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Unraveling Dynamic Recrystallization and Precipitate Re-formation in Friction Stir Welded 2195 Al-Li Alloy: A Pathway to High-Performance Joints

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Abstract: This study systematically investigates the friction stir welding process of 6.5 mm thick 2195 Al-Li alloy, with a focus on the microstructure evolution, precipitate behavior, and mechanical properties of the joints under different welding parameters. The intrinsic relationship between microstructure and mechanical performance is elucidated. The results show that high-quality joints can be obtained at a rotational speed of 450 rpm and a traverse speed of 250 mm/min. Under these conditions, the joint achieves an ultimate tensile strength of 430 MPa and an elongation of 6.38%, corresponding to 74.14% and 75.06% of the base material, respectively. In welding nugget zone, the T₁ phase undergoes dissolution and re-precipitation, accompanied by the presence of T_B and β' phases. The recrystallization mechanism is dominated by continuous dynamic recrystallization in WNZ. The fracture surface of high-quality joint exhibits a mixed morphology of fine dimples and cleavage facets, characteristic of mixed ductile-brittle fracture. This study provides a theoretical basis for process optimization and microstructure control in FSW of medium-thick 2195 Al-Li alloy plates.

Key words: 2195 Al-Li Alloy, Friction Stir Welding, Precipitate formation, Mechanical Properties

1 Introduction

During the development of aerospace technology, lightweighting has always been the focus of structural material research, and it also requires that aerospace components possess excellent comprehensive performance [1]. Fuel tank, as a key structure of the propeller, accounts for 60% of the weight of the rocket body. It not only stores liquid propellant but also bears most of the structural loads, and is the key factor determining the performance of the launch vehicle [2]. Therefore, the lightweight manufacturing technology of the fuel tank is the key to the manufacturing of large-

diameter rockets. One of the main ways to achieve lightweighting is to improve the structural materials. Aluminum alloy possesses excellent low-temperature toughness, mechanical properties, load-bearing capacity and manufacturing performance, and has a moderate cost. It is widely used in the manufacturing of low-temperature propellant tanks [3]. Compared with traditional alloys, Al-Li alloys exhibit more outstanding performance in terms of specific strength and specific modulus, which can achieve a weight reduction of approximately 10%-15% in aerospace structures while also increasing structural stiffness by 15%-20%. Therefore, it enjoys the reputation of being "one of the most promising aerospace metal materials in the future" [4].

Large fuel tanks are usually difficult to form in one piece and must be joined together by welding multiple components. When using traditional fusion welding techniques such as tungsten inert gas (TIG) welding and plasma arc welding (PAW) to weld Al-Li alloys, the weld seams are prone to defects such as hot cracks, and loss of Li elements, which adversely affect the performance of joints [5]. When using brazing, it is necessary to remove the dense oxide film on the surface of the Al-Li alloy, and the welded joint is prone to over-ageing after welding [6].

Friction stir welding (FSW), as a solid-state joining technology, relies on the heat generated by friction to cause plastic deformation in the welded base materials, enabling high-quality connections of aluminum and magnesium alloys [7]. At present, many scholars have conducted extensive research on the friction stir welding process of Al-Li alloys. Song et al. [8] studied the fatigue crack propagation behavior of the friction stir welded joints of 2195 Al-Li alloy, found that the fatigue life of the weld seam on the retreating side (RS) was generally longer than that on the advancing side (AS) and the direction of the crack would significantly affect the fracture mode. Liu et al. [9] analyzed the recrystallization mechanism of the welding nugget zone (WNZ) in Al-Cu-Li alloy under complete annealing conditions, and concluded that the fine equiaxed crystals were formed through continuous dynamic recrystallization (CDRX) and discontinuous dynamic recrystallization (DDRX). In addition, some researchers have also employed other forms of FSW techniques. Avettand Fènoël et al. [10] investigated the effects of pre-welding and post-welding heat treatments on microstructure of the stationary shoulder FSW (SSFSW) joint of AA2050 Al-Li alloy, and found that the shoulder affected zone presented a complex texture composition, while the stir pin affected zone exhibited a single fiber texture form that was close to $\langle 111 \rangle$. Goebel et al. [11] successfully achieved butt welding of 3 mm thick AA2198-T851 Al-Li alloy using semi-stationary upper shoulder bobbin tool friction stir welding (SS_UBT-FSW) process, and found that tensile strength of the joint reached 82% of the base material. However, regarding the process of microstructure evolution during FSW process of Al-Li alloys, the current research is still not deep enough.

2195 Al-Li alloy is an Al-Cu-Li-Mg-Ag alloy, the main precipitates include T_1 (Al_2CuLi), θ' (Al_2Cu), β' (Al_3Zr), and δ' (Al_3Li), etc [12]. Among them, T_1 phase is in the form of thin sheets and precipitates along the $\{111\}$ lattice plane of the aluminum matrix. It hinders the movement of dislocations and thereby inhibits coplanar slip, so it is regarded as the phase with the most significant strengthening effect [13]. Generally, precipitates in base materials may undergo processes such as dissolution, coarsening, precipitation and re-precipitation during FSW. There are many factors will affect the evolution of precipitates, such as welding thermal cycle, flow stress, strain rate, and so on [14]. However, after FSW process, whether the performance of the joint is still related to the T_1 phase remains a matter of much debate. Some researchers reported that there was no T_1 phase in the weld zone, while other researchers proved the existence of the T_1 phase. Chen et al. [14] used different process parameters to weld 7.5-mm-thick 2195-T8 alloy, and found that under high rotational speed conditions, there was T_1 phase in shoulder affected zone (SAZ), but it disappeared in pin affected zone (PAZ). Tao et al. [15] studied the distribution of precipitated phase in FSW joints of 2060 alloy plates with a thickness of 2 mm, and found that the T_1 , S' , and δ'/β' phase were completely dissolved in WNZ, and only δ'/β' phase precipitated and dissolved after welding. Gao et.al [16] observe T_1 phase at the bottom of the WNZ, and they believe that the high cooling rate at the bottom of the 12mm-thick plate and the dissolution of precipitates, which generate supersaturated solute atoms, provide sufficient kinetic conditions for the formation of T_1 phase with quasi-lattice structure. In conclusion, the distribution and formation process of precipitates in FSW joint of Al-Li alloy warrant further exploration.

This paper conducts a study on FSW process for 2195 Al-Li alloy with a thickness of 6.5mm. The microstructure evolution process and precipitates formation behavior in the joint are systematically investigated. Additionally, mechanical property tests are carried out on the joint to determine the influence laws of process parameters on the joint performance, and to explain the intrinsic relationship between it and microstructure. This research aims to reveal the intrinsic mechanism of the FSW process for Al-Li alloys, providing a theoretical basis for the manufacturing of large-scale structural components.

2 Experimental procedures

2.1 Experimental material and equipment

Base material selected for this project is 2195 Al-Li alloy. 2195 Al-Li alloy has the advantages of low density, high specific strength, high specific modulus and good low-temperature performance, so it is regarded as one of the main structural materials in aerospace field in the future [17]. The

chemical composition and mechanical properties are shown in Table 1 and 2 respectively. The sheet size is 125 mm × 75 mm × 6.5 mm.

Table 1 Chemical composition of 2195 Al-Li alloy (wt. %)

Cu	Li	Mg	Ag	Zr	Mn	Zn	Si	Al
4.0	1.0	0.38	0.42	0.12	≤0.25	≤0.25	≤0.12	Bal.

Table 2 Mechanical properties of 2195 Al-Li alloy

Tensile strength (MPa)	Yield strength (MPa)	Microhardness (HV)	Elongation (%)
580	530	170	8.5

Following the design of the welding tools used by Gao [16], the FSW tool was fabricated by H13 steel, stirring pin adopts a circular truncated cone form and machined with threads and triple facets topologies. The diameter of the shoulder, constrained end and free end of the stirring needle is 22 mm, 8.0 mm and 5.0 mm, respectively. The length of stirring pin is 6.25 mm. The surface of the shoulder is an inward-convex shape, with an inward-convex angle of 10 °, which is designed to expand the space for accommodating the softened materials that heat up below the shoulder and prevent material overflow. According to the relationship with the rolling direction (RD), the welding reference frame is welding direction (WD, along the RD), traverse direction (TD, perpendicular to RD), and normal direction (ND, perpendicular to the horizontal plane). The schematic diagram of FSW process is shown in Fig. 1.

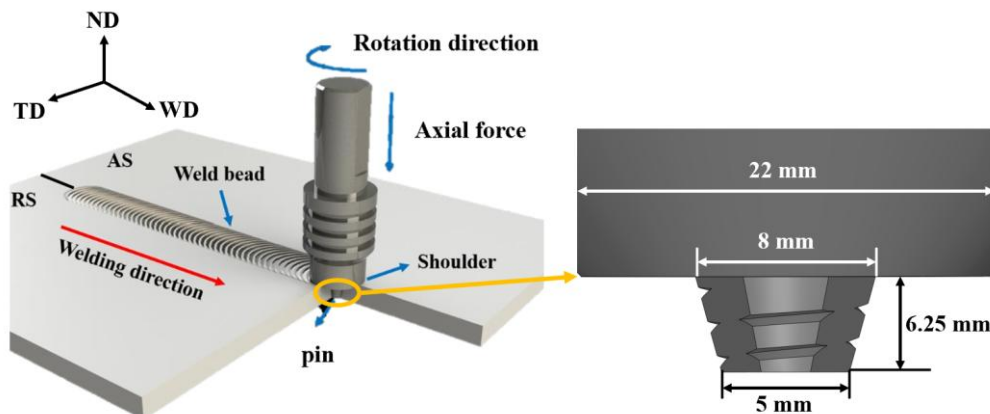


Fig. 1 Schematic diagram of FSW process

2.2 Experimental methodology

Before welding, the author first removed oxide on surface of the base material. After conducting preliminary experiments and explorations, we finally determined the process parameters for this project, as shown in Table 3. This paper mainly investigates the influence of traverse speed and rotational speed on joint formation, so plunging depth and tilt angle remain unchanged.

Table 3 Welding process parameters

Number	Rotational speed (rpm)	Traverse speed (mm/min)	Plunging depth (mm)	Tilt angle (°)
1#	450	150	0.1	2.5
2#	450	200	0.1	2.5
3#	450	250	0.1	2.5
4#	450	300	0.1	2.5
5#	350	250	0.1	2.5
6#	400	250	0.1	2.5
7#	500	250	0.1	2.5

After the welding test was completed, joints were cut into metallographic samples by electric spark cutting equipment, then grinding with abrasive paper and polishing with diamond polishing agent to mirror finish. The cross section was etched with a Keller's reagent (2.5 mL HNO₃ + 2 mL HCl + 1.5 mL HF + 95 mL H₂O) at room temperature for 20 ~ 30 s to reveal the microstructure for inspection by Olympus-GX51 optical microscope (OM). Specimens for electron backscatter diffraction (EBSD) were treated with electropolishing with a 30 vol% HClO₄ + 70 vol% CH₃OH solution at 20 V and -30 °C. Hitachi-3400 scanning electron microscopy (SEM) and its associated TSL-OIM system were used to perform EBSD tests on samples, the scanning step size is 0.5 μm. Aztec Crystal software was used to analyze data and information about orientation and grain size. Furthermore, a lower limit boundary misorientation cut-off of 2 ° was utilized to remove the spurious boundaries caused by orientation noise [18]. After samples for transmission electron microscopy (TEM) were mechanically ground, holes were punched at different positions. Then, a circular sample with a diameter of 3.0 mm was subjected to ion milling. Finally, the sample was photographed from different directions using the Talos F200S G2 field emission transmission electron microscope. The cutting positions of each sample are shown in Fig. 2(a).

Microhardness testing samples were cut perpendicular to the welding direction, the cutting position of the sample is shown in Fig. 2(a). A row of Vickers indentations was produced on specimen under a load of 100 gf with a dwell time of 10 s, and they were placed 3 mm from the top surface with a spacing of 0.5 mm between adjacent indentations, as shown in Fig. 2(b). Tensile tests were carried out on the INS5967 electronic universal material testing machine, with the crosshead speed of 1 mm·min⁻¹. For each joint, tensile strength and elongation were calculated by averaging the values of three specimens. Sample size and cutting position are shown in Fig. 2(a).

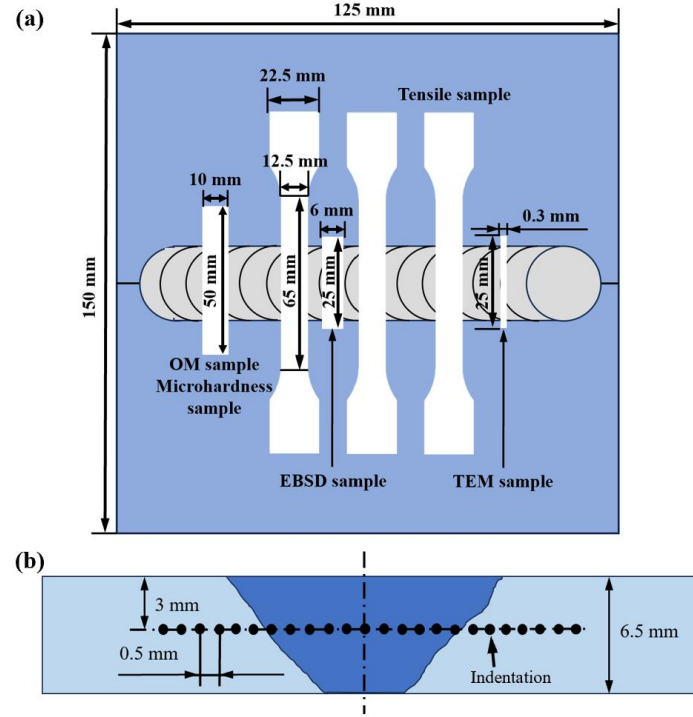


Fig. 2 Diagram of Sampling location

(a) Sample size and cutting position; (b) Indentation distribution

3 Results and discussion

3.1 Microstructure characterization of joints

The optical macrographs of joints in cross-section obtained with different process parameters are shown in Fig. 3. As can be seen from the figure, using the process parameters of each group, defect-free and well-formed joints can be obtained. The entire joints present a shape that is wider at the top and narrower at the bottom. When the rotational speed remains constant, as the traverse speed increases, the peak temperature drops, and the width difference between bottom and top of joints becomes larger. From 150 mm/min, it gradually transitions from a "dome shape" to a "trapezoidal shape" at 300 mm/min. Due to the small amount of oxide film on the joint surface being mixed into the weld seam, an "S" line feature is formed in WNZ [19]. Compared to base material, the grain size in joints is significantly reduced. When the welding speed is low, the boundary lines of each regions are not obvious. As the welding speed increases, heat input decreases and grains in PAZ experience a more rapid heating and cooling process, so the width at the bottom of joint gradually decreases. Under the condition of constant traverse speed, as the rotational speed increases, the range of heat affected zone (HAZ) will gradually extend to surrounding regions. This is because the shearing range of the stirring pin and heat input have both increased.

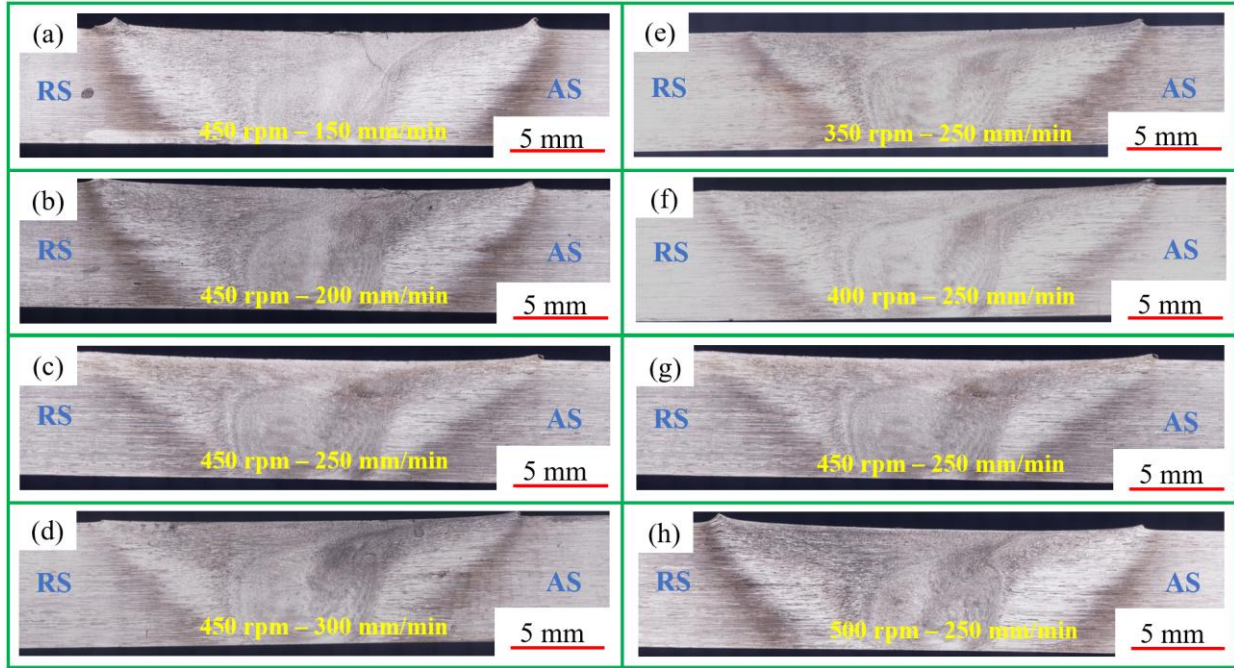


Fig. 3 OM of the cross-section overview of the joints with various process parameters:

- (a) 450 rpm-150 mm/min; (b) 450 rpm-200 mm/min; (c) 450 rpm-250 mm/min; (d) 450 rpm-300 mm/min;
 (e) 350 rpm-250 mm/min; (f) 400 rpm-250 mm/min; (g) 450 rpm-250 mm/min; (h) 500 rpm-250 mm/min

Microstructure of the joint obtained with the parameters of rotational speed of 450 rpm and traverse speed of 250 mm/min is shown in Fig. 4. The joint can be generally divided into four main areas: weld nugget zone (WNZ), thermo-mechanical affected zone (TMAZ), heat affected zone (HAZ), and base material (BM). From Fig. 4 (b), it can be seen that the base material presents as elongated plate, with fine precipitates at grain boundaries, which are typical characteristics of rolled state. Grains in WNZ undergo a fragmentation and recrystallization under the mechanical agitation of the stirring pin, so the grain size is tiny. Due to the fact that heat input mainly comes from the friction between the shoulder and the upper surface of the sheet [20], there are also differences in grain size at the top, middle and bottom of WNZ, as shown in Figs. 4 (c-e). Figs. 4 (f-h) indicates that the range of TMAZ is very limited, and there are almost no elongated deformed grains. By comparing advancing side (AS) and retreating side (RS), it can be found that the range of RS-TMAZ is relatively larger and the transition of the grain morphology on AS is more intense. Grains in HAZ were only affected by the welding thermal cycle and no dynamic recrystallization (DRX) occurred, so their size significantly increased. Additionally, from Fig. 4 (i-j), it can also be seen that some precipitates dissolve and the other aggregate and coarsen, the color difference between grains can also prove that solute atoms have diffused.

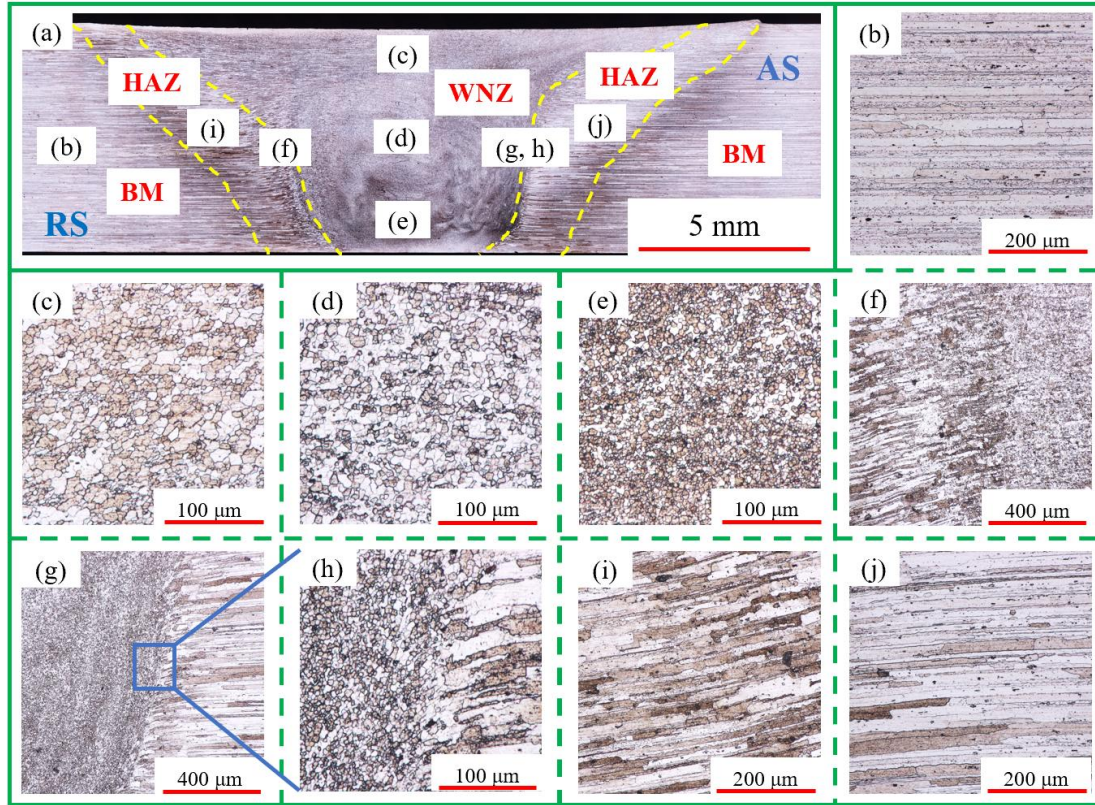


Fig. 4 Microstructure of the joint (450 rpm-250 mm/min)

(a) Macrostructure; (b) BM; (c) WNZ-top; (d) WNZ-middle; (e) WNZ-bottom; (f) RS-TMAZ; (g) AS-TMAZ (low magnification); (h) AS-TZ (high magnification); (i) RS-HAZ; (j) AS-HAZ

Inverse pole figures (IPF) and grain size distribution in BM, TMAZ and WNZ of 2195 Al-Li alloy FSW joint obtained with typical welding parameters (450 rpm - 250 mm/min) are shown in Fig. 5. Grains in BM are plate-like, representing the characteristics of rolled state, and orientations of each grain are basically consistent, with $\langle 111 \rangle$ and $\langle 101 \rangle$ being dominant. There are a few convex lens-shaped grains with a size less than 10 μm at grain boundaries of the plate-like grains (Fig. 5 (a)). The reason is that the internal storage energy of grains is high and a few nuclei form recrystallized grains at initial grain boundaries during subsequent natural aging process. Nevertheless, the coarse grains remain the main microstructure in BM, with an average size of approximately 45 μm (Fig. 5 (a')). During FSW process, grains in RS-TMAZ is mainly formed by the flow deposition of plasticized metal with a wide area, and they undergo an incomplete dynamic recrystallization process, presented in an equiaxed morphology (Fig. 5 (b)). By contrast, initial grains on AS undergo intense friction, resulting in a relatively narrow area of AS-TMAZ and a drastic change in grain orientation (Fig. 5 (c)). In TMAZ, grain size is significantly smaller than that of BM, and grain size of AS-TMAZ is slightly larger than that of RS-TMAZ due to higher peak temperature and faster grain boundary migration speed (Fig. 5 (b', c')). As for WNZ, grains underwent significant changes compared to BM. WNZ-top was mainly affected by the shoulder, while WNZ-middle was mainly

subjected to shear and friction from the side of stirring pin. The stirring pin has a taper angle with threaded feature, so the shear surface morphology is relatively complex. Finally, grain orientation still mainly consists of $\langle 1\ 0\ 1 \rangle$ (Fig. 5 (d, e)). WNZ-bottom was simultaneously subjected to stirring and shearing from the side and end face of the stirring pin, so grain orientation was the most disordered (Fig. 5 (f)). Average grain size decreased from WNZ-top to WNZ-bottom, and the smallest average grain size was less than $7\ \mu\text{m}$ (Fig. 5 (d'-f')).

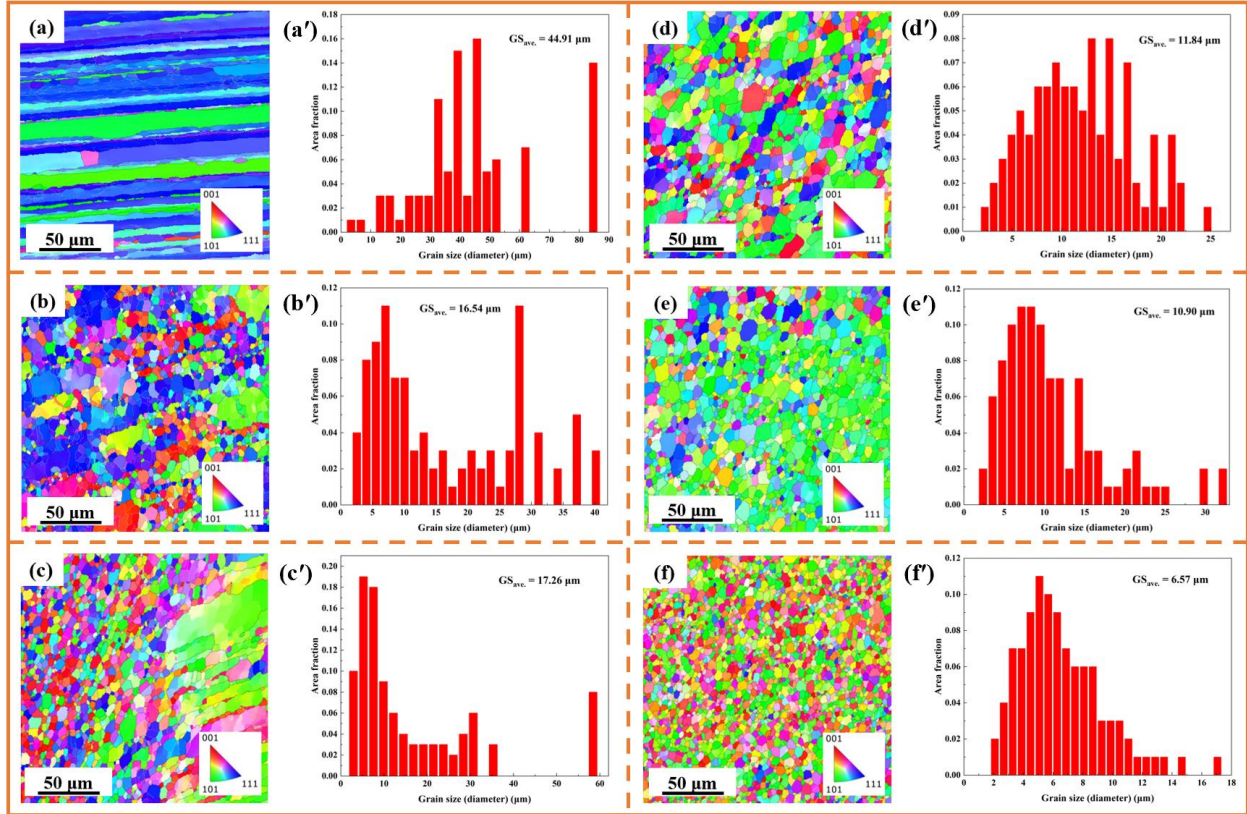


Fig. 5 IPF maps and grain size distribution of different regions in the joint (450 rpm-250 mm/min)
(a, a') BM; (b, b') RS-TMAZ; (c, c') AS-TMAZ; (d, d') SZ-top; (e, e') SZ-middle; (f, f') SZ-bottom

Microstructure in the middle of WNZ obtained with different process parameters is shown in Fig. 6. In order to systematically explain the relationship among the degree of deformation, grain size and peak temperature, an analysis was conducted based on the Zener-Hollomon formula [21]:

$$Z = \varepsilon \exp\left(-\frac{Q}{RT}\right) \quad (1)$$

$$d = C_1 Z^{C_2} \quad (2)$$

$$\ln d = \ln C_1 + C_2 \ln Z \quad (3)$$

where Z is Zener-Hollomon parameter, ε is deformation rate, C_1 and C_2 are both constants (C_2 is negative), R is gas constant, Q is activation energy, and T is deformation temperature. As traverse speed increases continuously, grain boundary migration rate drops, so DRX nuclei cannot grow in time, resulting in grain refinement. When traverse speed reaches 300 mm/min, although heat

input further decreases, the faster traverse speed also weakens the frictional deformation effect on base material, and decreasing strain rate leads to a reduction in the nucleation rate. Therefore, the grain size slightly increases. As the rotational speed gradually increased from 350 rpm to 500 rpm, the peak temperature of the welding thermal cycle kept rising, causing the grain size in WNZ to show a trend of increasing.

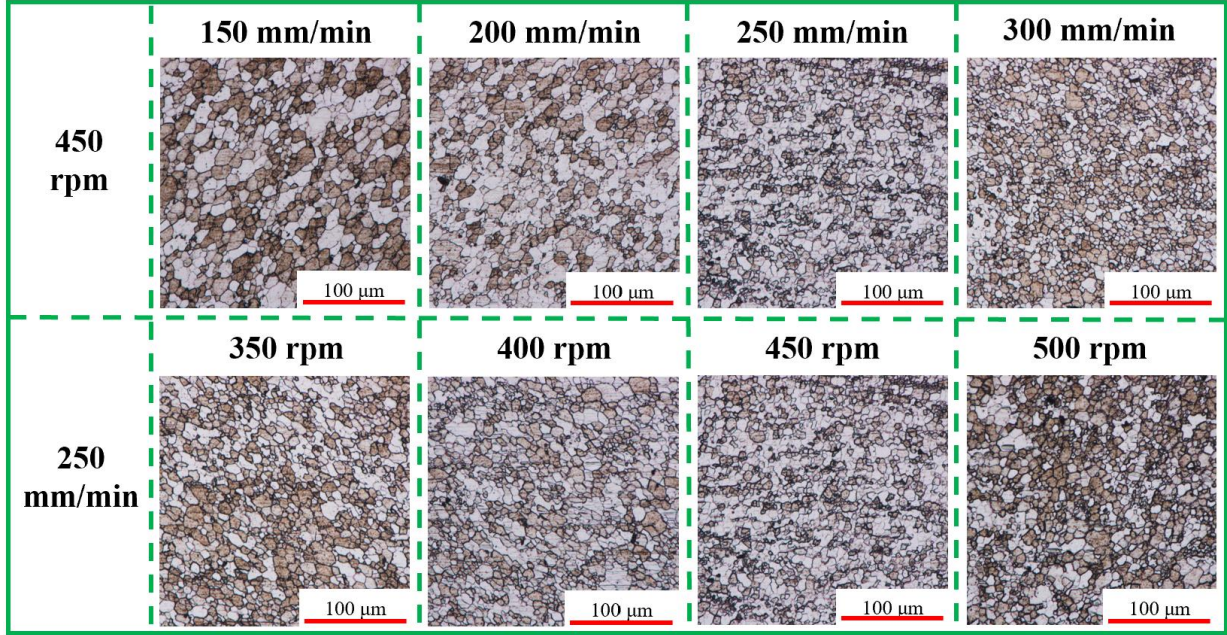


Fig. 6 Microstructure in WNZ-middle obtained with different process parameters

3.2 Mechanism of dynamic recrystallization

Dynamic recrystallization (DRX) can be described by the grain orientation spread (GOS) of the grains. The GOS value of grain i can be expressed by formula (4):

$$GOS(i) = \frac{1}{J(i)} \sum_j \omega_{ij} \quad (4)$$

where $J(i)$ represents the number of pixels in grain i , and ω_{ij} is the angle difference between the orientation of pixel j and the average orientation of grain i . Grains with $GOS(i)$ less than 2° are regarded as recrystallized grains, grains with $GOS(i)$ greater than 7° are considered as deformed grains, and another grains are classified as sub-grains [22]. To further identify the deformed grains, sub-grains, and recrystallized grains in each region of the joint, GOS and grain boundaries distribution are shown in Fig. 7. Grain boundaries with misorientation angle ranging from 2° to 10° are defined as low-angle grain boundaries (LAGBs), while grain boundaries with misorientation angle greater than 10° are defined as high-angle grain boundaries (HAGBs). Recrystallization degree of deformed grains and sub-grains is relatively low, thus there are lots of LAGBs within grains. As can be seen from Fig. 7(a), after solution treatment and aging, deformed grains basically disappear in BM. However, due to the main occurrence of recovery, so there are

still a large number of subgrains. The dynamic recrystallization process carried out thoroughly in WNZ, so the proportion of recrystallized grains in WNZ-middle and WNZ-bottom reaches more than 60% (Fig. 7(e-f)). Due to its direct contact with air, WNZ-top has a rapid cooling rate, so recrystallization process did not proceed fully. In TMAZ, the degree of DRX is low but the strain is high, resulting in subgrains that have not completed recrystallization nucleation being dominant, and there are still many deformed grains.

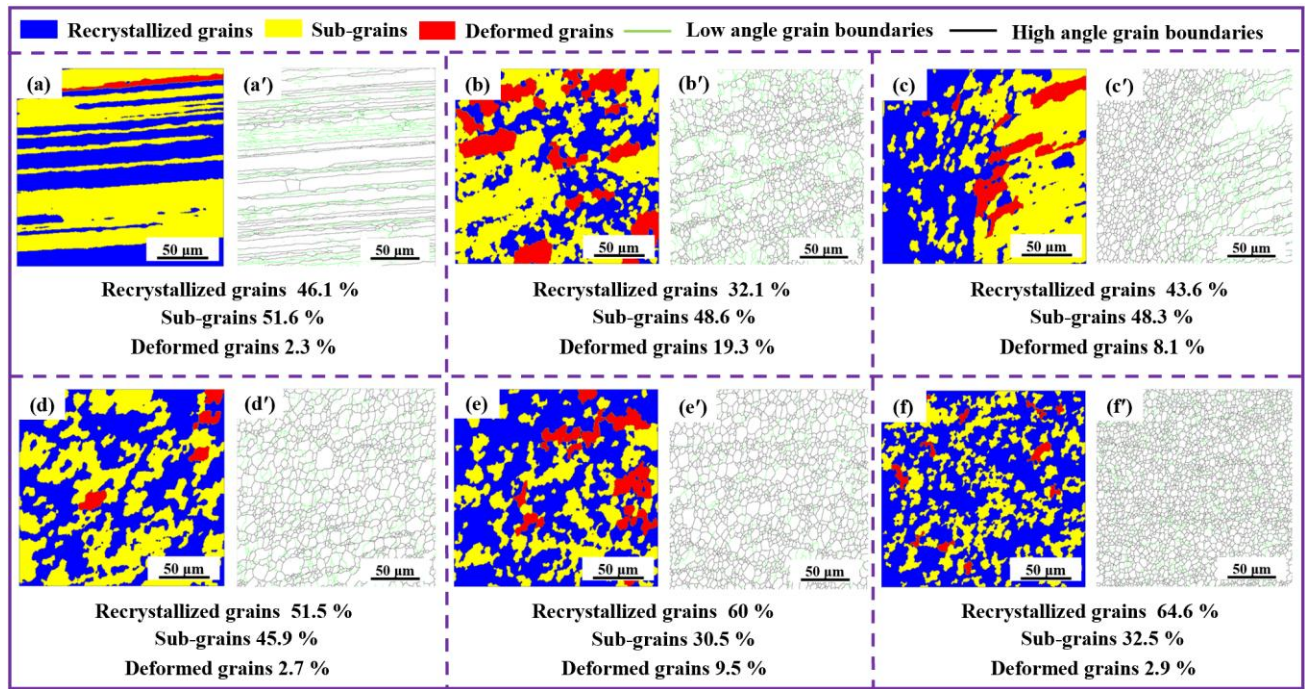


Fig. 7 GOS and grain boundaries distribution of the FSWed specimen for Fig. 5

(a, a') BM; (b, b') RS-TMAZ; (c, c') AS-TMAZ; (d, d') SZ-top; (e, e') SZ-middle; (f, f') SZ-bottom

Misorientation angle and axis vector distributions of grain boundaries in each region of the joint is shown in Fig. 8. As can be seen from the figure, the proportion of LAGBs in BM is relatively high. The reason is that high-density dislocations were generated during rolling process, then cancelled and rearranged during aging treatment. However, precipitates would hinder the migration of dislocation and grain boundaries, inhibiting static recrystallization, so a larger number of LAGBs were retained. In WNZ, intense stirring effect would prompts the original grains to break apart and reorganize. High-temperature condition promotes the dissolution of precipitated phases while facilitating the occurrence of continuous dynamic recrystallization (CDRX) and geometric dynamic recrystallization (GDRX). The high-density dislocations form LAGBs through slip and climbing, then absorb more dislocations, gradually transforming into LAGBs. Therefore, the proportion of LAGBs in WNZ is relatively low. The proportion of LAGBs at WNZ-top are higher than those in WNZ-middle and WNZ-bottom, which is due to high-density strain and a steep thermal gradient, ultimately resulting in the generation of high-density subgrains and dislocations [14]. At TMAZ and

WNZ-top, the proportion of grain boundaries with misorientation angle ranging from 2° to 20° shows a decreasing trend, which is consistent with the characteristics of CDRX [23]. However, in WNZ-middle and WNZ-bottom, the proportion of grain boundaries with misorientation angle ranging from 10° to 30° is similar, indicating that discontinuous dynamic recrystallization (DDR) also plays a role. The proportion of grain boundaries with a misorientation angle of approximately 55° is relatively high in both BM and WNZ. For BM, when the misorientation angle of grain boundaries is within the range of 50° to 60° , their axis vectors are concentrated near $\langle 2\ 2\ 3 \rangle$, while in each region of WNZ, corresponding axis vectors are mainly distributed near $\langle 1\ 1\ 1 \rangle$. In addition, 2195 Al-Li alloy contains much alloy elements such as Li and Cu, with relatively low stacking fault energy. It may form $\Sigma 3$ twins ($60^\circ/\langle 1\ 1\ 1 \rangle$), but no twin boundaries can be observed in grain boundary distribution map, thus ruling out this possibility. In WNZ, there are a few grains with misorientation angle ranging from 60° to 70° , but the distribution of their axis vectors is the most dense, mainly concentrated near $\langle 2\ 1\ 2 \rangle$.

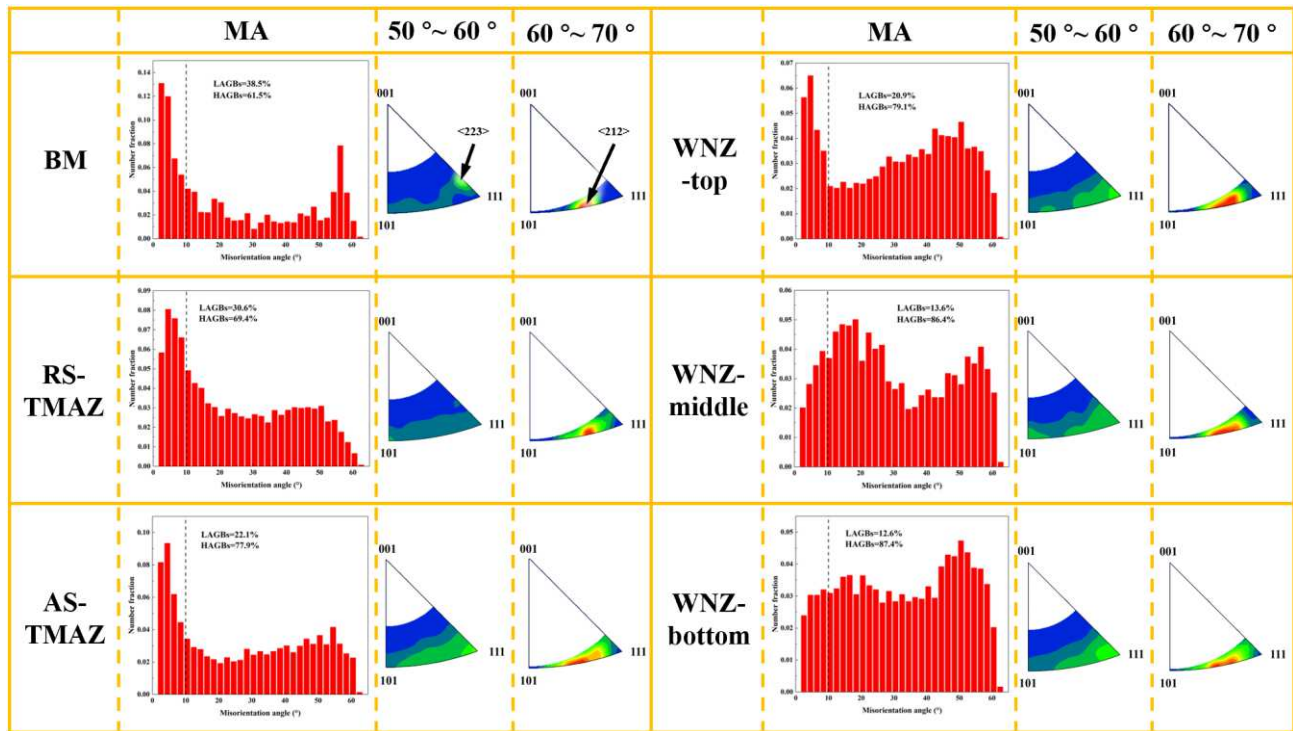


Fig. 8 Misorientation angle and axis vector distributions of the FSWed specimen for Fig. 5

Grains within base metal will undergo torsion during FSW process and forming deformation texture. Due to the occurrence of DRX, recrystallization texture will also be formed. Usually, there is an interaction between these textures, and their forms and intensities will to some extent affect the anisotropy of the material. To further analyze the evolution of texture, the $\{1\ 0\ 0\}$, $\{1\ 1\ 0\}$, and $\{1\ 1\ 1\}$ pole figure of BM, TMAZ-RS, and WNZ were compared and analyzed, as shown in Fig. 9. The shear deformation in the friction stir welding process is the main deformation mechanism [24].

Therefore, the shear plane normal direction (SPN) and the shear direction (SD) were used to describe shear textures in WNZ and TMAZ. The standard polar figures of several typical textures are shown in Fig. 9, too. It can be found that there are strong rolling textures in BM, mainly consisting of Goss $\{110\}\langle001\rangle$ and Copper $\{112\}\langle11\bar{1}\rangle$ texture, with the maximum intensity reaching 13.24, and forming specific orientation relationships. The angle between $\{110\}$ and $\{112\}$ lattice plane is close to 55° , so the distribution of misorientation angle forms concentrated distribution (Fig. 8). The texture of WNZ is mainly $B\{1\bar{1}2\}\langle110\rangle$ and $\bar{B}\{\bar{1}1\bar{2}\}\langle\bar{1}\bar{1}0\rangle$, as well as weak $C\{001\}\langle110\rangle$. This proves that FSW promotes the formation of shear textures in WNZ. The intensity of TMAZ is the weakest, and there are $B\{1\bar{1}2\}\langle110\rangle$, $\bar{B}\{\bar{1}1\bar{2}\}\langle\bar{1}\bar{1}0\rangle$ and Goss $\{110\}\langle001\rangle$. This is because grains in TMAZ near WNZ are strongly subjected to shear, forming shear textures, while grains far from WNZ are influenced by welding thermal cycle and undergo recovery, resulting in the weakening of original rolling texture.

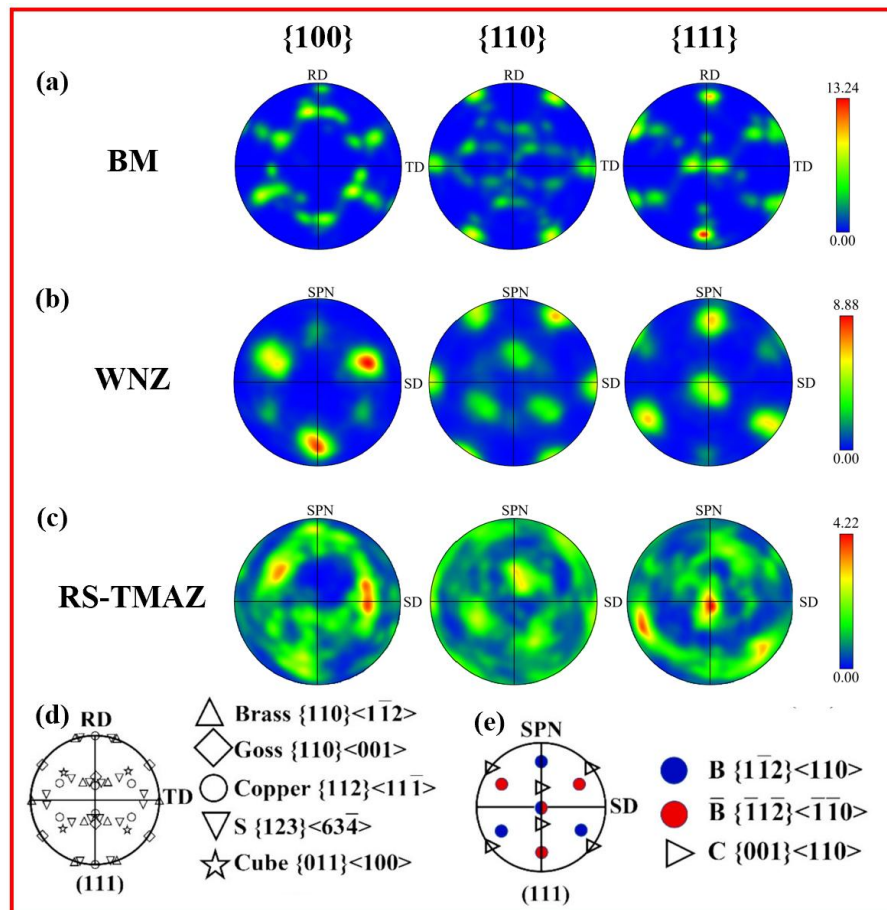


Fig. 9 Pole figures of the FSWed specimens

Specimen coordinate frame: (a) BM; Shearing coordinate frame: (b) RS-TMAZ; (c) WNZ;
(d) Standard pole figure in specimen coordinate frame; (e) Standard pole figure in shearing coordinate frame

3.3 Reprecipitated strengthening phases and dislocations

In order to determine the type, size and distribution of precipitates, transmission electron microscope was used to take bright-field (BF) images and selected area electron diffraction (SAED) patterns of the microstructure in different regions of the joint obtained with typical parameters. Energy dispersive spectroscopy (EDS) mapping was also performed on the characteristic areas. Precipitated phase was identified based on its morphological characteristics, diffraction patterns and element distribution. Fig. 10 displays the strengthening precipitates of BM. The shooting direction and SAED direction were both along $Al\langle 110 \rangle$ zone axis. As can be seen from Fig. 10, a large number of needle-shaped T_1 phases and θ' phases are distributed in Al matrix. Usually, the angle between two T_1 variants is approximately 109.4° , and the angle between T_1 phase and θ' phase is approximately 125.3° [25]. The length of T_1 phase is 60-100 nm, and the thickness is 2-5 nm [26]. The high-resolution TEM image of the T_1 phase is shown in Fig. 10 (b), and its SAED pattern is shown in Fig. 10 (c). Measured interplanar spacing of lattice is 0.375 nm, which is similar to the interplanar spacing (0.392 nm) of $(10\bar{1}1)$ marked in the SAED pattern [27]. θ' phase does not show the corresponding bright spot in the SAED pattern because its habit plane is $\{100\}$. Combining EDS mapping results of needle-shaped precipitates (Fig. 10 (e-f)), it can be seen that there is Cu atom enrichment in dark region. Comprehensive comparison can prove that the precipitated phases in 2195 Al-Li alloy are mainly T_1 phase and θ' phase.

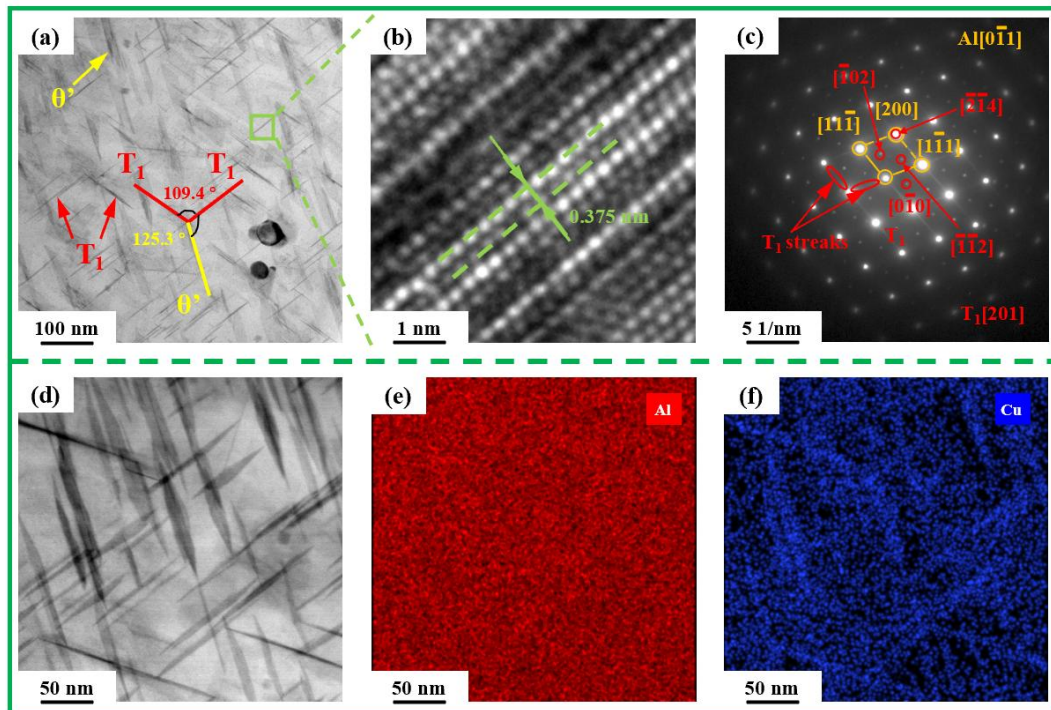


Fig. 10 Phase identification using TEM in BM: (a) Low-magnification BF image; (b) High-resolution TEM image of T_1 phase; (c) SAED pattern; (d) High-magnification BF image; (e) EDS mapping (Al); (f) EDS mapping (Cu)

Fig. 11 displays the strengthening precipitates of WNZ. The shooting direction and SAED direction were both along $Al\langle 110 \rangle$ zone axis, too. By magnifying grain A and adjacent region, as shown in Fig. 11 (b-c), T_1 phase with coherent relationship was observed near initial grain boundary. Its morphology is significantly different from BM, changing from thin sheet to granule. This may be due to the high temperature causing T_1 phase to dissolve back into matrix, and natural aging after welding causing it to precipitate again near the grain boundary [26]. By magnifying grain B and adjacent region, a small amount of short rod-shaped precipitates (with a size of approximately 40 nm) can be observed diffusely distributed within the grain, and there are always spherical precipitates (with a size of approximately 10 nm) accompanying them nearby. These can be considered as T_B (Al_7Cu_4Li) and β' (Al_3Zr) [28-30]. In addition, there are also many fine θ' phases in the grain, as shown in Fig.11 (e-f), which is also the result of re-precipitation. Moreover, many entangled dislocations can be observed near the grain boundary, indicating that the DRX process has not yet been fully completed.

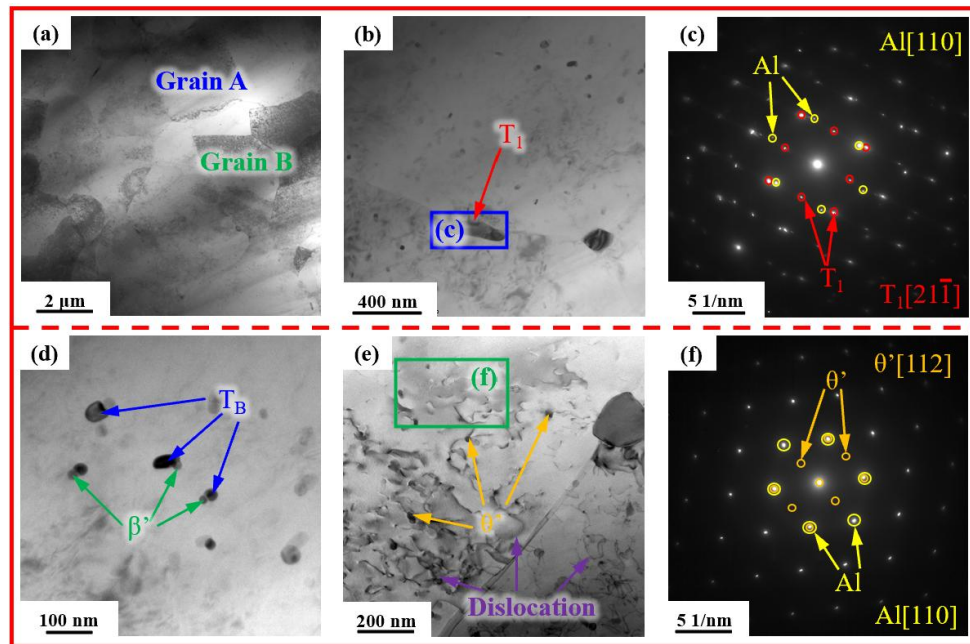


Fig. 11 Phase identification using TEM in WNZ: (a) Low-magnification BF image; (b) High-magnification BF image of grain A; (c) SAED pattern of region (c); (d) High-magnification BF image of grain B; (e) BF image of grain B (near grain boundary); (f) SAED pattern of region (f)

3.4 Mechanical properties

Vicker's microhardness distributions of joints obtained with different process parameters are shown in Fig. 12. As displayed in the figure, it can be seen that microhardness distribution of joints is overall in the shape of a “W”. Due to the presence of a large number of precipitates, strengthening effect of BM is significant where the microhardness is the highest, ranging from 165 to 170 HV. No matter what process parameters are applied, microhardness of HAZ is always the lowest, only around

110 HV. This is because precipitated phases are basically all dissolved, and at the same time, the grain size has also grown, resulting in fewer grain boundaries and easier dislocation activation. Microhardness of WNZ is relatively high, but the fluctuation range is large, ranging from 120 to 135 HV. This is because precipitates in this region do not all reprecipitate after dissolution. However, due to the finest grains, its hardness slightly higher than HAZ. As for TMAZ, grains undergo distortion and deformation, but still contain some fine grains, with microhardness similar to WNZ. Compared with AS, grains in RS are mainly formed by the backflow solidification of plasticizing materials, so growth time of over-aged phase is more sufficient and microhardness of grains in RS is relatively lower. The influence of traverse speed on distribution of microhardness mainly manifests in WNZ and the joint range. As traverse speed increases, the degree of grain growth in WNZ decreases, thus the microhardness shows an upward trend; in addition, the influence range of the welding thermal cycle gradually decreases, so the width of joint decreases. Compared with the change in welding speed, the change in the rotational speed has smaller impact on the microhardness of WNZ. When the rotational speed is 350 rpm, the dissolution of precipitates causes a decrease in microhardness; when the rotational speed is increased to 400 rpm, new strengthening phases such as δ'/β' will reprecipitate within the corresponding temperature range after cooling, which helps to increase microhardness. Further increase in rotational speed leads to excessive heat input, resulting in weakened refined crystalline strengthening and counteracting the precipitation strengthening, so overall hardness slightly decreases. In conclusion, refined crystalline strengthening and precipitation strengthening are the main mechanisms affecting the distribution of microhardness in the joint.

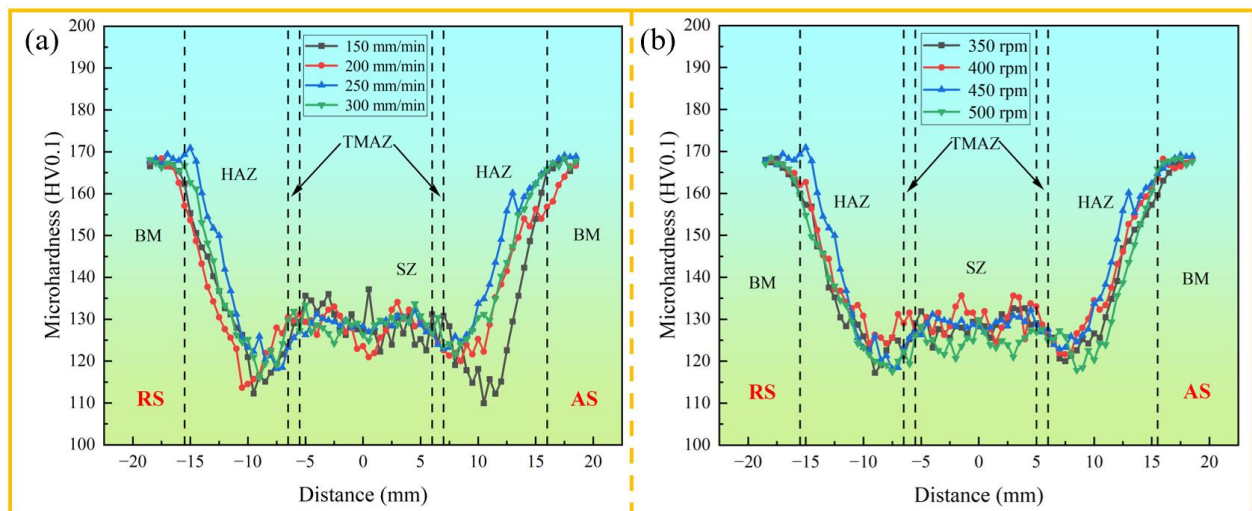


Fig. 12 Microhardness distribution of joints

(a) The rotational speed is constant at 450 rpm; (b) The welding speed is constant at 250 mm/min

Tensile properties of joints obtained with different process parameters are shown in Fig. 13. As indicated in the figure, the optimal process parameters are 450 rpm - 250 mm/min, which results in

an tensile strength reaching 430 MPa, accounting for 74.14% of the base material; and the elongation can reach 6.38%, which is 75.06% of the base material. In general, traverse speed has little effect on the performance of joints. When the traverse speed is lower than 250 mm/min, the grain size gradually decreases with traverse speed increasing, so tensile strength also increases. However, if the traverse speed is 300 mm/min, insufficient welding heat input leads to a decrease in the fluidity of the plasticized metal at the root, so AS of joint root is prone to form discontinuous kiss-bond. During the stretching process, microcracks start to occur as the initial source. The influence of rotational speed on the elongation is particularly significant. Especially when the rotational speed is 500 rpm, the degree of material plasticization is higher and friction between welding tool and base material decreases, there may be a delamination phenomenon in the root area. Stress concentration is prone to occur during the stretching process, and after microcracks extend upwards, the joint will break.

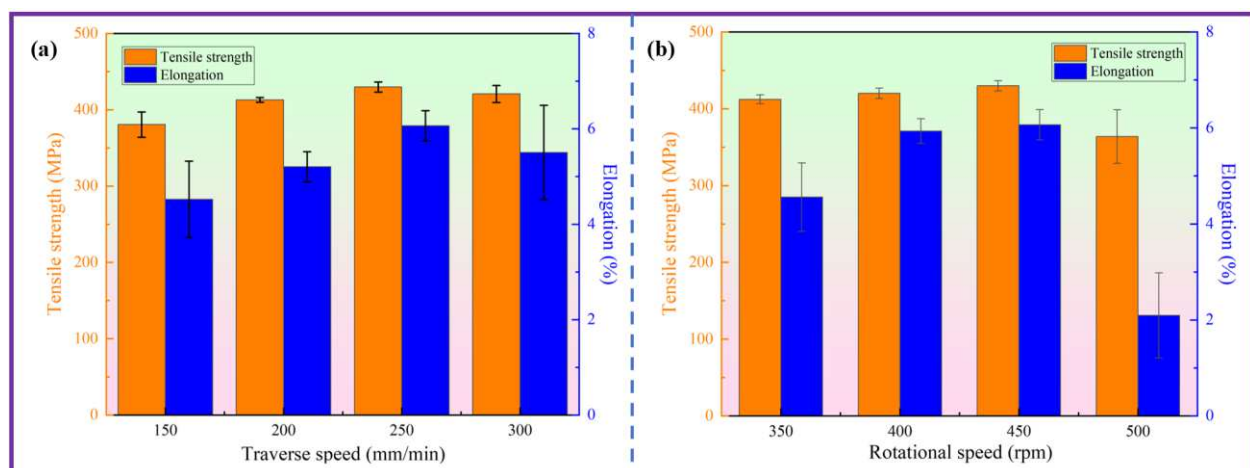


Fig.13 Tensile properties of joints

(a) The rotational speed is constant at 450 rpm; (b) The welding speed is constant at 250 mm/min

Necking phenomenon at fracture surface of 2195 Al-Li alloy FSW joints is not obvious. Fracture positions can be roughly divided into two categories, as shown in Fig. 14. The fracture path of high-quality joint are located in RS-HAZ (Fig. 14(a)), at a 45 ° angle to the surface of sheet, and the fracture surfaces are relatively flat. This is because the heat input on RS is lower compared to AS, and its cooling process is slightly slower, so the over-aged phase in RS-HAZ is more fully developed over a longer period of time with the lowest hardness. During the tensile process, it is prone to become the origin of microcracks. Surface morphology of this fracture is mainly composed of fine dimples and small cleavage facets (Fig. 14 (b,c)), clear tearing edges can also be observed, as well as granular precipitate phases, indicating that the fracture mode is mixed ductile-brittle fracture. The fracture positions of poorer-performing joint are located inside the WNZ (Fig. 14(d)), and the fracture path present a serrated shape, indicating that there are micro-layered tears inside WNZ. There is a bend at the junction of the shoulder affected zone and pin affected zone. This is because

there is size difference between surface and internal grains, resulting in a mismatch in strength. Surface morphology of this fracture is in the form of river-like pattern, with almost no dimples, indicating that the fracture mode is close to brittle fracture, as shown in Fig. 14(e,f).

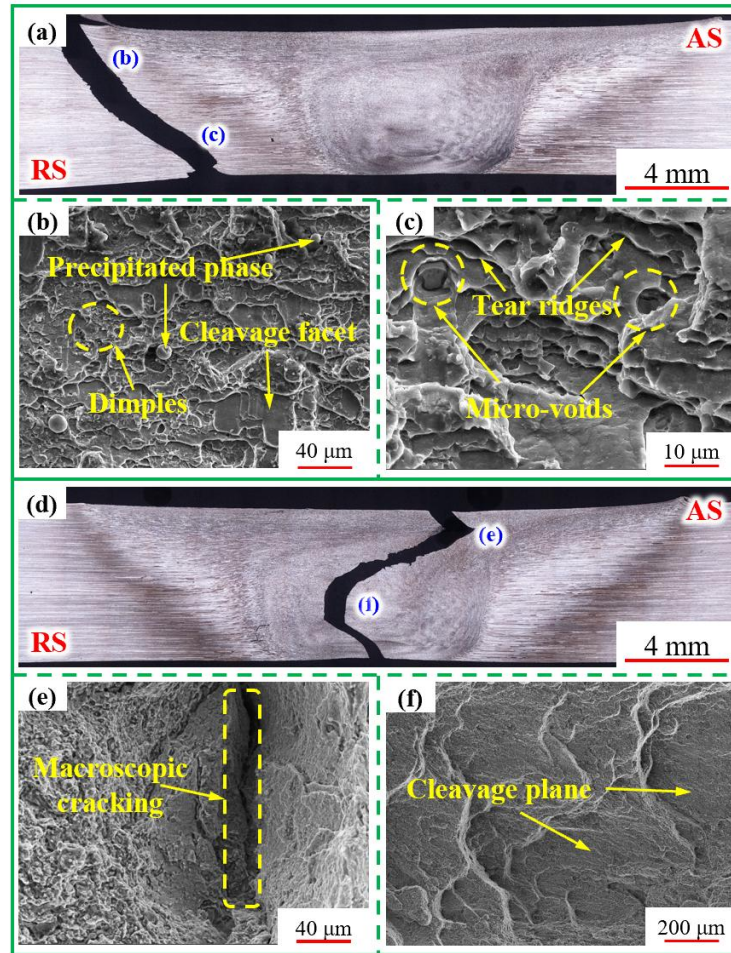


Fig. 14 Fracture path and morphology of joints

- (a) Fracture occurs in the HAZ; (b) SEM image (HAZ-top); (c) SEM image (HAZ-bottom)
 (d) Fracture occurs in the WNZ; (e) SEM image (WNZ-top); (f) SEM image (WNZ-bottom)

4 Conclusion

This study successfully obtained high-quality FSW joints of 2195 Al-Li alloy. The evolution process of microstructure in joint was deeply investigated, intrinsic relationship between mechanical properties and microstructure was summarized, and the fracture mechanism of the joint was analyzed. The main conclusions are as follows:

(1) By using FSW, defect-free joints with a thickness of 6.5 mm can be fabricated from the 2195 Al-Li alloy. When the welding parameters are set at 450 rpm and 250 mm/min, superior joints with excellent performance can be obtained. Its strength can reach 430 MPa, which is 74.14% of the base material, and the elongation can reach 6.38%, which is 75.06% of the base material.

(2) Among all of regions of joint, equiaxed grains of WNZ have the smallest size and the highest hardness, the degree of DRX is almost complete. The texture of WNZ is mainly $B \{ 1 \bar{1} 2 \} < 1 1 0 >$ and $\bar{B} \{ \bar{1} 1 \bar{2} \} < \bar{1} \bar{1} 0 >$. Grains in TMAZ are in a streamlined shape, and the transition of AS is more intense than that of RS, both forming shear textures. The grains of HAZ have significantly grown, and the grain orientation is basically consistent with that of the BM.

(3) Regarding precipitates, the strengthening phases in BM are mainly composed of T_1 phase and θ' phase; within the grains in WNZ, there are T_B phase and accompanying spherical β' phase, as well as reprecipitated fine T_1 phase, and there are also many dislocations near grain boundaries. In conclusion, refined crystalline strengthening and precipitation strengthening are the main mechanisms affecting mechanical properties of FSW joints of 2195 Al-Li alloy.

CRedit authorship contribution statement

Wenjing Gao: Conceive and design this study, and, Writing – original draft. Xu Liu: Perform the experiments, organize the data. Ruichen Liu: Perform the experiments, organize the data. Xin Liu: Perform the experiments, organize the data. Chenyu Wu: Perform the experiments, organize the data. Yuxin Li: Perform the experiments, organize the data. Kang Ma: Writing – review & editing. Wentie Niu: Writing – review & editing. Shikang Gao: Writing – review & editing. Li Zhou: Writing – review & editing. All authors read and approve the manuscript.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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