Original Article

Microstructural evolution and mechanical properties of refill friction stir spot welded alclad 2A12-T4 aluminum alloy

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1. Introduction

RFSSW is a new solid-state joining process invented and patented by GKSS-GmbH in 2005 aiming to eliminate the keyhole formed by the conventional FSSW [1]. The keyhole inevitably remains at the center of the nugget directly limits the widespread applications of FSSW process in transportation systems, particularly in automotive industry [2]. Firstly, the keyhole is an inherent defect itself, which will cause stress concentration and reduction of the effective connection area of the spot weld which in turn compromises the tensile properties of the joint [3]. Meanwhile, corrosion could take place preferentially at the keyhole because of rainwater remaining in the keyhole, where body paint barely reaches the bottom [4]. As a derivative of conventional FSSW, RFSSW has distinct advantage over conventional FSSW since the usually...
observed limitations for losing of effective bearing area and corrosion problems associated with presence of the keyhole have been eliminated [5]. Other advantages of this solid-state joining process are the absence of fusion related defects such as welding porosity, hot cracking and oxide inclusion [6]. For instance, resistance spot welding of high strength aluminum alloys (2XXX series and 7XXX series) is limited by the oxide inclusion that with higher melting temperatures than the substrate. In such situations, higher heat input is needed to melt down the oxide layer into the substrate. However, this also leads to further thermal softening of the alloys due to degradation of the hardening precipitates or even evaporation of the solution elements [7]. What is more, hydrogen cracking may occur during the consolidation process [8]. In comparison with the resistance spot welding or laser spot welding, RFSSW also consumes less energy but owns better surface finishing which leads to improved mechanical properties [9].

RFSSW process is finished by coordinated motion of a pin, a sleeve and a clamping ring. There are two process variants, i.e., the pin plunge and the sleeve plunge operation modus [10]. In both modus the plates are firstly fixed together by the clamping ring against the back anvill and both sleeve and pin start to rotate at a pre-set rotational speed to generate frictional heat on the upper plate surface. Following that, either the pin or the sleeve plunges into the plates, leading to formation of a volume of plasticized metal around the rotational component. In the pin plunge modus, the sleeve is concomitantly retracted backwards during the pin plunging to create a cavity to accommodate the plasticized metal displaced by the pin. After the pre-set plunge depth is reached the process is reversed, where both the sleeve and pin retract back to their original position. Sleeve retraction forces the plasticized metal entrapped in the sleeve cavity to refill back the hole left by the pin. By the end of the welding process the tool is retracted, leaving a flat surface connection with minimum material loss. In the sleeve plunge modus, the cavity is resultant from the pin retraction while the sleeve is forced into the plates. The following steps are equal to the pin plunge modus. Compared with the pin plunge modus, sleeve plunge RFSSW fabricates welds with larger bonding area and higher strength. However, it demands a bigger plunge force at the plunging stage.

The above actions induced severe plastic deformation and significant level of frictional heat, resulting in microstructure evolution throughout the joint and formation of macrostructure characteristics of bonding ligament and hook [10]. A number of studies have undertaken to investigate the grain structure evolution, macrostructure characteristics and mechanical properties of the RFSSW joint [9–21]. The main aluminum alloys investigated so far are AA2024 [11,12], AA5052 [13], AA5754 [14], AA6061 [15–17], and AA7075 [18–21] alloys. The region experiences an exceptionally high level of strain is termed as stir zone (SZ), where the grains are found to be very fine (e.g., <10 µm) [13]. Within the SZ, the retained interface is termed as bonding ligament. The studies relating the bonding ligament and tensile-shear properties [11,12,19] reported that the bonding ligament is a weak-bonded region at the lap interface. A continuous bonding ligament tends to compromise the tensile-shear properties of the RFSSW joint by providing an easier path for annular crack propagation and leads to separation of the plates by a shear mode. Besides, hook is a macrostructure feature characterized by deviation of the original lap interface, which is generally accepted to be a crucial feature affecting the tensile-shear properties of the joint since it can reduce the effective bonding area of the plate, diminish the integrity of the joint and boost crack nucleation [12–18]. For alclad aluminum alloys, tensile properties of the RFSSW joints are also dependent on the distribution of the alclad. The previous investigations [19,21] reported that deep plunge tends to make the alclad more dispersed within the SZ, consequently, resulting in favorable joints.

The precipitates and substructure distribution are important concerns to heat-treatable aluminum alloys, which play important role in affecting the hardness, corrosion susceptibility and various other material properties [22]. However, precipitates along with substructure evolution in aluminum alloys due to RFSSW is yet to be elucidated in open literature. There have been some literatures investigating the substructure development and precipitates evolution during friction stir welding (FSW) of heat-treatable aluminum alloys [23–28]. It was reported that welding parameters strongly influence the temperature history during FSW and the precipitates in heat-treatable aluminum alloys can coarsen or dissolve into the aluminum matrix depending on alloy type and maximum temperature [23,24]. As is known, the mechanical properties of the Al–Cu–Mg alloys are significantly depended on the metastable precursors of the equilibrium S (Al2CuMg) phases. The precipitation behavior of Al–Cu–Mg alloys is generally reported to be: supersaturated solid solution (SSS) → Cu–Mg co-clusters → S’ → S” (S) [25–28]. Gerlich et al. [25] examined the precipitates evolution of a AA2024 aluminum alloy during FSW using TEM. No precipitates were observed within the SZ, indicating that all the precipitates were dissolved into the α-Al matrix during FSW. Genevos et al. [26] obtained similar results from TEM characterization of AA2024-T351 FSW seam. Mohammadtaheri et al. [27] reported that the SZ experienced the highest peak temperature and the metastable precipitates S’ dissolved into the α-Al matrix during FSW of AA2024-T351 aluminum alloy. Yang et al. [28] reported the presence of low level substructure in the SZ and high level substructure in the thermo-mechanically affect zone (TMAZ) in AA2524-T351 aluminum alloy FSW joints.

A comprehensive understanding of microstructure evolution during RFSSW is necessary if the resultant microstructure and associated mechanical properties are to be optimized. In the present study, RFSSW is performed for 2A12-T4 aluminum alloy at various rotational speeds. The aim is to investigate microstructure evolution and correlate different microstructure to microhardness and tensile-shear properties of the joint.

2. Experimental procedure

Alclad 2A12-T4 aluminum alloy plates with a dimension of 80 mm × 30 mm × 2 mm were chosen as base metal, surfaces of these plates were covered with alclad layer to protect the inner alloy from corrosion. 2A12-T4 alloy is a ternary Al–Cu–Mg alloy with good tensile strength and medium corrosion resistance that is widely used in aviation industry for
Table 1 – Chemical compositions and mechanical properties of 2A12-T4 Al alloy.

<table>
<thead>
<tr>
<th>Chemical compositions (wt.%)</th>
<th>Mechanical properties</th>
</tr>
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<tbody>
<tr>
<td>Cu</td>
<td>Mg</td>
</tr>
<tr>
<td>4.66</td>
<td>1.69</td>
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Fuselages and other aircraft structures, whose chemical compositions and mechanical properties are listed in Table 1. Prior to welding, all plates were thoroughly milled by SiC papers and then cleaned with acetone to wipe off the oxidation layer. The plates were overlaid with an area of 30 mm × 30 mm and the welding was conducted at the center of the overlaid area, as shown in Fig. 1. The RFSSW tool consists of three independent components, i.e., a 17 mm diameter clamping ring, a 9 mm diameter sleeve and a 5.3 mm diameter pin. The welding process was performed on a self-developed RFSSW machine. In the present study, a sleeve plunge mode was adopted. Constant sleeve plunge depth of 2.5 mm and sleeve plunge rate of 1 mm/s were applied. The tool rotational speeds were 900, 1100, 1300, 1500 and 1700 rpm, respectively. The thermal cycles were determined by the K-type thermocouples that were fixed at the mid-plane of the upper plate, 6 mm away from weld center (located in the heat affected zone (HAZ)).

After RFSSW, the welding samples were cross-sectioned for metallographic analyses and mechanical tests. The specimens for microstructure examination were cold mounted, grinded with SiC papers to 2000 grit, polished using a diamond paste to a 0.05 μm finish and etched with Keller reagent (2.5 ml nitric acid, 1.5 ml hydrochloric acid, 1 ml hydrofluoric acid and 100 ml distilled water). Olympus GX51 optical microscopy (OM) along with Zeiss-MERLIN Compact scanning electron microscope (SEM) were used for macro/microstructure characterization. Foil disks of 3 mm in diameter for TEM were sectioned from the BM, HAZ, TMZ and SZ, subsequently electropolished using a twin-jet polisher containing a solution of 30% nitric acid in methanol at −30 °C and 10 V. Thin foils were analyzed by a JEM-2100 operating at 200 kV. The TEM image analyses were conducted on [001] Al zone axis orientation. Microhardness profiles were measured at the mid-thickness of the upper plate on the polished cross-sections by a microhardness tester (MICRO-S86). The spacing between the adjacent indentations was 0.5 mm, and the testing load was 1.96 N for 10 s. The tensile specimens were prepared by adding spacers 2 mm in thickness to each end of the welding sample, as shown in Fig. 1. Tensile-shear tests were performed in triplicate for each welding parameter using an Instron-1186 mechanical tester with a crosshead displacement speed of 1 mm/min.

![Fig. 1 – Schematic diagram for welding spot.](image)

3. Results and discussion

3.1. Thermal history

During RFSSW of heat-treatable aluminum alloys, heat input control is of great importance from several aspects such as formed macrostructure characteristics (i.e., the hook and bonding ligament) and evolved precipitates. Furthermore, investigation of thermal cycles enables welding engineers to obtain a basic correlation between processing parameters and peak temperatures experienced by the workpieces. According to Schmidt et al. [29], the friction-induced heat generation in conventional friction stir welding can be estimated by the following model:

\[ dQ = \omega dM = \omega r dF = \omega \pi r dA \]

where \( dQ \) – heat input of surface element (W); \( \omega \) – angular speed (rad/s); \( r \) – radius (m); \( M \) – torque (Nm); \( F \) – downward force (N); \( r \) – shear stress on the contact surface (Pa), which is a function of temperature and can be estimated from the material yield strength at \( T \) (i.e., \( r = \sigma_y(T)/\sqrt{3} \)); \( dA \) – area of surface element (m²). According to this model and the RFSSW tool geometry (see Fig. 2a), the heat generation between tip of the pin and the plasticized metal (\( Q_1 \)), between tip of the sleeve and the plasticized metal (\( Q_2 \)), and between sides of the sleeve and the plasticized metal (\( Q_3 + Q_4 \)) can be estimated. The total friction-induced heat input (\( Q_{total} \)) during RFSSW including heat from all the four interfaces and therefore can be estimated by the following formula:

\[ Q_{total} = \frac{2}{3} \pi r^2 \omega \left( r_{sleeve}^2 + 3r_{pin}^2(h_p + h_t) + r_{sleeve}^2 h_p \right) \]

where \( r_{sleeve}, r_{pin}, h_p \) and \( h_t \) are the radius of sleeve, radius of pin, sleeve plunge depth and pin retraction depth, respectively, as indicated in Fig. 2a. Using this relationship, it can be deduced that the increase in tool rotational speed causes an increase in heat input and thus a higher peak temperature, which agrees well with the results presented in Fig. 2b. Peak temperature recorded by the thermocouples increases from 314.9 °C to 407.3 °C as the tool rotational speed increases.

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from 900 to 1700 rpm. Meanwhile, the exposure time at elevated temperature (typically over 150 °C) extends from 12.08 to 15.67 s with tool rotational speed increasing from 900 to 1700 rpm. The severer thermal exposure at high rotational speed can lead to a higher precipitates deterioration level and formation of a softened zone within the weld, which will be discussed later. It is worth noting that the processing temperature below the sleeve and pin should be higher than 407.3 °C, where the exact temperature cannot be captured easily by the thermocouples since the RFSSW tool tends to destroy the thermocouples during welding. Experimentally captured or numerically simulated peak temperature, reported in Refs. [25,30,31], in the SZ during friction stir seam welding of AA2024 aluminum alloy (equivalent to Chinese grade Al 2A12) ranges from 433 °C to 515 °C. It is noted that the highest possible peak temperature in the SZ remains around the solidus temperature of AA2024 aluminum alloy (about 522 °C). In FSW, as temperature approaches the solidus temperature, shear strength of the plasticized metal decreases significantly, and incipient melting may also take place. Therefore, a drop in heat generation efficiency occurs and further temperature rise will not take place.

3.2. Microstructure characterization

Cross-sections of the RFSSW joints obtained at the various tool rotational speeds are presented in Fig. 3. It was found that the keyholes were refilled successfully in all the conditions. Basing on the cross-section characteristic, three distinct zones, i.e., the SZ, TMAZ and HAZ can be discerned, as shown in Fig. 3d. Among them, the SZ can be further divided into the pin stir zone (P-SZ) and sleeve stir zone (S-SZ). At the low rotational speeds, a continuous alclad layer retained in the P-SZ at the level of the plates bonding interface. The retained alclad layer (also termed as bonding ligament) is composed of pure aluminum, which tends to provide an easier path for crack propagation under tensile-shear loading due to its poor strength and should be classified as a weak-bonded region of the lap joint [21]. The length of the alclad layer is defined as \( H_1 \), which will surely affect the tensile-shear properties of the joint. That is, the larger \( H_1 \) is, the smaller effective bonding length at the lap interface and the lower tensile-shear properties. Hook is a morphology feature formed near the SZ/TMAZ interface by the bending of the primary interface due to shear action of the sleeve and squeezing of the SZ metal during the refill stage [12–14,16]. The height of hook \( (H_2) \) is defined as the vertical distance between the hook tip and the original interface, as indicated in Fig. 3d. A more sophisticated characterization of the hook interface is shown in Fig. 4. It can be figured out that the hook interface is characterized by the combination of two regions, i.e., the partially-bonded region with the original sheets interface partially interrupted and the metallurgical bonding partially generated (see Fig. 4b) and the unbonded region (see Fig. 4c). The bonding state at the hook interface changes from unbonded to partially-bonded progressively close to the SZ periphery where the thermal effect and deformation rate are more sufficient to break the original sheets interface and generate metallurgical bonding. Existence of the hook will reduce the effective thickness of the upper plate as well as induce stress concentration due to the notch effect [14]. The above factors make hook a crack sensitive region during the tensile-shear test. Therefore, the hook along with the alclad layer plays an important role in affecting the shear tensile properties of the RFSSW joint.

The effect of tool rotational speed on geometric parameters of the alclad layer and hook are quantitatively analyzed with the results summarized in Fig. 5. \( H_1 \) decreases with increasing tool rotational speed as the more sufficient stir effect under higher tool rotational speed leads to further dispersion of the alclad. The geometric morphology of the hook is related to the material deformation behavior during RFSSW. When the tool rotational speed is low, only a small amount of material of the TMAZ is fully plasticized due to the low heat input, and the vertical flow of the plasticized material is relatively weak, resulting in a low \( H_2 \). When the tool rotational speed is increased, more material can be fully plasticized and the
Fig. 3 – Cross-sections of the joints at different tool rotational speeds: (a) 900 rpm; (b) 1100 rpm; (c) 1300 rpm; (d) 1500 rpm; (e) 1700 rpm.

Fig. 4 – Interface characteristic of hook: (a) an overview; (b) the weak-bonded region; (c) the unbounded region.
upward flow of the plasticized metal is enhanced. Therefore, the H_2 is elevated significantly. While when the heat input
beyond a threshold value, further increase in tool rotational
speed will lead to a decrease in the plasticized material vis-
cosity and thus slipping between the tool and the material,
which is a common phenomenon called as stick/slip transition
reported by many researchers [19]. Consequently, when
the tool rotational speed further increases to 1700 rpm,
the impact imposed on the material deformation within the TMAZ
from the tool becomes weaker, resulting in a reduced H_2.

OM images in the different metallurgical zones are pre-
sented in Fig. 6. The 2A12-T4 aluminum alloy with a 60 μm
thick alclad layer used in the present study exhibits a typi-
cal hot rolled structure with the grains elongated remarkably
along the rolling direction (Fig. 6a and b), and a majority of
the fine particles that mainly precipitated along the grain
boundaries together with a few coarser dispersoid type par-
ticles are also discernable. According to Ref. [27], the latter
are thought to be the base metal constitute particles that are
hardly affected by the thermal cycles. The HAZ is far away
from the weld nugget and is only affected by thermal cycling,
and the grain structure in the HAZ is similar to that of the
BM, as shown in Fig. 6c. The TMAZ grain structure seems to be
twisted upward as a result of the deformation induced by the
shear action of the tool (Fig. 6d). As shown in Fig. 6e and f, the
SZ is characterized by the fine equiaxed grains in comparison
to the BM. The SZ located in the center is directly stirred by the
sleeve and pin, where the deformation rate and temperature
are high enough to induce dynamic recrystallization (a typical
metallurgical phenomenon formed in friction-based welding
processes) followed by evolution of a fine grained microstruc-
ture [11]. Furthermore, it is found that the grains of the sleeve
stir zone are finer than those of the pin stir zone, which can
be attributed to the higher strain rate and the resultant higher
recrystallization nucleation rate in that zone. On the other
hand, the prior intergranular particles are refined to a much
finer scale in the SZ owing to the intense stir/milling effect
exerted by the tool.

Grain evolution in the SZ and TMAZ under the different
tool rotational speeds is shown in Fig. 7. As the tool rotational
speed increases from 900 to 1700 rpm, the SZ exhibits an aver-
age grain size of 4.32, 5.33, 6.54, 6.88 and 7.20 μm, respectively,
which are plotted in Fig. 8a. For a number of aluminum alloys,
the recrystallized grain size, d_{rec}, and the peak temperature
are related as follows [25]:

\[ d_{rec}^{-1} = a + b \ln Z = a + b \left( \ln \frac{\sigma R_{sz}}{H_{sz}} + \frac{Q}{RT} \right) \]  

where a and b are constitutive constants; Z denotes
the Zener–Hollomon parameter; Q denotes the activation energy
for lattice diffusion (144 kJ mol\(^{-1}\)) for Al 2A12 [32]; T denotes
the absolute temperature in the SZ; R denotes the gas con-
stant; R_{sz} and H_{sz} denote the average radius and depth of the
recrystallized zone, respectively. R_{sz} is assumed to equal to
0.78 of the observed SZ boundary radius (based on Ref. [25]).
In the present case, R_{sz} and H_{sz} are determined to be around
3.6 mm and 2.9 mm, respectively. The constitutive constants
employed in all the calculations are those reported in Ref.
[27], namely a = −0.8417 and b = −0.0235. The absolute tem-
peratures in the SZ calculated by Eqs. (3) at the various tool
rotational speeds are plotted in Fig. 8b. It is observed that
the calculated temperature in the SZ is comparable to those
experimentally captured or numerically simulated peak tem-
perature reported in Refs. [25–28] and the SZ temperature
increases with increasing tool rotational speed. The higher
temperature provides more energy for grain growth and eventu-
ally leads to coarsening of the recrystallized grains in the SZ
(see Fig. 7a and c). In the TMAZ, grains become coarser at the
higher rotational speed for the same reason (see Fig. 7b and
d).

3.3. Precipitate and substructure

TEM characterizations of the BM are presented in Fig. 9a and
b. Plenty of fine precipitates with a length of 60–110 nm and
a width of 8–13 nm are discernable. The [001]\(_{Al}\) selected area
diffraction (SAD) patterns encompassing these fine precipi-
tates show faint reflections at the S\(^{\ast}\) positions, indicating that
such precipitates are the orthorhombic S\(^{\ast}\) which has a semi-
coherent relationship with the matrix. Moreover, few coarse
precipitates with a length up to 500 nm are distributed in the
BM. According to the EDS measurements in Fig. 9b, such par-
ticipates consist of Al, Cu, Mn and Mg elements, which can be
judged to be the residual phases that are stable under ele-
terature and have no strengthening effect on the matrix.
The precipitate evolution in the SZ, TMAZ and HAZ at
the different tool rotational speeds is noticeable. The HAZ is
characterized by the densely distributed needle-shaped pre-
cipitates (Fig. 9c and d). The [001]\(_{Al}\) SAD patterns obtained from
these precipitates revealed reflections at the S position,
indicating that the needle-shaped precipitates are S. With
increasing of the rotational speed, distribution density of the
S is reduced significantly due to the higher thermal exposure.
In the TMAZ, the amount of S precipitates decreased obvi-
ously in comparison with the HAZ, accompanied by which is
a precipitate morphology change from the needle-shaped to
lath-shaped (Fig. 9e and f), indicating dissolution and coarsen-
ing of S have occurred. As rotational speed increases from 900
to 1700 rpm, the dissolution of S is accelerated, which leads to
further decrease in distribution density of the S precipitates. In
the SZ, as presented in Fig. 9g and h, some coarse rod-shaped residual phases but no $S''$ and $S$ are observed.

The precipitation strengthening phase in Al–Cu–Mg alloys depends on the atomic ratio of Cu to Mg, the Cu:Mg atomic ratio of the present alloy is 2.75, hence the equilibrium phase is the orthorhombic $S$ with a composition of $\text{Al}_2\text{CuMg}$ [33,34]. The precipitation sequence of the present alloy is SSS (supersaturated solid solution) $\rightarrow$ Cu–Mg co-clusters $\rightarrow S' \rightarrow S'(S)$ [25]. As $S'$ have the same crystal structure as the $S$ with a slight difference in their lattice parameters, it is widely accepted that there is no distinction between the $S'$ and $S$ phases [35]. To help explain the precipitate evolution mechanism in the various regions of the RFFSW joint, computed vertical section of Al–Cu–Mg phase diagram at 95.5 mass% Al (based on Fig. 7 in Ref. [37]) superposed with the temperatures in the different zones of the RFFSW joint is presented in Fig. 10.

The BM plates used in the welding trials were processed by water quenching after solution treatment at 495°C for 30 min,
followed by natural aging for several months. With reference to Fig. 10, it can be noted that the elevated temperature during solution treatment can lead to dissolution of the equilibrium $S$ and its metastable precursors into the $\alpha$-Al matrix while precipitation of the equilibrium $S$ phase does not occur during the subsequent natural aging process due to the low temperature. Therefore, only metastable $S''$ phases are resolved from the Cu–Mg co-clusters, as revealed in the TEM observation of Fig. 9a where the BM presents plenty of $S''$ precipitates. From the detected peak temperature within the HAZ, a high tem-
temperature from 314.9 to 407.3 had occurred during the RFSSW process. The elevated temperature within the HAZ reaches that required for the S phase precipitation to start (260–320) and leads to the dense precipitation of S phase (see Fig. 9c and d) which is retarded by the low temperature of natural aging. The TMAZ is closer to the weld center and is subjected to severer thermal exposure. Therefore, overaging and coarsening of the S phase occurs within the TMAZ, leading to the reduced distribution density of the S precipitates as indicated in Fig. 9e and f. The SZ in the center of the weld undergoes the most severe thermal exposure, where the peak temperature has been calculated to lie in the range of 461.8–525.9 °C during the welding thermal cycle. Such a high temperature has exceeded the solution temperature of S, resulting in the

Fig. 9 – Precipitates evolution under different tool rotational speeds: (a) BM; (b) EDS of the dispersoids in the BM; (c) HAZ-900 rpm; (d) HAZ-1700 rpm; (e) TMAZ-900 rpm; (f) TMAZ-1700 rpm; (g) SZ-900 rpm; (h) SZ-1700 rpm.
diminishing of S and its metastable precursors as revealed in Fig. 9g and h.

The joint welded at the high rotational speed (i.e., 1700 rpm) are found to contain a black band pattern distributed along the streamline formed by the material flow during the weld stage, as shown in Fig. 11a. At higher magnification, some microcracks run along the crystal boundaries are discernable in the band pattern (Fig. 11b). SEM image in Fig. 11c reveals the presence of densely distributed bright particles in the micro-crack, which are identified as Al–Cu–Mg based according to the EDS measurements. This is seen to be microstructure evidence of liquation and microcracks in the band pattern are thus believed to be solidified defects. Although RFSSW is a solid-state welding process with the processing temperature lower than the melting point of the base metal, the high temperature in the SZ may cause localized melting of some precipitates and eutectics that have a low melting point [36,38]. The calculated temperature in the SZ under the high rotational speed has exceeded the melting point of S (see Fig. 10), hence it can be deduced that localized melting of S has occurred in the SZ. The liquated S gathers to the grain boundaries of the recrystallized grains, resulting in formation of liquid films around the grains. During the subsequent cooling process, micro-cracks initiate and propagate along the liquated grain boundaries under the tensile stress, and liquation cracks occur.

Besides the precipitates, dislocation distribution also plays a role in the strengthening of the Al–Cu–Mg alloy. Evolution of the dislocation configuration is presented in Fig. 12. Grains in the SZ are found to contain high density of dislocations (Fig. 12a). Material in the SZ experiences severe plastic deformation due to the intense stir action of the tool, leading to dislocation multiplication in the grains. As tool rotational speed increases, an obvious decrease in dislocation density achieved by rearrangement of the dislocations by polygonization can be observed in Fig. 12b. This result indicates that recovery has occurred in the SZ under the higher heat input. The TMAZ presents a significant dislocation density, as shown in Fig. 12c. Material in the TMAZ undergoes the shear effect of the sleeve and squeezing of the SZ metal. The above actions induced plastic deformation and can be responsible for the densely distributed dislocations in the TMAZ grains. With increasing of the tool rotational speed, dislocation density decreases due to further recovery, as shown in Fig. 12d. Grains in the HAZ (see Fig. 12e and f) are found to contain a relatively low density of intragranular dislocations as the HAZ is only affected by the thermal cycling with no new dislocations introduced during the welding process. In addition, no distinct difference in dislocation density is detected at the different tool rotational speeds. A possible explanation is that the densely distributed S precipitates exerted strong pinning action on the dislocation migration, hence the recovery was difficult to occur regardless of the change in heat input.

3.4. Mechanical properties of the joints

Fig. 13 presents the hardness profiles of the joints at different tool rotational speeds. The hardness profiles are either U-shaped or W-shaped depending on the tool rotational speed. For low rotational speed, e.g., 900 rpm, a W-shaped hardness profile is discerned; i.e., the hardness drops progressively before reaching a minimum of 116 Hv near the edge of the SZ, and rises above 134 Hv within the SZ. With increasing tool rotational speeds, e.g., 1500–1700 rpm, the hardness only recovers slightly to a level of 125 Hv within the SZ after reaching a minimum, indicating the formation of U-shaped hardness profiles. It is widely accepted that the thermal softening occurred in the weld zone of heat-treatable aluminum alloys is principally related to the dissolution and coarsening of the strengthening precipitates [15–18]. In the present weld, dissolution and coarsening of the fine S precipitates can be responsible for the hardness softening within the TMAZ. In the SZ, the S precipitates dissolve into the α-Al matrix due to the severe thermal exposure during welding. The dissolution of a sufficient amount of S causes supersaturated solute remaining in the SZ after a rapid cooling process. Therefore, the solution strengthening in the SZ is prominent. The high degree of supersaturation remaining along with the fine recrystallized grains and the high density of dislocations contribute to the hardness recovery within the SZ. At higher rotational speeds, the exposure time at the elevated temperature increases, resulting in coarsening of the recrystallized grains as well as dislocation rearrangement/recovery. Consequently, the hardness within the SZ drops.

Tensile-shear test results and the corresponding fracture positions of the RFSSW joints welded at the different tool rotational speeds are shown in Figs. 14 and 15, respectively. The tensile-shear failure load (TSFL) firstly increases with increasing tool rotational speed, reaching a maximum value of 10.03 kN at the 1300 rpm, then decreases as tool rotational speed further increases to 1700 rpm. Correspondingly, two kinds of fracture modes are identified, i.e., the shear fracture and shear-plug fracture. At the lowest rotational speed of 900 rpm, the TSFL is considerably low due to the continuously
distributed alclad layer at the lap interface. When the RFSSW joint bears tensile-shear load, an annular crack firstly initiates at the hook tip due to the imposed stress fields around the spot, followed by a crack propagation to the center of the weld nugget along the continuous alclad layer with poor strength. In the end, the upper and lower plates are sheared off from each other, as shown in Fig. 15a. As the rotational speed increases, the alclad at the weld periphery is more dispersed and thus crack propagation during the tensile-shear test cannot follow the alclad layer path as in the case of the shear fracture mode. Therefore, the fracture mode changes from the shear fracture into the shear-plug fracture, as shown in Fig. 15b and e, and the corresponding TSFLs are relatively high. In this case, the crack firstly initiates at the tip of the hook at the stretch face of the spot and propagates along the SZ/TMAZ interface until reaching the upper surface of the joint. Then the crack propagates
around the weld periphery to the extrusion face of the spot. Finally, the crack propagates through the SZ due to rotation of the weld nugget under the tensile-shear load, as shown in Fig. 15f, causing the shear-plug fracture of the RFSSW joint. In the shear-plug fracture mode, the diffusion bonding state at the TMAZ/SZ interface along with the microstructure in the SZ affect the TSFL [17,18]. With increasing of tool rotational speed, diffusion bonding strength at the TMAZ/SZ interface increases due to the higher temperature and longer diffusion time, leading to the increase in TSFL. However, when the rotational speed exceeds 1300 rpm, the effective bonding area at the TMAZ/SZ interface decreases due to the excessive upward bending of the hook; what is more, grain coarsening along with presence of the liquation cracks compromises the tensile properties of the SZ. The above two points can be responsible for the TSFL drop at the high rotational speeds.

4. Conclusions

The following conclusions can be drawn from the experimental results and analyses:

(1) The macrostructure characteristics of the present RFSSW joints are the bonding ligament and hook. A higher
rotational speed makes the distribution of the bonding ligament at the weld periphery more dispersed, which is beneficial for the tensile-shear properties of the joint. The interface of the hook is characterized by a combination of partially-bonded region and unbonded region, which makes it a crack sensitive area under the tensile-shear loading.

(2) The microstructure in the different zones exhibits variations in the grain sizes, precipitates distribution and substructure. A softened region consisted of SZ and TMAZ exists in all the RFSSW joints. Formation of the softened region can be attributed to coarsening and dissolution of the S precipitates. With the increase of the rotational speed, the hardness of the SZ decreases and the hardness profile of the joint changes from the W-shaped to U-shaped in the macroscopic level.

(3) At the high rotational speed of 1700 rpm, liquation cracks associated to the localized melting of S phase are identified in the SZ where the peak temperature has been calculated to reach 525.9 °C during the weld thermal cycle. Such
a high temperature has exceeded the melting temperature of S, which resulted in the localized melting of S and formation of the liqation cracks during the subsequent cooling process.

(4) With increasing the rotational speed from 900 rpm to 1300 rpm, the TSFL constantly increases to the maximum value of 10,030 N. Then it decreases as the rotational speed further increases to 1700 rpm. The shear-plug fracture mode is necessary for a high quality RFSSW joint.

Conflicts of interest
The authors declare no conflicts of interest.

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REFERENCES