Interface microstructure and formation mechanism of ultrasonic spot welding for Al-Ti dissimilar metals

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Abstract

The AA6061 Al and commercial pure Ti were welded by ultrasonic spot welding (USW). The focus of this investigation is the interface microstructure and joint formation. The Al-Ti USW joints were welded at the welding energy of 1100 J~ 3200 J. The joint appearance and interface microstructure were observed mainly by Optical microscope (OM) and field emission scanning electron microscope (SEM) The results indicated that good joint only can be achieved with proper welding energy of 2150 J. No significant intermetallic compound (IMC) was found under all conditions. The high energy barriers of Al-Ti and difficulties in diffusion were the main reasons for the absence of IMC according to kinetic analysis. The heat input is crucial for the material plastic flow and bonding area which plays an important role in the joint

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formation.

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1 Introduction

The light-weight structure notion is proposed for the performance enhancement and CO₂ emission reduction as the rapid development in the transportation, aerospace and automotive industries [1]. Titanium alloy is valued as a promising light-weight material with its high strength ratio, excellent thermal and corrosion resistance, but the high price limits its applications. More investigations have gravitated to the welding of Al and Ti dissimilar materials which is considered as a solution to balance property and cost.

Fusion welding is widely used in the industries but not very suitable for Al-Ti dissimilar metals. Intermetallic compounds (IMCs) which are characterized by brittle in most cases have a large tendency to form at the interface for Al-Ti fusion welding joints due to their high degree dissimilarity in metallurgical and physical properties [2-3]. In addition, problems of great deformation, internal stress concentration and grain coarsening which deteriorates the joint quality generally occur owing to high temperature.

Concerning brazing, Al-based, Al-Si, Zn-based filler metals for Al-Ti joining were studied [4-6]. The thick IMC layer formation and residual thermal stress are the main reasons for compromising the joint strength and restricting its application for
Al-Ti connection. Welding-brazing generally selects laser or arc as a heat source and provides suitable welding energy and faster heating rate. It has been regarded as an effective approach to suppress IMC thickness and control IMC morphology [7-8]. But requirements for high-level operation and strict filler selection decrease this method feasibility to some extent.

Friction stir welding (FSW) exhibits advantages on Al-Ti joining due to the low welding temperature and good ability of plastic material flow [9-10]. Zhao et al. found the thickness of IMC increased with the length of the probe and cleavage fracture gradually dominated the fracture of the joints due to the IMC formation [11]. Zhou et al. studied the rotation speed influence and found IMCs could form in the Al/Ti mixture and hook with the rotation speed 1400 rpm [12]. However, the probe wear problem during the process reduces joint stability and increases the cost [13]. Recently, Huang et al. [14] obtained excellent dissimilar Ti-Al joints with no probe worn by friction surfacing assisted hybrid friction stir welding. Another pressure welding method explosive welding (EXW) is preferred to fabricate multi-layer composites [15-16]. The interface of Al-Ti joints welded by EXW is characteristic of wavy or flat due to the enormous pressure and local high temperature. A subsequent annealing process is needed to improve the properties for this method [17].

Recently years, many investigations focused on using the ultrasonic spot welding (USW) to join aluminum and titanium dissimilar alloys because of its easy operation and high efficiency compared with other welding methods [17]. Ultrasonic vibration and clamping force are combined together to perform a solid state welding [18]. The
parameters optimization investigations regarded welding energy, welding force and displacement amplitude as crucial factors for the joint quality [19]. The relationship between summit load and welding time was studied by Zhang et al. [20]. Wang et al. [21] found adding pure aluminum interlayer could increase the adhesion effect. It is worth mentioning that no apparent IMC layer was detected at the interface by USW even under transmission electron microscope [22-23]. Previous studies have been proved the weldability of Al-Ti dissimilar alloys and the common phenomenon of no IMC formation at the interface by USW. However, most of investigations are about the influence and optimization of welding parameters, the mechanical properties and the fracture mode of welded joints. Interface microstructure and formation mechanism of USW for Al-Ti dissimilar alloys are barely clarified [24]. The influence of welding time to Al-Ti ultrasonic spot welds are introduced in our previous study [24]. This paper reveals the reasons of no IMCs at the interface with kinetic analysis basing on the microstructure studies and discusses the influence of welding heat input on joint formation.

2 Materials and Methods

1.5 mm thick AA6061 Al alloy and 1 mm thick commercial pure Ti sheets were used as the experimental materials. The nominal composition (wt.%) of AA6061 Al was 0.85 Si, 0.75 Mg, 0.7 Cu, 0.3 Mn, 0.25 Mn and Al balance. Output power of ultrasonic metal spot welder can reach 3.6 kW, and the vibration frequency is 20 kHz. The knurl pattern of sonotrode tip and the bottom anvil with a size of 10 × 10 mm are shown in Fig. 1(a). The specimens were cut as rectangles with dimensions of
65 mm × 20 mm and Al sheet was placed on top of Ti sheet with a 20 mm overlapped area, as described in Fig. 1(b). Specimens surfaces were ground with abrasive papers, cleaned with acetone and then dried in ambient air before welding. The sonotrode tip was set at the center of welds for the duration of USW process. The welding energy was selected ranging from 1100 J to 3200 J which is proportional to welding time. The vibration amplitude and clamping pressure were performed at 32 μm and 15 MPa, respectively. The K-type thermocouples of 0.5 mm diameter were inserted into welds center between Al sheet and Ti sheet to estimate the thermal cycle, the schematic is shown in Fig. 1(b).

![Fig. 1. Schematic illustration for (a) detailed pattern and dimension of the sonotrode tip (b) The configuration of welding sheets setup](image)

Metallographic specimens were cut perpendicular to the welding direction using electric discharge machine (EDM). The specimens were ground and polished with different grades abrasive papers and 1 μm diamond paste respectively, and then
etched with a 10% aqueous NaOH solution at 60 °C for 5 min, washed with a 5% nitric acid aqueous solution. Optical microscope (OM), field emission scanning electron microscope (SEM) equipped with an energy-dispersive X-ray spectroscopy (EDS) detector and FEI Tecnai-G2 F20 transmission electron microscope (TEM) were used to observed the joint appearance and interface microstructure. Thermodynamic and kinetic analysis were used to reveal the mechanism of no IMC formation.

3 Results and discussion

3.1 Joint appearance and microstructure

The joint morphology of the Al-Ti dissimilar alloys produced by USW with welding energy of 2150 J is shown in Fig. 4. The surface of aluminum and titanium base metal both had serrated indentation area which was 10 mm × 10 mm, as shown in Fig. 2(a). During the process, the sonotrode tip pressed into the aluminum sheet to conduct the ultrasonic energy. Friction was occurred between welding tip and aluminum, aluminum and titanium, titanium and bottom anvil by the high-frequency vibration of welding tip. There were no obvious welding defects at the joint appearance. Fig. 2(b) shows the cross-section of joint marked with yellow line as shown in Fig. 2(a). Compared with fusion welding which has high thermal input, no apparent fusion zone and heat affected zone were observed at USWed joints. According to the cross-section, it can be seen that the indentation of aluminum was deeper than titanium due to its inferior strength and hardness.
Fig. 2. Joint morphology at the welding energy of 2150 J:

(a) weld appearance of a typical Al-Ti USWed joint; (b) macrostructure of the joint cross-section marked by the dotted line in (a).

The cross-sections of joint were observed by OM as shown in Fig. 3(a)-(c) corresponding to the white rectangles marked in Fig. 2(b) respectively from top to the bottom. The microstructure change of aluminum side is mainly focused since the titanium side experienced no apparent deformation due to high strength. Plastic flows occurred at the upper surface of aluminum under the friction and compression of sonotrode as shown in Fig.3(a). Microstructure of the middle region of aluminum turned to the equiaxed from fibrous as shown in Fig. 3(b). The grains contiguous to the interface were refined by the intense friction between base materials as shown in Fig. 3(c). Above microstructure changes could be explained by the dynamic recovery and dynamic recrystallization of aluminum which were easily occurred under the welding temperature and high stress at the USW duration [25-26]. The refined effect also may be associated with the pinning effect by second phase particles (Mg-Si) that inhomogeneously distributed along the boundaries and within the grains.
Fig. 3. Typical joint microstructure from the up of aluminum to the interface

(a) the microstructure at position I in Fig. 4(b) (b) the microstructure at position II in Fig. 4(b)

(c) the microstructure at position III in Fig. 4(b).

The microstructure of interface varied with different welding energy. The observation result is exhibited in Fig. 4. Continuous gaps between Al and Ti could be observed from the Fig. 4(a). The welding energy in this condition was 1100 J which is not enough for the joint completing well. As the energy increased to 1625 J, the interface was partially joined as the Fig. 4(b) showing. With the energy of welding further increasing, the aluminum became more soften, and the connection between aluminum and titanium was more stable. The Al-Ti dissimilar alloys got a reliable joining under the sufficient welding energy of 2150 J, as shown in Fig. 4(c). Whereas, the excessive welding energy led to cracks formation at the interface so the strength of joint decreased, as shown in Fig. 4(d)-(e). Welding cracks were affected by the stress concentration resulting from the friction and fatigue effect by ultrasonic.
3.2 Interface structure

To verify the interface structure more precisely, SEM-EDS and TEM-EDS observations under a welding energy which is 3200 J are shown in Fig. 5 and Fig. 6, respectively. IMC is most likely observed under the maxim welding energy. The interface condition and elements distributions are studied.

The interface of Al/Ti dissimilar metals was obvious and no apparent reaction layer was found according to the SEM images. The wave-shape deformation at interface was observed as presented in Fig. 5(a). Sufficient welding energy is considered as the important factor to the interface deformation and well joining between dissimilar metals. The EDS line scan analysis approximately demonstrated atomic diffusion. Diffusion distance between Al and Ti was ~4 μm, as shown in Fig. 5(b). No IMC was detected even at the largest welding energy in this investigation.
The reasons are associated with kinetic analysis in the following.

Fig. 5. SEM-EDS results of the interface at a welding energy of 3200 J:

(a) the interface morphology (b) the elements distribution at the interface
Fig. 6. TEM-EDS analysis of the element at the interface at the welding energy 3200 J

(a) interface morphology (b) Al element (c) Ti element (d) Mg element (e) O element
(f) Si element (g) Fe element

The well bonding condition of Al-Ti joint was shown in Fig. 6(a)-(c) and element distributions of Mg, O, Si and Fe were shown in Fig. 6(d)–(g). The distributions of magnesium were consistent with oxygen according to Fig. 6(d) and (e). The segregation of Mg and O both occurred at interface. The concentrated magnesium with high activity led to the growth of oxygen absorption rate of Al. On the other hand, the friction between the faying surfaces of Al-Ti at the process duration caused the temperature high enough to form the oxide film of Mg, which is confirmed by Field [27]. Silicon originally in the aluminum base metal diffused toward the interface as shown in Fig. 6(f), the Fe element which is the highest content impurity element in the titanium was few at the interface as shown in Fig. 6(g).

The greater the negative chemical enthalpy of elements X and Y, the greater the chemical attraction between them. The chemical enthalpy between elements Si, Mg, Fe and base material Al, Ti are displayed at the Table 1. From the table, $\Delta H_{Ti-Mg}^{mix}$ (16 kJ·mol⁻¹) is higher than $\Delta H_{Al-Mg}^{mix}$ (-2 kJ·mol⁻¹) which reveals that the segregation of Mg was not attracted by Ti but was spontaneous to aluminum surface. The value of $\Delta H_{Ti-Si}^{mix}$ is -66 kJ·mol⁻¹, which is much lower than that of $\Delta H_{Al-Si}^{mix}$ (-19 kJ·mol⁻¹). The migration of silicon to the Al-Ti interface from aluminum alloy could be attributed to the greater affinity between Si and Ti. The difference of $\Delta$
and is not significant, so the driving force is insufficient for Fe diffusing to the Al-Ti interface that’s why there was few Fe at the interface. The analysis results are consistent with Fig. 6. Generally, elements segregation has negative effect on the mutual diffusions between Al and Ti.

Table 1. Chemical combination of alloying elements (Si, Mg, Fe) with matrix (Al, Ti)

<table>
<thead>
<tr>
<th>Diffusion couple</th>
<th>$\Delta H_{mix}^{m_{X-Y}}$ (kJ·mol$^{-1}$)</th>
<th>Diffusion couple</th>
<th>$\Delta H_{mix}^{m_{X-Y}}$ (kJ·mol$^{-1}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al-Ti</td>
<td>-30</td>
<td>Ti-Al</td>
<td>-30</td>
</tr>
<tr>
<td>Al-Mg</td>
<td>-2</td>
<td>Ti-Mg</td>
<td>16</td>
</tr>
<tr>
<td>Al-Si</td>
<td>-19</td>
<td>Ti-Si</td>
<td>-66</td>
</tr>
<tr>
<td>Al-Fe</td>
<td>-11</td>
<td>Ti-Fe</td>
<td>-17</td>
</tr>
</tbody>
</table>

The observation results indicate that there were no IMC at the interface. Thermodynamics and kinetics are discussed for analyzing this phenomenon. According to the Al-Ti binary phase diagram, many intermetallic compounds have the possibility to form between Al and Ti, including TiAl, TiAl$_3$ and Ti$_3$Al. The formation of reaction products is associated with free energy difference. The relationship between free energy difference and the temperature is as follows from the Gibbs-Helmholtz equation:

$$
\Delta_r G_m^\ominus = \Delta_r H_m^\ominus - T \Delta_r S_m^\ominus
$$

In the formula, $\Delta_r H_m^\ominus$ is the Standard molar reaction enthalpy at temperature T (kJ·mol$^{-1}$);

$\Delta_r S_m^\ominus$ is the standard molar reaction entropy change at temperature T (kJ·(K·mol)$^{-1}$).

$$
\Delta_r H_m^\ominus (T) = \Delta_r H_m^\ominus (298.15 \, K) + \int_{298.15 \, K}^{T} \Delta_r C_{p,m} \, dT
$$
$$\Delta_s^s_m(T) = \Delta_s^s_m(298.15\text{ K}) + \int_{298.15\text{ K}}^T \frac{\Delta r u_m}{T}dT$$

(4-3)

$$\Delta r u_m = C_{p,m}(Ti_{p,Al}) - [pC_{p,m}(Ti) + qC_{p,m}(Al)]$$

(4-4)

where, $\Delta r u_m$ is the constant pressure molar heat tolerance of reactants and products (kJ·(K·mol)$^{-1}$).

Fig. 7. Curves of Gibbs free energy of Ti-Al intermetallic compounds varying with temperature

The functions of Gibbs free energy of different IMC varying with the temperatures were provided by Kattner [28], as shown in Table. 2. Utilizing the data from articles, the Gibbs free energy of Ti-Al IMCs at the range of 273 K to 1300 K was shown in Fig. 7. The TiAl$_2$ has the largest free energy change as shown in Fig.9, but the TiAl$_2$ need to be formed on the basis of AlTi, so TiAl$_3$ has the largest possibility to form during the welding process. However, TiAl$_3$ wasn’t observed at all conditions so the following kinetic analysis is necessary.

Table 2. The functions of Gibbs free energy of IMC in the series of Ti-Al varies with the temperature

<table>
<thead>
<tr>
<th>Compounds</th>
<th>Free energy of formation(kJ · mol$^{-1}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>TiAl</td>
<td>-37445.1+16.79376*T</td>
</tr>
<tr>
<td>Ti3Al</td>
<td>-29633.6+6.70801*T</td>
</tr>
<tr>
<td>TiAl2</td>
<td>-43858.4+11.02077*T</td>
</tr>
<tr>
<td>TiAl3</td>
<td>-40349.6+10.36525*T</td>
</tr>
<tr>
<td>Ti2Al5</td>
<td>-40495.4+9.52964*T</td>
</tr>
</tbody>
</table>
At the range of USW temperature, the growth of TiAl$_3$ layer is determined by diffusion. The thickness of reaction layer and time satisfied the parabolic law at a certain temperature, usually simplified as follows:

$$\ln x = n \cdot \ln t + \ln K$$

(4-5)

Where, $K$ is the rate constant; $t$ is the reaction time and $n$ is the kinetic index.

The transformation rate depends on the frequency of the atom reaching the active state, as determined by the Arrhenius equation:

$$K = A \cdot \exp \left( -\frac{\Delta G^a}{RT} \right)$$

(4-6)

Where $A$ is the pre-factor; $\Delta G^a$ is the growth energy of the reaction layer; $T$ is the absolute temperature.

A series of metastable phases may appear during the transformation process as long as the Gibbs free energy of the phase transitions decrease. Fig. 8 shows a free energy diagram of a single atom transition from a metastable state to a lower energy steady state during phase transition where $G_1$ and $G_2$ are the metastable and steady-state Gibbs free energy, respectively. To complete this transition, the atom must receive more free energy to reach the activation state. Therefore, activation energy determines the difficulty of the whole diffusion system essentially. The USW of Al-Mg, Al-Fe and Al-Ti all belong to solid-solid reactions which are more difficult to form IMC compared with liquid-liquid or liquid-solid status, but IMCs aren’t observed only in Al-Ti interface [29]. That may be explained that Al-Ti has a diffusion activation energy with 250–300 k J/mol, which is greatly higher than that of Al-Mg and Al-Fe with diffusion activation energy of 60-70 kJ/mol and 190 kJ/mol respectively [30-31].
A solid solubility is prerequisite for atom diffusion. The calculations of dissimilar metals solubility are shown in Fig. 9. The solid solubilities of Al-Mg and Al-Fe are high, although the solubility of Fe in Al is low which is similar to that of Ti in Al, but the solubility of Al in Fe is the highest among them, which could also explain that IMCs exist in the Al-Mg and Al-Fe but except Al-Ti interface. A common way to diffuse in the multi-crystalline solids is grain boundary diffusion [32]. Relationships between the effective diffusion coefficient of polycrystalline solids and bulk diffusion coefficient and grain boundary diffusion coefficient are as follows [32]:

\[ D_{eff} = g \cdot D_{gb} + (1 - g) \cdot D_L \]  
\[ g = \frac{q \delta}{d} \]

Where \( D_{gb} \) is the grain boundary diffusion coefficient; \( D_L \) is the bulk diffusion coefficient; \( g \) is the grain boundary volume fraction; \( \delta \) is the grain boundary width; \( d \) is the grain size; \( q \) is the numerical factor of the grain shape.

It is indicated that the size and configuration of grains affect the effective diffusion coefficient. Diffusion rate increases with the grain refinements. According to the analysis of microstructure at the joint interface, the grain sizes of aluminum became coarse with the increasing heat input due to the decreasing effective diffusion coefficient. Hence, the IMC growth driving force is insufficient because the diffusion
amounts from Al to Ti are difficult to achieve the solid solubility limit during the short welding time.

Similar to elements segregation which would deteriorate the joint mechanical properties and prevent the growth of interaction layer to some extent, the residual oxide film formed at the titanium surface as a diffusion barrier hindered the elements movements at the initial stage of diffusion. Because the vibration amplitude of USW was small (<37 μm) while the welding tip width was of 10 mm, therefore the broken oxide film at the center of the joint couldn’t be moved outside the interface.

![Mutual solid solubility between Al-Ti, Al-Mg and Al-Fe](image)

**Fig. 9.** Mutual solid solubility between Al-Ti, Al-Mg and Al-Fe

### 3.3 Thermal cycle

At the USW process duration, the thermal cycle varying with the welding parameters has a closely relationship with the microstructure and mechanical properties of joints. The thermal cycles of the welds center corresponding to welding energy of 1100–3200 J are revealed in Fig. 10(a). All the temperatures of the center welds rose rapidly to the maximum value after that gently declined to the room temperature. To see the details more clearly, magnification of the thermal cycle curve
at the welding energy of 3200 J is shown in Fig. 10(b). Three parts are divided to describe the thermal cycle, which are temperature rise period, high temperature holding period and cooling period. The temperature of the workpiece first rose rapidly under a heating rate of 648 K/s due to friction and clamping pressure. It cost approximately 0.8-1 s to achieve the maximum temperature. When the time exceeded 1 s, the ability to generate heat and the ability to disappear heat reached equilibrium so the thermal cycle achieved temperature holding period. Once achieved the set time, the heat input stopped and the temperature of the workpiece started to descend. The high temperature holding period plays a critical role in the plastic deformation and element diffusion. The longer the high temperature holds, the more sufficient elements diffuse in theory.
Fig. 10. Thermal cycling curves at different welding energies: (a) the trend of thermal cycling during the welding process (b) a partial enlarged view of the thermal cycle variation curve at 3200 J welding energy

3.4 The welding joint formation mechanism

Welding heat input is considered as an important factor for the joint formation. Therefore, the mechanism of the joints formation by USW is explained by analyzing the joint formation with different heat inputs as follows.

1) When the heat input is low

During the welding process, the sonotrode and bottom anvil made jagged indentations on the surface of upper and bottom workpiece respectively to conduct the ultrasonic vibration to the interface of two sheets. The ultrasonic energy conducting by the sonotrode induced the intense mutual friction and pressure concentration on some projections of base material. When the heat input is low, the number of micro-bonding is small and the strength of biting facets is low. The micro-bonding is destroyed by the shear stress of ultrasonic vibration rapidly. Meanwhile the heat input is insufficient to break the oxide film so the dissimilar metals can’t contact directly.
Consequently, the metallurgical connections are difficult to form, as shown in Fig. 11(a).

(2) When the heat input is suitable

When the welding process improves, the micro-bonding regions has plastically deformed. The wave-like deformation at interface under the pressure and shear force generated by relative sliding and the swirls plastic flows at the micro-connections which improved mechanical connections are shown in Fig. 11(b). During the friction process by the ultrasonic vibration, the temperature rapidly increases while the deformation resistance of materials decreases with the area of friction expanding. At the same time, the oxide film is broken, displaced and dispersed. The atoms at the surface of Al-Ti dissimilar alloys are close to attract each other and the elements such as Al, Si diffuse quickly through these passages, and the joints with effective bonding areas are formed, as shown in Fig. 11(c).

Under the stress and shear force generated by ultrasonic vibration, the plastic flows continue and the broken oxide films disperse into the inner position of base material. As the friction process progressing, the number of points of biting increases with the effective connection area expanding continuously. When the bonding force at the weld is higher than shear stress caused by the ultrasonic mechanical vibration, the workpiece is no longer cut by the shear stress and a strong joint is formed. When the heat input reaches the optimum point, the joint strength is the highest, as shown in Fig. 11(d).

(3) when the heat input is too high

Since the defection of sonotrode netting, difference in the thickness of plates and weld deformation, the forces on the weld spot aren’t uniform during the welding which lead to the physical contact incomplete so the interspace may be exist between
the plates. As the heat input continuously increases, the diffusion distance extends, and the stress concentration arises by the shear stress and fatigue effect with high frequency ultrasonic vibration and the accumulated deformation resistance internal stress, which lead to the crack appears at the joint as shown at the Fig. 11(e).

Fig. 11. Schematic illustration for the joint formation under different heat input conditions

(a) insufficient welding heat input (b) schematic diagram of material plastic flow at the interface

(c) appropriate welding heat input (d) optimal welding heat optimal

(e) excessive welding heat input

4 Conclusions

The microstructure of the Al-Ti joints by USW was investigated, the reasons of no visible IMC were analyzed, the mechanism of joint formation was discussed, the important conclusions are following:

(1) 1.5mm thick AA6061 Al and 1 mm thick commercial pure Ti sheets can get a sound welded joint by USW. With the increasing welding energy, the base material
plastically deformed with grain gradually coarsening and recrystallization. Under the high stress and friction, the upper of aluminum and nearby the interface part had a plastic flow with grains refinement. The intermetallic compounds reaction layer was not observed under all welding parameters.

(2) The IMC of aluminum and titanium was difficult to form by USW due to the high energy barrier of Al-Ti essentially. Another important reason is that the solid solubility limitation of Al-Ti is inaccessible during the short welding process because of the elements segregation, residual oxide and decreasing effective diffusion coefficient with the coarsening grains on the Al side.

(3) During the USW process, the weld center temperature with different welding parameters rapidly rising to the maximum costing about 0.8-1 s, and then decreased to the room temperature slowly. The heating rate was 648 K/s. Temperature rise period, high temperature holding period and cooling period can be observed from the thermal cycle.

(4) Combined the action of static pressure and ultrasonic mechanical vibration, mechanical occlusion occurred between the surfaces of Al-Ti dissimilar alloys, and the interface was gradually expanded by discontinuous micro-connections until an effective joint formed. When the welding heat input was appropriate, the mechanical fitting and metallurgical bonding between the materials were excellent.
Acknowledgements

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