Optimization of EB keyhole welding of aluminum to titanium alloy

Petr Havlík, Jan Čupera, Jan Kouřil, Ivo Dlouhý, Rudolf Foret

In directly keyhole electron beam welding of Ti/Al alloys the heterogeneous interfacial reactions were found in the weldments. The intermetallic brittle phases (e. g. Ti₃Al) formed at the weld joint and affected mechanical properties. To improve the homogeneity, change of electron beam offset and using of filler metal were studied. The effects of attempts were investigated by structural analysis, analysis of chemical composition and by hardness measurement. The results indicate that the modification of welding process from directly electron beam welding to electron beam welding-brazing process was more beneficial than using of filler metal.

Keywords – electron beam welding, Al alloy AA6061-T651, Ti alloy Ti6Al4V, microstructure, EDS analysis.

Introduction

The joining of titanium alloys with aluminum could have a major application in the aerospace and automotive industry where high strength to weight ratio is the basic requirement. The increasing demands of manufacturing industries have resulted in a commercial interest in techniques capable of welding metal combinations that were previously thought to be unweldable. Although the joining of dissimilar alloys within the same family is readily accomplished using fusion welding processes (e.g. carbon steel to stainless steel), the effective joining of materials from different alloys is a great represent challenge to the materials engineering [1]. However, successful welding of titanium and aluminum alloys is challenge due to the differences in physical, chemical and metallurgical properties between the two alloys. The key issue is the formation of brittle intermetallic compounds (IMC) in Ti/Al joints [2].

The melting points of Ti and Al are 1667 °C and 660 °C, respectively. This wide difference in the melting points between the two metals leads to the difficulty joining by fusion welding processes. Moreover, the phase diagram of Ti-Al [3] shows a terminal solid solubility towards the Al side, whereas a solid solution up to about 12 at. % Al can be obtained towards Ti-rich alloy. Therefore, depending on the composition, the IMC, namely Ti₃Al, TiAl and TiAl₃, can be formed in the fusion zone during welding. As Al melt first by fusion welding, Ti dissolves in it and forms mostly TiAl₃. It has low strength compared with TiAl and Ti₃Al and almost no ductility up to about 620 °C. The formation of IMC thus leads to embrittlement and decrease in the mechanical properties. Therefore, the formation of IMC is undesirable during welding [4, 5].

Undesirable formation of IMC can be resolved by other joining techniques (e.g. riveting, clinching and screwing). However, these techniques require additional machining and use of extra material which could increase the resulting weight. The successful welding of lightweight Ti/Al structures requires the precision control of welding process during the formation of IMC phases occurred [5]. In recent years, lot of experiments about Ti/Al welding were already done. Jiangwei et al. [6] carried out diffusion welding of Ti/Al alloys, Dressler et al. [7] used friction stir welding and Majumdar et al. [4] performed crack-free Ti/Al welds by CO₂ laser. Direct keyhole laser welding of Ti/Al alloys was published by Tomashchuk et al. [8]. Further possibility is using of modified welding methods. Bang et al. [9] presented gas tungsten arc welding supported by hybrid friction stir welding of Ti6Al4V/AA6061-T6 alloys. Chen et al. [10] improved the reaction at Ti/Al interface by laser welding-brazing process. These methods relate to the effort to reduce dilution of base metals (BM) to avoid formation of IMC phases.

Electron beam (EB) welding appears to be suitable alternative for welding of Ti/Al joints. EB welding as well as laser beam welding brings several advantages. Among the benefits of EB welding belong high density (up to 10⁷ W/cm²), presence of vacuum, high welding speeds and precise control of welding process. The high energy density is required for
welding in keyhole mod. Keyhole mode allows formation of deep and narrow welds with limited deformation. High welding speeds leads to small interaction zone which promote high thermal gradients. Thermal gradients could affect the amount and type of emerging phases and could reduce mixing and diffusion phenomena of BM. Furthermore, the differences in thermophysical properties of welded materials could be diminished by offsetting of EB into one of BM. The resulted strength of joint depends on the thickness of IMC layer at the Ti/Al interface [8, 11].

Nevertheless, publications about EB welding of titanium to aluminum are limited. The weldability of Ti alloy Ti6Al4V and Al alloy AA6061 was studied in resent paper. The main objective was minimized dilution of BM by shifting EB into one of BM and by using of filler metal. The interface between Al and Ti alloys were analyzed in more details.

**Experimental materials and procedures**

For the experiments, Al alloy AA6061-T651 (6061) and Ti alloy Ti6Al4V (Ti64) were selected as BM. BM were delivered in form of sheets with thickness 8 mm (6061) and 8.5 mm (Ti64). Surfaces of samples were in same height during experiments. Chemical composition of BM is shown in Table 1. Structure of heat treated 6061 alloy (Fig. 1) was formed by large grains of α-Al solid solution, complex Q phases (Al-Mg-Si-Cu-Fe-Mn) and hardening β phases (Mg2Si). Ti64 alloy was delivered in mill annealed condition with structure formed by equiaxed grains of α phase with small amount of β phase (Fig. 2).

| Table 1
<table>
<thead>
<tr>
<th>Chemical composition of BM (in wt. %)</th>
</tr>
</thead>
<tbody>
<tr>
<td>6061</td>
</tr>
<tr>
<td>Ti</td>
</tr>
<tr>
<td>Bal.</td>
</tr>
</tbody>
</table>

Several approaches were tested in order to limit the formation of IMC. At firstly, the Ti/Al weld was performed without any modification (sample A). Influence of EB deviation into one of the BM was evaluated on samples B-D. The filler material in the form of thin Cu foil was used on sample E in order to reduce dilution of BM. EB welding was performed out by universal chamber machine K26 (EBG 60 – 150) from Pro-Beam company (Germany). The oscillation frequency of EB (f = 500Hz) was used during welding all samples. EB was focused on the surface of welded materials. EB was focused 3 mm below surface of welded materials in the case of sample E. Samples A–C were welded with lower accelerating voltage (Ua) in order to reduce the evaporation of alloying elements, especially from aluminum alloy. On samples D and E was used higher Ua which reduced the divergence of EB and size of FZ. For the same reasons, smaller diameter of EB spot was used on these samples. The remaining parameters used during the experiments are shown in Table 2.

**Table 2
Parameters of EB welding**

<table>
<thead>
<tr>
<th>Sample</th>
<th>Ua [kV]</th>
<th>Ia [mA]</th>
<th>v [mm/s]</th>
<th>Øspot [mm]</th>
<th>Offset [mm]</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>80</td>
<td>30</td>
<td>15</td>
<td>0.25</td>
<td>0</td>
</tr>
<tr>
<td>B</td>
<td>80</td>
<td>30</td>
<td>15</td>
<td>0.25</td>
<td>0.3 (Ti)</td>
</tr>
<tr>
<td>C</td>
<td>80</td>
<td>30</td>
<td>15</td>
<td>0.25</td>
<td>0.3 (Al)</td>
</tr>
<tr>
<td>D</td>
<td>120</td>
<td>20</td>
<td>15</td>
<td>0.20</td>
<td>0.4 (Al)</td>
</tr>
<tr>
<td>E</td>
<td>120</td>
<td>13</td>
<td>10</td>
<td>0.20</td>
<td>0</td>
</tr>
</tbody>
</table>

Samples for the metallographic analysis were ground, polished and etched in Kroll’s reagent (2 ml HF, 8 ml HNO₃ and 92 ml distilled H₂O). Microstructure analysis were carried out using by Zeiss Axio Observer Z1m light microscope. The scanning electron microscope Zeiss Ultra Plus equipped with energy-dispersive X-ray spectroscopy (EDS) detector Oxford was used for further evaluation.
of microstructure and chemical composition. Microhardness measurements (HV0.1) across the weld joints were carried out on Leco LM 274AT device.

**Results and discussion**

During macroscopic evaluation of sample A, the several individual areas were observed: aluminum base metal (Al-BM); aluminum weld metal (Al-WM); titanium base metal (Ti-BM); titanium heat affected zone (Ti-HAZ) and bulk of intermetallic phases (IMC). The similar results (Fig. 3) were observed on sample B and C, where was used offset of EB.

![Image](image.png)

**Fig. 2. Influence of EB offset on macrostructure of Ti/Al welds: a) 0.3 mm to Al, b) 0 mm and c) 0.3 mm to Ti.**

The diameter of EB spot and EB offset were insufficient to reduce dilution of BM. Although Tomaschuk [8] used high energy laser beam in combination with high welding speed to reduce the interaction time of molten BM on welded samples with thickness 2 mm. In the case of welding thicker sheets, high EB energy and welding speed \(v = 15\) mm/s still allows formation of IMC in the upper part of weld. Due to rapid solidification were presented pores in Al-WM which originated from entrapped vapors of alloying elements [12].

EB offset could controls proportion of individual areas in dissimilar Ti/Al welds, especially IMC and Ti-HAZ. However, nor in one case was not observed the continuous bulk of IMC across the entire weld thickness. It was caused due to different melting temperatures and thermal conductivity of BM. Therefore, upper part solidified after reaction of liquid Al and liquid Ti (L-L reaction). Lower part solidified during reaction of liquid Al and solid Ti (S-L reaction). In this part of weld, Ti64 was heated by thermal diffusivity of 6061 alloy during thermal cycle of EB welding – EB welding-brazing process was occurred [2]. Amount of bulk IMC in Al-WM decreased with EB offset to 6061 alloy (0.3 mm to Ti: 35.11% → 0 mm: 26.05% → 0.3 mm to Al: 15.35%). The greatest amount of IMC was reduced when EB was shifted by 0.3 mm into Al-BM (Fig. 3a). However, the cracks were emerged inside IMC when this offset was used.

Based on these results, the 0.4 offset into Al-BM was used on sample D (Fig. 4a). With this setting, EB brazing process was occurred in the entire thickness of the weld. Only Al-BM was melted and the formation of IMC was suppressed. Only isolate pores were observed inside Al-WM which were the same origin as the pores on samples A – C.

![Image](image.png)

**Fig. 3. Macrostructure of Ti/Al welds: a) 0.4 mm offset to Al and b) sample E.**

The use of filler metal in the form of thin foil did not lead to satisfactory results (Fig. 4b). Cu foil did not prevent the formation of IMC. Moreover, dilution of BM with filler metal caused sample breakdown during metallographic preparation. That indicates significant embrittlement of Ti/Al weld joints in comparison with welds without filler metal. Using of Cu foil support entrapping of pores in the fusion zone – up to 18% of the total weld area.
Fig. 5 presents results of EDS analysis across Ti/Al welds: a) sample A and b) sample D.

The thin layer of IMC growing from Ti-rich areas was observed at Ti-HAZ/Al-WM and IMC/Al-WM interfaces (Fig. 7a). Analysis of chemical composition revealed that layer slowly transforms from TiAl to TiAl3. TiAl3 phase was also observed in form of clusters and needles in Al-WM (Fig. 7b). Needles of TiAl3 grew from Ti/Al interface into Al-WM due to diffusion process from solid state. Tomaschuk et al. [8] demonstrated that phase TiAl3 is last solidified phase in Ti-Al system. TiAl3 was also observed during laser welding-brazing carried out by Chen et al. [10] which confirmed that TiAl3 was growing during S-L reaction. Higher amount of IMC resulted in poorer mechanical properties of weld joints [5, 10].

Evaluation of Ti-HAZ revealed microstructure identical with autogenous Ti EB welds. Microstructure was changed from $\alpha + \beta$ into mixture of martensitic $\alpha'$ phase, acicular $\alpha_{ac}$ phase and rest of undisolved phase $\beta$. Microstructure was fully transformed into martensitic structure at interface where L-L reaction occurred (Fig. 6a). Influence of EB offset was more pronounced on the microstructure and chemical composition of IMC phases. The chemical composition of observed IMC gained from EDS spot analysis is shown in Table 3. Type of IMC phases was determined by combination of results from EDS spot analysis and analysis of Ti-Al phase diagram [3]. Bulk of IMC was formed by Ti3Al (Fig. 6b) when EB welding was performed without offset or with offset into Ti-BM. Structure and chemical composition of IMC bulk were changed when used EB offset into Al-BM (Fig. 6c). That bulk of IMC was predominantly formed by TiAl phase. Large cracks were observed in bulk of TiAl which probably raised due to rapid cooling.

**Table 3**

Results of EDS spot analysis of IMC (in at. %)

<table>
<thead>
<tr>
<th>No.</th>
<th>Ti</th>
<th>Al</th>
<th>V</th>
<th>Fe</th>
<th>Si</th>
<th>Ti3Al</th>
<th>TiAl</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>70.41</td>
<td>26.20</td>
<td>3.02</td>
<td>0.23</td>
<td>0.14</td>
<td>Ti3Al</td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>33.49</td>
<td>64.27</td>
<td>1.56</td>
<td>0.18</td>
<td>0.49</td>
<td>TiAl</td>
<td></td>
</tr>
<tr>
<td>3</td>
<td>37.93</td>
<td>60.23</td>
<td>1.61</td>
<td>-</td>
<td>0.23</td>
<td>TiAl</td>
<td></td>
</tr>
<tr>
<td>4</td>
<td>24.72</td>
<td>73.65</td>
<td>1.08</td>
<td>-</td>
<td>0.55</td>
<td>TiAl3</td>
<td></td>
</tr>
<tr>
<td>5</td>
<td>28.11</td>
<td>69.84</td>
<td>1.29</td>
<td>0.20</td>
<td>0.43</td>
<td>Ti5Al11</td>
<td></td>
</tr>
</tbody>
</table>

Fig. 4 – microstructure of Ti/Al welds: a) Ti-HAZ (sample A); b) bulk of IMC (sample B) and c) bulk of IMC (sample C)

Fig. 5. - microstructure of Ti/Al welds – sample B: a) Ti-HAZ/Al-WM interface and b) needles and clusters of TiAl3 in Al-WM.
When offset 0.4 mm to Al-BM was used the formation of IMC was suppressed. Only a very thin IMC layer (t < 0.5 µm) was formed at the Ti/Al interface due to diffusional process (Fig 8a). Ti atoms dissolve into molten Al when EB passing and concurrently Ti react with Al atoms to give TiAl compounds [6]. Sample E contained small solidification cracks close to the Ti/Al interface where the thermal gradient was highest (Fig. 8b). Solidification cracks are one of typically defects in WM of aluminum alloys.

**Table 4**

Results of EDS spot analysis of Al-WM and Al-BM (in at. %)

<table>
<thead>
<tr>
<th>No.</th>
<th>Al</th>
<th>Mg</th>
<th>Si</th>
<th>Fe</th>
<th>Cu</th>
<th>Mn</th>
<th>phase</th>
</tr>
</thead>
<tbody>
<tr>
<td>6</td>
<td>99.29</td>
<td>0.62</td>
<td>0.09</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>α-Al</td>
</tr>
<tr>
<td>7</td>
<td>90.19</td>
<td>1.11</td>
<td>2.83</td>
<td>4.87</td>
<td>0.39</td>
<td>0.43</td>
<td>Q</td>
</tr>
<tr>
<td>8</td>
<td>73.16</td>
<td>-</td>
<td>8.38</td>
<td>15.28</td>
<td>0.78</td>
<td>1.59</td>
<td>Q</td>
</tr>
<tr>
<td>9</td>
<td>98.40</td>
<td>0.89</td>
<td>0.58</td>
<td>-</td>
<td>0.13</td>
<td>-</td>
<td>α-Al</td>
</tr>
</tbody>
</table>

Microstructure of Al-WM was composed of grains of α-Al solid solute, complex phase Q which precipitated on α-Al grains boundary, isles of Ti₃Al and dispersed rods of Ti₅Al₁¹ (Fig. 9a). Thin IMC layer which were created during S-L reaction was also observed on these isles. Isles of IMC were formed due to oscillation of EB and high thermal conductivity of Al-WM which prevents dissolving of IMC into Al-WM. Al-WM had very fine α-Al grains in comparison with Al-BM (Fig. 9b). This was caused by cooling rate of Al-WM and addition of small amount of Ti dissolved in Al-WM. Ti in small concentrations acts as inoculant in Al alloys [12]. Chemical composition of α-Al solid solute in Al-BM and in Al-WM were identical. Differences in chemical composition of Q-phase in Al-BM and Al-WM was revealed (Table 4).

Dilution of both BM with Cu filler metal resulted in complex microstructure of WM (Fig. 10). WM was composed from α-Al grains; dendritic Cu-rich areas and mixture of Ti₃Al₆ and Ti₅Cu₆ IMC. Although Cu is an important alloying element in Al alloys, the brittle IMC easily form between Ti and Cu owing to thermal effects because of the poor mutual solubility and the large physical and chemical differences between Ti and Cu [13]. Needle-like shape and amount of IMC supported the WM embrittlement which led to sample breaking during metallographic preparation.

**Fig. 11.** Results of microhardness across weld joints cross sections.

Development of HV0.1 microhardness across Ti/Al weld is shown in Fig. 11.
Increase of microhardness at Ti/Al interface depended on type of presented IMC. In the case of formation bulk of Ti3Al phase (sample A and B) was observed sharp change in hardness profile, which negative affected mechanical properties of weld joints. Maximal microhardness that arises between 550 and 730 HV0.1. Maximal microhardness reached 450 HV0.1 in case of formation bulk of TiAl phase in Al-WM on sample C. Similar course of microhardness was also measured on sample E due to the presence of Ti₅Al₆ and Ti₅Cu₆ compounds. Microhardness in Ti-HAZ and Ti-BM followed microstructure changes typically for Ti/Ti welds. In Al-WM was not observed drops in hardness in comparison with Al/Al welds. Conversely, microhardness of Al-WM is comparable to Al-BM.

Only EB welding-brazing process prevented formation of IMC in Al-WM. Average microhardness on sample E in Al-WM (76 ± 8 HV0.1) was lower than in Al-BM (88 ± 2 HV0.1) which corresponds with microhardness course in Al/Al welds. The resulted microhardness was determined by cellular microstructure of WM without presence of IMC.

Conclusions
Optimization of EB keyhole welding process of Ti64 alloy and 6061 alloy have been investigated in presented paper. The main results are summarized as follows:

Direct keyhole EB welding of Ti/Al welds with thickness up to 8 mm led to formation of bulk IMC in the upper part of weld due to reaction of molten Ti-BM and molten Al-BM (L-L reaction). EB welding-brazing process was occurred in the lower part of Ti/Al welds due to higher thermal conductivity of Al alloy which prevents the melting of Ti alloy throughout the joint thickness (S-L reaction).

Setting of EB offset up to 0.3 mm into one of BM did not reduced dilution of BM. Bulk IMC were created in the upper part of welds. However, EB offset controlled the chemical composition, amount, hardness and crack sensitivity of IMC bulk.

Bulk of Ti₃Al, layer of TiAl and Ti₅Al₁₁ phases were formed during L-L reaction. Formation of TiAl₃ phases at interfaces with Ti-rich areas was occurred during S-L reaction. Presence of needles and clusters of TiAl₃ and Ti₅Al₁₁ could affected mechanical properties of Al-WM.

The use of filler metal in form of thin Cu foil did not lead to satisfactory results. Mixture of Ti₅Al₆ and Ti₅Cu₆ compounds caused significant embrittlement of weld joint. The average porosity of WM reached 18% of the total WM area.

Formation of bulk, needles and clusters of IMC in weld joint was suppressed when 0.4 mm offset of EB into 6061 alloy was set. Under these conditions EB welding brazing process was occurred – only Al-BM was melted. Very thin layer (< 0.5 μm) of Ti₅Al₆ was observed at the Ti/Al interface. Measurement of microhardness and chemical composition across weld joint did not confirmed presence of IMC in Al-WM. Small solidification cracks were revealed in Al-WM close to the Ti/Al interface.

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