Influence of Pin Offset on Microstructure and Mechanical Properties of Friction Stir Welded Mg/Ti Dissimilar Alloys

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Abstract

The influences of pin offset on the formation, microstructure and mechanical properties of friction stir welded joint of Ti6Al4V and AZ31B Mg dissimilar alloys were investigated. The results show that sound joints are obtained at different offsets. With the offset decreasing from 2.5 to 2.1 mm, the number of Ti alloy fragments is increased, and the stir zone (SZ) is enlarged and the grains in SZ become coarser. A hook-like structure is formed at the Mg/Ti interface and its length is increased with the decrease in pin offset. The Al element has an enrichment trend at the Ti alloy side near the Mg/Ti interface when the offset is decreased, which is beneficial to the bonding of the interface. An Al-rich layer with a thickness of 3–5 μm forms at the offset of 2.1 mm. All the joints fracture at the interface and present a mixed ductile-and-brittle fracture mode. The joint tensile strength is increased with the offset decreasing from 2.5 to 2.1 mm, and the maximum tensile strength of 175 MPa is acquired at the offset of 2.1 mm.

Keywords Mg/Ti dissimilar alloys · Friction stir welding · Pin offset · Mechanical properties · Microstructure

1 Introduction

Magnesium (Mg) alloys are widely used in the fields of aerospace and automotive because of their lowest density among the commercially available alloys, while their corrosion resistance is poor [1–5]. Titanium (Ti) alloys have not only high specific strength but also an excellent corrosion resistance [6, 7]. The joining of these two alloys can make their respective advantages complement each other. Traditional fusion welding is a widely used method to achieve a welded joint by melting the base metal [8–11]. However, it cannot be applied in the welding of Mg and Ti alloys because of the large difference in the melting point between Mg and Ti alloys [12].

Friction stir welding (FSW) is a novel welding technology which joins materials by severe plastic deformation induced

by the rotating action of tool and the heat input generated by the friction between rotating tool and materials [13–17]. The base materials (BMs) remain the solid state during the FSW process so that the problem of the difference in melting point between dissimilar materials can be avoided [18–22]. Accordingly, FSW is widely used to obtain the joint of dissimilar alloys. Xue et al. [23] joined 1060 aluminum (Al) alloy and pure copper (Cu) by FSW and the hard Cu plate was placed at the advancing side (AS). They obtained sound joints by the sufficient reaction between Al and Cu at a small pin offset. Dressler et al. [24] joined Ti6Al4V and AA2024-T3 Al alloys with the pin center shifted toward to the Al side. They found the fracture surface was located at the joint interface, and the Al–Ti diffusion bonding was formed.

In fact, as a research hot spot, some studies on the FSW of Mg/Ti dissimilar alloys have been reported [1, 12, 25]. Compared with dissimilar alloys, pure Mg and pure Ti are more difficult to be welded since they cannot realize metallurgical reaction [1]. Consequently, their alloys are generally selected to be welded and investigated. Tanabe et al. [25] used the rotating tools with different shapes in the FSW process of CP Ti/AZ31B Mg alloys. They pointed out that compared with the cylindrical pin, the threaded pin could optimize the distribution of Ti alloy fragments and push more Mg alloy to the Mg/Ti joint interface, which was beneficial to

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the bonding of joint interface. Aonuma et al. [6] welded pure Ti and Mg alloys with different Al contents via FSW. They found that an Al-rich layer was formed at the joint interface, which played an important role in the interface bonding.

So far, only a few investigations on the FSW of Mg/Ti dissimilar alloys have been reported and no systematic investigation on the relationship between the process parameters and joint properties has been undertaken. In this study, the FSW experiments of Ti6Al4V/AZ31B dissimilar alloys with different pin offsets were carried out, and the influences of offset on microstructure and mechanical properties of the joint were investigated.

2 Experimental

Ti6Al4V alloy and AZ31B alloy plates with the thickness of 2 mm were used as BMs and their chemical compositions are listed in Table 1. They were shaped into rectangular plates with a dimension of 50 mm × 150 mm, and the surfaces were machined and degreased with acetone. A W-Re alloy tool with a concave shoulder of 12 mm diameter, and a

tapered-thread pin of 1.9 mm length was used. The diameters of pin bottom and pin tip were 6 and 4 mm, respectively. Figure 1a shows the schematic of the FSW process. The hard Ti alloy plates were fixed at the AS according to the following work. Xue et al. [23] pointed out that the weld quality was influenced by the fixed location when the dissimilar materials with different microhardnesses were joined by FSW, and fixing hard material at AS was beneficial to obtaining sound joint because the hard material fixed at retreating side (RS) was difficult to transfer to the AS. The tool tilting angle was kept at 2.5° during the welding process for reducing the advancing resistance. Argon gas was applied for protection because the Ti easily reacts with oxygen at a high temperature. On the basis of the previous study by Aonuma et al. [1], the welding speed and rotating velocity were 100 mm/min and 850 rpm. The pin was mainly placed at the Mg alloy side. According to the paper of Ramachandran et al. [26], the distance between the central axis of pin and the Mg/Ti butt interface was defined as the pin offset (Fig. 1a). The pin offsets of 2.5, 2.3 and 2.1 mm were used in this study.

The specimens were cut into the specified shapes by an electrical discharge cutting machine for metallurgical and

Table 1 Chemical compositions of AZ31B and Ti6Al4V alloys (wt%)

Materials	Chemistry compositions					
AZ31B	Al	Zn	Mn	Fe	Si	Mg
	3.00	0.88	0.88	0.018	0.015	Bal.
Ti6Al4V	Al	V	Fe	C	N	Ti
	6.15	4.00	0.50	0.10	0.05	Bal.

Fig. 1 **a** Schematic of the FSW process; **b** size of tensile specimen

mechanical properties examinations. The cross sections were etched by a solution (5 ml acetic acid, 4.2 g picric acid, 10 ml distilled water and 100 ml ethanol) for 12 s after manual and mechanical polish in accordance with the standard of metallography. The microstructures of the cross sections were observed by an optical microscope (OM), and the microhardness distributions were measured by a Vickers microhardness tester. During the microhardness test, the load, dwell time and distance between tested sites were 300 g, 10 s and 0.5 mm, respectively. The tensile tests were carried out by a universal testing machine at the speed of 3 mm/min under room temperature. Three tensile specimens were prepared for each pin offset, and Fig. 1b shows

the size of specimen. The joint interface was observed by a scanning electron microscope (SEM) equipped with an energy-dispersive X-ray spectroscopy (EDS). The fracture morphology was analyzed by the SEM.

3 Results and Discussion

3.1 Cross Sections

Figure 2 shows the cross-sectional macrostructures with different offsets. The cross sections present a typical basin-shape and can be divided into SZ, heat-affected zone

Fig. 2 Cross sections of the joints at different offsets: **a** 2.5 mm, **b** 2.3 mm, **c** 2.1 mm; **d–f** enlarged views marked in Fig. 2a–c

(HAZ), thermomechanically affected zone (TMAZ) and BM based on the condition of thermal and mechanical action. The Ti alloy fragments are mainly scattered at the AS of SZ. Due to the tapered-thread shape of the pin, these fragments present a downward flowing trend when the rotating tool rotates anticlockwise, which is beneficial to avoid the defect of kissing bonding formed near the joint bottom. A hook-like structure exists at the bottom of the joint interface. This structure generated by the extrusion of the pin stirring action can provide a mechanical interlocking and improve the joint tensile strength [27]. The TMAZ observed at the Mg alloy side is the transitional region between SZ and HAZ. However, it is not obvious at the Ti alloy side. In fact, the TMAZ is also difficult to be observed even in the FSW joint of similar Ti alloys which experiences a higher temperature and more severe deformation during the welding process [27]. The SZ has a fine microstructure due to the grain refinement, and the HAZ exhibits a coarse microstructure due to the grain growth. The remarkable difference in the microstructure between the SZ and HAZ easily leads to the occurrence of stress concentration at the transition region of them under the tensile load. Therefore, the TMAZ is often the weak region for the joint [28].

In order to analyze the influence of offset on the SZ area, the width of SZ is measured and the measured position is 1 mm away from the joint bottom as marked by the blue arrows in Fig. 2. When the offsets of 2.5, 2.3 and 2.1 mm are used, the SZ widths are 4.59, 4.76 and 5.26 mm, respectively. During the FSW process of dissimilar alloys involving Ti alloy and light alloys such as Al and Mg, more contacting region between the tool pin and Ti alloy always leads to a higher frictional heat generation. Li et al. [29] pointed out that the bigger contacting region between the tool and Ti alloy, the higher peak temperature was obtained during the FSW process of Al/Ti dissimilar alloys. Aonuma et al. [1] found that increasing the contacting region between the tool and Ti alloy could lead to an increase in heat generation during the FSW process of Mg/Ti dissimilar alloys. Therefore, the smaller pin offset results in the larger heat generation and then the higher welding temperature is acquired, which improves the plastic flow of material and results in a wider SZ.

Similarly, with the offset decreasing, more Ti alloy fragments are scattered in the SZ as shown in Fig. 2d–f. However, too much Ti alloy fragments may hinder the plastic flow of the Mg alloy and induce the formation of hole defect. Moreover, the stress concentration easily occurs around the big Ti alloy fragment and decreases the joint tensile strength under the tensile load due to the large difference in the stress–strain property between the Ti alloy fragment and the Mg alloy matrix [22]. The length of hook-like structure is enlarged at a smaller pin offset, and the hook-like structure

even reaches the middle of the SZ bottom when the offset of 2.1 mm is used.

3.2 Microstructures

The microstructures of the AZ31B alloy BM and the SZ center regions with different offsets are shown in Fig. 3. The grains of SZ present a fine-equiaxed morphology resulting from the dynamic recrystallization induced by the combined action of the high temperature and the severe intense plastic deformation [24].

Decreasing the tool offset can lead to a higher peak temperature in the SZ during the welding. Higher temperature can decrease the flow stress of material, which is beneficial to heightening the flow velocity and then attaining finer grains. However, higher temperature can lead to the coarsening of the grains. In Fig. 3, the grain sizes of the BM and the SZ center areas at the offset of 2.5, 2.3 and 2.1 mm are measured to be 41, 6.95, 8.6 and 9.7 μm , respectively. The grains of the SZ are refined greatly compared with that of BM. With the pin offset decreasing, the grains of SZ get coarsened gradually.

Since no metallurgical reaction occurs between Mg and Ti, Al element plays an important role in the metallurgical bonding at the Mg/Ti joint interface and its distribution can indicate the condition of the metallurgical reaction [22]. Figure 4a–c shows the distribution conditions of the Al element near the Mg/Ti joint interface at the offsets of 2.5, 2.3 and 2.1 mm.

In Fig. 4a, due to the different Al contents between the Ti6Al4V and AZ31B alloys, an obvious interface can be observed and the Al element at Ti alloy side exhibits a uniform distribution at the offset of 2.5 mm. With the offset decreasing to 2.3 mm, the enrichment of Al element begins to occur at Ti alloy side as displayed in Fig. 4b. With the offset decreasing to 2.1 mm, an Al-rich layer even forms at the interface in Fig. 4c. The enrichment is attributed to that the Al atoms driven by chemical potential tend to diffuse from the alloys to the interface with low Al and high Ti molar fraction [30]. In fact, because Mg and Ti cannot realize metallurgical reaction, the Al element plays an assisted role in the interface bonding of the joint. Due to the metallurgical reaction, the reasonable intermetallic compounds (IMCs) of Al and Ti can form at the interface and improve the joint bonding. Tan et al. [30] obtained the laser welded-brazed joint Mg/Ti dissimilar alloys using Mg alloys with different Al contents as the filler material. They reported the tensile-shear strength of the joint with the filler material of AZ91 filler which contained more Al element was 85% higher than that of the joint with AZ31B filler because more Ti_3Al was formed at the interface. In Fig. 4a–c, more and more enrichment of Al element occurs at the interface because the higher temperature improves the diffusion rate of Al

Fig. 3 Microstructures of **a** AZ31B alloy BM and SZs at offsets of **b** 2.5 mm, **c** 2.3 mm, **d** 2.1 mm

Fig. 4 Distributions of Al element near joining interfaces at offsets of **a** 2.5 mm, **b** 2.3 mm, **c** 2.1 mm; EDS results of **d** Point 1, **e** Point 2, **f** Point 3

atoms from Mg alloy to Ti alloy. As the thickness of the Al-rich layer is lower than the maximum favorable thickness of 10 μm [3], the formation of the Al-rich layer is beneficial to the improvement of the joint tensile strength. Spot scan was carried out to analyze the microstructural component, and Fig. 4d–f shows the proportions of Al, Mg and Ti elements at the Points 1–3 marked in Fig. 4a–c, respectively. With the offset decreasing, the diffusion rates of Mg and Al atoms at Mg alloy side near the interface are both improved because the decreased offset results in a higher welding temperature which plays an important role in improving the diffusion rate of atoms. When the offset is 2.5 mm (Point 1, Fig. 4d), the atom ratio of Al to Ti is 1:6.7 which is near the component in the Ti alloy BM. When the offset is 2.3 mm (Point 2, Fig. 4e), this atom ratio is increased to 1:3.2, which indicates the beginning of Al enrichment. When the offset decreases to 2.1 mm (Point 3, Fig. 3f), the atom ratio is increased to 1:2.7. The atom ratios of Al to Ti at the Point 2 and Point 3 are near the ratio of 1:3, which indicate that the Ti_3Al may be formed in the Al-rich region. The formation of Ti_3Al has also been found in the welding-brazing joint of Mg/Ti dissimilar alloys by Xu et al. [3] and Tan et al. [30].

3.3 Microhardness

The microhardness distributions of the joints with different offsets are displayed in Fig. 5. The measured line is 1.5 mm away from the joint bottom. The microhardness distribution presents a Z-like curve from Ti alloy side to Mg alloy side. The Ti alloy near the joint interface undergoes a work hardening due to the extruding and forging from the rotating tool [23]. Therefore, the maximum microhardness of the joint is located near the interface at the Ti alloy side and the average value is 354 HV which is higher than that of the Ti alloy BM. The lowest microhardness is located at the HAZ of Mg

Fig. 5 Microhardness profiles of the joint cross sections at offsets of 2.5 mm, 2.3 mm and 2.1 mm

alloy side because the grains in this region get coarsened during the welding.

According to the Hall–Petch relationship [31], the average microhardness in the SZ presents a decreasing trend with decreasing offset because of the coarsened grains. The microhardness in the SZ near the Mg/Ti interface is higher than that near the SZ/TMAZ boundary at the Mg alloy side due to the existence of Ti alloy fragments. The microhardness in the HAZ also decreases with decreasing offset because a higher heat input results in a longer grain growing time. In some FSW joints of dissimilar alloys, the microhardness in the local region of SZ occasionally presents a sudden increase due to the formation of the hard IMCs. Niu et al. [5] investigated the FSW of Al/Mg dissimilar alloys and found that the microhardness in the SZ of Mg alloy was much higher than that of Mg alloy BM because of the scattered Al–Mg IMCs. In the FSW joint of Mg/Ti alloy, no similar sudden increase appears in the SZ because the main elements of Mg and Ti cannot realize the metallurgical reaction.

3.4 Tensile Properties

Figure 6 exhibits the result of tensile test. The tensile strengths of the joints at the offsets of 2.5, 2.3 and 2.1 mm are 126, 151 and 175 MPa, respectively. Generally, a higher microhardness corresponds to a higher tensile strength. However, with the offset decreasing, the joint microhardness decreases and the joint tensile strength presents an increasing trend. It is because the lowest microhardness of the joint in this study does not correspond to the joint weakest position. As displayed in Fig. 7, the fracture occurs at the joint interface which is the weakest position of the FSW joint of Mg/Ti dissimilar alloys. For many FSW joints, the fractures occur near the boundary between the SZ and TMAZ due to their remarkable difference in the

Fig. 6 Tensile strengths of joints at different offsets

Fig. 7 Fracture position of the joint at 2.1 mm offset

microstructure [25]. However, because Mg and Ti cannot realize metallurgical reaction, the metallurgical bonding relying on the Al–Ti metallurgical reaction is relatively poor at joint interface and accordingly becomes the weakest region of the joint. As the offset increases, the interface bonding strength is heightened due to the Al–Ti metallurgical reaction. Moreover, the elongated hook-like structure leads to a longer joint interface with the decrease in pin offset, which is also beneficial to improving the joint tensile strength when the fracture position is located at the Mg/Ti joint interface. The two above-mentioned reasons make that the tensile strength of Mg/Ti dissimilar alloys joint increase with the decrease in pin offset.

3.5 Fracture Morphologies

Figure 7 presents the fracture location of the joint with 2.1 mm offset and the fracture path is marked by the white arrows. The joint completely fractures at the Mg/Ti joint interface. Although the hook-like structure provides the mechanical interlocking, a stress concentration also occurs at this region during the tensile test. Therefore, the crack initiates at the hook-like structure tip and then propagates to the top surface along the joint interface.

Figure 8a–d shows the fracture morphologies of the regions A–D marked in Fig. 7. Both ductile fracture mode and brittle fracture mode occur in the joint. In Fig. 8d, Ti alloy fragments can be observed at the fracture surface. The existence of the Ti alloy fragments can lead to the stress concentration and then facilitate the crack propagation. Figure 8e, f shows the enlarged views marked in Fig. 8a. It can be seen that the fracture surface presents different morphologies in different regions. Plenty of dimples are formed in the region displayed in Fig. 8e, which indicates a ductile fracture mode. However, no dimple is formed and the fracture surface is relatively flat in the region displayed in Fig. 8f, which presents a brittle fracture mode. These morphologies indicate that a mixed ductile-and-brittle fracture mode occurs at the joint interface.

Fig. 8 Fracture surface morphologies of joint at 2.1 mm offset: **a–d** regions A–D marked in Fig. 7; **e, f** enlarged views marked in Fig. 8a

4 Conclusions

The investigations of Ti6Al4V and AZ31B alloys FSW were carried out at different pin offsets. The following conclusions can be extracted based on the present study.

1. With decreasing offset, the width of the SZ is enlarged, the hook-like structure is extended and the plenty and size of Ti alloy fragments in the SZ are increased.
2. Because the decreased offset results in a higher heat input, the grains in SZ get coarsened gradually, and the diffusion rate of the Al atom is improved. An obvious Al-rich layer is formed at the Mg/Ti joint interface when the offset is 2.1 mm.
3. With decreasing offset, the microhardness in SZ at Mg alloy side shows a decreasing trend because of the coarsened grains. The minimum microhardness of the joint occurs in the HAZ at Mg alloy side and the maximum value occurs at the Ti alloy side near the Mg/Ti joint interface.
4. The tensile strength is increased with decreasing offset because the formation of Al-rich layer improves the interface bonding. A mixed ductile-and-brittle fracture occurs at the Mg/Ti joint interface and the maximum tensile strength of 175 MPa is obtained at the offset of 2.1 mm.

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