Original Research

Microstructures and mechanical properties of Ti$_3$Al/Ni-based superalloy joints arc welded with Ti–Nb and Ti–Ni–Nb filler alloys

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Abstract

Dissimilar joining of Ti$_3$Al-based alloy to Ni-based superalloy has been carried out using gas tungsten arc (GTA) welding technology with Ti–Nb and Ti–Ni–Nb filler alloys. The joint welded with the Ti–Nb filler alloy contained much less interfacial brittle phases than the one using the Ti–Ni–Nb filler alloy. The average room-temperature tensile strength of the joint welded with Ti–Nb was 202 MPa and the strength value of the one welded with Ti–Ni–Nb was 128 MPa. For both fillers, the weak links of the dissimilar joints were the weld/In718 interfaces. The presence of TiNi, TiNi$_3$ and Ni$_3$Nb intermetallic compounds in the joint welded with Ti–Ni–Nb induced microcracks at the weld/In718 interface and deteriorated the mechanical properties of the joint. And the adoption of the Ti–Nb filler alloy decreased the formation tendency of interfacial brittle phases to some extent and thus enhanced the tensile strength of the joint.

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Keywords: Ti$_3$Al; Ni-based superalloy; Arc welding; Interface; Intermetallics; Mechanical properties

1. Introduction

Ti$_3$Al-based alloy has attracted great interests in aerospace industry in the past decades and become one of the potential materials for aero-engine applications because of its low density, superior strength, high stiffness as well as good creep resistance at elevated temperatures [1,2]. Substituting Ti$_3$Al-based alloy for Ti-based alloy can enhance the working temperatures of aero-engine components and using Ti$_3$Al instead of Ni-based superalloy can realize weight reduction. Therefore, the welding technologies of Ti$_3$Al-based alloy to itself and to other materials are indispensable to be developed for realizing the practical applications of this intermetallic alloy.

The primary problem for the welding of Ti$_3$Al-based alloy is the generation of solid-state cracks in the joints [3,4]. Efforts have been made to join Ti$_3$Al with fusion welding technologies and some advances have been achieved [5]. For example, the room-temperature tensile strength of the laser welded Ti–24Al–17Nb joint was equal to that of the base material in Wu et al.’s research [6] and Lei et al. indicated that the strength value of the Ti–22Al–27Nb joint using laser beam welding was basically comparable to that of the base metal [7]. The joining between Ti–22Al–25Nb and TC11 alloys was carried out using electron beam welding technique and the room-temperature tensile strength of the joint was higher than that of the TC11 base alloy [8]. Feng et al. investigated the interfacial microstructure and mechanical properties of the dissimilar joint of Ti$_3$Al and TC4 alloys welded by electron beam welding process and the highest tensile strength of the joints could reach almost 92% of that of the Ti$_3$Al-based alloy [9].

These works published so far on the weldability of Ti$_3$Al-based alloy mainly belong to joining Ti$_3$Al to itself or to Ti-based alloy. Actually, the dissimilar joining of Ti$_3$Al-based alloy to Ni-based superalloy is more attractive for engineering applications due to not only the weight reduction effect but
also its high-temperature service potential. For example, Ti$_3$Al-based alloy can be used to manufacture the outer casing of advanced aero-engine compressor and its connection with the mounting edge usually made of Inconel 718 superalloy will need the dissimilar joining technology. However, there are few researches have been reported previously concerning the dissimilar joining of these two materials [10,11].

In fact, a reliable and strong joining of Ti$_3$Al to Ni-based superalloy appears to be extremely difficult owing to a lack of metallurgical compatibility between these two materials such as their different chemical compositions, different physical properties as well as the high reactivity between Ti and Ni [12–14]. The dissolution enthalpy of Ti in liquid Ni solvent is $-170 \text{kJ/mol}$ [15], indicating the high affinity of the two elements and kinds of Ti–Ni intermetallics such as Ti$_2$Ni, TiNi and TiNi$_3$ could form according to the Ti–Ni binary alloy phase diagram [16]. This could result in the strong tendency to form brittle intermetallic phases across the joint interface between Ti$_3$Al-based alloy and Ni-based superalloy, which is detrimental to the mechanical properties of the joint and could even induce cracking [17]. Chen et al. made an attempt to join a Ti$_3$Al-based alloy to a Ni-based superalloy using a Ti–Zr–Cu–Ni filler alloy, but the corresponding joint strength was only 86.4 MPa [10].

The present study aims to investigate the dissimilar joining of a Ti$_3$Al-based alloy to a Ni-based superalloy by gas tungsten arc (GTA) welding technology using Ti–Nb and Ti–Ni–Nb filler alloys. It is considered that the good ductility of Ti–Nb and Ti–Ni–Nb alloys is favorable to plastic deformation and residual stress relieving of the joints [18,19]. Microstructure evolutions along the dissimilar joints and the interface metallurgical behaviors have been analyzed. In addition, the mechanical properties of the joints have also been studied to clarify the relationship between microstructure and joint performances.

2. Experimental procedures

2.1. Materials

The Ti$_3$Al-based alloy (Ti–24Al–15Nb–1Mo at%) used in this research was composed of $\alpha_2$+B2+O three phase equiaxial grains [20] and was prepared by the following steps: vacuum-consumable electrode arc-melting, breaking down in the $\beta$/B2 phase fields, forging and rolling in the $\alpha_2$+B2 phase field and heat treating at 1253 K for 1 h followed by cooling in air. The other base material to be joined was Inconel 718 (In718), a $\gamma^\prime$ precipitation-hardened Ni-based superalloy. The chemical composition of the In718 alloy is listed in Table 1. Rectangular plates of 2.5 mm in thickness, 45 mm in width and 75 mm in length were machined from the Ti–24Al–15Nb–1Mo and In718 alloys for welding experiments. Oxide layers on the surfaces of the plates were removed by mechanical polishing and cleaning with acetone.

A Ti–Nb alloy and a Ti–Ni–Nb alloy were used as filler materials in this work, respectively and their chemical compositions are also given in Table 1. The diameter of the purchased Ti–Nb wire was 1.6 mm and the Ti–Ni–Nb filler alloy had the cross section geometry of $3 \times 3 \text{mm}^2$, which was machined from the purchased material.

2.2. GTA welding experiments

Before welding, the Ti–24Al–15Nb–1Mo and In718 plates were preheated to 573 K in an air heat treatment furnace. Then GTA welding between the dissimilar base alloys were carried out using the Ti–Nb and Ti–Ni–Nb alloys as filler materials, respectively. The processing parameters of the welding experiments were as follows: welding speed 60–100 mm/min, current 40–65 A DC and voltage 10–16 V. To prevent oxidation during the welding processes, flowing argon shielding was applied to the top, tail and bottom surfaces of the weld beads at a flow rate of 15 L/min, 20 L/min and 8 L/min, respectively. After welding, the samples were subjected to furnace cooling from about 573 K to room temperature at a cooling rate of 0.02–0.05 K/s.

2.3. Microstructure and mechanical properties

The as-welded samples were sectioned transversely to the welding direction and polished using standard metallographic techniques. Microstructure evolutions along the cross sections of the dissimilar joints were then investigated using a scanning electron microscopy (SEM; CS-3400) equipped with an X-ray energy dispersive spectroscopy (EDS). Micro-hardness values of various regions on the cross sections of the joints were measured using a Vickers hardness tester (450-SVD) with a load of 200 g and load time of 15 s. At least 5 indentations were performed at each region. Transverse tensile specimens were produced and tested at room temperature to evaluate the tensile strength of the joints. The fractured surfaces of the joints were examined under the SEM, EDS and X-ray diffractometer (XRD) to investigate the positions that the fractures occurred and to verify the phase constitutions of the weld/In718 interfaces.

Table 1
Chemical compositions (wt%) of the In718 alloy and the filler alloys.

<table>
<thead>
<tr>
<th>Material</th>
<th>Ti</th>
<th>Ni</th>
<th>Cr</th>
<th>Nb</th>
<th>Fe</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>In718</td>
<td>0.7–1.1</td>
<td>50.0–55.0</td>
<td>17.0–21.0</td>
<td>4.8–5.5</td>
<td>Balance</td>
<td>2.8–3.3</td>
</tr>
<tr>
<td>Ti–Nb</td>
<td>37.0–39.0</td>
<td>/</td>
<td>/</td>
<td>/</td>
<td>Balance</td>
<td>/</td>
</tr>
<tr>
<td>Ti–Ni–Nb</td>
<td>37.5–40.5</td>
<td>51.5–54.5</td>
<td>/</td>
<td>7.3–9.3</td>
<td>/</td>
<td>/</td>
</tr>
</tbody>
</table>
3. Results and discussions

3.1. Microstructure evolution of the Ti₃Al/In718 joint welded with Ti–Nb

Microstructure evolution along the cross section of the dissimilar Ti₃Al/In718 joint welded with the Ti–Nb filler alloy is shown in Fig. 1. Chemical compositions of the representative phases in the joint were inspected by EDS and the results were given in Table 2, along with the deduced phases. As seen in Fig. 1a, a sound joint of the Ti₃Al and In718 base alloys has been obtained using the Ti–Nb filler alloy. The weld with ~5 mm thickness is distinguished from the base materials and no interface cracks are visible.

Fig. 1b displays the back-scattered electron (BSE) image of the Ti₃Al/weld interface of the joint, which is free from a reaction layer. The left side in the picture is the Ti₃Al base alloy and the right side is the weld. It is seen that the weld is composed of dendritic crystals showing white contrast distributed in the dark matrix. From the EDS analysis results listed in Table 2, the dark matrix as marked by “1” is (Ti, Nb) solid solution dissolved with Ni, Fe and Al and the white phase labeled “2” is (Ti, Nb) solid solution dissolved with Cr and Al. During the welding process, elements from the base alloys diffuse and dissolve into the weld and solid solutions of multi-elements are formed.

Microstructure of the weld/In718 interface is presented in Fig. 1c. In contrast with the Ti₃Al/weld interface, this region

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**Table 2**

Chemical compositions of the representative phases in the Ti₃Al/In718 joint welded with the Ti–Nb filler alloy.

<table>
<thead>
<tr>
<th>Position</th>
<th>Composition (at%)</th>
<th>Deduced phase</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Ti</td>
<td>Ni</td>
</tr>
<tr>
<td>1</td>
<td>51.2</td>
<td>10.4</td>
</tr>
<tr>
<td>2</td>
<td>43.6</td>
<td>2.8</td>
</tr>
<tr>
<td>3</td>
<td>9.9</td>
<td>41.8</td>
</tr>
<tr>
<td>4</td>
<td>7.4</td>
<td>33.6</td>
</tr>
<tr>
<td>5</td>
<td>55.3</td>
<td>25.0</td>
</tr>
<tr>
<td>6</td>
<td>32.0</td>
<td>16.3</td>
</tr>
<tr>
<td>7</td>
<td>32.1</td>
<td>2.3</td>
</tr>
<tr>
<td>8</td>
<td>50.6</td>
<td>19.9</td>
</tr>
</tbody>
</table>
Fig. 1d is a magnification image of the region marked by a white square in Fig. 1c. Typical phases of the weld/In718 interface can be seen in this reaction region. Fig. 2 shows the XRD pattern of the fractured surface of the joint welded with Ti-Nb after the tensile test (presented below in section 3.3.2). According to the experimental results, all the fractures occurred at the weld/In718 interfaces during the tensile tests. Therefore it is considered that the XRD analysis of the fractured surface of the joint should be helpful to verify the phase constitution of the weld/In718 interface. Based on the Ti–Ni binary alloy phase diagram [16], the phase exhibiting dark contrast as marked by “5” in Fig. 1d is deduced to be Ti$_2$Ni intermetallics. And it is seen in Fig. 2 that the XRD peaks associated with Ti$_2$Ni are detected. This phase is a reactive product of Ti from the filler alloy and Ni from the In718 base alloy, and it has been also identified as one of the major phases generated in the joints when brazing TiAl with Ti–Cu–Ni filler alloys previously [21,22].

As deduced from the EDS analysis results, the light gray phase as marked by “6” in Fig. 1d is (Ti, Nb) solid solution dissolved with Ni, Fe and Cr and the white phase (labeled “7” in Fig. 1d) is also considered to be (Ti, Nb) solid solution, with Cr dissolved. The presence of these phases is further confirmed by the XRD analysis results, as the XRD peaks associated with (NbTi) solid solution are observed in Fig. 2. In addition, the bright precipitated particles (labeled “8” in Fig. 1d) are considered to be solid solution of elements Ti, Ni and Nb as presented in Table 2.

3.2. Microstructure evolution of the Ti$_3$Al/In718 joint welded with Ti–Ni–Nb

Fig. 3 presents the microstructure evolution along the cross section of the Ti$_3$Al/In718 joint welded with the Ti–Ni–Nb filler alloy and a full penetration weld has been obtained as seen in Fig. 3a. In Fig. 3b, it is observed that a transitional layer with the thickness of ~100 μm is formed at the Ti$_3$Al/weld interface during the welding process. According to the EDS analysis results listed in Table 3, the phase exhibiting white contrast that dominates in this layer (labeled “1” in Fig. 3b) is (Ti, Ni, Nb) solid solution, but dissolved with Al and the concentration of Nb increases as next to the Ti$_3$Al base alloy. It is noted that along the boundary of the interface close to the Ti$_3$Al side, some bright particles are precipitated within a thin area of 8–20 μm thick (labeled “2” in Fig. 3b). The EDS analysis results indicate that these precipitates are depleted in Ni, when Mo and Nb concentrate. They are (Ti, Nb) solid solutions dissolved with Al and Mo. Besides, this interface also contains a few crystals showing dark gray contrast, which are Ti$_3$Ni intermetallics (labeled “3” in Fig. 3b).

The microstructure of the Ti–Ni–Nb weld is shown in Fig. 3c and the corresponding EDS analysis results are also given in Table 3. The weld is characterized by a continuous network of bright structure distributed in the dark matrix and the chemical composition of this region is close to that of the original Ti–Ni–Nb filler alloy.

Fig. 3d displays the BSE image of the weld/In718 interface and microcracks can be observed within this region. Fig. 4 displays the XRD pattern of the fractured surface of the joint welded with Ti–Ni–Nb after the tensile test, in which the phases at the weld/In718 interface should be detected. This interface is composed of various reactive products of the In718 and Ti–Ni–Nb alloys. As presented in Table 3, the region next to the weld (labeled “6” in Fig. 3d) contains ~31 at% Ti and ~54 at% Ni as well as several atom percents of Nb, Fe and Cr. According to the Ti–Ni binary alloy phase diagram, in the compositional range with Ni between 50 at.% and 75 at.%, Ti–Ni alloys form a two phase equilibrium between TiNi and TiNi$_3$ [16]. Therefore, the phase constitution of this region is deduced to be TiNi+TiNi$_3$. It is seen in Fig. 4 that both the TiNi and TiNi$_3$ phases have been detected by XRD. In a previous study, Fukumoto [23] indicated that the reaction layer formed at the TiNi/TiNi interface consisted of TiNi$_3$ as the primary crystal and a TiNi+TiNi$_3$ eutectic microstructure, when TiNi alloy and stainless steel were friction welded using Ni interlayer. This is in accordance with the experimental observation in this work. It is noted from the EDS analysis results that, in this interface, as the position is closer to the In718 base alloy, the stoichiometric ratio of Ni and Ti as well as the concentrations of Fe and Cr in the gray matrix increase gradually. The location marked by “7” in the picture is also composed of TiNi+TiNi$_3$, but dissolved with 10 at% Fe and 10 at% Cr and the TiNi$_3$ concentration increases. For the
position adjacent to the In718 base alloy (labeled “8” in Fig. 3d), the stoichiometric ratio of Ni and Ti turns into 3:1, so the phase constitution is mainly TiNi3 and the concentrations of Fe and Cr become higher.

Additionally, lots of white cellular and dendritic crystals are embedded in the Ti3Al/TiNi–Nb intermetallic matrix (labeled “9” in Fig. 3d). Based on the EDS and XRD analysis results, these distributed precipitates are presumed to be a mixture of Ni3Nb + TiNi + TiNi3 intermetallic compounds dissolved with Fe and Cr, which are formed due to the reactions of Ni, Ti and Nb from the filler material and Fe as well as Cr from the In718 base alloy. It is recognized that when the In718 superalloy is exposed at temperatures above 923 K, the principal strengthening phase γ”-Ni3Nb with ordered bct DO22 crystal structure can be transformed to detrimental δ-Ni3Nb with the crystal structure of ordered fcc Li2 [24]. Therefore in this research, it is not strange that Ni3Nb phase is observed at the weld/In718 interface when the In718 and Ti–Ni–Nb alloys go through such a high temperature GTA welding process.

In this study, the microcracks observed at the weld/In718 interface could not be avoided when using the Ti–Ni–Nb alloy as filler material, which is attributed to the brittle compounds

Table 3

<table>
<thead>
<tr>
<th>Position</th>
<th>Composition (at%)</th>
<th>Deduced phase</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Ti 37.8, Ni 25.6, Nb 18.8, Fe 1.6, Cr 2.6, Al 13.6, Mo /</td>
<td>(Ti, Ni, Nb) solid solution dissolved with Al</td>
</tr>
<tr>
<td>2</td>
<td>Ti 37.0, Ni 4.7, Nb 35.4, Fe 0.6, Cr 1.6, Al 10.4, Mo 10.3, /</td>
<td>(Ti, Nb) solid solution dissolved with Al and Mo</td>
</tr>
<tr>
<td>3</td>
<td>Ti 59.5, Ni 29.8, Nb 3.3, Fe 1.7, Cr 0.9, Al 4.8, Mo /</td>
<td>Ti3Ni</td>
</tr>
<tr>
<td>4</td>
<td>Ti 40.0, Ni 36.6, Nb 15.1, Fe 1.7, Cr 4.8, Al 1.8, Mo /</td>
<td>Ti–Ni–Nb filler alloy</td>
</tr>
<tr>
<td>5</td>
<td>Ti 44.2, Ni 45.8, Nb 4.5, Fe 2.1, Cr 1.5, Al 1.9, Mo /</td>
<td>Ti–Ni–Nb filler alloy</td>
</tr>
<tr>
<td>6</td>
<td>Ti 30.9, Ni 53.7, Nb 4.9, Fe 4.4, Cr 4.6, Al 1.5, Mo /</td>
<td>TiNi + TiNi3</td>
</tr>
<tr>
<td>7</td>
<td>Ti 22.3, Ni 53.5, Nb 1.9, Fe 10.5, Cr 10.2, Al 1.6, Mo /</td>
<td>TiNi + TiNi3 dissolved with Fe and Cr</td>
</tr>
<tr>
<td>8</td>
<td>Ti 16.9, Ni 47.5, Nb 1.9, Fe 13.8, Cr 18.4, Al 1.5, Mo /</td>
<td>TiNi3 dissolved with Fe and Cr</td>
</tr>
<tr>
<td>9</td>
<td>Ti 21.1, Ni 31.9, Nb 13.0, Fe 14.8, Cr 18.1, Al 1.1, Mo /</td>
<td>Ni3Nb + TiNi + TiNi3 dissolved with Fe and Cr</td>
</tr>
</tbody>
</table>

Fig. 3. Microstructure evolution along the cross section of the Ti3Al/In718 joint welded with the Ti–Ni–Nb filler alloy: (a) Low magnification image, (b) Ti3Al/weld interface, (c) weld and (d) weld/In718 interface.
formed during the welding process. It is obvious that the presence of the TiNi, TiNi3 and Ni3Nb intermetallic compounds are detrimental to mechanical properties of the joint.

3.3. Mechanical properties of the Ti3Al/In718 joints

3.3.1. Micro-hardness

Vickers micro-hardness profiles across the dissimilar Ti3Al/In718 joints are presented in Fig. 5. The hardness values of the interfaces are higher than those of the base materials as well as the welds for both filler alloys. In the joint welded with the Ti–Nb filler alloy, the maximum hardness (822 HV) appears at the weld/In718 interface, where hard Ti2Ni intermetallics forms during the welding process. And the hardness value decreases gradually in the direction to the weld. At the Ti3Al/weld interface, the micro-hardness at the weld side (631 HV) is higher than that in the central part of the weld (540 HV) and this is attributed to the strengthening effect of multi-element dissolving [25,26].

For the joint welded with the Ti–Ni–Nb filler alloy, the hardness value exhibits similar variation trend. According to the micro-hardness measurement results, the 8–20 μm thick layer along the boundary of the Ti3Al/weld interface (as seen in Fig. 3b) shows a sudden increased micro-hardness (801 HV) as shown in Fig. 5b. At the weld/In718 interface, the micro-hardness increases sharply to 882 HV at the position near the In718 base alloy. This high micro-hardness in this interface is caused by the presence of the hard intermetallic compounds produced during welding, such as TiNi, TiNi3 and Ni3Nb, with Fe and Cr dissolved [27]. In the weld, the hardness is at a relatively lower level of 570–580 HV.

3.3.2. Tensile strength and fractographic observation

Table 4 shows the tensile strength of the GTA welded dissimilar Ti3Al/In718 joints. The average room-temperature tensile strength of the joints welded with the Ti–Nb filler alloy is calculated to be 202 MPa and the value of the joints achieved with Ti–Ni–Nb is 128 MPa. Fractured surfaces of the joints were inspected under the SEM. As shown in Fig. 6a, the fracture of the joint welded with Ti–Nb exhibits a quasi-cleavage feature. At the edges of the quasi-cleavage facets in the fractured surface, there exit ductile features such as some dimples and tear ridges as marked by white arrows in the picture. In contrast, the joint using the Ti–Ni–Nb alloy as filler material indicates a cleavage dominated fracture featured with cleavage facets and river markings, as shown in Fig. 6b. This reveals that the joint made with Ti–Nb has better ductility than the one made with Ti–Ni–Nb. However, both fractured surfaces contain secondary cracks, indicating the inherent brittleness of the joints.

Via macroscopic observation of the samples after the tensile tests, the fractures monotonously take place somewhere at the weld/In718 interfaces during the tensile tests for both the filler alloys. According to the EDS analysis results (as presented in Table 5) on the fractured surface of the joint welded with Ti–Nb, the chemical compositions of the positions labeled “1” and

<table>
<thead>
<tr>
<th>Filler alloy</th>
<th>Tensile strength (MPa)</th>
<th>Average value (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti–Nb</td>
<td>185, 217, 203</td>
<td>202</td>
</tr>
<tr>
<td>Ti–Ni–Nb</td>
<td>144, 115, 126</td>
<td>128</td>
</tr>
</tbody>
</table>

Fig. 4. XRD pattern of the fractured surface of the joint welded with the Ti–Ni–Nb filler alloy.

Fig. 5. Vickers micro-hardness profiles across the Ti3Al/In718 joints welded with (a) the Ti–Nb filler alloy and (b) the Ti–Ni–Nb filler alloy.
in Fig. 6a are very close to those of the phases labeled “7” and “6” in Fig. 1d, which further proves that the weld/In718 interface is the weak link of the joint welded with the Ti–Nb filler alloy.

For the joint welded with the Ti–Ni–Nb filler alloy, the average chemical composition of the fractured surface as given in Table 5 is close to that of the weld/In718 interface where the fracture occurs. This is in agreement with the observation of microcracks in this region. The formation of the TiNi, TiNi3 and Ni3Nb intermetallics at the weld/In718 interface is the main reason for the decrease of tensile strength of the joint welded with the Ti–Ni–Nb filler alloy. The interface characterized by those brittle compounds is the weak link of the joint where cracks initiate and propagate during the tensile test. As discussed above, when using the Ti–Nb filler alloy without Ni element, the amount of brittle compounds across the joint interface is much less than that in the joint welded with Ti–Ni–Nb. The tensile strength of the joint using the Ti3Al base alloy and microcracks are observable at the weld/In718 interface, which is attributed to the formation of TiNi, TiNi3 and Ni3Nb in this region during welding.

In this study, using Ti–Nb and Ti–Ni–Nb filler alloys, the Ti3Al and In718 base alloys have been successfully joined by GTA welding technology. For both fillers, the weak links of the joints locate at the weld/In718 interfaces, and adopting the Ti–Nb filler alloy could help to decrease the formation tendency of interfacial brittle phases and thus enhance the tensile strength of the joint. This study has verified the possibility of the joining between Ti3Al-based alloy and Ni-based superalloy by GTA welding technology. However, more research should be carried out to inspect the interfacial metallurgical behaviors of other elements and further improve the microstructure and mechanical properties of the joints between Ti3Al-based alloy and Ni-based superalloy [28,29].

### 4. Conclusion

Dissimilar GTA welding between Ti3Al-based alloy and Ni-based superalloy has been successfully achieved using the Ti–Nb and Ti–Ni–Nb filler alloys. Microstructure evolutions and mechanical properties of the joints have been investigated and the main conclusions are as follows:

1. Sound joints of the Ti3Al and In718 base alloys have been obtained using the Ti–Nb filler alloy. The Ti3Al/weld interface is free from a reaction layer and the phase constitution of the weld/In718 interface is mainly (Ti, Nb) and (Ti, Ni, Nb) solid solutions as well as some Ti2Ni intermetallics. For the joint welded with the Ti–Ni–Nb filler alloy, the weld shows good combination with the Ti3Al base alloy and microcracks are observable at the weld/In718 interface, which is attributed to the formation of TiNi, TiNi3 and Ni3Nb in this region during welding.
2. For both the two filler alloys, the joint interfaces exhibit higher micro-hardness than the base materials and the welds. The average room-temperature tensile strength of
the joint welded with the Ti–Nb filler alloy is 202 MPa and the strength value of the joint welded with the Ti–Ni–Nb filler alloy is 128 MPa. The weld/In718 interfaces are the weak links of the joints for both fillers.

(3) In this study, the possibility of the joining of Ti3Al-based alloy and Ni-based superalloy by GTA welding technology has been verified. With the Ti–Nb alloy as filler material, the formation tendency of interfacial brittle phases is decreased to a certain extent and the tensile strength of the dissimilar joint is improved.

References