Temperature Dependence of Tensile Properties and Fracture Behavior of Welded Joints of Titanium Matrix Composites in Gas Tungsten Arc Welding

Jianwei Mao¹, Liqiang Wang, Weijie Lu², Di Zhang and Jining Qin

State Key Laboratory of Metal Matrix Composites, Shanghai Jiao Tong University, 800# Dongchuan Road, Shanghai 200240, P. R. China

The characteristics of gas tungsten are weldments of in-situ reinforced titanium matrix composites were investigated, and the joint quality was evaluated by means of weld morphology, microstructure and tensile tests. With an increase of welding process parameters, sound welded joints were fabricated. The weld zone had a refinement microstructure and TiB₆ exhibited smaller sizes and dispersed distribution, forming a novel network structure in the weld. Tensile tests indicated that the weld presented excellent high temperature strengths, even superior to the base metal under the welding conditions. The welded joints displayed a slower downtrend in strength than that of base metal with an increase of temperature, and the elongation values of joints were higher than the base metal because of the refined grain microstructure and dispersed distribution of TiB₆ in the weld. The strengthening mechanism of high temperature properties of welded joints is ascribed to smaller sizes and network structure of TiB₆ in the weld. The fracture surfaces of butt joints for TMCs include the hybrid ductile of matrix and brittle fracture.

Keywords: titanium matrix composites, welding, microstructure, high temperature properties, fracture

1. Introduction

Titanium matrix composites (TMCs) are attractive candidates for structural materials for applications in aerospace, nuclear and military industries due to their superior stiffness, excellent specific strength and good strength retention at high temperatures.¹,² However, TMCs are more expensive and only used when the technical characteristics of TMCs justify the increased costs.³

TMCs can be shaped by means of many processes. Generally, these methods are more costly and sophisticated than those used for other popular structural materials.⁴ Traditional casting and processing techniques may be replaced by modern methods e.g., superplastic forming, diffusion bonding and joining methods and they are used to make complete structures (i.e., near the shape) or smaller components that are fabricated into a complete structure. To produce TMCs components, there is a balance between the cost of forming the complete structure by a single process and the cost of joining together simpler parts.⁵ In many cases, joining is indeed an important manufacturing process and shows higher suitability and lower costs than other processing methods.

Titanium alloys are generally joined using the fusion welding processes, and gas tungsten arc welding (GTAW) is the most commonly used precision arc welding process. GTAW can produce the same quality welds as those made by electron beam welding and laser beam welding at lower costs.⁶ GTAW has become a more reliable method for welding TMCs and elevated temperature behaviors of weldments at different temperatures.⁷ Moreover, in situ synthesized TiB used is thermodynamically stable and essentially insoluble in titanium,⁸ and there are no intermediate phases as well as reaction-free interface between TiB and titanium.⁹ These advantages offer great chances to obtain the desired joint using GTAW for TMCs. So far, there exist few published research reports on discussing welding TMCs using a GTAW technique. In this paper, an attempt is made to investigate gas tungsten arc welding TMCs and elevated temperature behaviors of weldments at different temperatures.

2. Experimental Procedure

2.0 mm thick annealed sheets of TMCs were selected as the base metal in this paper. The matrix alloy is near-α titanium alloy, whose chemical compositions are similar to IMI 834 alloy. Reinforcements TiB and La₂O₃ were in-situ synthesized based on the reaction: 12Ti + 2LaB₆ + 3[O] = 12TiB + La₂O₃. Theoretical volume fractions of TiB and La₂O₃ are 1.82 and 0.58%, respectively. The β transus temperature of TMCs is 1040°C. TMCs were melted in a vacuum arc remelting furnace. The ingots were hot-forged into billets at 1150°C, and finally hot-rolled into sheets with thickness of 2.0 mm at 1100°C. The sheets were annealed at 650°C for 2 h to relieve the residual stress, and then cut into (130 × 55 × 2.0 mm) size coupons (see Fig. 1(a)).

Before welding, all coupons were accurately descaled with molten alkaline-based salt bath at 400°C, then pickled with 20 vol% H₂SO₄ aqueous solution at 60°C and a solution composed of 2 vol% HF and 25 vol% HNO₃ in water at 50°C and finally cleaned with acetone. Without filler metal used, all the coupons were autogenously welded in the butt welding perpendicular to the rolling direction using an arc welding machine. The weld area was fully protected with purity argon as a shielding gas. The major welding parameters

¹Graduate Student, Shanghai Jiao Tong University. Present address: State Key Laboratory of Metal Matrix Composites, 800# Dongchuan Road, Shanghai 200240, P. R. China
²Corresponding author, E-mail: luweijie@sjtu.edu.cn
The upper reveals the backside appearance. Visibly, if the welding current is too high or the welding speed is too low, e.g., condition G81, G83 and G134, the weld will receive much higher heat input. As a result, the penetration depth will be too high so that the resulting weld tends to melt through the base metal being joined and presents excessive weld width. On the contrary, as the speed is too fast or the current is too low, heat input, which enables the heat transfer along the long direction of welds, definitely decreases and much less volume of base metal is heated. Therefore, inadequate penetration or discontinuity weld may appear in the weld. Increasing welding speed or reducing arc current can contribute to improve the weld quality. Sound welds are obtained under appropriate welding parameters, e.g., condition G82, G93 and G113 and bright silver welds reveals that the weld shielding is satisfactory. Metallographic examination shows that neither porosity nor welding cracks are found in any of the welds (see Fig. 2(b)).

3.2 Microstructure of the weld

In Fig. 3(a), the microstructure of base metal consists of the \( \beta \) phase distributed at the elongated \( \alpha \) phase grain boundaries. Some whisker substances are found dispersed in the matrix. As analyzed by Lu,\(^1\) they are TiB whisker (TiB\(_{\text{w}}\)) reinforcements with an aspect ratio (AR; length-to-diameter ratio) of 4–5, which were \textit{in-situ} synthesized based on the reaction referred above during the melting process of titanium alloys.

In Fig. 3(b), besides base metal (BM), each weld is composed of fusion zone (FZ) and heat affected zone (HAZ). During the welding, the welded joint undergoes a large temperature gradient from the weld pool to base metal and those spatial and temporal changes in temperature result in a non-uniform microstructure in the weld.\(^8\) Compared to base metal, the fusion zone typically exhibit columnar grains and the microstructure are refined. Especially in HAZ, the high temperature side close to FZ, named HAZ 1, shows coarse equiaxed grains, which is ascribed to the poor thermal conductivity of titanium and the exposure to temperature above the \( \beta \) transus for sufficient length of time. By contrast, the low temperature zone, named HAZ 2, exhibits no evident microstructure changes, similar to that of the base metal. In addition, TiB\(_{\text{w}}\) in HAZ 2 display smaller sizes and more dispersed distribution, and in FZ and HAZ 1, TiB\(_{\text{w}}\), having a higher AR of 13–15 than that of base metal, redistribute at \( \beta \) grain boundaries to form a novel network-like structure, as shown in Figs. 3(c)–3(d) and Figs. 4(a)–4(b). Obviously, the weld microstructure gradually transits from FZ to BM through the HAZ (Figs. 3(b)–3(d)), and this reveals that the weld bead and the base metal are fused well and the structure of tight junctions is favourable.

From Fig. 3(b), the columnar grains in FZ start by epitaxial growth from the HAZ 1 and develop towards the weld center along the reverse direction of rejection of heat, because the partially melted grains of HAZ 1 supply the non-spontaneous nucleation during the solidification of melting bath. As the \( \beta \) phase appears, the solute enrichment at the solid/liquid interface causes the constitutional supercooling, and this can enhance the growth of primary and secondary dendrite

<table>
<thead>
<tr>
<th>Condition</th>
<th>Welding current (A)</th>
<th>Welding speed (cm/min)</th>
<th>Shielding gas/L·min(^{-1})</th>
<th>Failure location of welded tensile specimens</th>
</tr>
</thead>
<tbody>
<tr>
<td>G00</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>---</td>
</tr>
<tr>
<td>G81</td>
<td>80</td>
<td>10</td>
<td>17</td>
<td>FZ</td>
</tr>
<tr>
<td>G82</td>
<td>80</td>
<td>20</td>
<td>16</td>
<td>HAZ</td>
</tr>
<tr>
<td>G83</td>
<td>80</td>
<td>30</td>
<td>15</td>
<td>BM</td>
</tr>
<tr>
<td>G93</td>
<td>90</td>
<td>30</td>
<td>17</td>
<td>BM</td>
</tr>
<tr>
<td>G113</td>
<td>110</td>
<td>30</td>
<td>18</td>
<td>BM</td>
</tr>
<tr>
<td>G134</td>
<td>130</td>
<td>40</td>
<td>17</td>
<td>BM</td>
</tr>
</tbody>
</table>

are shown in Table 1, and the parameters were selected to avoid macrodefects in the weld and excessive flash on the joint surface. Condition name GT81 means that the welding current and welding speed is 80 A and 10 cm·min\(^{-1}\) respectively and all others are similar.

After welding, samples for microstructural investigation were cut perpendicular to the welding direction. The samples were prepared via conventional metallographic techniques and etched with Kroll’s reagent for 20 s. Microstructural examinations were made using optical microscope and scanning electron microscope (SEM). The joints were sliced and etched with Kroll’s reagent for 20 s. Microstructural examinations were made using optical microscope and scanning electron microscope (SEM). The joints were sliced and etched with Kroll’s reagent for 20 s.

3. Results and Discussion

3.1 Macromorphology of the weld

Figure 2(a) shows the appearances of weld beads obtained at different welding process parameters. In each condition,
and dendrite arms. The formation of barrier with extra B at solid/liquid interfaces reduces the grain growth.\textsuperscript{12} Therefore, these combined effects will yield smaller grain sizes in the weld. Besides, the effective temperature of heat source for GTAW, higher than 3000°C,\textsuperscript{7} is much higher than the melting point of TiB\textsubscript{w} (2000°C),\textsuperscript{1} and TiB\textsubscript{w} can be entirely dissolved in the fused bath. However, the solid solubility of B in titanium is very low.\textsuperscript{11} Based on the Ti–B phase diagram, TiB\textsubscript{w} will precipitate after the formation of the β phases. Tamirisakandala\textsuperscript{9} believed that the coherent interface between β phases can offer an energetic priority nucleation site for TiB, and this causes the formation of numerous TiB nuclei. Rapid cooling of liquid phases will reduce the growth rate of TiB\textsubscript{w}. As a result, TiB\textsubscript{w} having smaller sizes finally precipitate and grow at β grain boundaries in the weld.

Further analysis reveals that, in Figs. 4(a)–4(b), the martensitic α′ transformed from the β grains, which corresponds to a structure quenched from the β phase above the beta transus\textsuperscript{13} constitutes the microstructure of the FZ and HAZ 1. The HAZ 2 further away from the FZ exhibits little microstructure changes, which is similar to that of the base metal. Interestingly, in HAZ 2, TiB\textsubscript{w} seems to be divided into two parts and then taken to the adjacent region (Fig. 4(c)) owing to the effect of welding thermal cycle,
and this roughly reveals the changes of TiB$_{w}$ in the weld during GTAW. HAZ 2 receives much less heat input than the FZ and HAZ 1, and TiB$_{w}$ cannot be directly dissolved into the matrix, but the diffusion of boron atoms becomes more intensified. Therefore, TiB$_{w}$ partly moves from one position to another during the welding. Evidently, the joint microstructure, the distribution and size of TiB$_{w}$ in each weld zone are not uniform, and this is chiefly ascribed to the thermal cycle process of GTAW and physical properties of titanium.\(^7\)

### 3.3 Elevated mechanical properties

Figure 5 displays the stress-displacement curves of joints and base metal at different temperatures, and the tensile results are displayed in Table 2. At 600 and 650°C, the ultimate tensile strengths (UTS) of base metal are 740 and 550 MPa, respectively. As seen in Fig. 5 and Table 2, all the butt-joints show good tensile strengths compared to the base metal at the same condition. At 600°C, the joint strengths are higher than 85% of the base metal strength. With increasing welding parameters, the butt-joint strengths are gradually improved. For example, in condition G81, the joint strength is 602 MPa, and then enhanced to 690 MPa, a 15% increase, in condition G82. The peak value of joint strength, up to 710 MPa, is obtained in condition G113, and it almost reaches 96% of the base metal strength.

At the temperature up to 650°C (Fig. 5(b)), UTS decrease for all joint samples, and the reduction in strength is 16% in average, which is ascribed to the great softening of the matrix at the high temperature. This phenomenon was also explained by the drastic reduction in matrix shear strength and the relief of stress concentration around TiB$_{w}$ caused by the dislocation climb and dynamic recovery process.\(^{15}\) Significantly, with the increasing of welding parameters, the

![Fig. 5 Stress-displacement curves of joints and base metal at different temperatures (a) 600°C; (b) 650°C.](image)

<table>
<thead>
<tr>
<th>Condition</th>
<th>UTS/MPa</th>
<th>Average UTS</th>
<th>EL/%</th>
<th>Average EL</th>
<th>UTS/MPa</th>
<th>Average UTS</th>
<th>EL/%</th>
<th>Average EL</th>
</tr>
</thead>
<tbody>
<tr>
<td>G00</td>
<td>755</td>
<td>740</td>
<td>11.0</td>
<td>9.6</td>
<td>560</td>
<td>550</td>
<td>24.5</td>
<td>23.9</td>
</tr>
<tr>
<td>G81</td>
<td>610</td>
<td>602</td>
<td>7.9</td>
<td>8.4</td>
<td>532</td>
<td>538</td>
<td>22.0</td>
<td>22.2</td>
</tr>
<tr>
<td>G82</td>
<td>660</td>
<td>659</td>
<td>13.4</td>
<td>13.6</td>
<td>560</td>
<td>552</td>
<td>32.5</td>
<td>32.3</td>
</tr>
<tr>
<td>G83</td>
<td>645</td>
<td>641</td>
<td>10.4</td>
<td>10.2</td>
<td>570</td>
<td>561</td>
<td>23.1</td>
<td>23.8</td>
</tr>
<tr>
<td>G93</td>
<td>696</td>
<td>691</td>
<td>15.6</td>
<td>15.2</td>
<td>578</td>
<td>584</td>
<td>26.2</td>
<td>25.5</td>
</tr>
<tr>
<td>G113</td>
<td>702</td>
<td>710</td>
<td>15.2</td>
<td>580</td>
<td>603</td>
<td>603</td>
<td>32.2</td>
<td>32.2</td>
</tr>
<tr>
<td>G134</td>
<td>712</td>
<td>699</td>
<td>7.8</td>
<td>7.6</td>
<td>586</td>
<td>595</td>
<td>24.8</td>
<td>24.4</td>
</tr>
</tbody>
</table>

*Room tensile properties of base metal and welded joint were 1220 MPa and 1150 MPa respectively.*
joint strengths reach the full strength, even superior to the base metal. Besides, the joint strengths display a slower degradation trend than the base metal. At 600°C, the average strength loss of joints is 38% less than 39% of base metal, and the loss, 49%, of joints is much lower than 55% of base metal at 650°C.

As described above, the weld microstructure has been refined during GTAW, and TiBw are provided with smaller sizes and dispersed distribution in the weld. According to the Hall-Petch relationship, the tensile strength of joints is largely improved by the grain refinement mechanism. The distribution of TiB along prior β grain boundaries promotes the microstructure refinement,10) and enhances the microstructural stability at the high temperature by inhibiting the grain boundary mobility via Zener pinning effect.17) The network-boundary strengthening mechanism is viewed as an effective factor in tailoring the high temperature properties,11) and thus the network boundary containing TiBw still plays an important role in strengthening the joint. In addition, TiB has good interfacial bonding with the titanium1) and strengthens the joint through the load transfer mechanism,13) by which TiB is loaded in tension through the shear stress along the well bonded whisker/matrix interfaces. More importantly, the load-bearing capacity caused by the network structure and grain refinement due to boron addition is retained during the high temperature tensile tests.18) Therefore, the joints present much higher strength compared to the base metal at the high temperature.

Whereas the joint strength degrades with increasing temperature, the ductility—as measured by tensile elongation over 20 mm gauge length—displays the opposite trend. The joint elongations are much larger than those of base metal and increase gradually with increasing temperature, which is strongly dependent on the grain refinement and network structure of TiBw in the weld. At 600°C, the joint elongations are improved with increasing welding parameters. In condition G113, the maximum value of joint elongation is obtained. However, with further increase in welding parameters, e.g., in condition G134, the joint elongations decreases. As the temperature increases to 650°C, the differences in elongation between the joints and base metal become smaller, and this result is mainly ascribed to the great softening effect of the matrix with an increase in temperature.10)

All the tested joint samples show that the failure always fails in the base metal (see Fig. 6 for an example) and the welded joints of TMCs exhibit a good continuity of mechanical properties related to the base metal. Based on the research results, we can conclude that GTAW is a reliable joining method to make high quality welds for in situ TMCs, and it does not undergo undesirable changes or interface reactions in the weld microstructure, which may cause a deterioration of mechanical performances found in fusion welding SiCp/Al composites by GTAW.19)

3.4 Fractography

Figure 7 shows the fracture morphologies of tensile samples at high temperatures in comparison with the base metal. At 600°C, the fracture surfaces of butt joints and base metal consist of two regions, i.e., an area composed of fractured TiBw and the other area in the matrix including dimples. Therefore, the fracture mechanism is a combination of dimple patterns, tearing ridges in the matrix and brittle fracture of TiBw. Evidently, the fracture surface of base metal displays hybrid tough fracture of dimples and tear ridges (Fig. 7(a)), and dimples of the joint fracture surface are tiny and uniform (see Fig. 7(c)). With increasing the welding parameters, the fracture tearing edge of joints is higher and the tearing dimple is deeper (Fig. 7(e)).

At 650°C, the basic fracture characteristics of joints are similar to those observed at 600°C. However, the fracture surface of joints shows sharper and deeper dimples compared to the base metal and this reveals that they have experienced a large plastic deformation of the matrix. With temperature increasing, the size and depth of dimples increase, which is associated with the improved plastic behaviors (see Fig. 7 and Table 2). When the welding parameters change from condition G82 to condition G113, larger and deeper dimples are found in the joint fracture and the number of dimples increases (Figs. 7(e)–7(f)). Cracked TiBw are observed in the joint fracture and this reveals that TiBw still undertake the load-bearing effect during the high temperature deformation.14) TiBw pull-out is also found in the fracture surface (Figs. 7(d) and 7(f)). Xiao13) proved that the interface bonding strength decrease with an increase of test temperature, and when the AR of TiB is lower than the critical aspect ratio (ARc), the interfacial debonding is dominant at high temperatures. In this paper, failure of joint samples always occurred in the base metal, and the AR 4–5 of TiBw in the base metal is lower than the ARc value (>5.6) at 650°C.20) Therefore, TiBw may debond with matrix at the interface.

In order to understand the fracture process of joints, the longitudinal subsurface near the fracture is examined (see Fig. 8). At 600°C, the prevailing features in the longitudinal sections of joints are still composed of fracture of TiBw. Small voids are generated at the crack of TiB and they do not link between voids (Fig. 8(a)). At 650°C, voids can be found not only at the crack of TiBw but also at the end of TiBw and in the matrix, they grow longer along the tensile direction and finally link with each other. It is clear that the extent of whisker/matrix interface decohesion is more serious with increasing temperature, and interfacial debonding dominates the fracture mechanism of joints above 650°C.

Two damage mechanisms, whisker fracture and interface decohesion for TiBw, are found in the joint fracture surface.
and they play different roles dependent on the plastic behavior of the matrix at different temperatures. At 600°C, the whisker/matrix interface bonds well (Figs. 7 and 8), and those interfaces provide precondition for the effective undertaking stress of TiB$_{2x}$. At the early stage of the plastic deformation, the increased stress in TiB$_{2x}$ primarily due to the ongoing strain hardening of the matrix, leads to the rupture of the critical whisker,$^{20}$ and thus the crack is generated on the
The load relaxed by cracked TiB<sub>2</sub> is mainly taken up by the surrounding matrix and the crack will further grow into the matrix. As the strain hardening capacity of matrix becomes saturated, the stresses loosened by cracked TiB<sub>2</sub> are transferred to the adjacent whisker. The increased stresses acted on adjacent whiskers result in more whiskers fracture. The final fracture of the joint is ascribed to ductile failure of the matrix based on nucleation and growth of voids correlated to the fractured TiB<sub>2</sub>. As the temperature is 650°C, the matrix becomes softer and the strain hardening rate degrades. Local stresses around TiB<sub>2</sub> result from the constraint caused by TiB<sub>2</sub> on the plastic flow of the matrix, are relaxed partially through climbing of dislocations, and they decrease, even to a level lower than the whisker fracture stress. This leads to a lower proportion of cracked TiB<sub>2</sub>. Besides, the whisker/matrix interface usually becomes much weaker at the high temperature, and the interfacial strength is distinctly less. So, interface decohesion occurs due to the mismatch between the whisker and the matrix strain, and the failure mode transition from whisker cracking to debonding is anticipated with increasing temperature. It is noted that the formation of voids is always correlated to cracked TiB<sub>2</sub>. This reveals that TiB<sub>2</sub> can impede the plastic deformation of the matrix. As a result, larger plastic deformation around TiB<sub>2</sub> can be generated, resulting in the formation of voids. With increasing temperature, the plastic deformation around TiB<sub>2</sub> is further enhanced, and voids grow longer along the tensile direction and the internal damage is more serious. It can be known that the joint failure above 650°C is mainly initiated by interface decohesion and voids propagate progressively with further plastic deformation. As the voids link wholly, the butt joint will fail. This damage mechanism is similar to that in TiC/TA15 composite at 600 and 700°C.

4. Conclusions

1. *In-situ* titanium matrix composites having 2.0 mm thickness can be successfully joined by GTAW without filler metal used.
2. In FZ, the microstructure exhibits refined columnar grains, and coarse equiaxed grains are found in HAZ 1. TiB whiskers, with smaller sizes, redistribute at β grain boundaries to form the network structure in FZ and HAZ 1, and they display a more dispersed distribution in HAZ 2.
3. The strengths of TMCs joints made by GTAW are more than 85% of the base metal strength. With an increase of test temperature, the welded joints exhibit a slower downtrend in strength and reach the full-strength, even higher than the base metal. The load-bearing effect and network boundary strengthening mechanism play a key role in high temperature properties of welded joints.
4. Fracture of TiB<sub>2</sub> followed by the ductile failure of the matrix is the main failure mechanism of the joint below 600°C, while interfacial debonding dominates the mechanism of the joint above 650°C.

Acknowledgments

We would like to acknowledge the financial support provided by the National Nature Science Foundation of China under Grant No. 51371114, the 973 Program under Grant No. 2012CB619600, the Excellent Academic Leaders of Shanghai under Grant No. 12XD1402800 and the Dawn Program of Shanghai Education Commission under Grant No. 10SG15.

REFERENCES