Original Article

The effect of temperature on microstructure and mechanical properties of Al/Mg lap joints manufactured by magnetic pulse welding

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ARTICLE INFO

Article history:
Received 14 November 2018
Accepted 7 May 2019
Available online 30 May 2019

Keywords:
Magnetic pulse welding
Magnesium alloys
Aluminum alloys
Intended service temperature

ABSTRACT

The microstructure and mechanical properties of Al/Mg welded joints after heat preservation at different temperatures were examined by optical microscopy (OM), scanning electron microscopy (SEM), energy dispersive spectrometry (EDS) and mechanical property test. The results showed that the treatment has no obvious effects on the morphology of the Al/Mg welding interface, but the morphology of the grain structure on the magnesium alloy side was improved. When the temperature was below 150 °C, there was no significant change in the microstructure and mechanical properties of Al/Mg welded joint. While, the intermetallic compound layer composed of Al12Mg17 was formed at the interface when the temperature was raised to 200 °C. The welding interface generated two intermetallic compound layers composed of Al12Mg17 and Al12MgG at the temperature reached to 250 °C. Intermetallic compounds severely affected the strength of the welded joint because of greater brittleness. In order to maintain reliable performance, the use temperature of the Al/Mg welded structural parts should not exceed 150 °C.

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1. Introduction

Aluminum and magnesium alloys are attractive for aerospace, shipbuilding, automotive, high-speed trains and other industrial applications due to their low density and high specific stiffness [1]. Their combination is an ideal lightweight structure, which can significantly reduce the weight of industrial equipment. However, it is difficult to join aluminum and magnesium because of the big difference in material physical properties and chemical properties [2].

Researchers have tried various methods for magnesium-aluminum welding, such as tungsten inert gas (TIG) welding [3], laser welding [4], cold metal transfer welding [5], vacuum diffusion welding [6,7], friction stir welding [8–10] and explosion welding [11]. Wang et al. [4] connected 6061 Al alloy and AZ31 Mg alloy successfully by laser welding, but they found intermetallic compounds (IMCs) such as γ-Mg17Al12 and β-Mg2Al3 at the bottom of the fusion zone which seriously affected the performance of welded joints. Wang et al. [5] achieved the bonding of dissimilar metals AZ31 Mg alloy and
1060 Al alloy with Al Si5 as the filler metal by cold metal transition welding. The multi-layered lamellar structure in the fusion zone near the magnesium side was mainly composed of Al, Mg solid solution and Mg17Al12 and Mg2Al3 IMC layers, which led to a decrease in the strength of the joint. Liu et al. [6] reported that the obvious diffusion zone consists of intermetallic compounds MgAl, Mg2Al3 and Mg2Al3 formed near the Mg/Al interface as a result of the vacuum diffusion bonding. In Jafarian et al’s study [7], they found that the crack, distortion and segregation produced using fusion welding can be avoided by using diffusion welding, but the interfacial transition zone composed of Al12Mg17 and Al13Mg2 was formed at the joints. Mohammadi et al. [10] managed to obtain FSW joints between AZ31B and Al6061 alloy sheets at different tool rotation. They found that the IMCs as Al12Mg2 and Al13Mg2 were identified in weld nugget and at Al/Mg interface. The interface bonding mechanism of explosively welded magnesium AZ31B-aluminum 7075 plate was investigated by Yan et al. [11]. The results showed that the “metallurgical bonding” of the explosive welding wavy interface was achieved by local diffusion and no intermetallic compounds were found. However, explosive welding has low degree of mechanization and poor working conditions, and it also will produce noise and other environmental pollutants. Therefore, many researchers were committed to finding a suitable method to make magnesium and aluminum achieves effective connection.

Magnetic pulse welding is a solid-state bonding process that achieves mechanical or metallurgical connection through the high-speed collisions of two materials [12]. Magnetic pulse welding has the advantages of high production efficiency, no pollution and easy automation. Moreover, the low heat input in the process can not only avoid defects effectively such as cracks and pores, but also suppress the formation of IMC [13,14]. There have been relevant researches on the aluminum-magnesium dissimilar alloys magnetic pulse welding in recent years [15-18]. Ben-Artzy et al. found that a thin melted layer exists at the bonding interface of the Al-Mg magnetic pulse welded joint and intermetallic compounds with different compositions were produced during the subsequent rapid cooling [15]. The structural and mechanical properties of the interface layer in Al/Mg joined by magnetic pulse welding were conducted an in-depth study by Stern and Aizenshtein [16,17]. Chen and Jiang [18] discussed the microstructure evolution during magnetic pulse welding of dissimilar aluminum and magnesium alloys. The dissimilar materials of aluminum and magnesium were bonded successfully by means of magnetic pulse welding, and the microstructure of the bonded joint was analyzed.

The previous study focus on the morphology and bonding mechanisms of magnetic pulse welding interfaces. However, the service situation of the Al/Mg structure bonded by magnetic pulse welding under the conditions of thermal load has not been reported systematically. As known from the literatures [19-21], the intermetallic compound layer was easily formed after the Al/Mg interface was heated, which would affect the serviceability of welded structures. So it is necessary to study the effect of temperature on the microstructure and properties of Al/Mg magnetic pulse welded joints.

In this study, AZ31B magnesium alloy and 1050 aluminum alloy were successfully joined by magnetic pulse welding, and then the sample was heated preservation treatment at different temperatures. The effect of temperature on aluminum-magnesium welded joints was discussed by analyzing the evolution of microstructure and mechanical properties of welded joints after being kept at different temperatures for a period of time.

2. Materials and methods

The materials used in MPW were AA1050 aluminum alloy and AZ31B magnesium alloy with the dimensions of 35 mm × 90 mm × 2.5 mm and 35 mm × 90 mm × 1 mm, respectively. The elemental composition of base metals is shown in Table 1. The original microstructure of AZ31B magnesium alloy is shown in Fig. 1.

The oxide layer at the surface of magnesium was removed by metallographic sandpaper and then washed using alcohol before magnetic pulse welding. The magnetic pulse welding equipment was the German company PST electromagnetic pulse connection device PS48-16 (the maximum voltage is 16 kV, the maximum energy storage is 48 kJ, the frequency range is 5-20 kHz). Magnetic pulse welding system includes a capacitor bank, coil and workpieces. The capacitor bank consists of an inductor-capacitor circuit and some impedance actuators. When the capacitors are discharged, the high-density current flows through the coil, which is regarded as the primary current. If there is a closed current path, the associated electromagnetic field will generate a strong secondary current through the flyer plate. The secondary current in the flyer plate is opposite to the primary current in the coil, so a strong repulsive Lorentz force between them, which causes the flyer plate to accelerate for the welding. The coil used in this study was a rectangular coil of 4 mm × 10 mm. The schematic diagram is shown in Fig. 2.

Collision speed and collision angle are two main parameters in magnetic pulse welding. The collision speed and

<table>
<thead>
<tr>
<th>Material</th>
<th>Al</th>
<th>Mg</th>
<th>Si</th>
<th>Zn</th>
<th>Mn</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>AA1050 Al</td>
<td>Bal.</td>
<td>&lt;0.05</td>
<td>&lt;0.25</td>
<td>&lt;0.05</td>
<td>&lt;0.05</td>
<td>0–0.4</td>
</tr>
<tr>
<td>AZ31B Mg</td>
<td>2.5–3.5 Bal.</td>
<td>0.08</td>
<td>0.6–1.4</td>
<td>0.2–1.0</td>
<td>0.003</td>
<td></td>
</tr>
</tbody>
</table>

![Fig. 1 – The original microstructure of AZ31B magnesium alloy sheet.](image-url)
collision angle are controlled by changing the discharge energy of the magnetic pulse equipment and the gap distance of the workpiece. In this study, the discharge energy was 30 kJ, 32.5 kJ, 35 kJ, 37.5 kJ, 40 kJ, the initial gap between flyer plate and base plate was chosen to be about 1.4 mm.

In order to study the service situation of the Al/Mg structure under the conditions of thermal load, the heat preservation treatment at different temperatures was performed on the magnetic pulsed welded structures. The temperature and holding time were selected at 50 °C, 100 °C, 150 °C, 200 °C, 250 °C, 300 °C, and 350 °C for 1 h. The samples were heated with the furnace heating method and cooled in the air atmosphere.

The specimens for microstructure observation were cut vertical to the weld direction. The cross section of the samples was grinded with metallographic sandpapers and then polished using diamond paste. The microstructure characteristics of the interface were evaluated by a quanta-200 environmental scanning electron microscope (SEM) equipped with the energy dispersive X-ray (EDS) detector. The fracture characteristics of the joints were observed by using SEM. The hardness measurements were carried out on polished samples along the transverse cross section using HVS-1000 digital micro Vickers hardness tester. Vickers microhardness was measured using a load of 500 gf and a dwell time of 10 s. All the values were an average of three indentations taken on the same sample. The shear strength was tested using the INSTRON 5985 universal testing machine. The shear speed was 2 mm/min and the average of the three test results was taken as the final result.

### 3. Results and discussion

#### 3.1. Effect of discharge energy on mechanical properties of Al/Mg joints

The shear test was carried out to evaluate the mechanical properties of Al/Mg joints under different discharge energies. Fig. 3 shows the load–displacement curves and the fracture energy of Al/Mg joints under different discharge energies. It can be seen from Fig. 3(a) that the maximum shear load first increased and then fell with the rise of discharge energy. The maximum shear load of Al/Mg joint reached the maximum value at discharge energy of 35 kJ. The fracture energy was used to quantify the energy absorption capacity of the Al/Mg joints under different discharge energies. The fracture energy $E$ was obtained from the integral of load–displacement curve. As shown in Fig. 3(b), the fracture energy of Al/Mg joints increased gradually with the discharge energy increased, but the increase of discharge energy had no significant effect on the fracture energy of Al/Mg joints after discharge energy reached 35 kJ.

Considering the maximum shear load and fracture energy, it is known that Al/Mg welded joint had the best mechanical properties at discharge energy of 35 kJ. Therefore, in this study, the Al/Mg welded joints used in exploring the service situation of the Al/Mg structure under thermal load conditions were obtained at discharge energy of 35 kJ.

#### 3.2. The effect of temperature on microstructure and mechanical properties of Al/Mg lap joints

##### 3.2.1. Macroscopic morphologies

Fig. 4 shows the macroscopic appearance of Al/Mg welded joint after heat preservation at different temperatures. There was no obviously change in the macro-morphology of the welded joints after heat preservation below 300 °C. The bulging pack was formed in the weld area of the Al/Mg joints after heat preservation at 300 °C and 350 °C for 1 h. It was found that the height of bulging pack in the weld zone became higher
as the temperature increased. This paper speculated that the bulging deformation occurred because the weld was very tight and the air was present inside the weld. When the heat preservation treatment was carried out, the volume expansion of air inside the weld resulted in the deformation of magnesium plate.

3.2.2. The microstructure of Al/Mg interface
The bonding interface of magnetic pulse welding usually shows three basic forms: the waveform junction area, the straight junction area and the fusion zone [22]. Fig. 5 shows the microstructure of Al/Mg magnetic pulse welding interface after heat preservation treatment at different temperatures. The photomicrograph (Fig. 5) demonstrates that the asymmetric wavy interconnect was obtained at the interface of magnetic pulse welding. Akbari Mousavi and Al-Hassani [23] stated that the shape of the wave was perfectly symmetric for metals with similar density and became more and more asymmetric as the difference in density increases. The morphology of Al/Mg interface presented the asymmetric waves on account of visible different density between Al and Mg alloy. No significant changes in the morphology of the Al/Mg interface were observed after heat preservation treatment at different temperatures. It can be seen from Fig. 5(a) that the grain structure on the magnesium alloy side had disordered grain morphology after the magnetic pulse welding. The heat preservation treatment improved the stress concentration at the bonding interface and the microstructure distribution on the magnesium alloy side. The microstructure of magnesium alloy side was recrystallized and the crystal grains were hexagonal structures after heating at different temperatures. There was a well-distributed diffusion layer can be seen on
the bonding surface after heating at 200 °C as seen in Fig. 5(e). It can be seen from Fig. 5(f) that the thickness of the interfacial diffusion layer became thicker as the temperature increased.

In order to further observe the evolution of the welding interface, the welding interface was studied by SEM and EDS. Fig. 6 shows the cross-sectional SEM micrographs of Mg/Al interface in the conditions of as-weld and heat treated at 50 °C, 100 °C, 150 °C, 200 °C, and 250 °C for 1 h, respectively. These images indicated that bonding at the Mg/Al interface had wavy morphology. The morphology of the interface obtained for 1 h of heat treated at 50 °C, 100 °C, 150 °C, 200 °C, and 250 °C were similar to these in the state after welding. It can be seen that there were no obvious changed and no clear diffusion layer formed in the bonding interface after the magnetic pulse welding after heat preservation treatment at 150 °C. There was a well-distributed diffusion layer can be seen on the bonding surface after heating at 200 °C as seen in Fig. 6(e). The Al/Mg welding interface would form diffusion layers of different structures at different temperatures and holding times [24]. It can be seen from Fig. 6(f) that the thickness of the interfacial diffusion layer became thicker as the temperature increased.

Fig. 7 shows the results of line scan composition analysis of Al-Mg elements for line a, b, c, d, e and f in Fig. 6(a–f) at the 1050/AZ31B composite interface after heat treatment at different temperatures. Fig. 7(a) is a line scan of the interface elements without heat preservation treatment. It can be seen that some elemental diffusion occurred in the bonding interface. The elements had no obvious over-gradients, but a certain slope was formed at the interface. It is indicated that the interface elements had diffusion under the action of high temperature and high pressure caused by plastic deformation and high-speed collision during the magnetic pulse welding process, and the diffusion of these elements led to the metallurgical bonding of the interface. Fig. 7(b–d) shows line scan of Al/Mg interface in the conditions of heat preservation treatment at 50 °C, 100 °C and 150 °C for 1 h, respectively. As it can be seen that the solid solutions were formed because of elements undergo further diffusion, but the concentration of the elements did not exceed their respective solid solubility. Fig. 7(e, f) is an elemental line scan of the composite interface after heat preservation treatment at 200 °C and 250 °C for 1 h, respectively. It was shown that the gradient distribution of Mg and Al concentration in the diffusing layer region had no trend of decline moderately but had a gentle intermediate downward trend. So it can be concluded that the material composition of the diffusion layer was different from the phase of AZ31B and 1050. It was considered that the diffusion of Mg and Al increased and exceeded the respective solid solubility as the heat preservation temperature increased. In order to determine the specific composition of the diffusion layer, EDS elements were scanned for A, B, C, and D in Fig. 6(e, f). The results of the scanning were shown in Table 2.

According to the composition ratio of atoms, the interfacial diffusion layer was intermetallic compound composed by Al$_{12}$Mg$_{17}$ heated at 200 °C for 1 h. The interfacial diffusion layers heated at 250 °C for 1 h were composed of Al$_3$Mg$_2$ and Al$_{13}$Mg$_{17}$ intermetallic compound. From the phase component analysis, it was found that the intermetallic compound layer near the Mg alloy side was Al$_{12}$Mg$_{17}$ and the intermetallic compound layer near the Al alloy side was Al$_3$Mg$_2$. The Al$_3$Mg$_2$ layer near the 1050-Al side was thicker than the Al$_{12}$Mg$_{17}$ layer near the AZ31-Mg side.

There are two important parameters in the diffusion process: diffusion coefficient $D$ and diffusion activation energy.

![Fig. 6](image-url) The SEM images of the interface Al/Mg joint after heat preservation at different temperatures (a) as-weld; (b) 50 °C, 1 h; (c) 100 °C, 1 h; (d) 150 °C, 1 h; (e) 200 °C, 1 h; (f) 250 °C, 1 h.

| Table 2 – The point of EDS elements at the interface Al/Mg corresponds to Fig. 6. |
|-----------------------------|-----------------------------|
| Al (at.%)  | Mg (at.%) |
| Point A  | 43.28 | 56.72 |
| Point B  | 42.36 | 57.64 |
| Point C  | 45.69 | 54.31 |
| Point D  | 62.36 | 37.64 |
Table 3 – Diffusion factor D₀ and diffusion activation energy Q corresponding to Al and Mg elements [25].

<table>
<thead>
<tr>
<th>Elements</th>
<th>Diffusion coefficient (D₀), m²/s</th>
<th>Atomic activation energy (Q), kJ mol⁻¹</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al</td>
<td>(1.7 \times 10^{-4})</td>
<td>(1.42 \times 10^5)</td>
</tr>
<tr>
<td>Mg</td>
<td>(1.5 \times 10^{-4})</td>
<td>(1.35 \times 10^5)</td>
</tr>
</tbody>
</table>

Q. The Arrhenius index relationship exists between diffusion coefficient, diffusion activation energy and temperature [25]:

\[ D = D₀ \exp(-Q/RT) \]  

where \(D\) is the diffusion coefficient, m²/s; \(D₀\) is the diffusion coefficient, m²/s; \(Q\) is the atomic activation energy, kJ mol⁻¹; \(R\) is the Boltzmann constant, 8.314 J mol⁻¹ K⁻¹; \(T\) is the heating temperature, K. The values of diffusion factor \(D₀\) and diffusion activation energy \(Q\) for Al and Mg elements are listed in Table 3. According to Eq. (1), it can be seen that the self-diffusion coefficient of Mg element is larger than the self-diffusion coefficient of Al element, indicating that the diffusion rate of Mg atom is greater than that of Al atom at the same temperature. At the same temperature, the rate of diffusion of Mg atoms into Al matrix is much larger than that of Al atoms to Mg matrix. Therefore, a large amount of Mg atoms is rapidly enriched at the interface of Al. The aluminum atoms reacted with a large number of magnesium eutectic to form the intermetallic compound \(\text{Al}_{12}\text{Mg}_{17}\) nuclei in the interface. As the Mg atoms continue to diffuse into the Al matrix, the \(\text{Al}_{12}\text{Mg}_{17}\) nuclei grow laterally along the interface and gradually become even city-wide to form a continuous \(\text{Al}_{12}\text{Mg}_{17}\) layer.

The schematic diagram of atomic diffusion at the Al/Mg interface was shown in Fig. 8. According to the literature [26], the free energy change per unit initial Al/Mg interface area, \(\Delta G_{d}\), can be approximated as:

\[ \Delta G_{d}(d) = G_{V} - \left(12/29G_{Al} - 17/29G_{Mg}\right)d + y_{\gamma/Al} + y_{\gamma/Mg} - y_{\gamma/Al} \]
\[ = \Delta G_{\text{vol}}(d) + y_{\gamma/Al} + y_{\gamma/Mg} - y_{\gamma/Al} \]

where \(d\) is the thickness of the layer of intermetallic, \(G_{V}, G_{Al}, G_{Mg}\) are the free energy of the \(\text{Al}_{12}\text{Mg}_{17}\) phase, the Al rich phase and the Mg rich phase, respectively. \(y_{\gamma/Al}, y_{\gamma/Al}\) and \(y_{\gamma/Mg}\) are the interfacial free energy of the \(\gamma/Al, \gamma/Mg\) and Al/Mg interface, respectively. The case of two intermetallic phases being present the expression for free energy change is:

\[ \Delta G(\gamma, \beta) = \left(\Delta G_{\text{vol}}(\gamma, \beta)\right) + y_{\gamma/Al} + y_{\gamma/Mg} + y_{\gamma/\beta} - y_{\gamma/Al} \]

where \(y_{\gamma/Al}\) and \(y_{\gamma/\beta}\) are the interfacial free energy of the \(Al/\beta\) and the \(\gamma/\beta\) interface, respectively, \(\Delta G_{\text{vol}}(\gamma, \beta)\) is the volume average free energy change in the intermetallic layers excluding the interface contributions, caused by the formation of the two intermetallic phases. Therefore, the Gibbs free energy required for the formation of \(\text{Al}_{12}\text{Mg}_{17}\) phase is lower than that required for the formation of \(\text{Al}_{3}\text{Mg}_{2}\) phase [26,27]. The \(\text{Al}_{12}\text{Mg}_{17}\) phase is more preferentially formed than \(\text{Al}_{3}\text{Mg}_{2}\) phase under the same conditions, so the interfacial diffusion layer heated at 200 °C for 1 h is only composed of \(\text{Al}_{12}\text{Mg}_{17}\) intermetallic compound. However, the growth rate of the \(\text{Al}_{3}\text{Mg}_{2}\) phase is much larger than that of the \(\text{Al}_{12}\text{Mg}_{17}\) phase when the \(\text{Al}_{3}\text{Mg}_{2}\) phase is formed the same as \(\text{Al}_{12}\text{Mg}_{17}\). Therefore, the \(\text{Al}_{3}\text{Mg}_{2}\) layer near 1050 side is thicker than the \(\text{Al}_{3}\text{Mg}_{2}\) layer near \(AZ31B\) side in the interface diffusion layer heated at 250 °C for 1 h.
3.2.3. The hardness at Al/Mg interface

Fig. 9 shows the microhardness profile across the AZ31B/1050 bonding interface after heat preservation treatment at different temperatures. The results showed that the microhardness of AZ31B Mg alloy changed obviously, while the microhardness of 1050 Al alloy did not change much after heat preservation treatment. On the AZ31B side, the microhardness in the vicinity of the bonding interface tended to increase and approached the peak due to the high dislocation density and strain hardening effect. As the heat preservation temperature increased, the microhardness of the magnesium alloy gradually decreased.

The microhardness at the interface had no obvious change at heating temperature below 150 °C. It was meant that no intermetallic compounds formed on the interface at the heat preservation temperature below 150 °C. The microhardness at the interface significantly increased at the heat preservation temperature of 200 °C. It can be speculated that there was an intermetallic compound formed at 200 °C. The intermetallic compound layer became thicker and thicker as the temperature increased, which was consistent with the previous micromorphology results.

3.2.4. The mechanical property of Al/Mg joints

The Al/Mg joint heated at 300 °C and 350 °C for 1 h cracked during the clamping process in the shear test. The comparison of the Al/Mg joints mechanical behavior after heat preservation treatment at different temperatures in the shear test was shown in Fig. 10. The mechanical properties of the Al/Mg joints did not change significantly at temperatures below 150 °C. However, the abrupt rupture of the Al/Mg interface led to a catastrophic failure at heat preservation temperature of 200 °C. The Al/Mg joint heated at 200 °C only can withstand a maximum loading force of 2432 N, which was only 40.7% of the original state joints. The Al/Mg joints had almost no strength heated at 250 °C.

The effects of temperature on the mechanical properties were shown in Fig. 11. As can be seen from Fig. 11, the Al/Mg joints in as-welded condition can withstand a maximum load force of 5974 N. They underwent plastic deformation and ductile failure before they ruptured. The Al/Mg joints heated at 50 °C, 100 °C, and 150 °C for 1 h can withstand a maximum loading force of 5912 N, 5880 N and 5593 N, respectively. It should be noted that the maximum shear force hardly changed when the holding temperature changed from room temperature to 150 °C (Fig. 11(a)). The maximum force dropped sharply as the temperature continued to rise, which was caused by the formation of intermetallic compounds at the interface. The fracture energy was used to quantify the energy absorption capacity of the Al/Mg joints after heat preservation at different temperatures. The fracture energy gradually decreased with the rise of the heat preservation temperature, as shown in Fig. 11(b).

3.2.5. The fracture of Al/Mg interface

All MPW lap joints in as-welded and heated conditions failed to break along the weld interface in the shear test. Therefore, the fracture surfaces on both Al-side and Mg-side were observed. SEM was used to study the fracture morphology of
the weld region. In order to determine the characteristic composition of the Al-side and Mg-side fracture surface, point 1–14 of the EDS elements were scanned in Fig. 12. The scan results were shown in Table 4. Fig. 12 shows the Al-side and Mg-side fracture profile of the Al/Mg welded joints in as-welded and heated at temperatures below 150 °C conditions. Fig. 12(a) and (b) were the fracture profiles of the Al side and the Mg side after the shear test in the welded state, respectively. It can be seen that the shearing interfaces on both sides were dimples of different sizes. The point sweep of the interface revealed that there was only aluminum element on the aluminum side interface, while magnesium and aluminum were present on the magnesium side interface. Dimples colonies were evidenced on the fracture surface of Al/Mg welded joints without heated (Fig. 12(a, b)), which showed the typical features of ductile fractures dominated by dimples.

The mechanical properties of Al/Mg joints heated at temperatures below 150 °C were comparable to those of joints as-welded. But the fracture morphology of Al/Mg welded joints had change after heated at 50 °C, 100 °C and 150 °C for 1 h. The fracture morphology of Al/Mg welded joints did not change significantly after heat preservation treatment at 50 °C for 1 h (Fig. 12(c, d)), only the number of dimples above the fracture surface decreased and the size of the dimples became smaller. The number of dimples decreased sharply and some of the tear ridges appeared on the fracture morphology of Al/Mg welded joints after heat preservation treatment at 100 °C for 1 h (as shown in Fig. 12(e, f)). The tear ridges and a small

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**Table 4** The point of EDS elements on the fracture surface of Al/Mg Joints corresponds to Fig. 12.

<table>
<thead>
<tr>
<th>Point</th>
<th>1</th>
<th>2</th>
<th>3</th>
<th>4</th>
<th>5</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al (at.%)</td>
<td>100</td>
<td>96.45</td>
<td>100</td>
<td>91.48</td>
<td>81.76</td>
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<tr>
<td>Mg (at.%)</td>
<td>–</td>
<td>3.55</td>
<td>–</td>
<td>8.52</td>
<td>18.24</td>
</tr>
<tr>
<td>Point</td>
<td>6</td>
<td>7</td>
<td>8</td>
<td>9</td>
<td>10</td>
</tr>
<tr>
<td>Al (at.%)</td>
<td>100</td>
<td>91.48</td>
<td>82.51</td>
<td>90.35</td>
<td>43.07</td>
</tr>
<tr>
<td>Mg (at.%)</td>
<td>–</td>
<td>8.52</td>
<td>17.49</td>
<td>9.65</td>
<td>56.93</td>
</tr>
</tbody>
</table>
The number of dimples also appeared on the fracture surface of Al-side and Mg-side after heating at 150 °C for 1 h (Fig. 12g, h).

Fig. 13 shows the fracture surface of Al-side and Mg-side after shear test of the joint after heating at 200 °C, 250 °C, 300 °C, and 350 °C for 1 h. Table 5 shows the EDS elements scan results of point a–g in Fig. 13. As the heat preservation temperature increased to 200 °C or 250 °C, there were a lot of cleavage steps and tear ridges at the fracture surface, indicating that the fracture of the joint was brittle fracture (Fig. 13a–d). The fracture morphology of Al/Mg welded joints was different from the others after heated at 300 °C and 350 °C for 1 h, and secondary cracks appeared on the fracture surface. According to EDS analysis, there were many oxides of aluminum and magnesium on the fracture surface.

The results of EDS scanning on the fracture surface revealed that the Al/Mg welded joint heated at 200 °C for 1 h broke at the intermetallic compound layer composed of Al12Mg17. According to the previous analysis of the weld interface, there were two layers of intermetallic compound in the Al/Mg welded joint heated at 250 °C for 1 h, which were Al12Mg17 layer and Al3Mg2 layer, respectively. It was found that Al/Mg welded joint heated at 250 °C failed at the Al3Mg2 layer by analyzing the EDS scan on the fracture surface. The thermal conductivity of the Al12Mg17 intermetallic phase is closer to the thermal conductivity of the matrix magnesium alloy [28]. Therefore, the stress concentration between the matrix magnesium alloy and the Al12Mg17 intermetallic compound diffusion layer is small. While the diffusion layer of the Al3Mg2 intermetallic compound and the Al12Mg17 intermetallic compound are respectively rich aluminum layer and rich magnesium layer, so the thermal conductivity difference is large. The phase region of Al3Mg2 intermetallic compound phase zone is more likely to initiate cracks [29]. In shear test, the welded joint failed at the Al3Mg2 intermetallic compound diffusion layer.

In summary, the shear failure modes of the Al/Mg welded joint after welding and heating can be mainly divided into two types and the failure diagram was shown in Fig. 14. The

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**Table 5** The point of EDS elements on the fracture surface of Al/Mg joints corresponds to Fig. 13.

<table>
<thead>
<tr>
<th>Point</th>
<th>a</th>
<th>b</th>
<th>c</th>
<th>d</th>
<th>e</th>
<th>f</th>
<th>g</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al (at.%)</td>
<td>45.35</td>
<td>39.13</td>
<td>57.05</td>
<td>58.83</td>
<td>36.79</td>
<td>22.83</td>
<td>34.67</td>
</tr>
<tr>
<td>Mg (at.%)</td>
<td>54.65</td>
<td>60.87</td>
<td>36.99</td>
<td>41.17</td>
<td>15.25</td>
<td>67.94</td>
<td>18.96</td>
</tr>
<tr>
<td>O (at.%)</td>
<td>—</td>
<td>—</td>
<td>5.96</td>
<td>—</td>
<td>13.86</td>
<td>9.24</td>
<td>26.34</td>
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<tr>
<td>C (at.%)</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>34.10</td>
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<td>20.03</td>
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</tbody>
</table>

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![Fig. 13](image1) The fracture morphology of the Al-side and Mg-side of the Al/Mg welded joints after heating at 200 °C, 250 °C, 300 °C, and 350 °C for 1 h.

![Fig. 14](image2) Schematic of interfacial failure: (a) ductile fracture and mixed fracture near the aluminum side of the weld (as-weld or heating below 150 °C); (b) brittle fracture through the transition layer (heating at 200 °C and above).
fracture mode of the Al/Mg welded joint without heating was a ductile fracture along the Al-side of the weld. The dimples were uniformly distributed on the fracture surface and the EDS elements on the fracture of the Al-side and the Mg-side were all scanned with Al (as shown in Fig. 12(a, b) and Table 4). The tear ridges appeared on the fracture surface in addition to the dimples, indicating that the fracture mode of the Al/Mg welded joint heated at less than 150 °C was a mixed fracture along the Al-side of the weld. The intermetallic compound layer was formed at the interface of Al-Mg welded joints heated at 200 °C and above. Because of the brittleness of intermetallic compound, the fracture mode of Al/Mg welded joint was brittle fracture through intermetallic compound layer. Combined with the mechanical properties and fracture morphology of the joint, the results showed that the formation of intermetallic compounds made the mechanical properties of the joint sharply decline.

4. Conclusions

The Al alloy and Mg alloy were successfully joined by magnetic pulse welding. The microstructure and mechanical properties of the as-weld and heated samples were studied. The evolution of the microstructure and properties of the Al/Mg welded joint after heat preservation treatment at different temperatures was discussed. The main results could be drawn as follows:

(1) The intermetallic compound layer composed of Al12Mg17 formed at the interface when the Al/Mg joint heated at 200 °C. The Al/Mg joint heated at 250 °C generated two intermetallic compound layers at the interface which were composed of Al12Mg17 and Al3Mg2, respectively.

(2) The formation of intermetallic compounds led to a sharp drop in the mechanical properties of the joints. The interface would be oxidized at the heat preservation temperature above 300 °C. The Al/Mg welded structural parts should be used at temperatures below 150 °C.

(3) The shear failure mode of Al/Mg welded joints is the ductile fracture (as-weld or heated at 50 °C) or mixed fracture (heated at 100 °C or 150 °C) along the Al-side of the weld. The shear failure mode after heating above 150 °C is brittle fracture through the intermetallic compound layer.

Data availability

The raw/processed data required to reproduce these findings cannot be shared at this time due to technical or time limitations.

Conflicts of interest

The authors declare no conflicts of interest.

Acknowledgements

This project is supported by the Foundation for Innovative Research Groups of the National Natural Science Foundation of China (51621004); the Key Research and Development Program of Hunan Province (2017GD2090).

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