Materials Research Express

OPEN ACCESS

Effect of heat input on microstructure and mechanical properties of GH159 and GH4169 dissimilar joints by laser beam welding

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Keywords: laser beam welding, dissimilar joint, laves phase, mechanical properties

Abstract
Co-based superalloy GH159 and Ni-based superalloy GH4169 have been successfully joined by laser beam welding and the effect of heat input on microstructure and properties of dissimilar joints were investigated systematically. The results showed that weld seams exhibited a nail shape and full penetration was attained at all dissimilar joints. Increasing grain size towards the fusion zone (FZ) was observed in the heat affected zone (HAZ) on the GH419 side while an increasing dissolution of γ′ and γ towards the FZ was attained at the HAZ on the GH4169 side. These resulted in decreasing microhardness towards FZ. Tensile failure was found in the FZ with the lowest microhardness. Meanwhile, the ultimate tensile strength (UTS) of the dissimilar joints increased with the decreasing of heat input. The high UTS of dissimilar joint with low heat input can be ascribed to the lower volume fraction of the Laves phase and the smaller dendrite arm spacing.

1. Introduction

One of the most vital and crucial components in the aero-engines is the high pressure rotor bolt assembly which exerts as a connector between the turbine disc and the shaft. It consists of spring plates and bolts, and the permanent connection between them is achieved by laser beam welding (LBW). The requirements for bolt assembly material are to provide sufficient mechanical strength, good creep and oxidation resistance, especially for long-term service at temperatures of 650 °C. All these demands can be met by the GH4169 alloy and thus it has been used in the practice production.

GH4169 is a precipitation-hardened Ni-based superalloy that is widely used in the aerospace industry due to its outstanding mechanical strength, good corrosion resistance and weldability [1–4]. However, recent studies have shown that the main strengthening phase γ′ coarsens and gradually transforms into δ phase after long-term thermal exposure of GH4169 at 650 °C, finally giving rise to its properties degradation [5, 6]. In addition, with the development of aerospace industry, it is of importance to develop the high-strength and large-size bolt assembly. To meet these requirements, GH159 of deformation-hardened cobalt-based superalloy is used in the aviation field. After deformation, its tensile strength can reach 1916 MPa at room temperature, and the tensile strength still remains at 1564 MPa at 595 °C [7–9]. Considering the cost, GH4169 is still used for the spring plate material, which involves the LBW of dissimilar materials.

The main challenges that occur during the LBW of GH4169 are the formation of brittle phases, cracks in the fusion zone (FZ) or heat affected zone (HAZ), and porosity [10–19]. According to the available literature, these metallurgical problems are unavoidable and can only be improved. There are scanty reports on the LBW of GH159. These challenges become manifold when attempting to weld dissimilar materials due to the big divergence of chemical composition and thermo-physical parameters, which may tend to aggravate the issues. Karadge et al. [20] attempted to achieve the jointing between single crystal CMSX4 and Ni-based superalloy RR1000 and pointed out the difficulties of welding two different Ni-based alloys with very different structures.
The chemical compositions of GH159 and GH4169 and the welding parameters are shown in tables 1 and 2, respectively. The component parts of the bolt assembly, bolts and leaf springs, have different heat treatments due to different requirements. Therefore, a solution heat treatment at 1050 °C for 8h at 620 °C followed by furnace cooling at a cooling rate of 55 °C h⁻¹ to 620 °C and maintained for 8h at 620 °C followed by air cooling) were conducted on GH4169 alloy before LBW. The dimensions for the weld sample as shown in figure 1(a). In order to investigate the influence of heat input on the microstructure and properties of LBW dissimilar joints of GH159 and GH4169, the welding process was conducted using a continuous laser CO₂ welding machine.

The macrostructure and microstructure evolutions of weld joints were characterized by optical microscopy (OM) and a field emission gun scanning electron microscope (SEM). The Image-Pro Plus software was employed to quantitatively calculate the grain size and the volume fraction and size of precipitates. The samples for OM and SEM observation were prepared using mechanical grinding and polishing methods. Next, the samples were chemically etched in a mixed solution containing 5g CuSO₄ + 100 ml HCl + 5ml H₂SO₄ [23]. The detailed microstructures were examined by transmission electron microscope (TEM) and TEM foils were prepared using standard techniques [24]. Bright field (BF) images, dark field (DF) images, and diffraction patterns were recorded using a TEM at an acceleration voltage of 200 kV to observe the fine microstructural details.

Non-standard micro-sized tensile specimens were prepared from the weld joints the dimensions of which are illustrated in figure 1(b). Similar micro-tensile specimens were also successfully used for the determination of tensile properties of diffusion bonded Ti-alloys and laser beam welded steel joints [25–32]. Tensile tests were performed using a single column desktop electronic universal testing machine equipped with a laser extensometer at a cross head speed of 1 mm min⁻¹ at room temperature to evaluate the mechanical properties of
various weld joints. To ensure the accuracy of the experimental results, each specimen was subjected to three instances of tensile testing. The hardness values of the specimens were obtained at a load of 1000 g for 10 s by using a Vickers hardness tester. The hardness test needs to start at 0.2 mm away from the weld surface and end at 0.2 mm away from the other side of the weld surface, in which test one point every 0.1 mm.

3. Results and discussion

3.1. Effect of heat input on macrostructure of weld joints

Figure 2 shows the macrostructures of the dissimilar joints under different heat inputs. No welding defects such as cracks and metal splashes were observed in all the weld joint surfaces under the heat input of 35.2 J mm\(^{-1}\) \(\sim\) 44.8 J mm\(^{-1}\), as shown in figures 2(a)–(d). All weld seams exhibited a nail shape and the porosity was observed in the bottom of the weld seams as shown in figures 2(e)–(h). The presence of porosity is mainly related to the high Ni content in the base metal [22]. In addition, the width and depth of the weld seam were measured and the results are shown in figure 3. As the heat input increased from 35.2 J mm\(^{-1}\) to 44.8 J mm\(^{-1}\), the width and depth of weld seams increased from 2050 \(\mu\)m and 1769 \(\mu\)m to 2212 \(\mu\)m and 2038 \(\mu\)m, respectively. These results indicate the heat input has significant influences on the weld seam geometry.

3.2. Evolution of microstructure of weld joints

It can be clearly observed from figure 4 that the weld joint could be obviously divided into 5 zones, namely base metal (BM) GH4169, HAZ of GH4169 side (HAZ B), FZ, HAZ of GH159 side (HAZ A) and BM GH159. The phase composition in each area one by one was confirmed through the SEM and TEM analyses, as described in the following paragraphs.
3.2.1. Microstructure of the BM

Figure 5(a) shows the microstructure of BM GH4169 after solution treatment at 1050 °C and double aging. SEM results indicate that BM GH4169 consisted of equiaxed grains and a few blocky particles in the grain boundaries. The blocky particles were determined to be MC carbides through the EDS analysis, as shown in figure 5(b). Similar results were also observed in Refs [17, 33, 34]. In addition to MC carbides, TEM images also show that nanometer precipitates were appeared in the BM GH4169 (figure 5(c)). Further confirmed by the selected area electron diffraction (SAED), the BM GH4169 contained strengthening phase γ' and γ'' (figure 5(d)). It can be seen from figure 5(e) that the γ' precipitated phase was dispersefully distributed in the matrix.

The situation is absolutely different in the microstructure of BM GH159 (figure 6(a)). Fine recrystallization grains (arrow 2), MC carbides and a much coarser grain with intersecting network of fine platelets (arrow 1) were observed in BM GH159. Further confirmed by SAED obtained along the most favorable orientation [011][9], BF and DF images, the fine platelets were twins (figures 6(b)–(d)). GH159 is mainly strengthened by the cold deformation which forms intersecting network of fine twins. The intersecting network of thin twins, which function as ‘cells’ or ‘subgrains’ and divide the coarse grain into many small parts, which is equivalent to refining the coarse grain. Furthermore, it is difficult for dislocations to move long distances in this intersecting network of fine thin twins.

3.2.2. Microstructure of the HAZ

Figure 7(a) depicts the microstructure of the HAZ A. Compared with the BM GH159, static recrystallization and the incremental grain size towards the FZ was clearly observed in HAZ A (The direction of the arrow in figure 7(a) is away from the FZ). Disappearance of cold deformation microstructure obviously indicates that the temperature during welding process near the weld line was more than the GH159 static recrystallization temperature of 920 °C [35]. However, MC carbides were still present close to the weld line after welding (figures 7(a)–(b)), namely, MC carbides did not dissolve after welding. This reveals the temperature near the
Figure 5. Microstructures of base metal GH4169: (a) SEM image, (b) EDS analysis of position b in (a), (c) TEM BF image, (d) a selected area electron diffraction (SAED) pattern of [001] zone of the γ matrix containing γ′ and γ′ superlattice reflections, and (e) Dark field (DF) image of γ′ using (100) reflection.

Figure 6. Microstructures of base metal GH159: (a) SEM image, (b) TEM image, (c) a SAED pattern of the corresponding area in (b), (d) DF image.
weld line during welding was lower than the solvus temperature of MC carbides ($\sim 1260$ °C)$^{[36]}$. Consequently, the temperature near the weld line during welding process is estimated to be in the range from $920$ °C to $1260$ °C. But, this temperature range is higher than the solvus temperature of strengthening phases $\gamma'$ (910 °C$^{[37]}$) and $\gamma'$ (900 °C$^{[38]}$) in GH4169. Consequently, it is expected that the strengthening phases near the weld in HAZ B have been completely dissolved during the welding process. As expected, no strengthening phases $\gamma'$ or $\gamma'$ were found in the area close to the weld line (as shown in figures 7(c)–(d)). Consequently, it is inferred that the increasing dissolution of strengthening phases $\gamma'$ and $\gamma'$ towards the FZ was attained at in the HAZ B. This conclusion is also supported by the finding of Ye et al$^{[39]}$.

3.2.3. Microstructure of the FZ

Figure 8(a) displays a typical SEM micrograph of dissimilar joints of GH159 and GH4169. The microstructure from the weld line to the weld center changed from dendritic structure to equiaxed grains as shown in figures 8(b) and (c). The irregular phases that appeared in the FZ were characterized by the SEM/EDS spectra analysis. Figure 8(d) verifies that irregular phases mainly contained Nb and Mo elements. The precipitates were further confirmed to be Laves phases by SEAD as shown in figure 8(f).

The elemental distributions of the dissimilar joints are shown in figure 9. It is distinct that under the minimum and the maximum heat inputs the amounts of elements Ni, Fe and Nb decreased from the GH4169 side to the GH159 side. However, the change of the amounts of Co and Ti was the opposite. In particular, under the minimum heat input, the element on the interface increased abruptly, while under the maximum heat input, the element on the interface increased slowly. Therefore, it can be concluded that with the increasing of heat input, element diffusion increases.

Figure 10 illustrates the dendritic microstructure of the FZ of different heat inputs. Obviously, the increasing of heat inputs caused a distinct increase in the dendrite arm spacing. To further determine the relationship between the heat inputs and the microstructure of the FZ, the dendrite arm spacing of the FZ was quantitative calculated. As shown in figure 11, the measured dendrite arm spacing increased from $2.72 \pm 0.62 \mu m$ to $3.94 \pm 0.63 \mu m$ as the heat input increased from 35.2 J mm$^{-1}$ to 44.8 J mm$^{-1}$. The dendritic arm spacing is mainly affected by the cooling rate. The relationship between the cooling rate and the dendritic arm spacing can be described by equation (1)$^{[40]}$

$$\lambda = K \cdot R^{-n}$$

(1)
where $K$ is the dendrite arm spacing and $R$ is the cooling rate. Equation (1) indicates that the cooling rate decreases with an increase in the dendritic arm spacing.

Figure 12 represents the distribution of the Laves phase in the FZ under different heat inputs. It indicates that the number of Laves phases increased with increasing heat input. Particularly, the morphology of the Laves phase was semi-continuous in figure 12(c) and highly interconnected in figure 12(d). Figure 13 gives the quantitative result on size and volume fraction of the Laves phase in the FZ under different heat inputs. With the increasing of heat inputs, the size of Laves phase increased slightly. However, the volume fraction of Laves phase varies greatly with the heat input. As shown in figure 13(b), the volume fraction of Laves phase was $1.86 \pm 0.15\%$, $2.88 \pm 0.09\%$, $4.93 \pm 0.31\%$, and $6.38 \pm 0.32\%$. The relationship between the heat input and the volume fraction of Laves phase can be explained as following: the increasing of the heat input due to the decreasing cooling rate corresponding to a high microsegregation. The extent of microsegregation with high heat input is higher that results in a large number of Laves phases.

### 3.3. Distribution of hardness across the profile of the weld joint

The above results indicate that the heat input has a significant effect on the microstructure of the dissimilar joint, which may lead to the change of mechanical properties. Figure 14 shows the microhardness changes of the dissimilar joints under different heat inputs and the average hardness values of each zone are listed in table 3. It is observed that the microhardness value of the whole weld joint decreased when the heat input increased at a same measuring position $0.2$ mm from the weld surface in the weld joint, indicating that the heat input exerts a
remarkable effect on the microhardness of the dissimilar joint of GH159 and GH4169. As can be observed from the curve in figure 14, the microhardness values decreased continuously when the measuring positions were gradually close to the center of the FZ. The lowest microhardness value in the whole dissimilar joint was acquired at the FZ. Thus, the following discussion mainly focuses on the relationship between heat input and FZ hardness.

The microhardness of a material is generally defined as its resistance to plastic deformation. It can be seen from table 4 that under different heat inputs, the average microhardness values of fusion zone were 254.0 HV ($35.2 \text{ J mm}^{-1}$), 241.1 HV ($38.4 \text{ J mm}^{-1}$), 230.0 HV ($41.6 \text{ J mm}^{-1}$) and 220.2 HV ($44.8 \text{ J mm}^{-1}$), respectively. With the increasing of heat input, the microhardness of the FZ decreased. Laves phase is not only brittle phase but also its

Figure 9. Linear scanning (EDS) of the weld joint.
precipitation depletes major strengthening elements Nb, Mo and Ti, thus weakening the \(\gamma\) matrix and making it soft. The volume fraction of Laves phase increased with increasing heat input and, consequently, a decrease in the content of elements Nb, Mo and Ti in the \(\gamma\) matrix. Moreover, the relationship between the microhardness of FZ and the secondary dendrite arm spacing can be explained by the Hall-Petch equation \(39\):

\[
HV = C + k\lambda^{1/2}
\]

where \(C\) and \(K\) are material constants and \(\lambda\) is secondary dendrite arm spacing. From equation \(2\), it can be seen that the hardness value is inversely proportional to the secondary dendrite spacing. Namely, the hardness value decreases with increasing dendrite arm spacing. The results of dendrite arm spacing in section 3.2.3 show that when the heat input increased from 35.2 J mm\(^{-1}\) to 44.8 J mm\(^{-1}\), and the dendrite arm spacing increased from 2.72 \(\mu\)m to 3.94 \(\mu\)m. Therefore, with increasing the heat input, the secondary dendrite arm spacing increases and the hardness value decreases. Obviously, the changes of microstructure in the FZ can serve as the compelling evidence of the changes of the microhardness in the FZ (see figures 11 and 13).
The width of HAZ with different heat input could be estimated by hardness diagram distribution. The widths of HAZ B and HAZ A were 0.8 mm and 0.6 mm (35.2 J mm\(^{-1}\)), 1.0 mm and 0.8 mm (38.4 J mm\(^{-1}\)), 1.5 mm and 0.9 mm (41.6 J mm\(^{-1}\)), 1.5 mm and 0.9 mm (44.8 J mm\(^{-1}\)), respectively. Obviously, the width of the HAZ B was larger than that of HAZ A and it could be attributed to the different thermal conductivity of two materials. The width of the HAZ can be roughly estimated using the following equation:

\[
W_{\text{HAZ}} = \sqrt{a t}
\]

where \(W_{\text{HAZ}}\) is the width of the HAZ, \(a\) is the thermal conductivity of the material, \(t\) is heat conduction time. The thermal conductivity of GH159 and GH4169 is 11.0 W m\(^{-1}\) \(\cdot\)°C \([40]\) and 14.7 W m\(^{-1}\) \(\cdot\)°C \([41]\), respectively. Big divergence in the thermal conductivity can be conducted on two materials. Therefore, the width of the HAZ B of is larger than that of HAZ A. The continuously decreasing of microhardness values in the HAZs along FZ are mainly contributed to increasing dissolution of the strengthening phases \(\gamma'\) and \(\gamma''\) and the grain size along FZ.

**Figure 12.** SEM micrographs showing the microstructure in the center of FZ with different heat inputs: (a) 35.2 J mm\(^{-1}\), (b) 38.4 J mm\(^{-1}\), (c) 41.6 J mm\(^{-1}\) and (d) 44.8 J mm\(^{-1}\).

**Figure 13.** The statistical results of average diameter and volume fraction of Laves phase under different heat inputs: (a) the average diameter of Laves phase and (b) the volume fraction of Laves phase.
3.4. Tensile testing results

Figure 15 depicts the engineering stress–strain curves and the corresponding mechanical properties are presented in Table 4. The ultimate tensile strength (UTS) in dissimilar joints of GH159 and GH4169 under different heat inputs was 923 MPa (35.2 J mm$^{-1}$), 903 MPa (38.4 J mm$^{-1}$), 839 MPa (41.6 J mm$^{-1}$), 756 MPa (44.8 J mm$^{-1}$), respectively. The UTS in dissimilar joints of GH159 and GH4169 under four heat inputs can meet the minimum requirements for using even without post-weld heat treatment. The elongations of all the dissimilar joints in time of tension tests which follow ASTM E8/8M standard were found to be above 15%. In addition, all the tensile fractures were observed in the FZ with the lowest microhardness. LBW is the last step in the production process of bolt assembly, so no post-weld heat treatment can be carried out to improve its mechanical properties. Heat input determines the final mechanical properties of the bolt assembly. Therefore, the following analysis mainly focuses on the relationship of heat input–UTS–FZ microstructure.

As we all know, chemical composition, grain size and precipitation of nanoscale γ′ and γ (solid solution strengthening, grain boundary strengthening and precipitation strengthening) have a remarkable effect on the strength of superalloys [39]. It can be seen from section 3.2.3 that there was no precipitation of strengthening phases γ′ and γ in the FZ, and the change of heat input only affects the dendrite arm spacing and the volume fraction of the Laves phase. Therefore, in this study, only the influences of solid solution strengthening and grain boundary strengthening on strength of FZ are considered [42, 43]:

![Figure 14. The microhardness profiles of GH4169 and GH159 dissimilar joints under different heat inputs.](image)

### Table 3. The average Hardness values at each zone in figure 14.

| Sample no # | Heat input (J mm$^{-1}$) | GH4169 (HV) | HAZ B (HV) | FZ (HV) | HAZ A (HV) | GH159 (HV) |
|-------------|--------------------------|-------------|------------|---------|------------|------------|
| 1           | 35.2                     | 295.5       | 254        | 349.5   |            |            |
| 2           | 38.4                     | 426.4       | 276.6      | 241.4   | 330.6      | 483.6      |
| 3           | 41.6                     | 276.7       | 230        | 328.5   |            |            |
| 4           | 44.8                     | 234.3       | 220.2      | 322.9   |            |            |

### Table 4. Tensile test results of GH159 and GH4169 dissimilar joints at room temperature.

| Sample no # | Heat input (J mm$^{-1}$) | UTS (MPa) | Elongation (%) | Fracture location |
|-------------|--------------------------|-----------|----------------|------------------|
| 1           | 35.2                     | 923       | 38.6           | FZ               |
| 2           | 38.4                     | 903       | 26.3           | FZ               |
| 3           | 41.6                     | 839       | 20.7           | FZ               |
| 4           | 44.8                     | 756       | 18.6           | FZ               |
The resulting strength $\sigma_r$ is the sum of the effects of the intrinsic strength $\sigma_0$, grain boundary strengthening $\sigma_{gb}$, and solid solution strengthening $\sigma_{ss}$. Each of these contributions affects the strength in the FZ of the GH159 and GH4169 dissimilar joints. Thus, each microstructural feature will be discussed below to investigate the heat input-structure-property.

According to figure 11, the measured dendrite arm spacing increased from $2.72 \pm 0.62 \mu m$ to $3.94 \pm 0.63 \mu m$ as the heat input increased from $35.2$ J mm$^{-1}$ to $44.8$ J mm$^{-1}$. The dendrite arm spacing increases with the increasing of heat input. According to Hall-Petch equation, the contribution of grain size is:

$$\sigma_{gb} = \frac{\kappa}{D^{1/2}}$$

where $D$ is the grain size, and $\kappa$ is the Hall-Petch slope. According to equation (5), grain boundary strengthening is inversely proportional to grain size, namely, the effect of fine grain strengthening decreases with the increasing of grain size. Therefore, it can be concluded that with the increasing of heat input, the dendrite arm spacing increases and the fine grain strengthening effect become weaker.

Laves phase is the main precipitation phase in the FZ, but it has no direct effect on the strength. Its precipitation depletes major strengthening elements Nb, Mo and Ti in the matrix. These elements are known to be as the most significant solid solution strengthening elements in superalloy [44]. Therefore, only the solid solution strengthening induced by these elements is considered in this work. In this study, the model established by Gyepen is adopted because of its simplicity and widespread application in nickel-based superalloys [44–46]. The solid solution strengthening can be defined as

$$\sigma_{ss} = (1 - f) \left( \sum_i \alpha_i^2 c_i \right)^{1/2}$$

where $\alpha_i$ is the strengthening constant of atomic species i and $c_i$ is the atomic concentration of atomic species i. $(1 - f)$ is the modifying factor, which is used to account for the solid solution strengthening confined only to the matrix. The following equation can be obtained: $(1 - f) = (1 - f_{Laves})$, $f_{Laves}$ is the volume fraction of the Laves precipitates. According to equation (6), as the volume fraction of Laves phase increases, the strengthening effect of solid solution weakens. According to the quantitative statistical results in figure 14(b), the volume fraction of Laves phase was $1.86 \pm 0.15\%$ (35.2 J mm$^{-1}$), $2.88 \pm 0.09\%$ (38.4 J mm$^{-1}$), $4.93 \pm 0.31$ (41.6 J mm$^{-1}$), $6.38 \pm 0.32$ (44.8 J mm$^{-1}$), respectively. Laves phase increased with the increasing of heat input. Therefore, it can be concluded that the solid solution strengthening effect weakens with the increasing of heat input.

On the basis of this detailed microstructure study, each individual strengthening contribution in equation (1) has been discussed. It can be concluded that the decreasing of UTS in FZ could be attributed to the higher volume fraction of Laves phase and the larger dendrite size.

Figure 15. The engineering strain stress curves of GH4169 and GH159 dissimilar joints under different heat inputs.
4. Conclusions

In summary, different heat inputs were applied to the LBW of dissimilar GH159 and GH4169 superalloys. The macrostructure and microstructure evolution were examined, the microhardness distribution and the tensile mechanical properties of dissimilar joints GH159 and GH4169 were tested. The main conclusions are as follows:

(1) Weld seams exhibited a nail shape and full penetration was attained at the heat input from 35.2 J mm\(^{-1}\) to 44.8 J mm\(^{-1}\). With the increasing of heat input, the width and depth of weld seams increased. Heat input has significant influences on the weld seam geometry.

(2) The microstructure result showed that FZ consisted of dendritic microstructure and Laves phase. The microstructure of GH4169 consisted of NbC, and the strengthening phases \(\gamma'\) and \(\gamma''\). But increasing dissolution of these two phases towards the FZ was attained at in the HAZ B. Static recrystallization and the incremental grain size towards the FZ was observed in the HAZ A.

(3) The width of the HAZ B was larger than that of HAZ A and it could be attributed to the different thermal conductivity of two materials. The significant reduction in the microhardness at the HAZs could be attributed to the dissolution of the main strengthening phases and the disappearance of intersecting network of fine twins.

(4) Tensile failures were occurred in the FZ with the lowest microhardness. With the decreasing of heat input, the UTS of the dissimilar joints increased. The high UTS of dissimilar joints with low heat input can be ascribed to the lower volume fraction of the Laves phase and the smaller dendrite arm spacing.

Acknowledgments

This research was funded by the National Natural Science Foundation of China (Grant Nos. 51774103 and 51974097), the Program of ‘One Hundred Talented People’ of Guizhou Province (Grant No. 20164014), Guizhou Province Science and Technology Project (Grant Nos. 20175656, 20175788, 20191414, and 20192162).

Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

Conflicts of 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.

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References

[1] Hong J K, Park J H, Park N K, Eom I S, Kim M B and Kang C Y 2008 Microstructures and mechanical properties of Inconel 718 welds by CO\(_2\) laser welding J. Mater. Process. Technol. 201 515–20
[2] An J, Wang L, Song X, Liu Y, Gai Z and Cao X 2018 Improving mechanism of both strength and ductility of GH4169 alloy induced by electric-pulse treatment Mater. Sci. Eng. A 724 439–43
[3] An XL, Zhang B, Chu CL, Zhou L and Chu Paul K 2019 Evolution of microstructures and properties of the GH4169 superalloy during short-term and high-temperature processing Mater. Sci. Eng. A 744 255–66
[4] Xia Y X, Shu X D, Zhu D B, Pater Z and Bartnicki J 2021 Effect of process parameters on microscopic uniformity of cross wedge rolling of GH4169 alloy shaft J. Manuf. Process. 66 145–52
[5] Lu X D, Du J H and Deng Q 2013 High temperature structure stability of GH4169 superalloy Mater. Sci. Eng. A 559 623–8
[6] Yu Z S, Zhang J X, Yuan Y, Zhou R C, Zhang H J and Wang H Z 2015 Microstructural evolution and mechanical properties of Inconel 718 after thermal exposure Mater. Sci. Eng. A 634 55–63
[7] Han G W, Jones J P and Smallman R F 2003 Direct evidence for Suzuki segregation and Cottrell pinning in MP159 superalloy obtained by FEG(S)TEM/EDX Acta Mater. 51 2731–42
[8] Gu J, Guo L, Gan B, Bi Z and Song M 2021 Microstructure and mechanical properties of an MP159 alloy processed by torsional deformation and subsequent annealing Mater. Sci. Eng. A 802 140676
[9] Lu S Q, Shang B Z, Luo Z J, Wang R H and Zeng F C 2000 Investigation on the cold deformation strengthening mechanism in MP159 alloy Metall. Mater. Trans. A 31 5–13
Qu F, Lu Z, Xing F and Zhang K F 2012 Study on laser beam welding/superplastic forming technology of multi-sheet cylinder sandwich structure for Inconel718 superalloy with ultra-fine grains Mat. Des. 39 151–61

Rakumkar D, Patel S D, Praveen S S, Choudhury D J, Prabaharan P, Arivazhagan N and Xavior M A 2014 Influence of filler metals and welding techniques on the structure–property relationship of Inconel 718 and AISI 316L dissimilar weldments Mater. Des. 62 175–88

Zhang Y C, Li Z G, Nie P L and Wu Y X 2013 Effect of heat treatment on niobium segregation of laser-cladded IN718 alloy coating Metall. Mater. Trans. A 44 708–16

Sui S, Chen J, Ming X L, Zhang S P, Lin X and Huang W D 2017 The failure mechanism of 50% laser additive manufactured Inconel 718 and the deformation behavior of Laves phases during a tensile process Int. J. Adv. Manuf. Technol. 91 2733–2740

Cam G and Koçak M 1998 Progress in joining of advanced materials Part 2: Joining of metal matrix composites and joining of other advanced materials Sci. Technol. Weld. Join. 3 159–75

Cam G and Koçak M 1998 Progress in joining of advanced materials Int. Mater. Rev. 43 1–44

Cam G, Fischer A, Ratjen R, dos Santos J F and Koçak M 1998 Properties of laser beam welded superalloys inconel 625 and 718 of 7th European Conf. on Laser Treatment of Materials (ECLAT ’98), Werkstoff-Informationsgesellschaft mbH (Frankfurt, Hannover, Germany) pp 333–8

Mei Y, Liu Y, Liu C, Chong L, Yu L, Guo Q and Li H 2016 Effect of base metal and welding speed on fusion zone microstructure and HAZ hot-cracking of electron-beam welded Inconel 718 Mat. Des. 89 964–77

Lin J, Wang X, Lei Y, Wei K and Guo F 2021 Study on hot cracking in laser welded joints of inconel 718 alloy foils J. Manuf. Processes 64 1024–35

Ye X, Zhang P L, Zhao J and Ma P 2018 Effect of macro- and micro-segregation on hot cracking of Inconel 718 superalloy argon-arc multilayer cladding J. Mater. Process. Tech. 258 231–8

Karadge M, Preuss M, Withers P J and Bray S 2008 Importance of crystal orientation in linear friction joining of single crystal to polycrystalline nickel-based superalloys Mater. Sci. Eng. A. 491 446–53

Rakumkar K D, Des S, Phani P K V, Rajendran R and Narayanan S 2019 Microstructure and properties of Inconel 718 and AISI 416 laser welded joints J. Mater. Process. Technol. 266 52–62

Kuo T Y and Jeng S L 2005 Porosity reduction in Nd–YAG laser welding of stainless steel and Inconel alloy by using a pulsed wave J. Phys. D: Appl. Phys. 38 7222

Odaçbaş M, Unlu N, Göller G, Kayali E S and Ersulu M N 2013 Assessment of the effects of heat input on microstructure and mechanical properties in laser beam welded Haynes 188 undermatched joint Mater. Sci. Eng. A. 559 731–41

Williams D B and Carter C B 2007 Transmission Electron Microscopy (Boston: Werkstoff-Informationsgesellschaft mbH)

Cam G, Bohm K H, Müllauer J and Koçak M 1996 The fracture behavior of diffusion-bonded duplex gamma TiAl, JOM, The Journal of the Minerals Metals and Materials Society 48 68–6

Cam G, Clemens H, Gerling R and Koçak M 1999 Diffusion bonding of fine grained gamma–TiAl sheets Metalld. 90 284–8

Cam G, Erism S, Yeniç and Koçak M 1999 Determination of mechanical and fracture properties of laser beam welded steel joints Weld. J. 78 193s–01s

Cam G, Yeniç, Erism S, Ventzke V and Koçak M 2011 Investigation into properties of laser welded similar and dissimilar steel joints Sci. Technol. Weld. Join. 3 177–89

Cam G, Müllauer J and Koçak M 1996 Diffusion bonding of two phase gamma–TiAl alloys with duplex microstructure Sci. Technol. Weld. Join. 2 213–9

Rao D, Heerens J, Pinheiro G A, Santos J D and Huber N 2010 On characterisation of local stress–strain properties in friction stir welded aluminum AA 5083 sheets using micro-tensile specimen testing and instrumented indentation technique Mater. Sci. Eng. A. 520 5018–25

Yang B, Xuan F Z and Chen J K 2018 Evaluation of the microstructure related strength of CrMoV weldment by using the in situ tensile test of miniature specimen Mater. Sci. Eng., A. 736 193–201

Ahn J, He E, Chen L, Dear J, Shao Z and Davies C 2018 In-situ micro-tensile testing of AA2024-T3 fibre laser welds with digital image correlation as a function of welding speed Int. J. Lightweight. Mater. Manuf. 1 79–88

Chen Y T, Yeh A C, Li M Y and Kuo S M 2017 Effects of processing routes on room temperature tensile strength and elongation for inconel 718 Mater. Des. 119 235–41

An X L, Zhang B, Chu C L, Zhou L and Chen P K 2019 Evolution of microstructures and properties of the GH4169 superalloy during short-term and high-temperature processing Mater. Sci. Eng., A. 744 255–66

Lu S Q, Shang B Z, Luo Z J, Wang R H and Zeng F C 1999 The effect of the thermal exposure on microstructure of MP159 alloy J. Mater. Sci. 34 5449–56

Hardy M C, Zärbel B, Shen G and Shankar R 2004 Developing damage tolerance and creep resistance in a high strength nickel alloy for disc applications TMS Superalloys 12 pp 83–90

Donachie M J and Donachie S J 2002 Superalloys a Technical Guide 23 23–4

Wlodzek S T, Kelly M and Alden D 1992 The Structure of N18 Superalloys 45 467–76

Ye B R, Li H Y, Ding R G, Doel T and Bowen P 2020 Microstructure and microhardness of dissimilar weldment of Ni-based superalloys IN718-IN713LC Mater. Sci. Eng. A. 774 138894

Nie P, Ojio O A and Li Z 2014 Numerical modeling of microstructure evolution during laser additive manufacturing of a nickel-based superalloy Acta Mater. 77 83–95

Damodaram R, Raman S, Satyanarayana D, Reddy G M and Rao K P 2014 Hot tensile and stress rupture behavior of friction welded alloy 718 in different pre-and post-weld heat treatment conditions Mater. Sci. Eng., A. 612 414–22

Theska F, Stanoevic A, Oberwinkler B, Ringer S P and Primig S 2018 On conventional versus direct ageing of alloy 718 Acta Mater. 156 116–24

Myhr O R, Andersen S J and Grong O 2001 Modelling of the age hardening behaviour of Al–Mg–Si alloys Acta Mater. 49 65–75

Roth H A, Davis C L and Thomson R C 1997 Modeling solid solution strengthening in nickel alloys Metall. Mater. Trans. A 28 1329–35

Joseph C, Persson C and Colliander M H 2017 Influence of heat treatment on the microstructure and tensile properties of Ni-base superalloy Haynes 282 Mater. Sci. Eng., A. 679 520–30

Galindo-Nava E L, Connor L D and Raea C M F 2015 On the prediction of the yield stress of unimodal and multimodal γ/Nickel-base superalloys Acta Mater. 98 377–90