Hybrid laser welding of dissimilar aluminum alloys: welding processing, microstructure, properties and modelling

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\section{Introduction}

Laser-arc welding is increasingly applied for joining aluminum alloy (AA), as this advanced technology is characterized with high-welding speed and low heat input \cite{1,2}. Up to date, extensive efforts have been made toward understanding the microstructural evolution in different zones of the joint \cite{3–5}, mechanical properties \cite{6–8}, fatigue properties \cite{9–12}, corrosion resistance \cite{13,14}, and the dynamic behavior of welding pools \cite{15}. These studies demonstrated that laser-arc welded components of Al alloys have smaller residual stress, greater tensile and fatigue strength than those produced via traditional welding methods (e.g. metal insert gas welding). Consequently, the laser-arc welding is considered as a promising method for joining similar Al alloys.

It should be mentioned that the focus of previous studies was on the investigation of laser-arc welded joints consisting of similar Al alloys, though joints with dissimilar Al alloys become more and more popular in modern structures of automobiles and high-speed trains \cite{16,60}. The welding of dissimilar Al alloys, especially in the case associated with 7xxx serials Al alloys, is usually conducted by friction stir welding (FSW) \cite{17–19}. This is because that FSW is advantageous over laser-arc welding in obtaining stronger joints of 7xxx Al alloys. The joint efficiency of FSW AA7N01 joints can reach as high as 83\% \cite{20}, whereas only about 63\% can be achieved with laser-arc welded AA7N01 joints \cite{13,21}. During laser-arc welding of AA7N01, strengthening elements like Mg and Zn can be vaporized, resulting in loss of strength. Therefore, FSW is a preferable method to manufacturing structures containing 7xxx serials Al alloys. However, FSW is not the best choice in welding Al-Mg alloys like AA5083. The joint efficiency of FSW AA5083 joints is reported to be around 59.3–88\% \cite{22,23}. This maximum value is the same as that in laser-arc welded AA5083 joints (88\%) \cite{5}. Also, the joint efficiency in dissimilar FSW joints consisting of AA5083 and AA7xxx is about 77–87\% \cite{24}, which is smaller than that in laser-arc welded AA5083 joints. The existing results suggest that FSW may be not a preferred welding method or has no obvious advantages in the case of joining AA5083 and other types of Al alloys. Under this circumstance, laser-arc welding turns out to be an alternative to join AA5083 with other types of Al alloys, given that FSW is difficult to produce complicated structures like T-shaped or pressure vessels.

\section*{ARTICLE INFO}

\textbf{Keywords:}
Laser-arc welding
Dissimilar Al alloys
Microstructure–properties relationship
Theoretical modeling

\section*{ABSTRACT}

Hybrid laser-arc welding of aluminum alloys has been receiving more and more attention in academic and industrial communities. The outstanding mechanical and fatigue properties of laser-arc welded joints with similar Al alloys have been demonstrated in literatures. Very few studies, however, reported the microstructures and properties of laser-arc welded joints with dissimilar Al alloys, while they are increasingly applied in modern engineering structures. Here, we study laser-arc welding of two dissimilar Al alloys, AA5083 and AA7N01. Their weldability, microstructures and properties are particularly investigated. Results show that sound welded joints of dissimilar Al alloys can be obtained via adjusting welding parameters. Microstructural characterization via SEM, EBSD and TEM reveals that microstructure in the fusion zone (FZ) is featured by large precipitates, low dislocation density, and coarsened grains with an average size of 66±57 μm. The tensile strength of FZ of coarsened microstructure is degenerated to 230 MPa, which is the lowest among the three zones. The fatigue strength of the joints is 110 MPa, about 40\% of their tensile strength. Pores and inclusions are the main sources for the deterioration of fatigue strength, as evidenced from SEM observations. In addition, a strength model is successfully built and utilized to simulate the yield strength, strain rate hardening and work hardening behavior of fusion zone.

https://doi.org/10.1016/j.jmapro.2020.03.048

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Received 3 January 2020; Received in revised form 19 March 2020; Accepted 25 March 2020

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Unlike laser-arc welding of similar Al alloys, several issues need to be addressed for laser-arc welding of dissimilar Al alloys:

1. Porosity, which is notorious for deteriorating the mechanical and fatigue properties of joints. It is known that high-Mg-contained Al–Mg alloy is prone to form pores since Mg is easily evaporated in welding [25]. Normally, by adjusting welding parameters to a certain configuration, macroscale pores in joints can be significantly reduced. The effectiveness of this method is yet unknown when laser-arc welding of Al–Mg alloys with other types of Al alloys. Moreover, micro pores are found to play a detrimental role in worsening the fatigue strength of welded joints, but their influences on the joints of dissimilar Al alloys have not been sufficiently investigated.

2. Solidification cracks, which are more detrimental than pores with respect of worsening the properties of joints. Solidification cracks are normally found in the joints of high-strength Al alloys such as Al–Zn–Mg alloy. The tendency of forming solidification cracks will slow down along with the decrease of the solidification rate. However, the solidification rate is difficult to be controlled as dissimilar Al alloys have different thermal conductivities. This hints that the solidification cracks may be a critical issue when laser-arc welding of dissimilar Al alloys.

3. Besides the problems (1) and (2) mentioned above, microstructure and mechanical properties are our focuses, which are the fundamental factors influencing the appropriate usage of laser-arc welded joints with dissimilar Al alloys in engineering structures. FZ is found to be the softest zone in joints of similar Al alloys such as Al–Mg and Al–Mg–Si alloys [26], because microstructure in FZ features coarsened grains and large precipitates [27]. But heat affected zone (HAZ) in joints of Al–Zn–Mg alloys is reported to be the softest zone [28]. When welding of dissimilar Al alloys, the heat input and cooling rate will be different from those in welding of similar Al alloys. This difference in heat input and cooling rate can significantly influence the microstructure obtained. Therefore, the microstructure and resultant properties are worth investigating.

The alloys of AA5083 (Al–Mg alloy) and AA7N01 (Al–Zn–Mg alloy), the most frequently used Al alloys in high-speed train bodies, are adopted as samples for producing joints by laser-arc welding technology. The weldability is firstly investigated to avoid macroscale defects such as pores and cracks. Then, microstructures, mechanical properties in local areas, and fatigue resistance are characterized by experiments to obtain a complete picture of the microstructure and property evolution of the joints. Finally, a quasi-static strength model is built and applied to predict the mechanical behavior of the joints. The microstructure-properties relationship is then discussed based on this model.

2. Experimental methodology

2.1. Hybrid laser-MIG welding

Apparatus of hybrid laser-MIG welding used in this work was described elsewhere [5,21,26]. For the sake of readers’ convenience, the relevant information of the apparatus is briefly presented here. The laser power was provided by IPG YLS-4000 fiber laser which had a 1.07 μm wavelength and 200 μm core diameter. Part of laser beam can be reflected by Al alloys during welding, and such reflection can damage the laser equipment. Thus, an inclined angle of 15° was applied between the laser beam and the vertical direction. Arc welding heat source was supplied by a MIG welding machine, named EWM PHOENIX 421 EXPERT force Arc MIG. The distance and angle between the arc and laser heat source were 2 mm and 35°, respectively. During welding, pure argon was used as shielding gas to prevent the formation of coarsening pores. 4 mm-thick AA5083 and AA7N01 were as base metals (BM) and were joined to achieve butt-type joints. The filling material was 1.6 mm-diameter ER5356 which was an Al-Mg alloy. Before welding, BMs were mechanically cleaned to remove contaminations of oxide surface and oils. Based on our previous studies [5,21,26], three welding parameters (i.e. laser power, arc welding current and welding speed) were adjusted to produce sound joints.

2.2. Microstructure characterization

To examine the evolution of microstructures in joints, samples along the cross-section of the joint were cut by an electrical discharged machining. The procedure of preparing metallurgy samples was similar to that in Ref. [29]. Briefly, the sample was mechanically grinded with different graded SiC papers (400 grit, 800 grit, 1200 grit, and 2000 grit), and polished with diamond paste to achieve mirror-like surface. The final step of polishing lasted for about one hour to remove the deformed surface layer with the aid of 0.02 μm colloidal silica solution. One of well-prepared samples was etched with Keller’s reagent for optical observation, others were used for electron backscatter diffraction (EBSD) testing which was performed in a scanning electron microscope (SEM), named Zeiss Ultraplus SEM. The parameters for EBSD test were the same as those in Ref. [30]. The voltage and current for EBSD observations were 20 kV and 4 nA, respectively. The scanning step size ranged from 0.4 μm to 4 μm.

The TEM sample was first grinded using 800 grit, 1200 grit and 2000 grit SiC papers. During this process, water was used as polishing solution. When the sample’s thickness was around 100 μm, the center of the sample was polished with diamond paste (1 μm grade) until the thickness was around 20 μm. Then, the 20μm-thick sample was put in a Gatan precision ion polishing system and polished with argon ions until the thickness of the center reached 100 nm. The TEM experiment was carried out in a JEOL 2100F microscope at the acceleration voltage of 200 kV.

2.3. Mechanical properties characterization

Mechanical properties of the welds were studied using three testing methods, which were microhardness testing, onset tensile testing, and local area tensile testing. The microhardness was measured on the cross section of the weld, as shown in Fig. 1a. The maximum load and holding time at this load in microhardness testing were respectively 3 kg and 15 s. Two types of tensile testing samples were machined, as illustrated in Fig. 1a, one was for determining the tensile strength of the whole joint, another was for obtaining strength in local places (i.e. FZ, HAZ and BM). The tensile testing sample was prepared via electric discharge machining (EDM) following the standard ASTM: B557-10, and the dimensions of samples for tensile testing are shown in Fig. 1b and c. These two types of tensile tests were performed with a stretching rate of 0.5 mm/min. At least three samples were tested for the tensile examinations, and the averaged value is taken as the results reported in this paper.

2.4. Fatigue characterization

The specimen for fatigue testing was made by the EDM method. The dimensions of the specimen are shown in Fig. 2. Before testing, all specimens were carefully grinded with fine sandpapers. High-frequency fatigue testing machine was used in the experiment within the frequency range from 97 Hz to 104 Hz. Tension-tension cyclic loading was applied on the sample with a stress ratio of 0.1. Fatigue tests were performed with the maximum stress ranging from 150 MPa to 110 MPa to find fatigue limits. The fatigue limit was considered as the stress at which the specimen was not broken after 1 × 10^7 cycles. SEM was employed for observing the surface of fractured samples and exploring cracking initiation and propagation.
3. Results

3.1. Weldability

It is known that defects like pores produced in welding can deteriorate the mechanical and fatigue properties [8,13,14,31]. Bunaziv et al. [7] reported that pores can reduce the strength of welded AA5085 joints when the porosity was as high as 50%. And the elongation of the joints is reduced when there are pores in joints [32]. Further, several studies [14,33,34] have reported that the fatigue strength of laser-arc welded Al alloys can be seriously softened due to the existence of pores, since these pores can act as stress concentration site and initiate fatigue cracks. Thus, efforts were made to adjusting welding parameters for sound welded joints without macropores, based on the work in Ref. [26]. Some results are shown in Fig. 3. Clearly, the macropores can be avoided by decreasing the welding speed. The mechanism behind controlling the porosity in hybrid laser welded joints has been discussed in Ref. [26]. Solidification cracks are not observed during welding, as demonstrated in Fig. 3. Moreover, it can be seen from Fig. 3 that the optimal welding parameters are: \( P_{\text{laser}} = 1.3 \, \text{kW}, I_{\text{arc}} = 130 \, \text{A}, \nu = 0.64 \, \text{m/min}. \)

3.2. Microstructures

3.2.1. Results from observation of OM, SEM, and TEM

Fig. 4 displays the optical microscope (OM) images for different zones in AA5083-AA7N01 joints (referred as AA5-7 joints hereafter). Dendrites, whose shapes are similar to those in Ref. [13], can be seen in FZ. These dendrites are parallel to each other with a space of \( \sim 5.8 \, \mu \text{m} \). The space of dendrites depends on cooling rate [35], a higher cooling rate results in a smaller dendritic space. Moreover, the orientation of dendrites may be related to the grain orientation, as demonstrated in Fig. 4a. Different microstructures can be seen in HAZs of AA7N01 and AA5083. For simplicity, the HAZ on the side of AA7N01 and AA5083 is referred as HAZ-7 and HAZ-5, respectively. No clear microstructure information such as grains size and precipitates can be obtained in the OM image of HAZ-5 (Fig. 4b), whereas coarsened phases and grain boundary are clearly observed in HAZ-7 (Fig. 4c). Similar to HAZ-7, the AA7N01’s microstructure can be illustrated via the etching process (Fig. 4d), and grain boundary and fine phases can be seen in Fig. 4d. No specific information about the microstructure is provided in BM AA5083 from the OM image, as demonstrated in Fig. 4e. The etching process also shows the presence of fine precipitates in BM AA5083.

<table>
<thead>
<tr>
<th>Welding parameters</th>
<th>Images from X-ray detection</th>
</tr>
</thead>
<tbody>
<tr>
<td>Laser power: 1.3kW</td>
<td>![Image of weld bead]</td>
</tr>
<tr>
<td>Arc current: 130 A</td>
<td>![Image of weld bead]</td>
</tr>
<tr>
<td>Welding speed: 1 m/min</td>
<td>![Image of weld bead]</td>
</tr>
</tbody>
</table>

![Image of weld bead]
The solution used in this study is Keller’s reagent, which may not be good for illustrating the microstructure of Al–Mg alloys, instead another etching agent (e.g. Weck’s reagent) may be more applicable [36].

To find out the chemical compositions of the particles in FZ, energy dispersive spectroscopy (EDS) is applied, and the results are shown in Fig. 5. There are two types of particles in FZ, one is a coarsened rod-like particle, another is much finer. The distribution of chemical elements in the coarsened particle is illustrated in Fig. 5c–e, which shows that Al is not abundant while two other elements (Mg and Zn) are rich in this particle. Judging from the combination with the EDS results (Fig. 5f), the coarsened particle is likely to be Mg2Zn which is the dominant particle in Al–Zn–Mg alloys. The fine particle has the chemical elements of Al, Mg, Cr, Mn, as shown in Fig. 5g. The elements in particle-free space (marked as “3” in Fig. 5a) are Al and Mg, as shown in Fig. 5h.

TEM images of particles in FZ are illustrated in Fig. 6. Small and coarsened rod-like particles are found in TEM images, a finding that is consistent with that from the SEM observation. With the help of selected area diffraction pattern (SADP), the small particle is found likely to be Al13Cr2, whereas the coarsened one may be Cu16Mg6Si7. Interestingly, the size of particles in this study is smaller than that in Refs. [8,13,14,30].

3.2.2. Results of EBSD

As shown in Fig. 4, some details of the microstructures of HAZ-5 and BM AA5083 are not prevailed from OM observation. Therefore, EBSD testing was conducted to provide further information of the microstructures in the welded joint. Inverse pole figures (IPF) of the joint are shown in Fig. 7. Equiaxed grains are found in FZ, HAZ-5, and BM AA5083, as shown in Fig. 7a–c. The intercept measurements show that the grain size in FZ, HAZ-5, and BM AA5083 is 66 ± 57 μm, 12 ± 10 μm, 10 ± 9 μm, respectively. Grains in HAZ-7 and BM AA7N01 have large length and small width, as displayed in Fig. 7d and e. Kernel average misorientation (KAM) in the welded joint is displayed in Fig. 8, and the dislocation density is evaluated from KAM via the following equation [14,37,38]:

\[
\rho = \frac{\theta_{KAM}}{|b| \times \mu \times n}
\]

where \(\rho\) presents the density of dislocations in the area considered, \(\theta_{KAM}\) the KAM angle, \(b\) the Burgers-vector magnitude which is 0.3 nm here, \(\mu\) the step size during EBSD examination, which varies from 0.5 μm to 4 μm, \(n = 2\) the number of nearest neighbours. The calculated results are listed in Table 1, which shows that the FZ has the lowest dislocation density with a value of \(1.8 \times 10^{12} \text{ m}^{-2}\). This value is smaller than that in Refs. [30,37]. The difference may be due to the cooling rate which plays a key role in generating dislocation densities, as argued by Gao et al. [39] and Bu [40].
Fig. 5. (a) SEM image of FZ, (b) the enlarged view in the marked phase in (a), (c)-(e) element distribution in the selected phase, (f)-(h) are EDS results respectively for the phase marked as “1”, “2” and “3” in (a).

Fig. 6. TEM images of (a) small particles, (b) coarsened rod-like particle.
3.3. Mechanical properties

Fig. 9 is the microhardness profile for the AA5–7 joint. The smallest hardness is found in the FZ, which is consistent with the work of Zhan et al. [41]. The hardness of HAZ-5 is slightly smaller than that of BM AA5083, while the hardness of HAZ-7 is obviously smaller than that of BM AA7N01. Since there is not much difference in terms of dislocation density and grain size of HAZ-7 and BM AA7N01, the worsened hardness in HAZ-7 may be because the phase in HAZ-7 become larger and have smaller density.

Fig. 10 shows tensile strength of the joints of dissimilar Al alloys, joints of similar Al alloys, and BMS. For simplicity, joints of similar AA5085 and AA7N01 are referred as AA5–5 joints and AA7–7 joints, respectively. Interestingly, AA5–7 joints and AA5–5 joints have very close tensile strength, that is an averaged tensile strength of 274 MPa and 273 MPa, respectively. AA7–7 joints have a higher tensile strength of 288 MPa. For BM, tensile strength of AA5083 is weaker than that of AA7N01. From the results of onset tensile testing, it seems that the strength of dissimilar Al alloys joints is determined by the BM that has relatively weaker strength, a finding which is consistent with the work.

![Fig. 7. IPF images of (a) FZ, (b) HAZ-5, (c) BM AA5083, (d) HAZ-7, (e) BM AA7N01.](image1)

![Fig. 8. KAM images of (a) FZ, (b) HAZ-5, (C) BM AA5083, (d) HAZ-7, (e) BM AA7N01; (f) distribution of KAM in different zones.](image2)

<table>
<thead>
<tr>
<th>Local area</th>
<th>FZ</th>
<th>HAZ-5</th>
<th>BMAA5083</th>
<th>HAZ-7</th>
<th>BMAA7N01</th>
</tr>
</thead>
<tbody>
<tr>
<td>KAM (°)</td>
<td>0.75</td>
<td>0.65</td>
<td>0.75</td>
<td>1.15</td>
<td>0.55</td>
</tr>
<tr>
<td>$\rho_{GND}$ ($10^{12} \text{ m}^{-2}$)</td>
<td>1.8</td>
<td>6.4</td>
<td>7.3</td>
<td>11</td>
<td>10.8</td>
</tr>
</tbody>
</table>

Table 1: KAM and dislocation density for local areas obtained from EBSD test.
of Wang et al. [42]. Stories are different if BMs are 6xxx Al alloys [16,29], as particles in HAZ on the side of 6xxx Al alloys would be dissolved and coarsened during welding thermal cycles.

SEM images of fractured surface show that there are ample dimples on the fractured surface of AA5–7 joints (Fig. 11a), while particles can be found at the bottom of the dimples for BMs (Fig. 11b and c). The local tensile strength of FZ, HAZ and BM is shown in Fig. 12, which clearly displays that the softest region is in FZ. The tensile strength of HAZ-5 is only slightly worsened compared to that of BM AA5083. However, the tensile strength of HAZ-7 is obviously weaker than that of BM AA7N01. The results of local tensile testing and microhardness testing illustrate that FZ has the lowest strength in the joint, a finding which is consistent with the work in Refs. [26,29].

### 3.4. Fatigue resistance

Fig. 13 is the fatigue life of joints under different applied loadings. For comparison, the fatigue lives of AA7–7 and AA5–5 joints are also depicted. As expected, the fatigue lives of all the three kinds of joints decrease with increasing applied loadings. And the conditional fatigue endurance (i.e. \( N_{f} \geq 10^7 \)) of AA5–7 joints is 110 MPa, which equals to that of AA7–7 and AA5–5 joints. The conditional fatigue strength of AA5–7 joints is about 40% of their tensile strength.

SEM images of fractured surface are shown in Fig. 14. Cracks initiating from inclusions are clearly observed, as demonstrated in
Fig. 13. Results of fatigue testing.

Fig. 14a, and secondary cracks along the grain boundary or penetrating grains are also found in propagation sites. Fatigue striations also distribute in part of the propagation sites (Fig. 14b). As illustrated in Fig. 14c, pores act as preferable sites for crack initiation, and these cracks then travel towards the inner side of samples. The propagation site is covered by cleavage river patterns with a small quantity of dimples (Fig. 14d). As evidenced in Fig. 14a and c, sites with inclusions and pores are the primary places of fatigue crack initiation (FCI). This finding is consistent with the works in Refs. [9,14,26,43,44]. As shown in Fig. 14e and f, stress concentration is usually found near the pores and inclusions, and the cracks will form in these high-stress-concentration places under cyclic loadings. The propagation of cracks may depend on several factors, such as the stress state and crystals where the crack tip lies in. Evidences have been founded from EBSD results which show that cracks can travel along the grain boundary or penetrate grains [34,45]. Since the crack’s propagation is not the focus of this study, the detailed behavior of fatigue cracking is not discussed here.

4. Discussion

4.1. Modelling the strength in softest zone

The results of mechanical testing show hierarchy mechanical properties in the AA5−7 joint with the FZ exhibiting lowest strength. Generally, the static strength of a welded joint is determined by the softest zone, since the plastic strain prefers to accumulate in the softest
zone [8,46,47]. A model, which can determine the strength of the softest zone of a joint, is crucial and can be used to predict the onset strength of the joints, as demonstrated by Yan et al. [8,29]. In this section, a strength model is built and used to predict the yield strength, strain rate hardening, and work hardening behavior of FZ.

Macroscopic yield strength of a metallic material benefits from strengthening mechanisms [48] such as intrinsic strengthening ($\sigma_i$), precipitation strengthening ($\sigma_p$), dislocation interaction strengthening ($\sigma_d$), solid solution strengthening ($\sigma_s$), and grain boundary strengthening ($\sigma_{gb}$). To obtain the macroscopic yield strength ($\sigma_{yield}$), a linear additive law is used according to Refs. [49–54], i.e.,

$$\sigma_{yield} = \sigma_i + \sigma_p + \sigma_d + \sigma_s + \sigma_{gb}$$  \hfill (2)

Since the fusion zone is in as-welded state, meaning that the precipitation strengthening is weak, Eq. (2) can be rewritten as:

$$\sigma_{yield} = \sigma_i + \sigma_p + \sigma_d + \sigma_s$$  \hfill (3)

The intrinsic strengthening of pure Al is about 10 MPa [14,50], considering the solid solution strengthening contributed from some minor elements (e.g., Si, Cu), the intrinsic strengthening of Al alloys is normally in the range of 43–70 MPa [51,52,55]. Thus, in the present case, $\sigma_i$ is reasonably taken as 53 MPa.

The grain boundary strengthening is given by [14]:

$$\sigma_{gb} = c^*d^{-2/3}$$  \hfill (4)

where $c$ = 0.14 is a constant [56], $d$ is the grain size. From the EBSD results, the grain size ($d$) is 66 $\mu$m. Therefore, $\sigma_{gb}$ is 12.3 MPa.

The dislocation strengthening is expressed as [53]:

$$\sigma_p = M^*\tau^*|b|_G^*c^1$$  \hfill (5)

where $M$ is the Taylor factors whose value is taken as 3 in this study, $\tau$ = 0.5 is a constant [57], $|b|$ = 0.3 nm the magnitude of, Burgers vector, $G$ = 27 GPa the shear modulus in Al alloys. The dislocation density $\rho$ is $1.8 \times 10^{12}$ m$^{-2}$ obtained from EBSD results. Taking these values into Eq. (5), the $\sigma_p$ is obtained as 5.43 MPa.

Since only Mg is found in solid solution areas in FZ, as confirmed from EDS results, the solid solution strengthening from only Mg is considered, and is estimated by [57]:

$$\sigma_s = K^*C^{2/3}$$  \hfill (6)

where $K$ = 29 MPa( wt%$^{2/3}$) is a constant [57], $C$ is the concentration of Mg, which is 1.12 wt% from EDS results. $\sigma_s$ is obtained accordingly as 31 MPa.

Substitution of the values of each aforementioned strengthening into Eq. (2), $\sigma_{yield}$ is solved as 112 MPa, which is close to the experimental value (107 ± 7 MPa). This implies that the model described in Eq. (2) provides an acceptable estimation of the yield strength of FZ in AA5–7 joints.

Once the yield strength is determined, a quasi-static strength model is built to simulate the strain-rate hardening and work hardening behavior of FZ. Flow stress in a metallic material can be expressed as:

$$\sigma_f = \sigma_{yield} \left( \frac{\varepsilon_f}{\varepsilon_0} \right)^n$$  \hfill (7)

where $\sigma_f$ is the flow stress, $\sigma_{yield}$ is the calculated yield stress (i.e. 112 MPa) from Eq. 2, $\varepsilon_f$ is the applied strain, $\varepsilon_0$ is the reference strain, $m$ is the strain hardening exponent. By fitting the experimental results of local area tensile testing, the optimal value of $\varepsilon_0$ and $m$ are obtained as 0.026 and 0.22, respectively. The fitted stress-strain curves from the quasi-static strength model is shown in Fig. 15, which demonstrates that the mechanical behavior of FZ is well simulated by the present strength model. It is worth to note that there is a mismatchment in elastic period between experiments and modelling, as depicted in Fig. 15. This mismatchment is also reported in other studies [8,58]. They argued that the instrumental factors (e.g. the resolution of measuring displacement, machine compliance) would have an influence on obtaining accurate stain of early deformation. However, the yield stress, and hardening behavior are well simulated by this model.

4.2. Evolution of microstructures and mechanical properties

The microstructure in each zone of AA5–7 joints is influenced by the zone’s maximum temperature during welding. As shown in Fig. 5–7, large grains with coarsened phases are found in FZ as materials in FZ experience quick solidification during welding. Besides, some strengthening elements (Mg and Zn) are vaporized in welding process [31]. According to the strength model presented in section 4.2, large grain size and the loss of strengthening elements can cause the reduction of the strengthening contribution from grain boundary and solid solution strengthening. Consequently, the strength in FZ is deteriorated.

Materials in HAZ-7 are subjected to a maximum temperature of 600 °C which declines quickly with the distance from the fusion line increasing [59]. Generally, $\eta'$ and $\eta''$ phases start to dissolve and grow larger if the temperature is higher than 380 °C [28], thus the micro-structure of HAZ-7 is presented by lower-density and coarsened phases, as evidenced in TEM characterizations in Refs. [42,59]. It is noted that, from EBSD results the grain size in HAZ-7 seems to be the same as that of BM AA7N01, therefore, the worsened strength of HAZ-7 is mainly attributed to the change of precipitates.

Though materials in HAZ-5 may also experience high annealing temperature in welding, such material is solid-solution strengthened, meaning that the change of precipitates have minimum or negligible effect on worsening mechanical properties. Considering that the strengthening elements of HAZ-5 is unable to lose during welding, the change of strength in HAZ-5 is thus only determined by the change of grain size and dislocation density. EBSD results show that the grain size in HAZ-5 is slightly enlarged compared to that in BM AA5083, whereas the dislocation density almost keeps unchanged. Therefore, the slightly reduced strength of HAZ-5 is due to the coarsening of grain size.

5. Conclusions

In this study, hybrid laser-arc welding is used to join two dissimilar Al alloys, AA5083 and AA7N01. The microstructures, mechanical and fatigue properties of the welded joint are investigated via experiments and theoretical modelling. Following summaries can be made from discussions above:

(1) Laser-arc welding is successfully applied to join dissimilar Al alloys with negligible or even no defects of macro pores and solidification cracks in the joints.

(2) Microstructure characterization show that FZ features large precipitates, low dislocation density, and coarsened grains with a size of 66 ± 57 $\mu$m.
(3) The tensile strength of AA5−7 joints is 274 MPa, a value close to the tensile strength of AA5−5 joints (273 MPa). The results from local area tensile testing show that the FZ has the lowest strength in AA5−7 joints, a finding which is consistent with that from micro-hardness testing. The fatigue strength of AA5−7 joints is 110 MPa, about 40% of their tensile strength. Pores and inclusions are the main reason for the deterioration of the fatigue strength.

(4) Based on the strengthening mechanism, a strength model is effectively built and used to predict the yield strength of FZ, with the predicted strength being 112 MPa which is close to the experimental value (107 ± 7 MPa).

(5) A quasi-static strength model is built and applied to precisely simulate the strain rate hardening and work hardening behavior of FZ.

(6) AA7N01 and its weldment are prone to corrosion attack, which can seriously worsen the service-life and reliability of the welded structure, future work can be conducted to investigate the performance of laser-arc welded AA5083-AA7N01 joints in corrosive environment.

6. Data availability
All data has been displayed in this paper.

Declaration of Competing Interest
None.

Acknowledgments
This project was funded by China Postdoctoral Science Foundation (Grant No. 2019M661269). The authors acknowledge the facilities, and the scientific and technical assistance of the Australian Microscopy & Microanalysis Research Facility at the Centre of Advanced Microscopy.

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