Tailored microstructure and robust joint of Inconel 718/316L bimetallic multi-material fabricated by selective laser melting

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Abstract

In this study, selective laser melting (SLM) technology was used to fabricate Inconel 718/316L bimetallic multi-material with robust bonding strength, and a deep insight into the microstructural morphology, mechanical property and its strengthening mechanism of the joint was taken. The transition region with a wide of approximately 150 μm was defined and showed a dominating columnar region which was embedded in dispersed Laves phase occupied the molten of Inconel 718 closed to the joint. X-ray diffraction (XRD) pattern detected the strong peaks of γ', γ'' and a weaker peak of d phase precipitates. Electron backscatter diffraction (EBSD) analysis showed that a distinct grain coarsened region existing and Inconel 718 region had a strong fabric texture with a <001>// Z (BD) orientation. The shear strength of the as-built joint was calculated to be 449.5 MPa, which was comparable to the nickel/steel multi-materials formed by other traditional processing technologies.

Introduction

Bimetallic multi-materials with excellent interfacial strength formed by specific metallurgical combination process, exhibited greatly improved integral physicochemical and mechanical properties without losing the remarkable performance of every single alloy [1–2]. Therefore, the bimetallic multi-materials have attracted widespread attention in the automotive, shipbuilding, electronics and military industries [3–6].

Typical technologies involved in bimetallic multi-materials fabrication could be broadly divided, according to the status of both metals during the processing period, into three categories: solid-solid [7–9], liquid-liquid [10–11] and liquid-solid [12–13] combination, both of which have been successfully applied to manufacture components of bimetallic multi-materials with the most known metal couples of steel-copper [14], steel-aluminum [15], steel-titanium [16], copper-aluminum [17] and steel-nickel [18]. The process of solid-solid combination was relatively complicated and had certain limitations on the shape of the profile, suitable to the preparation of plate and wire materials generally, and the final parts usually showed huge internal stress. Li et al. [19] produced SS400 steel/1050 aluminum multi-material parts by hot rolling and further proved the bimetallic components possessed the strongest bonding strength when the processing temperature was 400°C. Hou et al. [20] adopted the fiction stir welding technology to form 6061 aluminum/pure copper bimetallic joints and obtained the relatively high maximum tensile strength of 152 MPa. Liquid-liquid combination required high-quality equipment, and the convection of the two metal liquids was difficult to control during the compounding process, resulting it hard to be widely used in actual production.

Liquid-solid combination technology took advantage of solid metallurgical bonding, caused by the inter-atomic diffusion happened when liquid phase metal solidified on the contact surface of solid phase metal substrate, exhibiting more superior bonding performance than the ordinary mechanical combination. And notably, liquid-solid combination had lower requirement towards profile and formability of raw materials and could be applied to the multi-material fabrication in many alloys. Traditional liquid-solid methods involved extensively in liquid-solid casting composite method, and Liu et al. [21], for
example, successfully fabricated Mg/Al bimetallic multi-material by liquid-solid compound casting with the maximal bending strength of 23.06 MPa.

In addition to the liquid-solid casting composite method, selective laser melting (SLM), an appealing additive manufacturing (AM) technology, was based on layer-by-layer accumulation, melting, and solidification of powder particles to form metal parts with complex shapes and fine structures [22–23]. Specifically, powder particles were melted into liquid droplets, which further converged into separate molten pools under the action of high-energy laser beams and existed in the solidified metal surface of the previous layer. Total progress of element diffusion and new phase generation between molten pools and the previously solidified layer could be considered to be a liquid-solid combination spontaneously [24–25]. Recently, researches on fabricating bimetallic multi-materials by SLM has received great attention due to its potential to form complex, large-scale structure components with far-ranging raw materials, including metal, ceramic, polymer compound and so on. Representative works like, Chen et al. [14] produced firmly bonded 316L/CuSn10 lattice structure multi-materials with the size of 5 × 5 × 5 cm; Koopmann et al. [26] reported a multi-material process allowing for realization of a zirconia-alumina ceramic coating on a steel substrate to introduce a novel ceramic/steel sandwich system by SLM. Bartolomeu et al. [27] used the SLM and pressure assisted injection techniques to obtain Ti6Al4V-PEEK multi-material structures and guarantee a substantial improvement on the wear resistance of Ti6Al4V-PEEK hip implants.

Nickel-based superalloys were often valued for their excellent tensile strength, creep properties, and resistance to thermal corrosion and oxidation at high temperatures [28–29]. Austenitic stainless steels (SS) featured high toughness, excellent corrosion resistance, room temperature (RT) mechanical strength and relatively low cost [30–31]. As a result, Nickel/steel bimetallic multi-materials will satisfy the severe and diverse working condition while reducing the raw materials cost of nickel alloys. Hinojos et al. [32] explored the feasibility of joining Inconel 718 with 316L Stainless Steel by utilizing EBM, further suggesting the potential to manufacture nickel/steel multi-material components applied in nuclear applications. Su et al. [33] reported an Inconel-steel functionally graded materials (FGMs) with linear composition gradients of nickel (0%, 5%, 10%, 20% and 100%) fabricated by LMD additive manufacturing. However, relevant research about SLM Inconel/steel bimetallic multi-materials has rarely been reported, restricting further expansion of multi-material systems.

In this research, the optimal SLM parameter group in fabricating Inconel/steel multi-materials were found through the optimization experiment of the forming process window. Then the multi-material samples were formed to reveal the interface bonding characteristics and comprehensive mechanical properties, such as, hardness variation, tensile and flexural strength, and shear properties. This study will provide technical and theoretical support for the application of Inconel/steel multi-materials formed by SLM.

Materials And Tests
In the present study, the SLM Inconel 718/316L multi-material specimens were produced by using self-developed HK M125 equipment (Wuhan Huake 3D Technology Co., Ltd., China) equipped with a maximum 500 W single mode fiber laser. Based on the previous research [14, 22–23], the optimal SLM process parameters of both 316L and Inconel718 for the following optimization tests were shown in Table 1.

Table 1
SLM process parameters of single 316L and Inconel 718 alloy.

| Item                      | 316L | Inconel 718 |
|---------------------------|------|-------------|
| Laser power (W)           | 320  | 300         |
| Laser scanning speed (mm/s)| 650  | 650         |
| Scanning space (mm)       | 0.14 | 0.11        |
| Layer thickness (mm)      | 0.05 | 0.05        |

During the SLM experiments, the nominal chemical composition of the gas atomized spherical 316L and Inconel 718 alloy powders were given in Table.2 and Table.3, which were chosen to be the row materials with an average size of about 33.6 µm and 41.3 µm, respectively.

Table 2
Chemical composition of the 316L powder.

| Element | Cr   | Ni   | Mo | Mn | Si  | others | Fe | Content (wt. %) |
|---------|------|------|----|----|-----|--------|----|-----------------|
|         | 17.52| 12.43| 2.6| 0.06| 0.92 | 0.24   | Bal.|                 |

Table 3
Chemical composition of the Inconel 718 powder.

| Element | Cr   | Fe   | Mo | C  | Co  | Al  | Ti  | others | Ni | Content (wt. %) |
|---------|------|------|----|----|-----|-----|-----|--------|----|-----------------|
|         | 19.43| 18.52| 3.37| 0.04| 0.04| 0.51| 1.06| 5.19   | Bal.|                 |

The vector angle of the laser scanning direction between the layers was 67 degrees. The building platform was pre-heated to 200°C. The process was conducted under constant flow of ultra-high purity Ar gas to minimize the oxygen contamination. To investigate the mechanical properties with regard to different forms of combination, tensile specimens were built vertically and horizontally as shown in Fig. 1(a), and Fig. 1(b) was the detailed dimensions of the tensile specimen (horizontally combination (Type (1))) and vertically combination (Type (2))).

The SLM Inconel 718/316L multi-material samples were cut into 10 mm × 10 mm × 6 mm blocks by wirecutting using EDM process. The bonding interface of the samples was polished with silicon carbide sandpaper of 240, 400, 800, 1000, 1500 and 2000 mesh in turn. Finally, a diamond suspension (Shanghai Laboratory Testing Technology Co., Ltd.) with a particle size of 2.5 µm was used for polishing. The
polished samples were analyzed by XRD 7000S X-ray diffractometer (Shimadzu, Japan) with a Cu tube at 40 kV and 30 mA. The diffraction angle of 2θ varied from 20° to 100° with a step size of 10°/min. The metallographic etching solution (60% vol. glycerol, 33.3% vol. HCl and 6.7% vol. HNO₃) was used to electrolytic etch the samples for 5 to 10 seconds. The metallurgical characteristics in the bonding area were observed through an optical microscope (OM, Olympus BX60). A field emission scanning electron microscope (SEM, JSM-7600F, JEOL, Japan) was adopted to observe the microstructure and fracture morphology. The electron backscatter diffraction (EBSD) aiming at characterizing the crystallographic orientations was carried out under an HKL Nordlys orientation imaging microscope system (Oxford, Oxford Instruments, UK), in which the step size was set to 1 micron, and data analysis processing was performed using HKL Channel 5 software.

The overall density of multi-material samples were measured according to Archimedes principle. The variation trend of Vickers hardness in the interface transition area was measured using a Vickers hardness tester (Wolpert Wilson Instruments 432 SVD, China) at a set load of 29.4 N. A Zwick/Roell Z020 universal testing machine was used to perform the tensile test under the calibration conditions at room temperature, load of 90 KN and strain rate of 1 mm/min. The three-point bending test was operated on the Zwick/Roell Z020 universal testing machine, in which the speed of the indenter was set to 1 mm/min. The shear test was carried out on the electro-hydraulic servo universal testing machine controlled by ZY-213 microcomputer, and the elevating rate of the indenter loading force was set to 300 N/m.

**Result And Discussion**

**3.1 Parameter optimization of SLM**

To obtain the best joint bonding properties, it is necessary to set up a multi-parameter window to screen out the best process parameters (laser power and scanning rate) combination of the two alloys. Based on Table 3, the parameters for forming 316L alloy were unified (320 W and 650 mm/s), while a total of nine sets of parameters (laser power: 250 W, 300 W and 350 W; scanning speed: 600 mm/s, 650 mm/sand 700 mm/s, respectively) were set to fabricate Inconel 718 components. It could be seen from Fig. 2(a) that the specimens under all nine combinations could be formed into a well-bonded multi-material, indicating the wide forming compatibility of steel and nickel alloys. Therefore, forming the respective dense parts became a priority, and finally, the parameter combination when forming the experimental samples (Fig. 2(b)) was selected as: 320 W and 650 mm/s of 316 alloy; 300W and 650 mm/s of Inconel 718.

**3.2 Phase and microstructure analysis**

Figure 3 depicted the XRD pattern of the SLM multi-material sample. There was almost exact matching of the characteristic peaks of the γ/γ'/γ'' phase. Besides, the weak diffraction (2θ = 46.3°) marked in the (211) plane could be identified in the specimen, which confirmed the existence of precipitated phase in
the nickel alloy. Weak diffraction peaks of carbides (NbC) were also detected. At the same time, the strong diffraction peaks corresponding to the \( \gamma \)-Fe crystal planes (111), (200) and (220) could be clearly identified in the sample. In addition, no peaks of the martensite phase were detected.

Figure 4 showed the OM and SEM images of microstructure on XZ plane for the as-built sample. Figure 4(a), the OM image tested on the XZ plane of polished specimen, exhibited the combined effect and the metallurgical defects of 316L and Inconel 718. The OM image (Fig. 4(b)) after metallographic corrosion showed that the XZ plane could be divided into three distinct regions (distinguished by white dotted lines): the 316L region, the transition region and the Inconel 718 region, where the thicknesses of transition region accounted for approximately three scanning layers, which could be estimated as 150 µm. Overall, the as-built sample possessed solid metallurgical bonding as the transition area with clear boundaries existed between 316L and Inconel 718 region. Subtle pores could be observed in the 316L region while cracks were the dominant defects occurred in the transition region (defined as continuous and separate cracks, respectively), which was usually attributed to the insufficient relief of residual thermal stress [34–36].

Figure 4(c) and Fig. 4(d) were the typical SEM images corresponding to area A and area B marked by green dotted line in Fig. 4(b). From Fig. 4(c), columnar dendrite and cellular dendrite patterns occupied the separate molten pool of Inconel 718 in the transition region while the proportion of columnar dendrite far exceeded that of cellular dendrite. At the same time, columnar dendrite region \( \text{\textbullet} \) and \( \text{\textbullet} \) with different growth direction, which turned to be perpendicular to the boundary of molten pool, could be explained by the preferential growth of grains under different temperature gradients [37]. Moreover, common Laves phases of Inconel-based alloys were also found in columnar region with the long chain shape [38]. The formation of the Laves phase consumed the useful alloying element Nb in the matrix, thereby inhibiting the precipitation of the strengthening phases \( \gamma' \) and \( \gamma'' \) [39]. Secondly, the brittle Laves phase provided conditions for the nucleation and growth of cracks under the action of residual stress or other external stresses, resulting in a significant decrease in the tensile properties, fracture properties and fatigue properties of the shaped parts [40]. In addition, the formation of low-melting eutectic Laves phase between dendrites was likely to cause thermal cracking in the additive manufacturing process.

An elevated percentage of cellular dendrite could be observed in Fig. 4(d), which totally located in Inconel 718 region. An overlapping region connected the cellular region \( \text{\textbullet} \) and \( \text{\textbullet} \), with the average grain size calculated to be 1.2 and 0.7 µm, respectively. It had been reported that the fine equiaxed cellular dendrites were beneficial to reduce the segregation of Nb and form separated Laves phase particles, thus enhancing \( \gamma'' \) phase strengthening phenomenon [41].

As shown in Fig. 5, to measure the grain orientations and textures of Inconel 718/316L multi-material sample in the as-built condition, EBSD mappings, as well as Z (BD) direction inverse pole figure (IPF) color mappings were constructed from the EBSD data to visualize both the grain morphologies and the crystallographic orientations. Figure 5(c) indicated the variation tendency of grain size related to the 316L region, transition region and Inconel 718 region along the Z (BD) direction. Typical grain morphologies of
equiaxed fine grain and columnar grain were observed in 316L and Inconel region, while a visible grain coarsening transition region appeared between these two regions, and the average grain size were calculated to be 0.96 µm, 3.13 µm and 28.5 µm along the building direction. The coarsened grains proved to be the result of continued growth of 316L steel grains, which was attributed to the re-melting and recrystallization caused by the heat input effect when forming the Inconel 718 part and the continued growth of the grains caused by the temperature gradient and cooling rate declining [42] and finally, exacerbating the weakening of the interface performance.

The IPFs indicated that the Inconel 718 region of as-built sample had <001>// Z (BD) as the preferential crystallographic orientation. Because the texture intensity in <001> direction was measured to be 16.25, characterized by strong fiber texture. The texture orientation of the Inconel 718 depended on the competition between the local heat flow direction and the multiple preferential growth directions of the material itself. Primary columnar dendrites that were parallel or nearly parallel to the heat flow direction (<100> orientation) would preferentially grow and other columnar grains, deviating from it greatly, would be eliminated [43]. Meanwhile, the texture of the grain coarsening transition region exhibited a weak {001} <100> cubic texture belonging to the main FCC metal recrystallization texture, and its texture intensity was merely 3.32 (Fig. 5(g)).

### 3.3 Mechanical properties

Figure 6 illustrated the corresponding relationship between the micro-hardness variation of the Inconel 718/316L multi-material interfacial bonding region and the indentation image of sample after Vickers hardness test. At least three points were selected from each region, Fig. 6(b) showed the ten representative indentation points. The micro-hardness values of 316L region, transition region and Inconel 718 region were measured to be 383.7HV, 303.8HV and 320.7HV, respectively. Compared with the Inconel 718 region, the lower hardness value of the transition region with its microstructure dominated by columnar dendrite grains of Inconel 718 was attributed to the suppression of the γ" phase strengthening [41] caused by the formation of the Laves phase and the segregation of Nb as well as the crack initiation, which had been discussed in Fig. 4(c).

Figure 7 showed the tensile samples before test and the stress-strain curves in four states. The average ultimate tensile strength (UTS) of 316L and Inconel 718 samples were measured to be 1007.5 ± 21.3 MPa (Fig. 7(a)) and 986.2 ± 33.5 MPa (Fig. 7(b)), with elongation of 12.3 ± 0.7% and 22.4 ± 1.3%, respectively. In order to investigate the performance of the overall mechanical properties when the multi-material samples were subjected to different direction of loading force, relevant test on tensile specimens in horizontal and vertical combination were carried out. As shown in Fig. 7(c), the average UTS and elongation of horizontally combined Inconel 718/316L samples were 795.4 ± 2.3 MPa and 16.3 ± 0.9%. A reduction of approximately 21.1% and 19.3% occurred in average UTS comparing to the parent 316L and Inconel 718 alloy. This phenomenon could be explained as: (i) initial cracks existed in the transition region (Fig. 2(a)) propagated and grew rapidly when the samples were subjected to external load, and these places became the preferential fracture failure sites as a result of stress concentration; (ii) a lack of
γ" phase strengthening [41], as well as a weaker fine grain strengthening [44] led by coarser average grain size in transition region, which had been discussed above.

**combined Inconel 718/316L; (d) vertically combined Inconel 718/316L.**

Figure 8 demonstrated the tensile fracture morphology of horizontally combined Inconel 718/316L samples in details. Clear and defect-free boundaries found in transition region (Fig. 8(b)) further confirmed a dense metallurgical bonding happened. Typical cleavage and tongue patterns of brittle fracture were observed around the transition region, and the fracture mechanism could be concluded as: the rough and interconnected long-chain Laves phase particles in the sample provided a favorable location for the micro-pore nucleation and macroscopic crack propagation [45] and finally, leaded to brittle trans-granular fracture.

The average UTS and elongation of vertically combined Inconel 718/316L samples were 867.2 ± 21.3 MPa and 20.5 ± 0.3%. However, the fracture sites (as shown in Fig. 9) of all the samples were not located in the interface bonding area which meant the decline in average UTS of vertically combined sample was not related to the bonding effect of 316L and Inconel 718, and indicated a strong metallurgical bonding obtained in Inconel 718/316L bimetallic multi-material.

Three-point bending test could be used to characterize the deformation ability of multi-materials subjected to Z-axis load and reflect the internal defects of the multi-material parts. Long strip specimens of 80 mm in length, 10 mm in width and 10 mm in thickness (the thickness of 316L and Inconel 718 section was 5 mm each) were tested at a rate of 1 mm/min by an indenter with a radius of 10 mm in a three-point bending tester. And bending deformation occurred until the specimens were bent to broken, with the support span of 54 mm. Figure 10 demonstrated typical bending stress-strain curves for each type of specimen. The results of flexural strength and elongation were taken as the average values of three bending stress-strain curves. It could be seen that the flexural strength of 316L, Inconel 718, Inconel 718/316L of 316L at the bottom and Inconel 718 at the bottom were 2455.8 ± 39.7 MPa, 2443.9 ± 31.8 MPa, 1623.6 ± 27.7 MPa and 2462.9 ± 40.7 MPa, respectively.

**Inconel 718/316L specimen of 316L at the bottom; (d) Inconel 718 at the bottom.**

Moreover, Fig. 11 showed typical object diagram of three-point bending test specimens with curved fracture cracks. The elongation of 316L specimen, Inconel 718 specimen, Inconel 718/316L specimen of 316L at the bottom and of Inconel 718 at the bottom were 1.09 ± 0.13 mm/mm, 2.88 ± 0.14 mm/mm, 0.31 ± 0.05 mm/mm and 1.90 ± 0.16 mm/mm, respectively. The average flexural strength and elongation obtained in Inconel 718/316L of 316L at the bottom were the minimum compared with that of the other three types. The results showed that defects such as cracks at the joints of multi-material samples did have a detrimental effect on the overall bending performance of the components. And when the sample received a bending load, the crack was more likely to grow towards the 316L side, further causing bending failure. At the same time, no macro cracks appeared along the direction of the bonding interface.
even if the multi-material specimen failed to bend, which meant these two specific alloys were tightly bonded.

The joint shear strength of two SLM specimens and the entity graph before stretching were shown in Fig. 12. The samples dimensions were $30 \times 10 \times 10$ mm, which 316L and Inconel 718 accounted for 5 mm, respectively. The elevating rate of the indenter loading force was 300 N/m. The shear strength results were obtained when the 316L and Inconel 718 portions were completely separated. Sample 1 and sample 2 achieved the shear strength of the SLM joint to approximately 461 and 438 MPa. Typical Inconel/steel multi-materials formed by specific method and its shear strength [46–49] were concluded in Table 4. The average shear strength of the obtained sample was 449.5 MPa. Comparing the table data, it could be found that the shear strength of Inconel718/316L multi-material obtained by SLM was relatively high, which fully proved the high metallurgical bonding of dissimilar alloy joints.

| Fabricating method         | Experimental subjects | Post-processing | Shear strength (MPa) | Ref  |
|---------------------------|-----------------------|-----------------|----------------------|------|
| Induction brazing         | Inconel X-750         | 304 SS          | –                    | 483  | [46] |
| Gas tungsten arc welding  | Inconel 617           | 310 SS          | –                    | 445  | [47] |
| Explosive welding         | Inconel 625           | P355NH          | Stress relief annealing | 572  | [48] |
|                           |                       |                 | Normalizing          | 591  |      |
| Infrared brazing          | Inconel 601           | 422 SS          | –                    | 362  | [49] |
| Selective laser melting   | Inconel 718           | 316L SS         | –                    | 449.5| Our work |

**Conclusion**

The forming defects of multi-material joints, the formation and distribution of strengthening phases and joint bonding strength were the research focuses of this paper. Despite the presence of continuous or separated micro-cracks in the interface transition region, the shear strength of the joint indicated that the multi-materials with a strong metallurgical bond reached the welding level. However, the formation of the grain coarsening region revealed by EBSD, the suppression of $\gamma'$, $\gamma''$ phase precipitation strengthening phase revealed by SEM due to the presence of the Laves phase and the detrimental effects of the brittle
Laves phase itself were important reasons for restricting the further improvement of joint strength. This research showed that by controlling the continued growth of grains in the joint area and regulating the content of the Laves phase, the formation of defects such as micro-cracks could be reduced and the interface bonding performance could be boosted in the future.

Declarations

Competing interests

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.

Authors' contributions

S-FW and K-YC were in charge of the whole trial; YZ together with Y-SC wrote the manuscript; YL and JG assisted with sampling and laboratory analyses. All authors read and approved the final manuscript.

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