumulation in the first copper layer, reducing the thermal stresses and temperature gradients (Guan *et al.*, 2021). Techniques such as using larger particle sizes, thinner powder layers, and preheating the build platform can be effective in improving the energy absorption, especially for metals with high reflectivity and thermal conductivity. Additional strategies, such as reducing the scan speed, increasing the laser power, and reducing the hatch spacing, can help overcome these challenges, ensuring consistent energy absorption and improved bonding with materials such as copper alloys (Gusarov and Smurov, 2010; Liu *et al.*, 2014).

3.1.4 Melt pool dynamics

Building on prior knowledge from single-material LPBF, melt pool forces, such as Marangoni flow, influence intermixing through surface tension differences in thermal gradients and solid-liquid interfaces (Hondros *et al.*, 1998; Semak *et al.*, 2006; Clark *et al.*, 2020). The recoil pressure from metal vaporization causes localized movements within the melt pool (Cullom *et al.*, 2021; Mollamahmutoglu *et al.*, 2022). Furthermore, capillary and thermocapillary forces through solid-liquid interfaces can push the molten metal towards the melt pool center (He *et al.*, 2023), and buoyancy forces can also have a similar effect (Chen *et al.*, 2021b). Some of these forces are reported to be less eminent owing to the small melt pool sizes and the short interaction time of the material with the laser (Chen *et al.*, 2021b). The effectiveness of these forces is highly dependent on the process parameters, intrinsic material properties, and powder characteristics (Tey *et al.*, 2016; Wei *et al.*, 2019; Tan *et al.*, 2020). multimaterial LPBF has additional influencing parameters such as the interface type (i.e. direct, gradient, or with an interlayer) and interface orientation (vertical, horizontal, or with an angle). There are currently a few papers discussing the melt-pool dynamics of multimaterial LPBF interfaces as sharp and gradient interfaces (Sun *et al.*, 2020; Gu *et al.*, 2021; Tang *et al.*, 2022; Huang *et al.*, 2025). Studying the melt pool forces can help to design better intermixing at the interface and increase the bonding between materials.

The Marangoni number in a single-material LPBF is characterized based on equation (1) (Limmaneevichitr and Kou, 2000):

$$Ma = -\frac{dg}{dT} \frac{L \Delta T}{\mu a} \quad (1)$$

where the Marangoni number (Ma) is directly proportional to the surface tension differences in the temperature gradients ($\frac{dg}{dT}$, N/mK), the length over the temperature gradient (L , m), and the temperature difference along the melt pool and the solid interface (T , K), and inversely proportional to the dynamic viscosity of the melt pool (μ , Kg/ms), and the thermal diffusivity (a , m²/s). A higher Marangoni number leads to stronger convection flow at the edge of the melt pool (Rankouhi *et al.*, 2022). This higher number of multimaterial LPBF can increase the elemental intermixing at the melt pool. The thermal diffusivity a and dynamic viscosity μ are constant values for alloys, but the temperature differences along the interface, surface tension and the length over the temperature gradient can be engineered by optimizing the process parameters. Also, In multimaterial LPBF, the optimized laser beam shape and scanning strategy can increase the elemental intermixing at the interface melt pool. The direction of the vortices for surface tension differences in the temperature gradients ($\frac{dg}{dT}$) in multimaterial LPBF influences the flow. If it is negative, the flow is outwards, resulting in a broader melt pool with a reduced depth (Figure 6a), such as 316 L and copper alloy C52400 (Figure 6b) (Bai *et al.*, 2020), 316 L and Hastelloy X (Figure 6d) (Rankouhi *et al.*, 2022), AlSi10Mg and TiC nanocomposites (Yuan and Gu, 2015), and AlSi10Mg and Ti6Al4V vertical interfaces (Figure 6c) (Wu *et al.*, 2022). This flow can be engineered for horizontal interfaces using beam shaping from a Gaussian laser beam to a ring beam, potentially turning the Marangoni flow positive and changing the flow inward. This inward flow could benefit the elemental intermixing of the alloys.

Owing to gravitational forces and dilution, self-grading can occur in vertical multimaterial LPBF interfaces. Dilution is the ratio of the cross-sectional area of the melted substrate to the melted clad from the previous layer, as in equation (2):

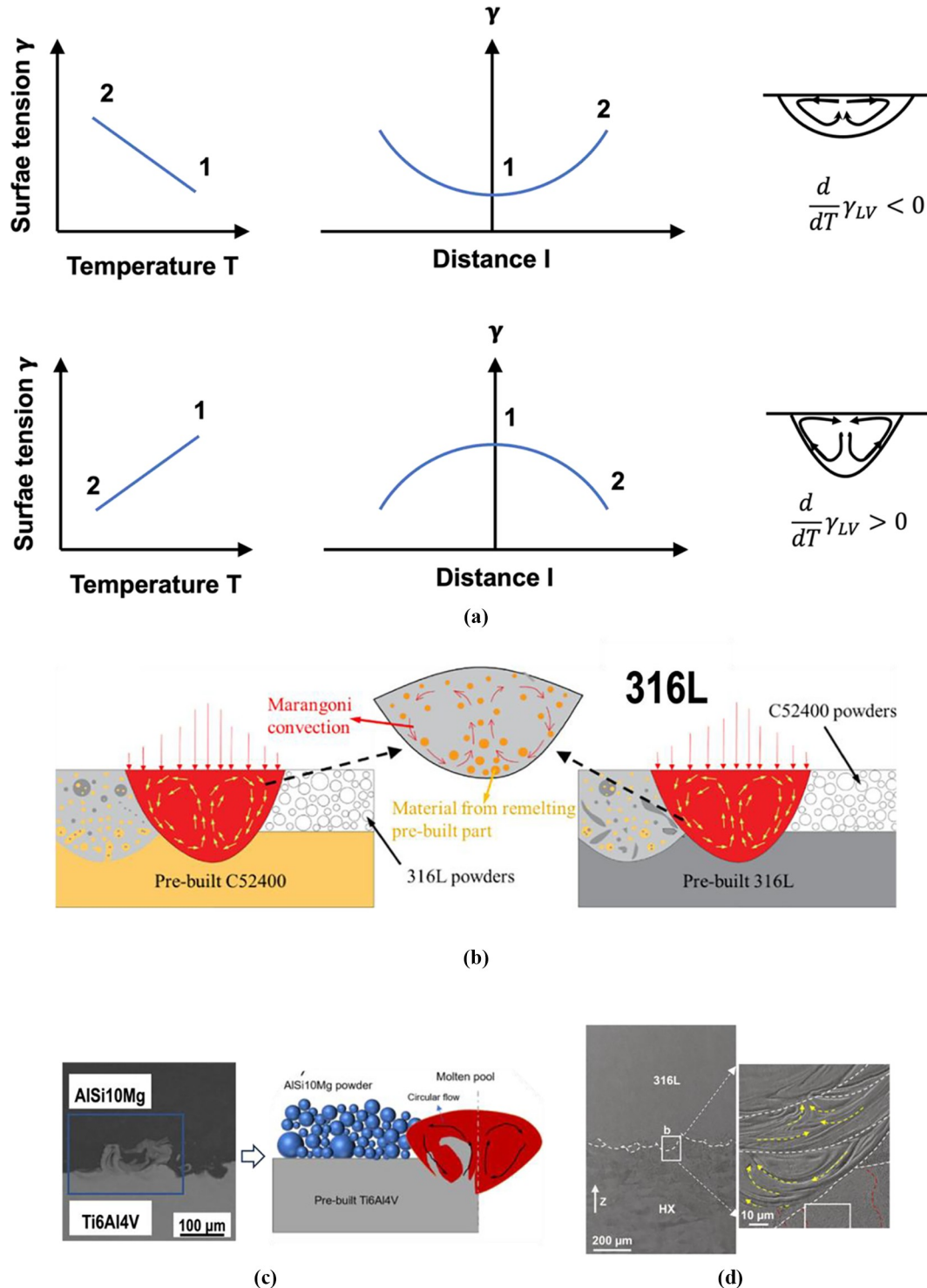
$$D = \frac{A_{sub}}{A_{sub} + A_{clad}} \quad (2)$$

A_{sub} and A_{clad} are the substrate area and cladding in the previous layer, respectively. This remelting of the fraction of the previous layers is inevitable in the LPBF processes to avoid delamination and achieve the required proper bonding between layers. This self-grading can benefit vertical multimaterial LPBF interfaces by eliminating the stress concentrations at the interface if the risk of brittle phase formation is controlled (Pulugurtha *et al.*, 2009; Liang *et al.*, 2014; Reichardt *et al.*, 2021)

3.1.5 Volumetric energy density and thermal models

Volumetric energy density (VED) is an empirical formula used to measure the local energy input of a laser. VED significantly influences the density, roughness, and uniformity of the layers in AM (Wang *et al.*, 2016; Caiazzo *et al.*, 2020). A higher VED improves the wetting of the molten metal on the surface, enhancing flowability (Tang *et al.*, 2020a). It also reduces the surface roughness and avoids lack-of-fusion (Gu *et al.*, 2014). However, excessively high VED cause unwanted porosity, cracking, and increased surface roughness (Wang *et al.*, 2016;

Figure 6 Schematic illustration of the effect of surface tension gradients on Marangoni flow within the melt pool. (a) Negative surface tension gradient (top), positive surface tension gradient (bottom), (b) 316 L/C52400 (steel-copper) multimaterial melt pool exhibiting a negative surface tension gradient. (c) Schematic representation and corresponding SEM image of an AISi10Mg/Ti6Al4V (aluminum-titanium) multimaterial melt pool with negative surface tension gradients, and (d) SEM image of a 316 L/HX (steel-Hastelloy) multimaterial melt pool with negative surface tension gradients



Source: Figure courtesies of [Bai et al. \(2020\)](#), [Wu et al. \(2022\)](#), and [Rankouhi et al. \(2022\)](#)

Vilanova *et al.*, 2020). For horizontal multimaterial LPBF interfaces, different VEDs must be selected based on their thermophysical differences. Materials with higher melting points demand a higher VED, creating a challenge at the interface where materials mix. As discussed in Section 2.1.1, gradients can mitigate the challenges of the thermophysical differences at the interface. Therefore, VED requirements can be smoothly transitioned from the first to the second material.

VED formula is as follows [equation (3)]:

$$VED \left(\frac{J}{mm^3} \right) = \frac{P(W)}{v \left(\frac{mm}{s} \right) * t(mm) * H(mm)} \quad (3)$$

where P is the laser powder, v is the scanning speed, t is the layer thickness, and H is the hatch spacing (Caiazzo *et al.*, 2020). In the current multimaterial LPBF literature, to study the impact of VED on hardness, Figure 7(a) can be used. This Figure shows the hardness trend with the VED moving from the first to the second material for direct bonding without considering the distances. This diagram uses one hardness value for the interface and two hardness values for a single material. We expect a smooth transition in hardness from one material to another with an average interface hardness value between two single materials. If the hardness value at the interface is higher than the average of the single materials, it might indicate the formation of brittle intermetallic phases, causing stress concentrations and interfacial cracks. This higher interface hardness is visible for the Ti6Al4V-Cu10Sn and pure Fe-AlSi10Mg pairs, and for both interfaces, the formation of various brittle intermetallic phases is reported [Figure 7(b) and (c)] (Demir and Previtali, 2017; Wei *et al.*, 2022).

3.2 Challenges with multimaterial LPBF interfaces

3.2.1 Formation of brittle intermetallic phases

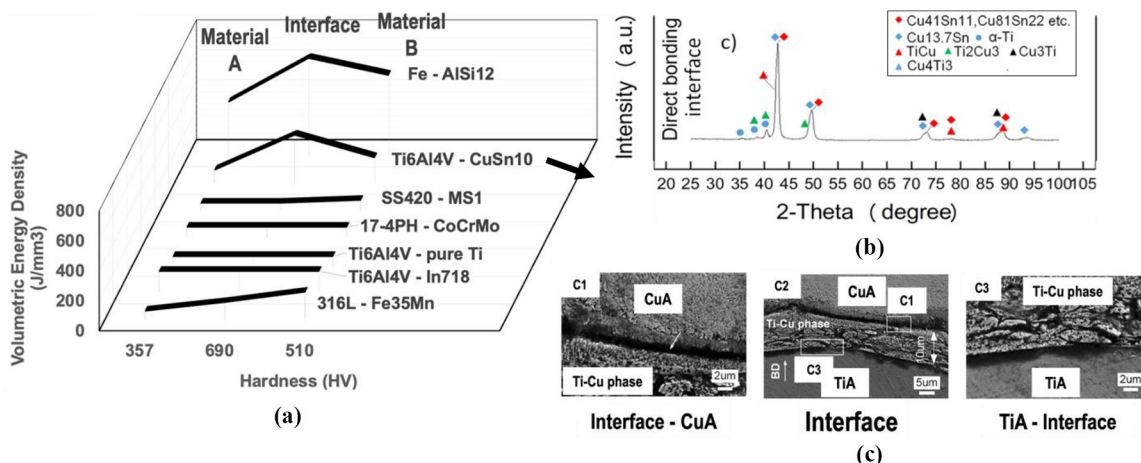
In multimaterial LPBF, thermodynamics at the interface can result in the formation of new intermetallic secondary phases. In most cases, these phases have higher hardness and melting

points, which can cause residual stress accumulation and defects, such as cracks and debonding (Figure 7). It is necessary to apply characterization tests like X-ray diffraction analysis (XRD) and electron backscatter diffraction (EBSD) or energy-dispersive X-ray spectroscopy (EDS/EDX) for multimaterial LPBF interfaces to identify the formed intermetallic phases, providing information on their types, amounts, shapes, and sizes.

Three complementary approaches can be considered to predict and prevent the formation of brittle phases at the interface:

- 1 **Nonequilibrium phase diagram simulations:** With new material libraries, it is possible to predict the formation of brittle intermetallic phases at the interface of the multimaterial LPBF pairs. LPBF is a rapid solidification process that leads to nonequilibrium states. Scheil-Gulliver solidification simulations, using tools such as CALPHAD, provide nonequilibrium calculations (Gulliver, 1913; Bocklund *et al.*, 2020b). These calculations are based on thermodynamic modeling of the crystal structure of each phase (gas, liquid, solid solution, and solid compound) to determine the Gibbs free energies at specific temperatures and pressures (Ohtani, 2006; Xu *et al.*, 2016). This approach creates a thermodynamic database of multicomponent material systems. It helps to identify the driving forces behind intermetallic phases, the kinetics of precipitation nucleation, and solute segregation/partitioning during gradient material solidification. For example, the composition and phase fractions of two FGMs have been calculated for CP Ti/Invar and Ti6Al4V/Invar using Scheil-Gulliver (Bocklund *et al.*, 2020a). They predicted phase fractions is reported to be in a good agreement with EBSD phase maps, showing validity of the predictions by this model for FGMs produced by AM. This tool not only predicts the formation of intermetallic compounds but also helps in the elemental redesigning of problematic alloys. Furthermore, it enables the optimization of process parameters to control the cooling rates and

Figure 7 (a) The impact of VED on hardness values for multimaterial LPBF pairs. Data courtesies by Demir and Previtali (2017), Tan *et al.* (2020), Borisov *et al.* (2021), Demir *et al.* (2022a), Steponavičiūtė *et al.* (2022; Wei *et al.* (2022a), (b,c) example XRD and SEM images illustrating the interface of the Ti6Al4V-CuSn10 pair (c2). C1 and C2 show brittle Ti-Cu phases



Source: Figure courtesies of Wei *et al.* (2022b)

temperature gradients derived from simulations, helping mitigate the formation of these phases. (Jägle, 2016; Reichardt et al., 2021; Dzobgiewu and du Preez, 2021; Liu et al., 2024).

- 2 *Alloy re-designing:* Even with optimized process parameters, the formation of intermetallic phases in certain material combinations is inevitable. Adding or removing particular amounts of elements in an alloy may reduce the formation of harmful brittle intermetallic phases at the interface. For instance, adding 0.02–0.05 Wt.% Zr to the Al–Ni alloying system can decrease the formation of the brittle polycrystalline Ni_3Al phase by 40%–50% (Liu, 1986). Similarly, the brittle Co_3V hexagonal phase can be transformed into a ductile FCC phase by substituting Co with Ni and Fe (Lin et al., 1992). Therefore, some intermetallic phases could be eliminated with slight changes in the composition of the alloys involved.
- 3 *Interface layer addition:* Another method to prevent brittle intermetallic phases is to add an interlayer compatible with both alloys at the interface. In DED multimaterial processes, alloying combinations such as Ti with Ni or Fe systems have been reported as very challenging for obtaining an interface without intermetallic phases. In these cases, interlayers can form solid solutions with the same thermophysical properties as the base alloys without intermixing (Reichardt et al., 2021). This layer can be added to multimaterial LPBF interfaces if the technology for powder deposition systems allows it. Zhang et al. (2023) added a 1-mm-thick copper interlayer between Ti6Al4V and AlCuMg. This reduced the formation of brittle Ti–Al phases such as γ -TiAl by replacing them with more ductile Cu–Al and Cu–Ti phases. The interlayer allows the transition from Al–Cu–Mg to Ti–6Al–4V through a series of more ductile phases (α -Al $\rightarrow$ Al₂Cu $\rightarrow$ AlCu₂Ti $\rightarrow$ CuTi₂ $\rightarrow$ α' -Ti).

3.2.2 Microstructure and melt pool

Grain growth during single-material LPBF is typically columnar. Owing to the complex temperature gradients and solidification rates in the build process, preventing epitaxial columnar grain growth and the resulting highly textured microstructure are challenging. Columnar grain growth in LPBF is one of the main obstacles to achieving multidirectional properties in parts (Liu et al., 2022; Wei et al., 2023). Cracking and microcracking along grain boundaries, at the interface, or near the interface is a common issue reported in multimaterial LPBF, contributing to the premature failure of components under mechanical stress (Duval-Chaneac et al., 2021; Chen et al., 2022; Schanz et al., 2022; Wei et al., 2022; Wu et al., 2022). These defects can stem from the concentration of intermetallic phases (Wei et al., 2022), residual stress due to porosities from partially melted powder (Wu et al., 2022), differences in thermal material properties, such as thermal expansion coefficients between different materials (Wei et al., 2021), lattice mismatch between dissimilar compositions (Wu et al., 2022), and lack of fusion during the process (Tan et al., 2020; Müller and Woizeschke, 2021; Chen et al., 2023). Another thermodynamic challenge in multimaterial LPBF is enthalpy of mixing. A positive enthalpy suggests that the

mixture releases heat into the melt pool, thus aiding the process. Conversely, a negative enthalpy indicates that the mixture absorbs heat, potentially hindering the intermixing at the interface. Schwendner et al. (2001) studied this effect using direct laser deposition with two alloys: Ti-10%Cr, with a negative enthalpy of mixing (−12.6 kJ/g atom), and Ti-10%Nb, with a positive enthalpy of mixing (+4.2 kJ/g atom). They found that negative enthalpy of mixing led to a more homogenous intermixing zone and finer grains. In contrast, the positive enthalpy of mixing causes poor intermixing, coarser grains, and slower solidification.

Managing the energy-density input can reduce these defects. An insufficient laser energy can lead to a lack of fusion, where the molten pool does not overlap sufficiently with adjacent pools, leaving unmelted particles at the boundaries. In multimaterial LPBF, differences in melting points and absorptivity can cause inhomogeneous mixing, evaporation of low-melting-point materials, and insufficient melting of high-melting-point materials. This can result in elemental segregation and powder infusibility, thereby deteriorating the mechanical properties of the finished part. Additionally, differences in the density and viscosity can lead to elemental segregation, with denser elements sinking to the bottom of the melt pool. In gradients, placing high-density alloys at the bottom and low-density alloys at the top can help overcome this problem. Homogenization heat treatment and optimized laser parameters can also solve this issue, although their effectiveness is limited when there are significant differences in melting points (Guan and Wang, 2023).

Liquid metal embrittlement (LME) has been reported in Fe–Cu multimaterial LPBF interfaces. LME is driven by Galvele's atomic surface mobility (ASM), where liquid copper diffuses into iron grains, causing the Kirkendall effect and resulting in grain boundary cracks filled with liquefied copper. LME occurs when the base metal temperature is higher than the melting point of copper, and there are no intermetallic phases between the solid and liquid materials (Schanz et al., 2022).

4. Conclusion and future works

This review presents a summary of the current status of multimaterial LPBF. It offers foundational guidelines for understanding the complexities of multimaterial interfaces, with the aim of supporting advanced applications. An overview of current multimaterial AM techniques and the motivations for transitioning to LPBF technology was followed by a detailed discussion of the developed powder deposition methods for multimaterial LPBF. This assessment facilitates the informed selection of methods by evaluating the advantages and limitations of each approach. It then addresses the metallurgical aspects of multimaterial LPBF interfaces, covering key topics such as thermophysical properties, process parameters, nonequilibrium phase simulations, melt pool dynamics, and energy density considerations. After outlining the technological context, it delves into the metallurgical characteristics of these interfaces, thereby providing a critical summary of the current understanding.

This study emphasizes the importance of metallurgical understanding of the interface and pairs based on multi-

functionality. It is important to study prior knowledge of other AM and dissimilar welding techniques to find overlapping challenges and possibly mitigate them using the advantages of LPBF. Moreover, differences in intrinsic material properties such as thermal conductivity, thermal expansion, and melting point, play an important role in interface design. This allows us to initially select the most appropriate interface type of direct bonding, gradients, or interlayer addition. Additionally, process parameters such as layer thickness, scanning strategy, laser power, scanning speed, hatch spacing, and laser beam shaping play an important role at the interface and melt pool forces. These parameters can be used to design the melt-pool forces to obtain the maximum intermixing between the two materials at the interface. Powder characteristics, including the composition, particle size, distribution, shape, density, and flowability of both materials at the interface, also play an important role. Common challenges for multimaterial LPBF interfaces include microstructural defects, melt-pool-induced defects, and unwanted brittle intermetallic phases. Solutions to address these challenges involve nonequilibrium phase diagram simulations, proper interface-type selection, better alloy design, and the use of interlayers at the interface. This understanding allows customized solutions to be obtained through the theorization of mixing mechanisms in the chosen alloys.

The emerging direction, based on the current literature, is towards gradients. Most of the thermal and physical differences can be mitigated using gradients, and controlling the brittle phase formation through cooling rates and temperature gradients from nonequilibrium phase diagram simulations at the interface can mitigate the typical intermetallic phases, because adding an interlayer in each problematic interface is a challenging limitation for designers. This may not be entirely feasible for complex and fine multimaterial LPBF parts.

Alloy redesigning can be beneficial. While conventional standard alloys are advantageous in multimaterial LPBF owing to extensive data availability and established standards, the interface between different materials can result in new compositions that may not yet be characterized in conventional manufacturing or other AM technologies. Consequently, compositions that are specifically tailored for compatibility using insights from nonequilibrium phase simulations are crucial for advancing multimaterial LPBF applications in the industry. Moreover, multimaterial LPBF parts must withstand long-term stresses in the long run and exhibit good fatigue and creep properties. Further research on the heat treatment and fatigue performance of the pairs is necessary for reliable industrial applications. Heat treatment, in particular, is challenging for multimaterial LPBF interfaces as different heat treatment is required for both sides of the pair for the interface itself. This requires further attention in the future, leading to improved design and manufacturing processes, ultimately resulting in more reliable and durable products.

Exploring the unique functionalities of multimaterial LPBF can lead to new products with multi-functional performance, especially when combined with alloys that possess high strength, ductility, and corrosion resistance. This can be particularly valuable for industries such as medical applications. Emphasizing improvements in the interface properties, such as gradient transitions, remelting, and

optimized scanning strategies, will be essential for future industrial applications.

Overall, the findings of this study provide researchers and industry professionals with valuable insights into the challenges and potential solutions to multimaterial LPBF. This study highlights strategies for integrating diverse functionalities while addressing critical microstructural and mechanical issues.

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