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Gao, Jian, Han, Quanquan, Soe, Shwe, Wang, Liqiao, Zhang, Zhenhua, Zhang, Han, Song, Jun, Liu, Yue, Setchi, Rossitza and Yang, Shoufeng 2024. Laser additive manufacturing of TiB₂-modified Cu₁₅Ni₈Sn/GH3230 heterogeneous materials: Processability, interfacial microstructure and mechanical performance. Materials Science and Engineering: A 900 , 146496. 10.1016/j.msea.2024.146496

Publishers page: <http://dx.doi.org/10.1016/j.msea.2024.146496>

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Laser additive manufacturing of TiB₂-modified Cu₁₅Ni₈Sn/GH3230 heterogeneous materials: Processability, interfacial microstructure and mechanical performance

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Abstract

Cu/Ni heterogeneous materials integrate excellent thermal conductivity and high-temperature mechanical properties, enabling them to be widely used in the aerospace domain. Differences in the thermal and physical properties of the Cu and Ni materials, however, make them difficult to be processed using the laser powder bed fusion (LPBF) additive manufacturing process. This study systematically examines the effects of various LPBF process parameters on microstructure, element diffusion, bonding strength and microhardness at the Cu/Ni interface, as well as investigating the mechanisms of defect formation within Cu/Ni heterogeneous materials. The results indicate that a reasonable control of laser energy input ($< 100 \text{ J/mm}^3$) facilitates the Cu/Ni components

through strong interfacial metallurgical bonding without pore defect formation. Compared to single-material Cu alloy, the ultimate tensile strength (UTS) of the horizontally bonded Cu/Ni specimen increased by 55.25%, without significant reductions in elongation. The vertically bonded Cu/Ni tensile specimen fractured in the middle of the Cu region rather than the interfacial region, indicating superb interfacial bonding strength. Another advantage lies in the enhancement of thermophysical properties, with a 109.5% increase in thermal conductivity achieved in the LPBF-fabricated Cu/Ni heterogeneous materials compared to the single-material Ni alloy. Quasi-static compression experiments indicated that the Cu/Ni lattice structure could absorb more energy when compressed parallel to the build direction (BD), compared to being perpendicular to the BD. This study provides guidance for the design and manufacture of high-performance Cu/Ni heterogeneous components via LPBF.

Keywords: additive manufacturing; laser powder bed fusion; Cu/Ni heterogeneous materials; interfacial microstructures; mechanical properties

1. Introduction

Compared to single metal or alloy materials, metal heterogeneous material components have become promising candidates within advanced industrial fields due to their enhanced and tailorable thermal, physical and mechanical properties [1–3]. Although metal heterogeneous materials may be fabricated by various processes, such as welding and vapor deposition [4], these traditional approaches cannot produce structures with the complex geometries that will meet the demands of advanced components. In recent years, the rapid development of additive manufacturing (AM) has provided greater design freedom and excellent formability for metallic heterogeneous material fabrication [5–7], which cannot be fulfilled by traditional routes. The AM preparation of multi-functional and high-performance metal components

integrated with heterogeneous materials has thus become an attractive research topic in advanced material research [8]. Laser direct energy deposition (LDED) and laser powder bed fusion (LPBF) are currently the two mainstream metal AM technologies [9]. LDED is more convenient in the high-efficiency preparation of metal heterogeneous materials due to its forming mechanism; as a result, the majority of studies related to the AM fabrication of metallic heterogeneous materials focus on the LDED process [10–12]. Compared with LDED, LPBF may be used to fabricate components with much finer dimensional accuracy, internal channels and lattice structures [13–15]. The use of LPBF to manufacture metal heterogeneous materials thus shows great potential in various engineering applications [16,17].

In recent years, the LPBF of metallic heterogeneous materials has mainly been used to explore alloy compatibility and to link the formability of different alloy systems. The most frequently studied metallic heterogeneous material system is copper/steel [5,18–20], followed by nickel/steel [10,21], but neither system can meet the extreme service temperature requirements of items such as aerospace engines. Copper/nickel heterogeneous materials, in contrast, integrate the good thermal conductivity of copper alloys and the good high-temperature mechanical properties of nickel-based superalloys, making it an ideal material system for aerospace applications [21]. Few studies have investigated Cu alloy and Ni-based superalloy combinations via LPBF, however, which is mainly due to the high laser reflectivity of copper and the very large differences in thermophysical properties of copper and nickel. Onuike et al. [22] studied the manufacturing of Inconel 718–Grcop84 components through the LDED process. Their microstructure analysis showed that the direct deposition of Grcop84 on the Inconel 718 substrate resulted in severe balling phenomena and difficulty in bonding. They found that introducing an intermediate layer of 50% Grcop84 and 50% Inconel 718 between Inconel 718 and Grcop84 could achieve good metallurgical bonding, which greatly improved thermal conductivity. Their study did not consider the effects of process parameters on

Cu/Ni interface defects, interfacial microstructure and bonding strength.

The rocket engine combustion chamber is composed of an inner lining of copper alloys and an outer lining of nickel-based superalloys. Very thin internal flow channels are generally required between the inner and outer linings since they must survive extreme environments providing challenging thermal and structural loads [23]. Complex high-resolution lattice structures are the nonnegligible advantages of LPBF-fabricated Cu/Ni heterogeneous materials, although very few studies have investigated Cu/Ni lattice structures fabricated via LPBF. These lattice structures composed of Cu/Ni heterogeneous materials offer great potentials in manufacturing internal flow channels used for the rocket engine combustion chambers. It is thereby important to fabricate these structures via LPBF and study their mechanical performance, including the processability, compression behaviour and deformation mechanism.

The mechanical performance of fabricated Cu/Ni heterogeneous materials via LPBF depends on the behaviour of both the Cu and Ni components. The high laser reflectivity and thermal conductivity, however, may significantly reduce the processability of Cu material fabricated via LPBF, while the crack defects formed within LPBF of nickel-based superalloys significantly impair their mechanical properties. The authors' previous study revealed that the laser absorption rate and mechanical properties of LPBF-fabricated Cu₁₅Ni₈Sn alloys can be significantly improved by adding 1wt.% submicron TiB₂ particles [24]. Previous study also demonstrated that the addition of 1wt.% submicron TiB₂ to GH3230 promotes grain refinement to achieve LPBF-fabricated crack-free and high-performance nickel-based alloys [25].

Based on the above analysis and our previous research, 1wt.%TiB₂-modified Cu₁₅Ni₈Sn powder and 1wt.% TiB₂-modified GH3230 powder were employed in the present study to investigate the LPBF of Cu/Ni heterogeneous materials. The addition of 1wt.% TiB₂ to Cu alloy was intended to improve the laser absorption rate of Cu₁₅Ni₈Sn while eliminating the crack defects in the LPBF of GH3230 [24,25]. This work systematically studies the effects of various

process parameters on the microstructure, element diffusion, bonding strength and microhardness at the Cu/Ni interface in order to reveal the defect formation mechanisms at the interface of Cu/Ni heterogeneous materials fabricated via LPBF. The processability and compression behaviour of Cu/Ni lattice structures with strut-based structures – including body-centred cubic (BCC) and surface-based (gyroid) lattices – were also studied in depth.

2. Materials and methods

2.1. Material preparation

Gas-atomised GH3230 powder (average particle size: 28.2 μm) with a diameter between 16 and 45 μm , supplied by Praxair Technology Ltd. (Indianapolis, IN, USA), was used in this study. The element composition (wt.%) of GH3230 powder is 22.18Cr–14.57W–0.46Al–0.08C–1.73Co–1.8Fe–0.02La–0.44Mn–2.29Mo–0.01N–0.0131O–0.0014S–0.37Si and 0.09Ti, with the balance Ni. Gas-atomised Cu15Ni8Sn powder (average particle size: 35 μm) with a diameter between 15 and 53 μm was supplied by the Vilory Corporation (Jiangsu, China). The Cu15Ni8Sn powder has an element composition (wt.%) of 14.96Ni–8.01Sn–balance Cu. Sub-micrometre TiB_2 particles (average particle size: 600 nm), supplied by CWNano (Shanghai, China), were used in this study. 1wt.% TiB_2 -modified GH3230 and 1wt.% TiB_2 -modified Cu15Ni8Sn were prepared by a high-speed mixer (SpeedMixer DAC 800.1 FVZ, Nordrhein-Westfalen, Germany). Our previous study includes details on the specific powder modification process [24,25]. The 1wt.% TiB_2 -modified GH3230 powder and 1wt.% TiB_2 -modified Cu15Ni8Sn powder used in the present study are shown in Fig. 1. As shown in Fig. 1c/d, the majority of TiB_2 particles uniformly adhered to the surface of the Cu15Ni8Sn and GH3230 powder. In fact, very few large-scale TiB_2 particles scattered amongst the metal powder after the high-speed mixing process.

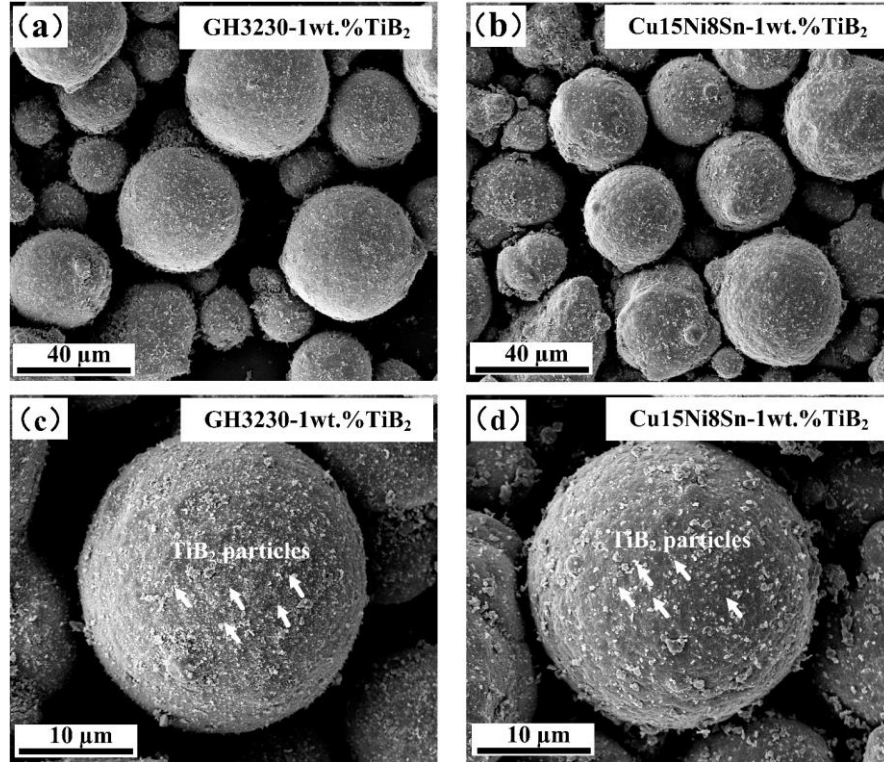


Fig. 1: Morphology of powder: (a/c) 1wt.%TiB₂-modified GH3230 powder; (b/d) 1wt.%TiB₂-modified Cu15Ni8Sn powder

2.2. Laser powder bed fusion (LPBF) process

TiB₂-modified Cu15Ni8Sn/GH3230 heterogeneous materials were manufactured using Mlab cusing 200R (Concept Laser GmbH, Germany), equipped with a 200 W continuous wave fibre laser with a spot size of 75 μm. Cubic specimens with dimensions of 7 × 7 × 7 mm³ were used to optimise the laser scanning speed parameter. Fabricating the copper alloy first and then the nickel-based alloy may lead to horizontal crack defects at the interface because of the high laser reflectivity and thermal conductivity of copper [26]. Therefore, the manufacturing scenario in this study is to fabricate the nickel alloy first and the copper alloy afterwards to achieve good interfacial bonding. In fact, after the fabrication of GH3230 material, the GH3230 powder inside both the build and powder chambers were manually removed prior to filling the TiB₂-modified Cu15Ni8Sn powder in both chambers. Note that the powder exchange process can only be completed manually due to the apparatus's limitations at this stage.

The GH3230 was re-melted three times prior to the manufacture of Cu15Ni8Sn material because the re-melting process enables homogenize the elemental distribution, reducing residual stress and un-melted defects at the interface [27,28], thereby improving the interfacial bonding performance [29].

Table 1 lists the process parameters used to manufacture each material. The density of the Cu/Ni samples was measured using the Archimedes principle to determine the optimal parameters. Thermal conductivity samples ($10 \times 10 \times 3 \text{ mm}^3$ cubes), tensile samples and lattice structure samples were fabricated under optimal conditions. Fig. 2 shows the geometric schematic as well as detailed dimensions of the tensile and lattice structure samples.

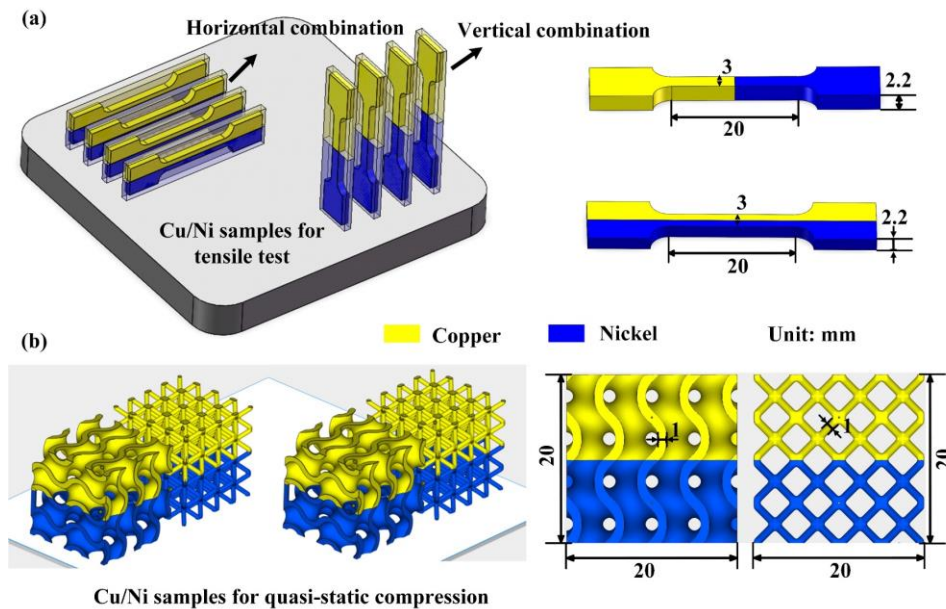


Fig. 2: Geometric schematic and detailed dimensions of LPBF-fabricated Cu/Ni (a) tensile and (b) quasi-static compression samples

Table 1: Process parameters for LPBF process optimisation

LPBF parameters	GH3230	Cu15Ni8Sn
Laser spot diameter	75 μm	75 μm
Scanning strategy	Chessboard	Chessboard
Rotation angle	67°	67°
Laser power	190 W	190 W
Layer thickness	40 μm	30 μm
Hatch spacing	100 μm	100 μm
Scanning speed	800 mm/s	300, 350, 400, 450, 500, 550, 600, 650, 700, 750, 800, 850 mm/s

2.3. Material characterisation techniques

Cu/Ni cubic samples were vertically sectioned parallel to the build direction (BD) before being ground and then polished. A 6XB (SOIF, China) optical microscope (OM) and a JSM-7800F field emission high-temperature scanning electron microscope (SEM; JEOL, Japan) equipped with a NordlysMax3 EBSD system (Oxford Instruments, UK) were employed to examine the interface microstructure and defect morphology of the Cu/Ni samples. The scanning voltage for SEM was set to 10 kV. The EBSD scanning area was $300 \times 1400 \mu\text{m}$, with a $0.6 \mu\text{m}$ scanning step. The EBSD data were analysed using the HKL Channel 5 (Oxford Instruments, UK) software package. The phase composition was detected by an X-ray diffraction (XRD) instrument (SmartLab 3 kW, Japan) with Cu-K α radiation set at 40 kV in the 2θ range from 20° to 100° , with a step size of 0.013° and a count time of 1 s per step.

2.4. Mechanical characterisation techniques

An HVS-1000a microhardness tester (Huayuzhongxin, Laizhou, China) with a 200 g load and an indentation time of 10 s was employed to measure the interface hardness. Three measurements were conducted and averaged to determine the Vickers microhardness. Tensile testing was carried out with an electronic universal testing machine (FBS-100KNW, FBS, Shenzhen, China) with a strain rate of $0.6 \times 10^{-3} \text{ s}^{-1}$ at room temperature. Quasi-static compression testing was performed using a universal testing machine (Z250, ZwickRoell, Germany). The thermal conductivity at room temperature was analysed using a thermal conductivity instrument (LFA 1000, Linseis, Germany).

3. Results

3.1. Relative density and defects

Fig. 3 shows the interfacial microstructure of the TiB₂-modified Cu/Ni heterogeneous material at different laser scanning speeds under OM inspection. Two common types of defects found near the interface area included keyholes and cracks, which were significantly affected by scanning speed. The cracks appeared to primarily form at the Cu zone, beyond the heterogeneous boundary. The Ni sections were fabricated first, employing the optimal process parameters to achieve nearly full-dense build. The Cu parts were then fabricated on the top of the solidified Ni components to achieve the parameters optimization by employing different scanning speeds varying from 300 to 850 mm/s. Note that, open pores mainly formed under the low scanning speed conditions (<550 mm/s).

In general, when utilising a high laser power and low scanning speed, excessive energy may lead to the formation of deep keyholes due to thermal capillary force, Marangoni convection and recoil pressure during the melting process of metal powder [30–32]. With an increased scanning speed, the quantity and size of the keyholes at the interface were found to have reduced, and could even be completely eliminated. At a scanning speed of 300 mm/s, the crack length was measured to be around 166 μm ; increasing the scanning speed resulted in further reductions in the length of the crack. It should be noted, however, that although the length of the interfacial cracking in heterogeneous materials can be shortened by increasing the scanning speed, the cracks cannot be completely eliminated by this strategy. The low scanning speed causes an excessive energy input and a high thermal gradient between the Ni and Cu materials, together with the differences in thermophysical properties, resulting in high residual stress during the forming process, which further causes interfacial cracking [26,33]. It is also worth noting that cracking may also

form due to stress concentrations caused by irregular pores. Increasing the scanning speed contributed to a reduction in the pores and crack defects at the interface of Cu/Ni heterogeneous materials, which we expected would improve the bonding strength of the interface.

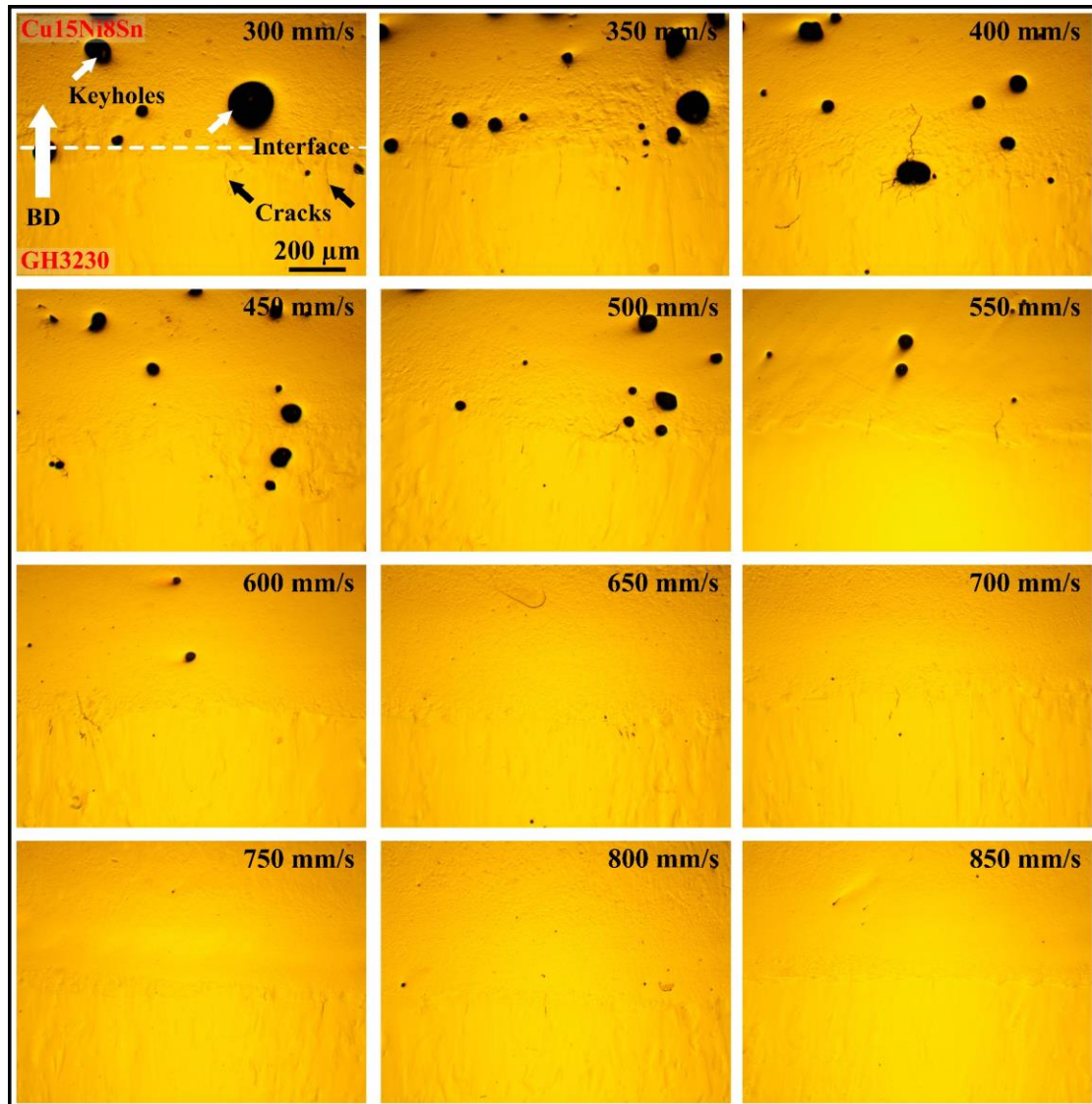


Fig. 3: Microstructure of TiB₂-modified Cu/Ni heterogeneous material interface under different laser scanning speeds using optical microscopy (OM)

Fig. 4 shows the densities of Cu/Ni heterogeneous samples under different laser energy density (LED) values. (The definition of LED may be found in the literature [34].) The absolute densities of the Cu/Ni heterogeneous samples showed a decreasing trend with increasing LED. When the LED was over 100 J/mm³, a large number of keyhole defects formed, resulting in low absolute

density values. When a relatively low LED ($< 100 \text{ J/mm}^3$) was employed, the keyhole defects could be completely eliminated, and the size of crack defects could be also suppressed. Based on the defect evolution of the interface microstructure and density, the optimal LED and scanning speed values were determined to be 84.44 J/mm^3 and 850 mm/s , respectively.

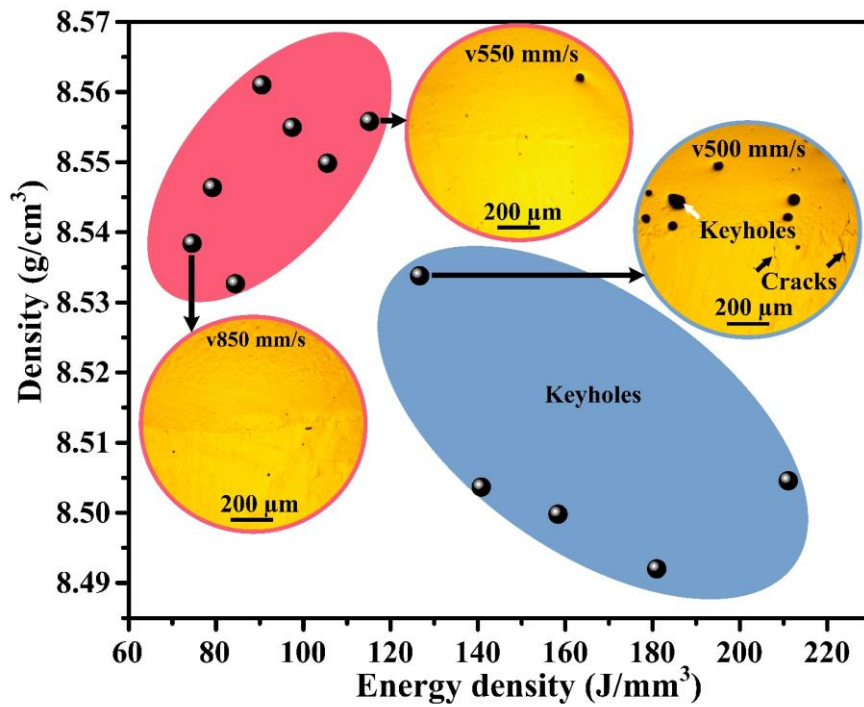


Fig. 4: Densities of TiB_2 -modified Cu/Ni heterogeneous material under different laser energy densities (LEDs)

Fig. 5 shows the material distribution close to the interface of the Cu/Ni heterogeneous samples fabricated via LPBF under two different scanning speeds ($v550$ and $v850$, respectively). The interface between the Cu and Ni heterogeneous materials may be classified into three distinct regions during energy-dispersive X-ray spectroscopy (EDX) mapping of the primary characteristic elements. Region I is located underneath the interface, mainly at the Ni matrix. Region II is located near the top of the interface, called the “diffusion region,” where the two materials transit to each other; this effect is primarily caused by the agitation of the molten pool induced by Marangoni convection [35]. Region III is mainly found at the Cu matrix. The interfaces’

melting depth/width values of the heterogeneous v550 and v850 specimens were measured to be 58.1/100 and 23.8/105, respectively.

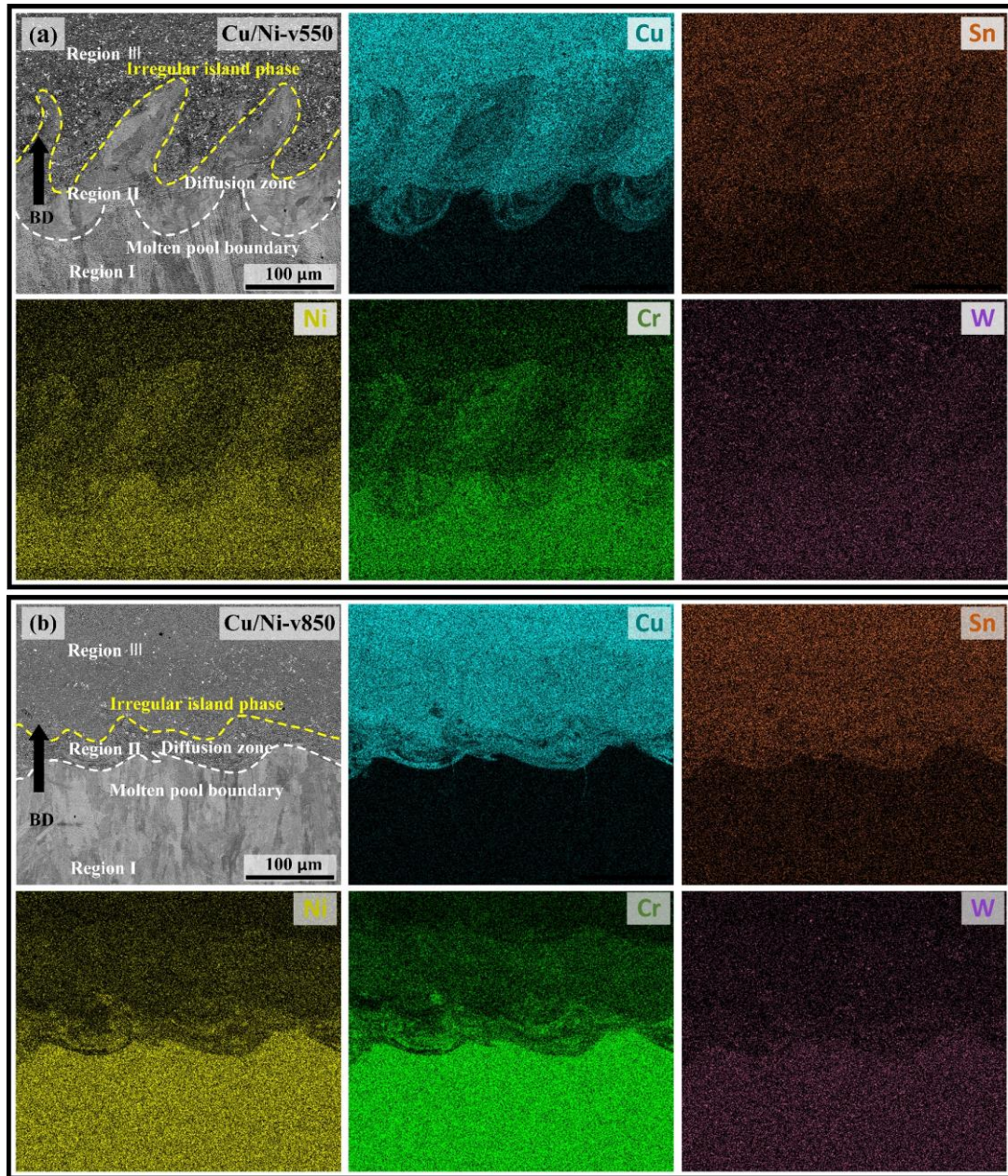


Fig. 5: Scanning electron microscope (SEM) images and energy-dispersive spectroscopy (EDS) mapping analysis of TiB₂-modified Cu/Ni heterogeneous material sample interface: (a) v550 sample; (b) v850 sample

The gradient transition of the main characteristic elements at the interface of Cu/Ni could be observed in the EDX mapping images, indicating that the elements in the two materials were strongly mixed. A slower scanning speed most likely caused deeper laser penetration into the molten pool of the Ni

material, such that a stronger agitation effect occurred, caused by Marangoni convection. Due to the differences in the surface tension and melting point between Cu and Ni, irregularly shaped island phases were formed at the interface, also caused by Marangoni convection. The heights of the irregular island phases of v550 and v850 heterogeneous samples were 107.1 μm and 42.9 μm , respectively. The EDX mapping also showed that the irregular island phase above the interface was enriched in the Ni, Cr and W elements and depleted in the Cu and Sn elements. During the process of high-energy laser melting, the coupled effects of surface tension and buoyancy led to convection inside the molten pool [35,36], inducing the elemental exchange at the interface material. Notably, this Marangoni-induced island phase formed an “interlocked” structure at the interface, which likely has a beneficial effect on the bonding strength of the interface.

3.2. Phase and microstructure constitution

Fig. 6 shows the XRD patterns of the Cu/Ni heterogeneous sample interface. The diffraction characteristic peaks of the hetero-material interface almost completely match those of the single materials Cu and Ni. In general, Cu and Ni have similar lattice constants, which also makes the XRD characteristic diffraction angles of Cu₁₅Ni₈Sn, GH3230 and their heterogeneous materials similar. In contrast to the single-material diffraction peak, however, the characteristic diffraction peak of the heterogeneous material interface showed two diffraction peaks at $2\theta = 31.71^\circ$ and $2\theta = 45.42^\circ$; these two diffraction peaks are not included in the XRD analysis software JADE when finding the phase to match peaks. Considering the semi-quantitative analysis of XRD, further characterisation tests will be conducted in future work to identify the composition of the phase. The particle size of the added TiB₂ particles was at the submicron scale, with 1wt.% content. Due to the limited spatial resolution of XRD, the diffraction peak of the TiB₂ phase was not detected in the

heterogeneous material sample. As shown in the Fig. 6b, a high-magnification BSE image was used to reveal the TiB_2 phase in the fabricated specimens, the added submicron TiB_2 particles were uniformly distributed in the GH3230 matrix. Previous studies also confirmed that the added TiB_2 particles can be uniformly distributed in the matrix materials [24,37].

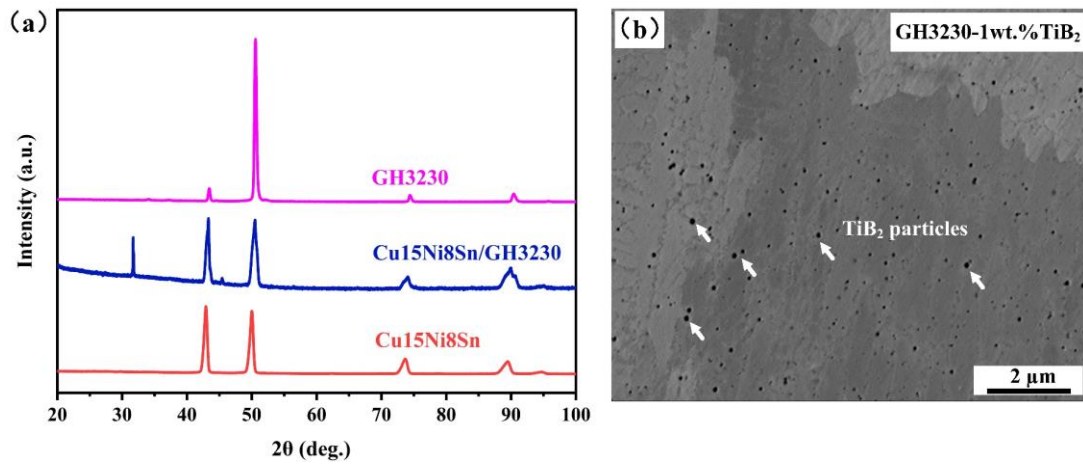


Fig. 6: (a) XRD patterns of TiB_2 -modified Cu/Ni heterogeneous material; (b) Backscattered electron (BSE) SEM images showing the distribution of TiB_2 particles

Fig. 7 shows the SEM images of the interface of TiB_2 -modified Cu/Ni heterogeneous material fabricated via LPBF. When a slow scanning speed of 550 mm/s was employed, cracks formed and were observed in the diffusion area of the diffused elements at the interface of heterogeneous materials, with lengths ranging from 60–200 μm (Fig. 7a). Under a relatively high scanning speed of 850 mm/s, the fusion boundary within the interface bonding region was more obvious, while maintaining superior interface quality (Fig. 7b); the crack length observed under high-magnification SEM was about 25 μm. In addition to the observed zigzag microstructure near the Cu region, a few nearly spherical Ni-rich particles were also observed (Fig. 7a/b). Due to the differences in thermophysical properties and the inconsistency of shrinkage between Cu and Ni, thermal penetration cracks were generated on the Ni side [38].

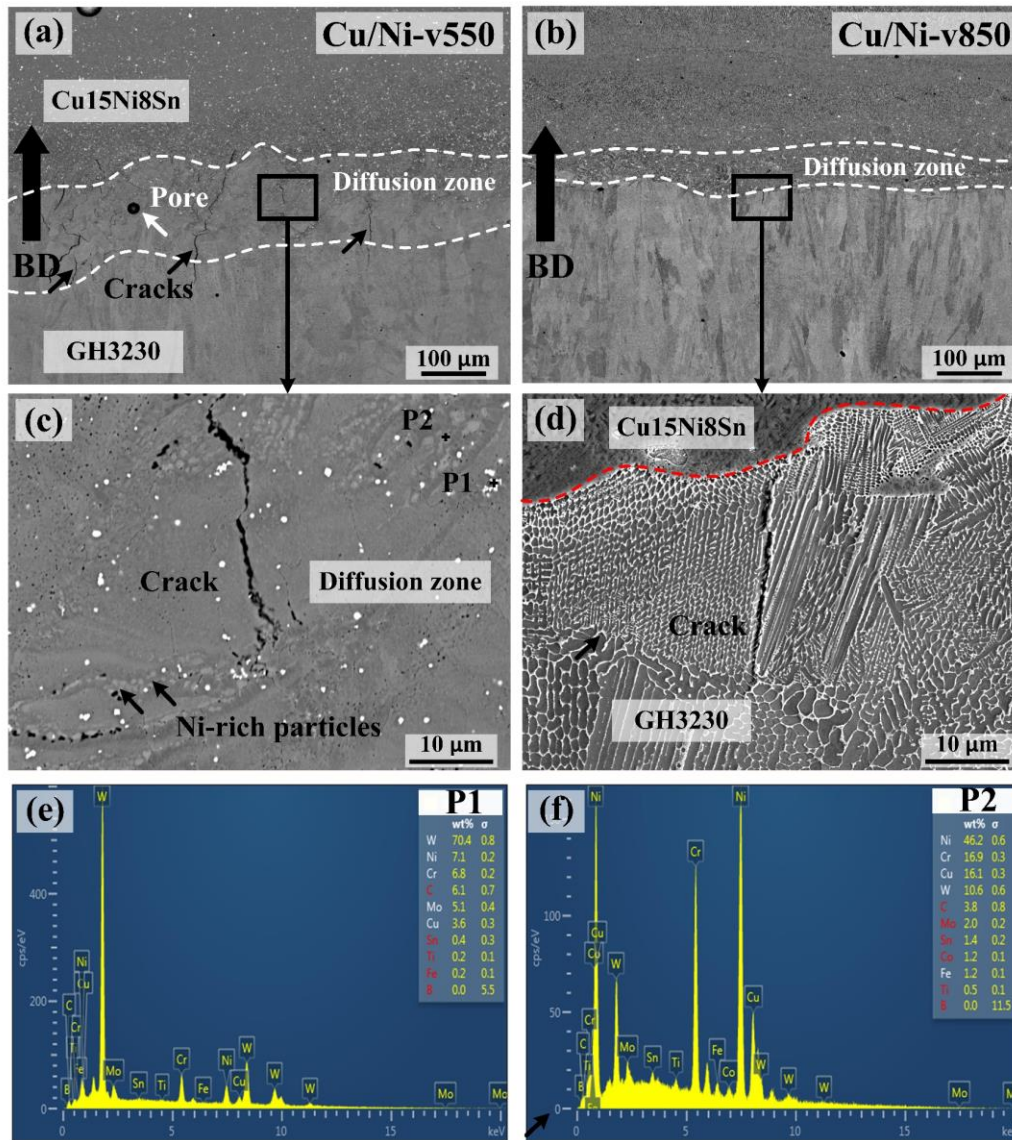


Fig. 7: SEM and EDS images of TiB₂-modified Cu/Ni heterogeneous material samples: (a–d): low- and high-magnification SEM images of v550 and v850; (e & f): EDS point scan of points 1 and 2 in v550

Notably, the high-magnification SEM also indicated a large number of nano-scale white spherical particles generated near the Cu side above the interface. The EDS point scanning image (point 1) showed that the white spherical particles contained a high concentration of W, Cr and C. Previous studies have reported that when GH3230 is heat-treated, a large number of W- and Cr-rich M₆C phases precipitate [39]. Due to the high thermal conductivity of Cu, the continuous thermal cycle generated by laser energy input essentially heat-treats the heterogeneous material interface all the time. GH3230 thus

undergoes intrinsic heat treatment to precipitate M_6C carbides. Considering the point scanning results and LPBF process characteristics, we can reasonably speculate that the two additional diffraction peaks noted in the XRD pattern of TiB_2 -modified Cu/Ni heterogeneous material were attributable to the M_6C carbide phase [39].

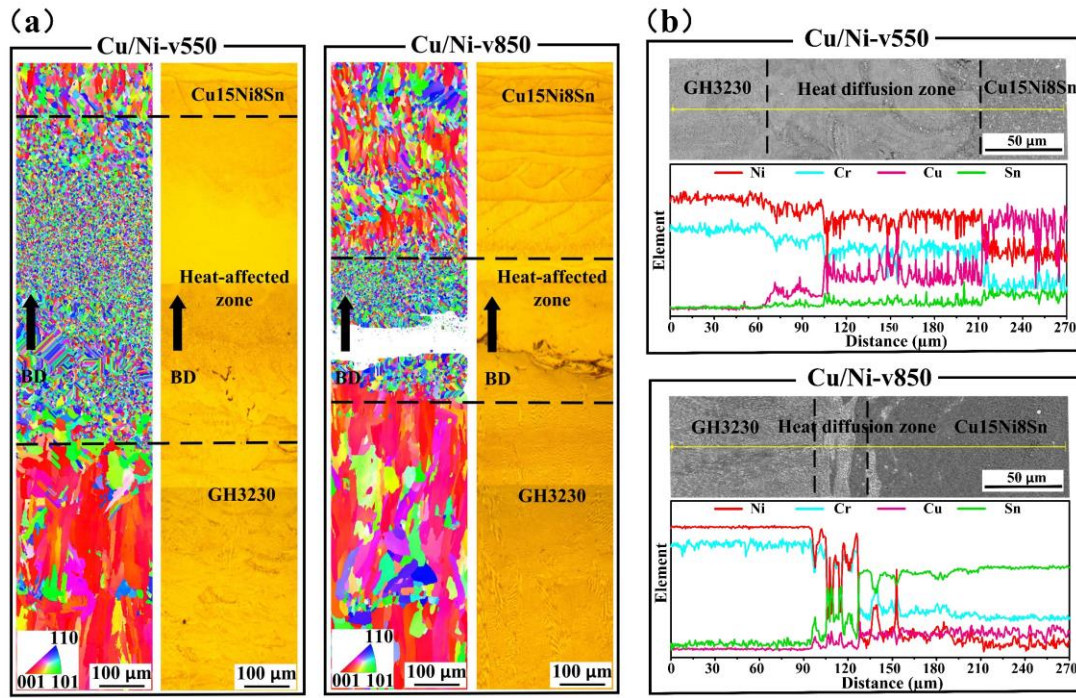


Fig. 8: (a) Grain size distribution images of the Cu/Ni heterogeneous material sample interface (v550 and v850); (b) element diffusion images of the Cu/Ni heterogeneous material sample interface (v550 and v850)

Fig. 8 shows the gradient evolution of grain and element diffusion of the v550 and v850 heterogeneous samples. The grain evolution of the Cu/Ni heterogeneous material along the BD included columnar grains in the Ni region and ultrafine equiaxed grains in the heat-affected zone (HAZ) above the interface-columnar grains in the Cu region. The widths of the HAZ above the interface of the v550 and v850 samples were determined to be about 750 μm and 310 μm, respectively. It is worth noting that the large difference in thermal expansion coefficients between Cu and Ni, together with the non-equilibrium metallurgical process of rapid melting/solidification in the LPBF process generated very high residual stress in the HAZ, resulting in a relatively low

EBSD resolution rate of the HAZ.

The thermal diffusion zone lengths of the v550 and v850 samples were found to be 150 μm and 30 μm , respectively. The difference in the length of the thermal diffusion region was probably caused by the difference in the laser energy density. During the initial layers of the forming interface, the material underneath the interface diffused into the molten pool due to remelting. As the process continued, the depth of the molten pool no longer penetrated the material underneath the interface, eventually leading to the termination of the diffusion zone.

The heterogeneous precipitation includes Ni-rich particles and Cu-rich particles in the HDZ and the ultrafine grains generated in the HAZ. It is known that Cr and W are the main solid solution strengthening elements of GH3230 [40], the remelting process together with Marangoni flow may lead to the diffusion of the elements in GH3230 above the interfacial zone. The intrinsic heat treatment during the LPBF process also caused the precipitation of Cr- and W-rich M₆C phase from the GH3230 side. This unique grain size gradient and heterogeneous microstructure ensured strong bonding of the Cu/Ni samples. It is worth noting that the element gradient transition and M₆C precipitated phase in the diffusion zone may also benefit the metallurgical bonding performance of the Cu/Ni interface.

3.3. Mechanical properties

Fig. 9 shows the microhardness values found in the interfacial bonding region of the Cu/Ni samples (v550 and v850). The average microhardness values within the Ni region in the v550 and v850 samples were determined to be 374.5 HV_{0.2} and 372.1 HV_{0.2}, respectively. Similarly, the average microhardness values within the interface bonding regions in the v550 and v850 samples were determined to be 291.6 HV_{0.2} and 314.3 HV_{0.2}, respectively. The average microhardness values of the Cu area in the v550 and v850 samples were

determined to be 214.4HV0.2 and 204.1HV0.2, respectively. The Vickers microhardness values of the Ni and Cu regions were comparable to those found in the literature [24,25]. The microhardness of the interfacial bonding region was higher than that of the Cu region, which could be attributed to the ultrafine grain strengthening caused by the HAZ and the element diffusion strengthening caused by Marangoni convection.

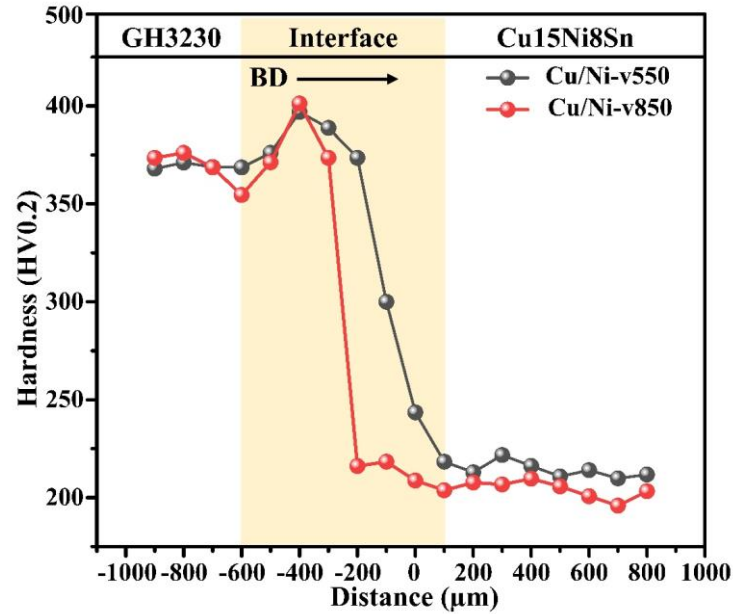


Fig. 9: Microhardness variation curve of the TiB_2 -modified Cu/Ni heterogeneous material interface

Tensile tests were performed on both horizontally bonded and vertically bonded Cu/Ni samples (v550 and v850) to evaluate the interfacial bonding strength of LPBF-fabricated Cu/Ni heterogeneous materials. The stress-strain curves of the tensile tests are shown in Fig. 10. The yield strength (YS), ultimate tensile strength (UTS) and elongation at fracture of the horizontally bonded Cu/Ni–v850 samples were determined to be 674.42 MPa, 867.80 MPa and 12.57%, respectively. Compared to the horizontally bonded single-material Cu–v850 sample, the YS and UTS of the Cu/Ni–v850 sample increased by 55.91% and 52.25%, while elongation was not sacrificed much. As shown in Fig. 10b, the UTS of the LPBF-fabricated Cu/Ni heterogeneous material in this study was about 300 MPa higher than that found in the literature [21][41][42][42][43][44].

The elongation of the horizontally bonded v850 sample was higher than that of the v550 sample. This effect occurred because of the formation of much longer cracks at the interface of the v550 sample. When the tensile force direction was perpendicular to the crack extension direction, the crack propagation accelerated fracture failures during the tensile process. As shown in Fig. 10a, the vertically bonded Cu/Ni tensile specimen fractured in the middle of the Cu region, away from the interface. Both the microstructure and tensile analyses thus indicated that the Cu/Ni heterogeneous material fabricated via LPBF in this study provided excellent interfacial bonding strength.

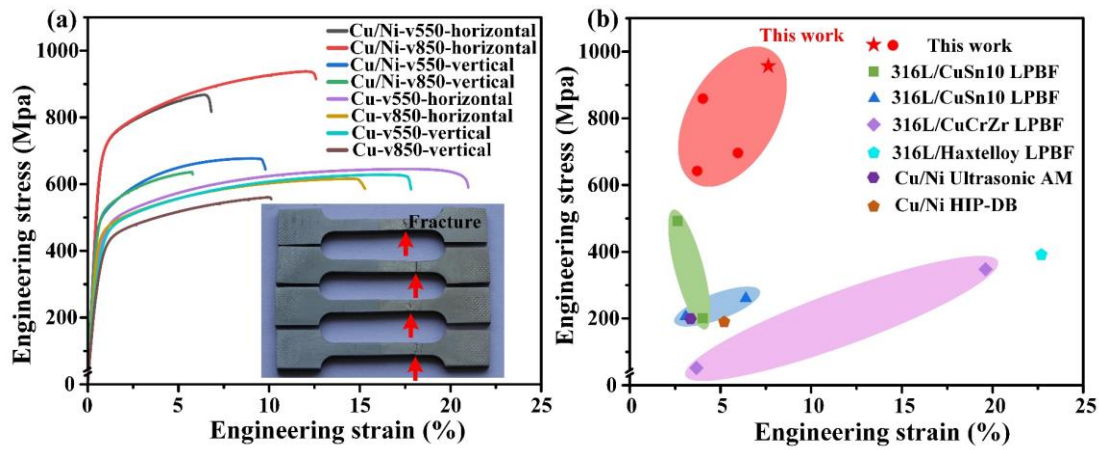


Fig. 10: Stress-strain curves of TiB₂-modified Cu/Ni heterogeneous materials

Fig. 11 shows the fracture morphologies of the horizontally bonded Cu/Ni heterogeneous material. The samples may be divided into three different regions: the Ni region, the interface fusion region and the Cu region. The width of the fusion zone at the fracture interface of the Cu/Ni heterogeneous materials was significantly affected by the laser energy input. As shown in Fig. 11a/c, clear cleavage planes were observed on the fracture surface in the fusion zone. Due to the difference in thermophysical properties between the Cu and Ni materials, the residual stress and the degree of hardening in the interface fusion zone was high. Cracks were initiated at the interface and propagated along the tensile direction, resulting in cleavage planes [45]. This result indicated the typical characteristic of brittle fracturing in the fusion zone. Compared to the horizontally bonded v850 sample, the fracture morphologies of the v550 sample

showed many opened cracks (about 60–120 μm in length), leading to sacrifices in its mechanical properties (Fig. 11b/d).

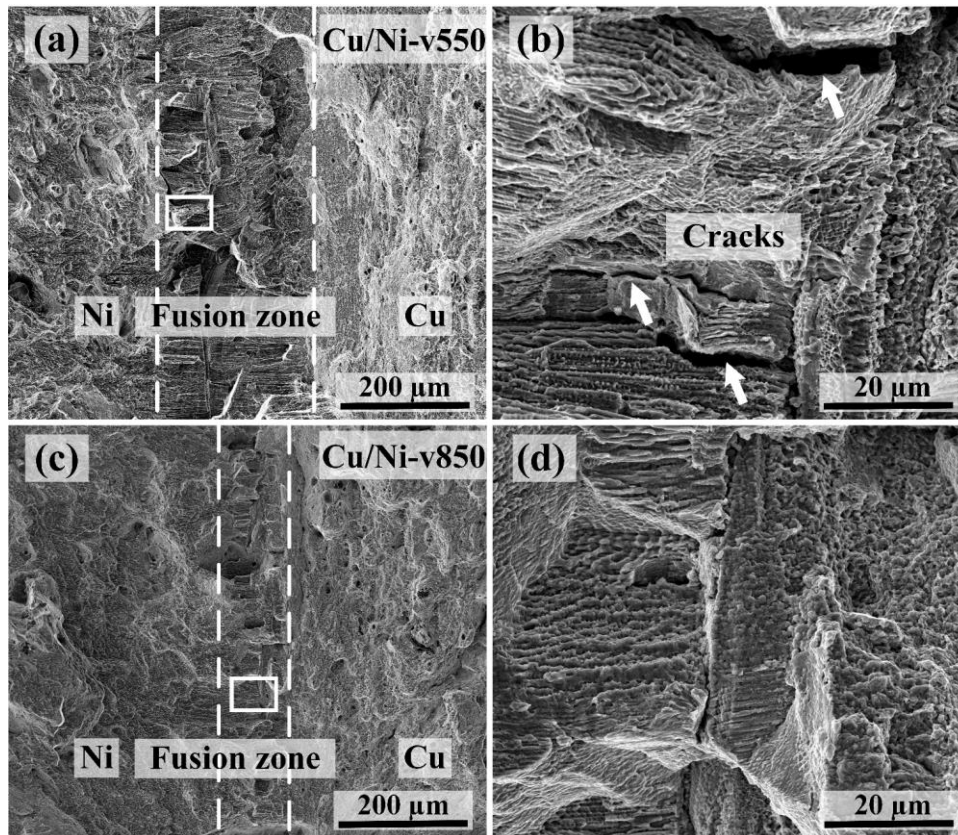


Fig. 11: SEM fracture images of tensile samples of horizontally combined TiB_2 -modified Cu/Ni heterogeneous materials

3.4. Compression behaviours of Cu/Ni lattice structures

Fig. 12 shows the SEM morphologies of different types of Cu/Ni lattice structures. Macroscopic defects such as warpage were not observed in either the strut-based BCC or the surface-based gyroid heterostructure (Fig. 12a/c), indicating that the LPBF-fabricated Cu/Ni lattice structures exhibited proper processability. The powder-sticking defect in the Cu portion was more serious than in the Ni portion. This effect was due to the high thermal conductivity of Cu alloy, which in general allows laser energy to be quickly and efficiently transferred out during shaping. As a result, more heat accumulates on the surface of the lattice structure, causing more of the particle-sticking

phenomenon.

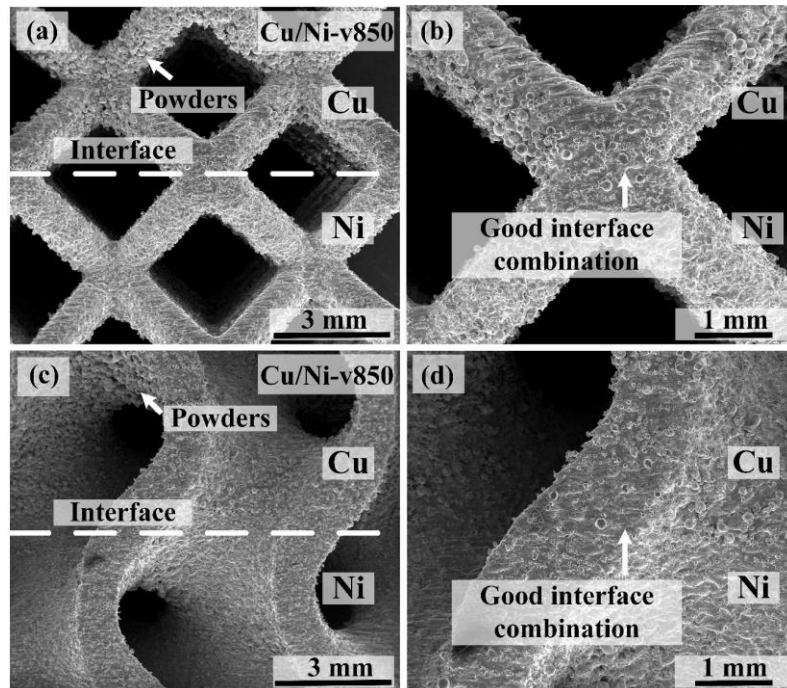


Fig. 12: SEM images of different types of TiB₂-modified Cu/Ni heterogeneous lattice structures fabricated via LPBF

Fig. 13 shows the stress-strain curves of the strut-based BCC and surface-based gyroid heterogeneous lattice structures under different compression directions. The compression curves of the Cu/Ni heterostructure mainly included three stages. In stage I, the stress increases linearly with the strain, which is the linear elastic stage of the lattice structure. Stage II, the fluctuation collapse plateau stage, is related to the yield deformation of the Cu and Ni portion. Stage III represents the densification of the whole Cu/Ni lattice structure. The Cu/Ni lattice structure (BCC-v and gyroid-v), when compressed perpendicular to the BD, exhibited a higher elastic modulus and peak value of the first stress plateau. The stress plateau of the Cu/Ni lattice structure that was compressed parallel to the BD showed a gradual rise, indicating that the Cu/Ni lattice structure that was compressed parallel to the BD exhibited layer-by-layer compaction deformation characteristics.

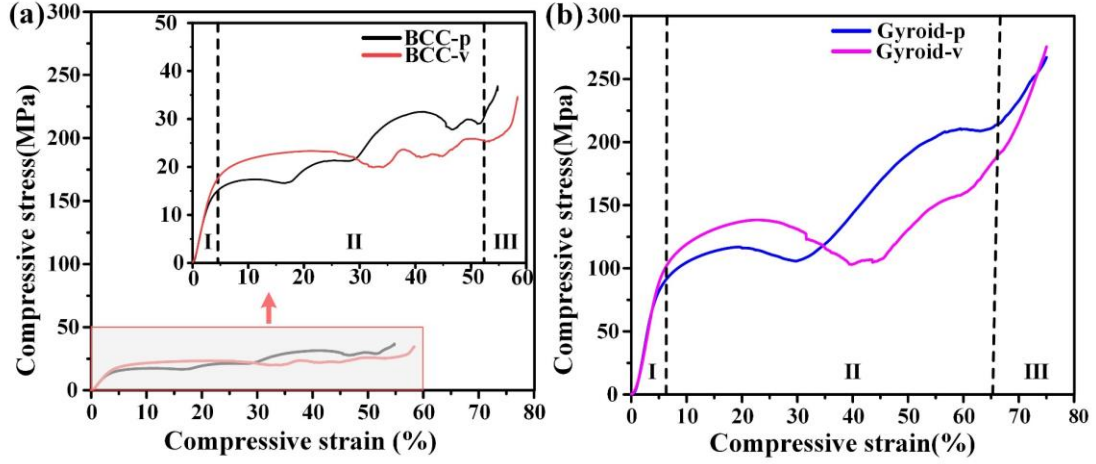


Fig. 13: Compressive stress-strain curves of different types of TiB₂-modified Cu/Ni heterogeneous material lattice structures: (a) BCC lattice; (b) gyroid lattice

The energy absorption characteristic is an important evaluation index for the application of the lattice structure within the aerospace and automotive fields, mainly including energy absorption and energy efficiency. The energy absorption per unit volume (W_v) is calculated as follows [46]:

$$W_v = \int_0^{\varepsilon} \sigma(\varepsilon) d\varepsilon \quad (1)$$

where ε represents the strain, and $\sigma(\varepsilon)$ represents the stress relative to ε strain. The energy efficiency (η) is the ratio of the absorbed energy to a given strain (ε) divided by the corresponding stress ($\sigma[\varepsilon]$), which may be expressed as follows [47]:

$$\eta = \frac{\int_0^{\varepsilon} \sigma(\varepsilon) d\varepsilon}{\sigma(\varepsilon)} \quad (2)$$

This equation is often used to determine the point (ε_d) where η reaches its maximum; the point represents the start of lattice structure densification during compression deformation [47].

Fig. 14 shows the energy absorption characteristics of BCC and gyroid lattices under different compression directions. Under low strains (< 40%), the Cu/Ni lattice structure was able to absorb more energy when compressed perpendicular to the BD. This result occurred because the Cu/Ni lattice structure exhibited higher stiffness and strength under this condition. The

energy absorption efficiency curve is mainly divided into two stages. The first is the linear growth stage, which is mainly related to the yield platform; the linear growth vertex corresponds to the end position of the yield platform. The second stage occurs after the energy efficiency reaches its maximum peak, which decreases with increasing strain. The strain corresponding to the highest vertex of the energy absorption efficiency curve represents the onset of lattice structure densification during compression deformation. We may therefore conclude that the Cu/Ni lattice structure that was compressed perpendicular to the BD offered higher energy absorption efficiency compared to the parallel condition.

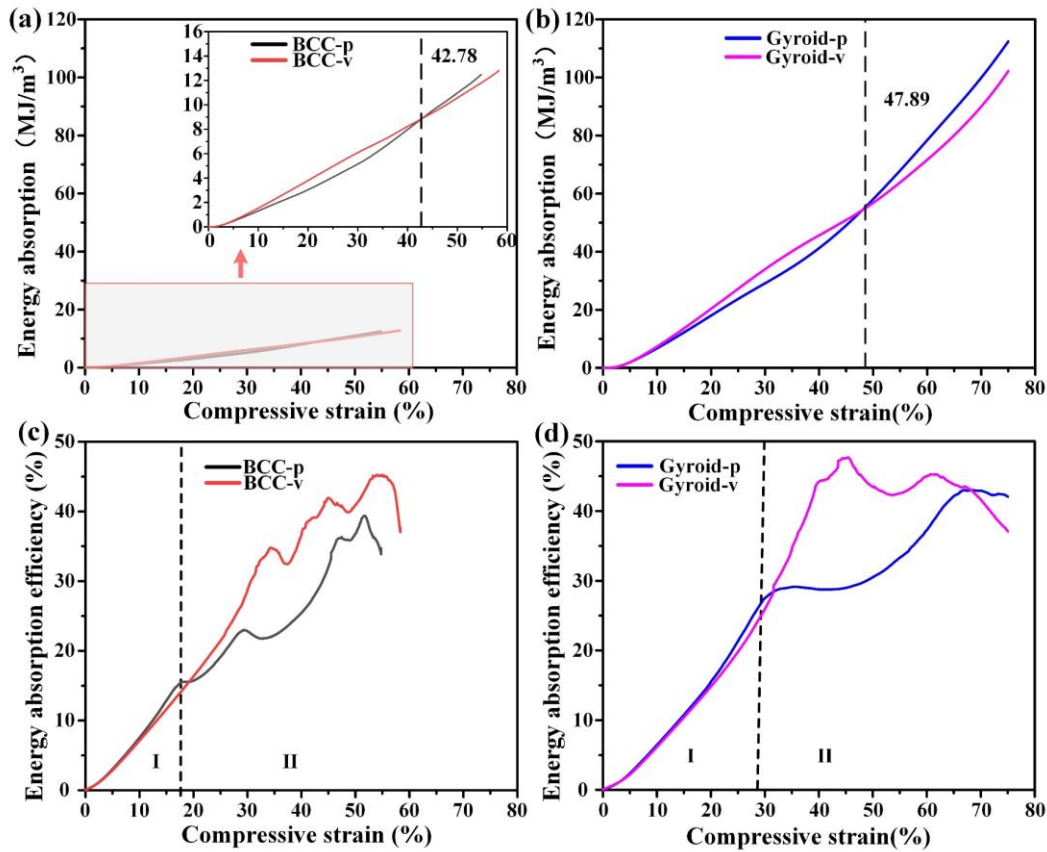


Fig. 14: Energy absorption characteristics of different types of TiB₂-modified Cu/Ni lattice structures under different compression directions: (a) energy absorption of body-centred cubic (BCC) structure; (b) energy absorption of gyroid structure; (c) energy absorption efficiency of BCC structure; (d) energy absorption efficiency of gyroid structure

Fig. 15 shows fracturing and collapse photographs of the BCC and gyroid lattices captured at several levels of overall strain (10%, 20%, 30%, 40%, 50%

and 30%) under different compression directions. Both the BCC-p and gyroid-p lattice structures exhibited layered compaction compression deformation behaviour when compressed parallel to the BD. At lower strain conditions, the Cu portion deformed into the yield stage first, while the Ni portion was still in the elastic deforming stage. When the strain was increased to 30%, the Cu portion became compacted and entered the densification stage. When the stress reached the yield strength of Ni, the Ni portion started to deform. Once most of the struts had made contact with other struts, the stress rose rapidly, and the Ni portion also became compacted as it entered the densification stage.

Compression of the BCC-v and gyroid-v lattice structures perpendicular to the BD also exhibited similar deformation behaviours. At lower strains, the Ni portion exhibited the stiffness of the overall structure. When the strain was increased to 30%, BCC-v and gyroid-v began to show inhomogeneous deformation characteristics along the left-diagonal 45-degree angle shear fracture (Fig. 15). When the compressive strain was increased to 30%, the lattice structures showed the asymmetric deformation characteristic of shear fracturing along the left diagonal of 45°. This result occurred due to the asymmetric material properties of the cross-section of the lattice structure parallel to the direction of compression. When the stress was high, the deformation of the Cu portion on the cross-section was large, such that the fracture failure of the Cu portion occurred first. This behaviour also explains the stress-drop at the end of the yield plateau of the BCC-v and gyroid-v curves (Fig. 13).

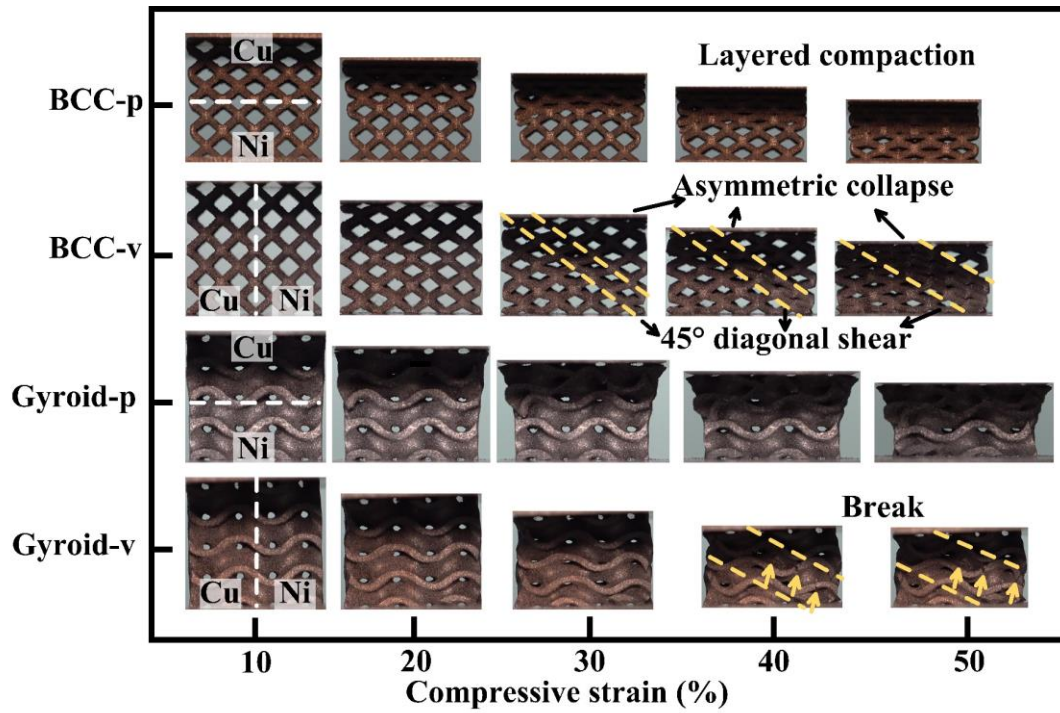


Fig. 15: Deformation processes of the lattice structures of BCC and gyroid heterogeneous materials under different compressive directions

3.5. Thermal diffusivity of Cu/Ni heterogeneous materials

Fig. 16 shows the thermal conductivity of single-material Ni and Cu, as well as Cu/Ni heterogeneous material, all at room temperature. The thermal conductivity values of single-material Ni and Cu/Ni heterogeneous materials were measured to be 11.58 ± 0.36 W/mK and 24.26 ± 0.94 W/mK, respectively. Previous studies have demonstrated that the thermal conductivity of metals is affected by imperfections within the material [48].

At room temperature, the thermal conductivity of Cu/ Ni heterogeneous material is 109.5% higher than that of single-material Ni. In the present study, such conductivity also implied that the LPBF-prepared Cu/Ni heterogeneous material exhibited a good metallurgical bonding interface.

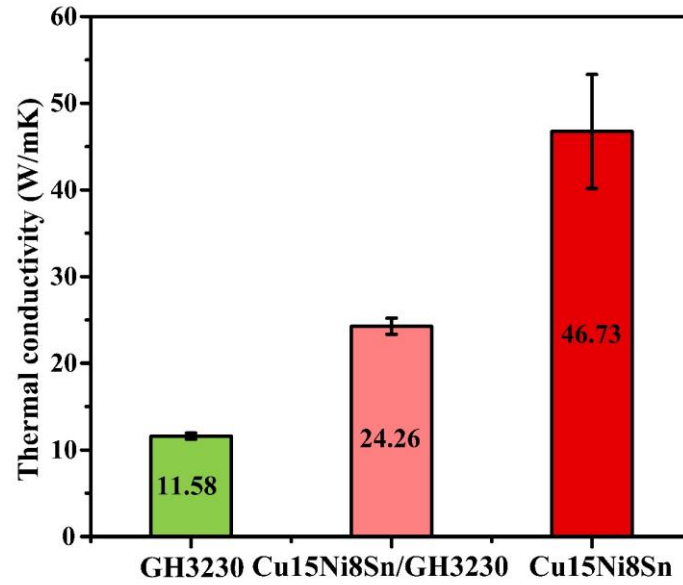


Fig. 16: Thermal conductivity of Ni, Cu and their heterogeneous materials fabricated via LPBF at room temperature

4. Discussion

4.1. Pore formation mechanism of interface defects

Interface defect control is crucial in manufacturing multi-material components using additive manufacturing. The open pore defect is a key defect in parts manufactured by the LPBF process. Because interface defect' size, morphology and position are the main factors to affect their mechanical properties, understanding the pore formation mechanism is crucial to eliminating pore formation during the LPBF of heterogeneous materials. Qi et al. [49] divided the melting modes into the keyhole, transition and conduction modes. Fig. 17 shows a schematic diagram of the thermal diffusion zone of the Cu/Ni interface at different scanning speeds. When the laser processed the first layer of Cu15Ni8Sn powder that was deposited on the solidified GH3230, the molten pool includes both remelted GH3230 and Cu15Ni8Sn. The difference in surface tension and melting point between Cu and Ni exacerbated the convection of the molten pool [35][50]. Because the LPBF process exhibited

very high cooling rates (10^6 K/s) and high actual dynamic undercooling [51], the miscible Ni-Cu materials can be easily subcooled to the metastable miscibility gap and eventually separated into Ni-rich and Cu-rich phases [52]. Compared to the Cu-rich phase, the aggregated Ni-rich phase offered a lower critical nucleation energy for heterogeneous nucleation and solidified first within the heat diffusion zone [53]. When the laser processed the Cu15Ni8Sn powder, the diffusion degree of GH3230 and Cu15Ni8Sn caused by Marangoni convection tended to slow down, resulting in a reduction of the Ni-rich phase.

Generally, an extensive high laser energy input value results in a deep and narrow molten pool morphology, called the keyhole mode. The molten pool depth to width (D/W) ratio, also known as the molten pool cross-sectional aspect ratio, can be used to identify the keyhole mode ($D/W \geq 0.55$) and conduction mode ($D/W < 0.25$) [36]. In the present study, the D/W ratio of the molten pool at the Cu/Ni interface of the v550 sample was calculated to be ~ 0.58 , indicating that the keyhole mode occurred during LPBF when the laser scanning speed was less than 550 mm/s.

Under the keyhole mode, the Marangoni force pushes Cu into the Ni region within the melt pools, while the high recoil pressure maintains a deep keyhole [36]. A greater penetration depth (molten pool depth) thus is achieved in the pre-solidified Ni [20], resulting in a deeper element diffusion zone (~ 200 μm) at 550 mm/s (Fig. 5). The multiple reflections of the laser beam in the keyhole also improve the laser absorption rate of the LPBF-formed Cu alloy as well as promoting material migration and element exchange between Cu/Ni, which is beneficial for interface bonding [54]. An overly high energy input, however, may lead to pinch-off at the bottom of the keyhole, which may remain in the diffusion region (Fig. 3) [55]. A low laser energy input can result in the “conduction mode,” with a semicircular melt pool geometry (at 850 mm/s, with a D/W ratio of melt pools of ~ 0.21 ; see Fig. 5). The wide, shallow molten pool results in a smaller penetration depth and a shorter element diffusion zone (Fig. 5).

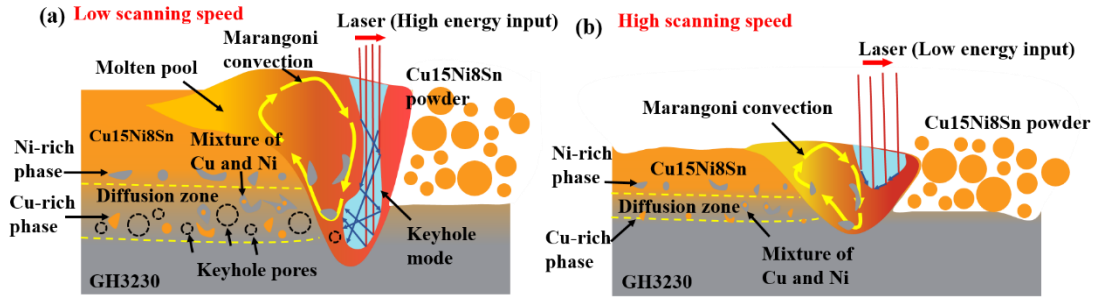


Fig. 17: Schematic diagram of the thermal diffusion zone of the Cu/Ni interface at (a) low and (b) high scanning speeds

In summary, the keyhole mode, caused by high energy density, can increase the laser absorption rate, which promotes Marangoni turbulence and increases the area of the Cu-Ni mixing zone, thereby promoting the mixing of dissimilar materials and strengthening the interfacial bonding [20,54]. The keyhole mode, however, may cause keyhole-induced pores at the interface.

4.2. Crack formation mechanism of interface defects

The crack is another kind of defect found at the interface of heterogeneous materials fabricated via LPBF. Marangoni convection is the main driving force for mass/heat transfer at the interface of LPBF-fabricated heterogeneous materials [36,56]. The surface tension of the liquid Cu and Ni in the molten pool during melting creates Marangoni convection. The higher energy input causes more molten Ni to be stirred into the molten pool, resulting in the formation of a solid solution rich in Cu and Ni in the molten pool. Owing to the extremely high cooling rate (up to 10^6 – 10^8 K/s [57]) of LPBF and the fact that the melting point of Ni is higher than that of Cu, the Ni stirred into the molten pool may solidify. The solidified phase sequence in the molten pool thus goes from the Ni-enriched phase to the Cu-enriched phase [20]. Due to the mismatch of thermophysical properties between Cu and Ni (Fig. 16), the thermal expansion of Cu is higher than that of Ni. The solidification and shrinkage of the Cu-enriched phase thus causes the Ni in the thermal diffusion zone of the interface to become torn, generating vertical cracks [58].

According to Chen et al. [38], four types of cracks are generated at the interface of copper/steel heterogeneous materials fabricated via LPBF. The types of interfacial cracks of Cu/Ni fabricated via LPBF found in the present study were similar to those noted by Chen et al. [38]. Extensive laser energy input leads to three types of cracks at the Cu/Ni interface. Type I consists of cracks inside the diffusion zone (Fig. 7a/c), while type II cracks extend from the diffusion zone to the Ni substrate (Figs. 3 and 7) and also include copper penetration cracks (Fig. 5b). Type I/II cracks are thermal solidification cracks. The high- and low melting point phases form a solid-liquid coexisting “mushy zone” during melting. The low melting point phase shrinks and generates tensile stress during solidification, causing solidification cracks [59]. This kind of cracking may be eliminated by properly controlling the energy input. Type III cracks are penetrating microcracks, which are produced by the liquid copper infiltration of the grain boundaries of the nickel matrix under the action of capillary forces [60]. Penetration microcracks cannot be eliminated by optimising process parameters, but their length can be shortened by process parameter optimisation.

5. Conclusion

In this study, Cu/Ni heterogeneous materials and their lattice structures – including body-centred cubic (BCC) and gyroid lattices – were fabricated via the laser powder bed fusion (LPBF) technique. This paper has disclosed the relationships of various processing parameters on the processability, interfacial metallurgical defects and thermophysical properties of LPBF-fabricated Cu/Ni heterogeneous materials. The following conclusions may be drawn from this study's results:

1. An appropriate laser energy input ($< 100 \text{ J/mm}^3$) can eliminate interfacial pore defects and suppress crack scales, thereby achieving high density and

strong interfacial metallurgical bonding of Cu/Ni heterogeneous materials. Cracks at the Cu/Ni interface include solidification cracks and copper penetrating cracks. Solidification cracks are generated by the thermal stress of the Cu/Ni material fabricated via LPBF, which can be suppressed and eliminated by reducing the input energy.

2. The ultimate tensile strength (UTS) of the horizontally bonded Cu/Ni heterogeneous material was measured to be 867.80 MPa. Compared with single-material Cu, the UTS of the horizontally bonded Cu/Ni tensile specimen was found to have increased by 55.25%, while the elongation had not significantly reduced. The vertically bonded Cu/Ni tensile specimen fractured in the middle of the Cu region away from the interface, indicating strong interfacial bonding strength. This excellent interfacial bonding strength was mainly attributable to the grain boundary strengthening and fine grain strengthening provided by the ultrafine grains in the interface thermal affected zone, as well as the element diffusion caused by Marangoni flow.
3. The results of quasi-static compression experiments indicated that the Cu/Ni lattice structure could exhibit higher yield strength and ultimate compressive strength when compressed perpendicular to the build direction (BD), but the lattice exhibited uneven fracturing along the 45° diagonal. Compared to compression perpendicular to the BD, the Cu/Ni lattice structure was found to have absorbed more energy for protection when compressed parallel to the BD.

Data Availability

The raw/processed data required to reproduce these findings cannot be shared at this time as the data also forms part of an on-going study.

CRedit authorship contribution statement

Jian Gao: Conceptualization, Methodology, Writing – original draft. **Quanquan Han:** Conceptualization, Funding acquisition, Writing – review & editing. **Shwe Soe:** Resources, Investigation. **Liqiao Wang:** Visualization, Data curation. **Zhenhua Zhang:** Investigation. **Han Zhang:** Data curation. **Jun Song:** Data curation. **Yue Liu:** Methodology. **Rossitza Setchi:** Writing – review & editing. **Shoufeng Yang:** Methodology, Writing – review & editing.

Declaration of Competing Interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

Acknowledgements

The authors appreciate the financial support from the National Natural Science Foundation of China (Grant No. 52005295), the Natural Science Foundation of Shandong Province (Grants No. ZR2023YQ053, ZR2022ZD08, ZR2022ME175, ZR2023QE287), Shenzhen Science and Technology Program (Grant No. GJHZ20220913142607014) and the Beijing Natural Science Foundation (Grant No. 3232007).

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