Effects of Welding Speed and Post-weld Hot Rolling on Microstructure and Mechanical Properties of Friction Stir-Welded AZ31 Magnesium Alloy

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Abstract

The as-rolled AZ31 Mg alloy sheets are welded by friction stir welding (FSW) in this study, followed by annealing and hot rolling with different reductions. The effects of welding speed and the rolling reduction on the microstructure, tensile properties, microhardness and fracture surfaces of the specimens are investigated. The results show that the microstructures of the FSW joint in different regions become more and more consistent as the hot rolling reduction increases. The mechanical properties of the FSW joints and the joint coefficient are improved with increasing rolling reduction; however, the hardness value in the stirred zone decreases firstly and then increases. When the rolling reduction is 50%, the tensile strength of FSW joint is close to that of base metal. The sample L₃ achieves the best comprehensive properties, with the yield strength, ultimate tensile strength and elongation of 162, 287.9 MPa, and 17.9%, respectively.

Keywords Magnesium alloy · FSW · Post-weld hot rolling · Microstructure · Mechanical properties

1 Introduction

Magnesium alloys have many advantages of low density, high specific strength, good thermal conductivity, damping capacity and convenient recycling, etc., which are the lightest metal structural materials in engineering applications. As green engineering materials in twenty-first century, the magnesium alloys have been widely used in the passenger vehicle, aviation aerospace, 3C industry and other fields [1–6]. Reliable welding technology has significant effects on applications. Many porosities and welding deformation emerged easily due to traditional welding method, which hinders the application of magnesium alloy in various industry fields.

As a solid-phase joining technology, FSW is the welding process of using a frictional heat in a plastic state. Without metal melting in the friction process, FSW has the advantages of less defect, high quality, etc. due to low welding temperature and small welding deformation. It is particularly suitable for the welding of light metals such as aluminum alloy and magnesium alloy [7–10]. FSW has been widely used to join several kinds of Mg alloys, such as Mg–Al–Zn [11–14], Mg–Al–Mn [15], Mg–Al–Ca [16], Mg–Zn–Zr [17, 18] and Mg–Zn–Y alloys [19].

Many previous studies indicate that the strengths of FSW Mg or Al alloys are commonly less than those of base metal, which are detrimental to the applications of FSW Mg or Al alloys [12, 13, 20–24]. In order to improve the mechanical properties of the welded joints of the magnesium alloy, it is necessary to reduce the microstructure difference between the welding joint and the base metal. It is reported that the post-weld hot rolling (PWHR) process of welding plate by tungsten inert gas welding can help solve the above problems due to the fine grain strengthening and work hardening [25]. The existing reports about the FSW of magnesium alloy are focused on the influence of processing parameters on the microstructure and mechanical properties of welded joint [26, 27]. However,

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the effect of PWHR after FSW is rarely reported. Therefore, the effects of welding speed and PWHR on the microstructure and mechanical properties of AZ31 magnesium alloy FSW joint were studied in this paper.

2 Experimental Procedures

Seven-millimeter-thick hot-rolled commercial AZ31 Mg alloy (Mg-3%Al-1%Zn) sheets were used in this study. Each sheet was machined into dimensions of 7 mm thickness, 100 mm length, and 70 mm width. Before welding, the plates were polished by abrasive paper and cleaned with acetone. A simple FSW tool was used with a concave shoulder 15 mm in diameter, and a cylindrical threaded pin with a diameter of 6 mm and a height of 5.7 mm. In this study, the constant tool rotation rate was 880 rpm and traverse speeds were 60 and 80 mm/min, respectively. They are referred to as FSW-L joints and FSW-H joints, respectively. The FSW tool was rotated in the counter-clockwise direction. The tilt angle was 2.5°. Welding direction was kept along with the rolling direction of the initial plates.

In order to eliminate the welding residual stress and make the following hot rolling feasible, the welded sheets were annealed at 350 °C for 2 h after welding. After annealing, the welded sheets were subjected to a subsequent hot rolling under different rolling passes and reductions at 350 °C. Before rolling, the samples were preheated at 350 °C for 15 min and re-heated to 350 °C (the rolling temperature) among each pass. The direction of the PWHR was the same with the rolling direction parallel to the welding direction (WD) of FSW specimens, as shown in

Fig. 1 Schematic illustration of the subsequent rolling process

Fig. 1. The samples and main processing parameters are shown in Table 1.

The samples for optical microscopy (OM) were etched with a solution consisting of 4.2 g picric acid, 10 ml acetic acid, 10 ml H₂O, and 70 ml ethanol. Optical microscope (OM: OLYMPUS PMG3) was used to structural observation. Electron backscatter diffraction (EBSD) was used to analyze the texture evolution near stirred zone (SZ) side region of the PWHRed AZ31 sheet. The EBSD detector was an HKL Channel 5 Systems equipped by a field-emission gun scanning electron microscope (SEM) (FEI Nova400). The Vickers hardness tests were conducted on the middle of cross-sectional plane along the transverse direction by a Vickers indenter with a 100 g load for 5 s. The orientation of the tensile specimens in respect of the plates and their dimensions are shown in Fig. 2. Tensile tests have been performed at room temperature on a SANS (CMT-5100) tensile machine, and the average value of the three tensile tests was adopted for the tensile properties of each joint. The direction of base metal (BM) tensile specimen was the same as that of the corresponding FSW specimens. The fracture surfaces of tensile specimen were observed by SEM (TESCAN VEGA 3 LMH).

3 Results and Discussion

3.1 Microstructural Characterization

High magnification images of the four zones of Fig. 3 are presented in Figs. 4 and 5. Figure 4 illustrates the optical metallographic photos of different regions in sample L₀. The microstructure of the FSW joint is formally divided into four zones: base metal (BM), heat-affected zone (HAZ), thermo-mechanically affected zone (TMAZ) and the stirred zone (SZ). Figure 4a shows the microstructure of the BM, and the inhomogeneous equiaxed grains in BM exhibit a typical characteristic of rolled AZ31 alloy sheet [12]. The average grain size of BM is about 8.87 μm. Figure 4b manifests the micrograph of TMAZ and the grain size is about 6.91 μm, which consists of some coarse grains and numerous recrystallized grains. It indicates that TMAZ suffers dynamic recrystallization (DRX), but the process of DRX in TMAZ is incomplete due to the insufficient heat or intense plastic deformation, which is consistent with the previous studies [8, 28]. A typical microstructure of HAZ is displayed in Fig. 4c, and the microstructure of HAZ appears similar to the BM. The average grain size of HAZ is about 9.75 μm. It can be seen that the grain size in HAZ is larger than that of in BM. The reason for grain growth may due to influence of the heat during FSW process in this region. The stir zone is characterized by equiaxed grains, as it is presented in Fig. 4d–f,

Table 1 Processing parameters of samples (PWHR parameters at 350 °C)

Sample L	Sample H	Rolling pass	Final reduction in thickness (%)	Final thickness (mm)
L ₀	H ₀	–	–	7.0
L ₁	H ₁	1	14	6.0
L ₂	H ₂	2	32	4.8
L ₃	H ₃	3	50	3.5

Fig. 2 Schematic illustrations of the tensile specimen **Fig. 3** Cross-sectional macrograph images of **a** FSW-L₀, **b** FSW-H₀

which can be attributed to the DRX caused by the combination of frictional heating and severe plastic deformation during FSW.

The grain size in the top, middle and bottom layers of all the SZs (Fig. 4d–f) are about 6.73, 6.08 and 5.72 μm, respectively. The average grain size of the SZ is 6.18 μm (Table 2). The grain size in stir zone decreases gradually.

The reason for the maximum grain size in the top layer of the SZ is that the top layer of the SZ is subjected to the double action of the squeezing friction of the shoulder and the stirring needle, which produces a lot of heat in the area and needs more heat dissipation time. The residence time at high temperature is long, which leads to the material in the region having sufficient time for dynamic recrystallization and grain growth. This result is consistent with a previous report [29]. In contrast, the grain size in the middle layer of the SZ is smaller. However, in the bottom layer of the SZ, the less heat generated from smaller diameter of the stirring needle and the plate takes away a part of the heat. Therefore, the finer grain is formed because the temperature is relatively low and the energy and time for grain growth are not enough.

Figure 5 illustrates the Optical micrographs of different regions in sample H₀. Average grain sizes of BM, HAZ, TMAZ and SZ are about 9.1, 9.5, 6.7 and 5.53 μm, respectively. Under a constant speed of 880 rpm, the grain size of the SZ decreases from about 6.18 μm to about 5.53 μm as the welding speed increases from 60 to 80 mm/min (Table 2). When the rotational speed is constant, the welding heat input and the welding temperature decrease with the increase in welding speed, the trend of grain recrystallization and the grain size of the SZ are decreased. Afrin et al. [12] argue that it is also possible that the increase in the welding speed leads to increase in the deformation of the material, which promotes the recrystallization nucleation rate and reduces the grain size.

With the change of process parameters of friction stir welding, the grain size of the SZ during the friction stir welding process is also changed. The relationship between the stirring friction process parameters and the temperature of the SZ was studied by Commin et al. [27]. It can be obtained by Eq. (1):

$$\frac{T}{T_m} = K \left(\frac{\omega^2}{V \times 10^4} \right)^\alpha \quad (1)$$

where T is the temperature of the SZ, T_m is the melting point of the alloy, and K and α are the constants. According to previous studies [27, 30], T_m , K and α can be taken as 610 °C, 0.8052 and 0.0442 for AZ31 alloys, respectively. The maximum temperature of the joint L₀ and the joint H₀ is calculated to be 415 and 399 °C, respectively. It can be seen that the amount of heat input during the friction stir

Fig. 4 Optical micrographs of different regions in sample L_0 : **a** BM, **b** TMAZ, **c** HAZ, **d** top layer of the SZ, **e** middle layer of the SZ, **f** bottom layer of the SZ

Fig. 5 Optical micrographs of different regions in sample H_0 : **a** BM, **b** advancing side of the HAZ, **c** retreating side of the HAZ, **d** advancing side of the TMAZ, **e** retreating side of the TMAZ, **f** SZ

welding is reduced as the welding speed increases, so that the grains of the stirring zone are relatively refined. This is consistent with the above metallographic analysis.

Moreover, according to Fig. 5b, c, the HAZ grain sizes on the advancing and retreating sides are about 9.6 and 9.3 μm , respectively. In contrast, the grain size of the HAZ

on advancing side (AS) is larger than that of the retreating side (RS). The main reason is that the AS is subjected to a larger plastic shear force and frictional resistance compared with the RS, so that more heat by friction stir is produced at the AS. As a result, the temperature of the AS is higher than that of the RS resulting grain growth. In Fig. 5d, e, it

Table 2 Grain sizes of the different stir zones

Sample	Grain size of top layer (μm)	Grain size of middle layer (μm)	Grain size of bottom layer (μm)	Average grain size (μm)
L ₀ ($W = 880$ rpm, $V = 60$ mm/min)	6.73	6.08	5.72	6.18
H ₀ ($W = 880$ rpm, $V = 80$ mm/min)	5.99	5.4	5.2	5.53

can be seen that the grains of the TMAZ region at the AS are more homogeneous and smaller than those of the TMAZ region at the RS. The plastic softening layer lied at the leading edge of stirring pin at the AS undergoes the shearing action of the stirring pin and flows toward the RS along the side surface of stirring pin. The softening layer that has flowed to the RS plays the role in the lubrication at the retreating side softening layer/pin interface. Thus, the agitation intensity formed by stirring pin at the RS is reduced and the deformation degree of TMAZ region at the AS is higher than that of TMAZ region at the RS, leading to more abundant recrystallization at the AS.

The grain size of the TMAZ region at the AS is smaller than that of TMAZ region at the RS because the part of the heat of plastic softening layer is also brought to the TMAZ region at the RS, resulting in the accumulation of the heat of the TMAZ region at the RS. Therefore, the duration of the high temperature thermal cycle at the RS is longer than that of the AS, so leading to the higher degree of grain growth at the TMAZ region at RS. Similar result has been reported in the literature [31].

During 1–3 rolling passes process, microstructure evolution occurred in the different regions as illustrated in Figs. 6 and 7, respectively. It is found that the microstructures of BM, HAZ, TMAZ and SZ changed after each pass. In contrast to Figs. 4 and 5, it can be seen from Figs. 6a and 7a that no recrystallization occurs in each region except part of the BM when rolling reduction is 14% after annealing, which may be due to that the reduction in rolling cannot provide sufficient driving force to meet the conditions of initial DRX. At the same time, the grain size in each region has increased significantly, which is due to the annealing and preheating treatment before hot rolling. When the reduction in thickness increases to 32%, recrystallization occurs in each region. Figures 6b and 7b show that different degree of DRX has taken place in these four areas, owing to the reduction in rolling can provide sufficient driving force to meet the conditions of initial DRX. The grain size in each region is significantly reduced. Comparing Figs. 6b and 7b, the DRX of the sample H₂ is more significant than sample L₂, and the extent of grain size reduction is greater in each region. When the rolling reduction is 50%, the sufficient driving force is provided because the amount of deformation increases and recrystallization occurs in some of the grains

furthermore, so that the average grain sizes of these four regions are nearly the same.

Figure 8 shows variation of the average grain sizes in these four areas with rolling reductions. When rolling reduction is 14%, the grains in different regions grow significantly, but the growth speed of BM grains is smaller than that of SZ. The analysis shows that some fine grains appear locally in BM (Figs. 6a, 7a), which indicates that partial recrystallization has started to take place in the region. Part of the absorbed energy during the homogenization annealing and preheat before hot rolling is used for recrystallization and then for grain growth. However, in the process of welding, SZ experienced high temperature and large strain, severe plastic deformation occurred. The original coarse grains are significantly broken, and the dynamic recrystallization is sufficient. The absorbed energy in the subsequent heat treatment is mainly used for grain growth. The grains of SZ are therefore relatively large, as shown in Figs. 6a and 7a. Moreover, the absorbed energy in the subsequent heat treatment at HAZ and at TMAZ is mainly used for grain growth as well. Therefore, the grain size of HAZ, TMAZ and SZ in Figs. 6a and 7a is much coarser than those of BM.

When the rolling reduction is 32%, the average grain size in each region is drastically reduced. When the rolling reduction is 50%, the average grain sizes in the HAZ, TMAZ and SZ regions are further reduced. However, the average grain size of BM increased which shows that the recrystallized grains have grown. This phenomenon indicates that the BM undergoes a DRX process which includes deformation, nucleation and grain growth during PWHR. Similarly, DRX also occurs in HAZ, TMAZ and SZ. Therefore, the average grain sizes in each region after three passes of hot rolling are almost same.

Figure 9 shows the SEM images of the SZ after 1–3 rolling passes. A large number of grainy $\beta\text{-Mg}_{17}\text{Al}_{12}$ is distributed within the matrix, whereas only a few of $\beta\text{-Mg}_{17}\text{Al}_{12}$ is distributed along the grain boundary (Fig. 9). It can be found that the coarser $\beta\text{-Mg}_{17}\text{Al}_{12}$ particles distribute at the grain boundaries and the finer particles distribute within the matrix, which are gradually dissolved into the matrix with the rolling reduction increases. It is obvious that the number of the $\beta\text{-Mg}_{17}\text{Al}_{12}$ phases decreases as the rolling reduction increases. Thus, another

Fig. 6 Optical micrographs of the PWHRed sample L_0 with different rolling reductions: **a** 14%, **b** 32%, **c** 50%

microstructural characteristic is few β - $Mg_{17}Al_{12}$ phases remained in the PWHRed AZ31 sheets in addition to significant grain refinement.

3.2 Tensile Properties

Figure 10a, b shows the tensile stress–strain curves of the FSW joints before and after rolling. The tensile properties are shown in Table 3. As shown in Fig. 10a, b, the tensile strength of L_0 and H_0 is less than that of the base materials.

Table 3 displays the yield strengths of sample L_0 and sample H_0 , approximately reached to 90 and 87 MPa respectively, which are lower than the yield strength (~ 129 MPa) of the corresponding BM. The tensile strengths of sample L_0 and sample H_0 are about 190.8 and 179 MPa, respectively, also lower than the tensile strength of BM (~ 217.5 MPa). In addition, the elongations of sample L_0 and sample H_0 are approximately 7.4 and 7.3%, respectively, which are lower than the elongation ($\sim 15.3\%$) of the corresponding base metal. The joint

Fig. 7 Optical micrographs of the PWHRed sample H_0 with different rolling reductions: **a** 14%, **b** 32%, **c** 50%

coefficients of sample L_0 and sample H_0 are 87.7 and 82.3%, respectively. Similar results have been reported in the studies [20, 23, 24].

It could be found that the yield strengths of sample L_1 and sample H_1 reach ~ 122 and ~ 109 MPa, respectively. The tensile strength increases by ~ 74.6 and ~ 67.5 MPa, respectively. Elongations of the L_1 and H_1 are about 16.1 and 13.8% when the rolling reduction is 14%, which is lower than that of the corresponding BM (~ 31.8 and 30.9%). The joint coefficients are even increased to

91.9 and 84.4%. When the rolling reduction is 32%, the joint tensile properties and joint coefficient of PWHR gain further improvement, meanwhile the joint of L_2 is slightly higher than that of H_2 . The yield strength, tensile strength, elongation and joint coefficient of the joint L_2 and the joint H_2 are increased to ~ 153 , ~ 278.3 MPa, 17.2, 92.8% and ~ 152 , ~ 276.2 MPa, 17.1, 90.5%, respectively. When the rolling reduction is increased to 50%, the tensile properties and joint coefficient of the joint obtain a further improvement, and the joint of L_3 owns the highest yield

Fig. 8 Relationship between the average grain size and rolling reduction: **a** sample L, **b** sample H

Fig. 9 SEM images at SZ of **a** 14%, **b** 32%, **c** 50% PWHRed AZ31 sheets

Fig. 10 Stress-strain curves of the tensile test: **a** sample L, **b** sample H

strength, tensile strength and joint coefficient, which are ~ 162 , ~ 287.9 MPa and 94.4%, respectively. In addition, the elongation of H₃ reaches a maximum value of 18.7%. Therefore, the sample L₃ achieves the best comprehensive properties, with the yield strength, ultimate tensile strength and elongation of 162, 287.9 MPa, and

17.9%, respectively. Elongation of the H₃ specimen is higher than the other specimens due to its finest grain size.

Before rolling, when the rotational speed keeps constant, the tensile properties of the joint are reduced as the welding speed increases. The welding speed increases from 60 to 80 mm/min, and the tensile strength of the joint decreases

Table 3 Tensile properties of the FSW joints and corresponding base metals (φ = UTS of FSW joints/UTS of BM)

Weld joint	YS (MPa)	UTS (MPa)	EL (%)	BM	YS (MPa)	UTS (MPa)	EL (%)	φ (%)
L ₀	87	190.8	7.4	L ₀ -BM	129	217.5	15.3	87.7
L ₁	122	265.4	16.1	L ₁ -BM	156	288.8	31.8	91.9
L ₂	153	278.3	17.2	L ₂ -BM	162	300.1	34.5	92.8
L ₃	162	287.9	17.9	L ₃ -BM	177	305.1	32.8	94.4
H ₀	91	179.0	7.3	H ₀ -BM	129	217.5	15.3	82.3
H ₁	109	246.5	13.8	H ₁ -BM	159	292.1	30.9	84.4
H ₂	152	276.2	17.1	H ₂ -BM	165	305.0	33.6	90.5
H ₃	158	284.7	18.7	H ₃ -BM	180	308.1	32.5	92.4

from 190.8 to 179.0 MPa. Since the number of rotations of the agitator per unit length decreases as the welding speed increases, the welding heat input decreases accordingly. The yield strengths of samples L₀ and H₀ are much lower than the corresponding yield strength of the parent material. Research [32] has explained that the SZ side is composed of easy to activate extension twinning (EAET) and easy to activate basal slip (EABS) regions. Extension twinning and basal slip are easy to be activated in the two regions, which causes relatively lower yield strength (YS) for the FSW AZ31 alloy.

In general, the mechanical properties of joint are determined by grain size, dislocation density, amount and morphology of second phase and crystallographic texture, etc. The mechanical properties of joint and joint coefficient have been improved gradually with the increase in rolling reduction, which can be rationalized on the basis of several factors.

In addition, the significantly refined grains are obtained due to the occurrence of DRX during the PWHR, which is beneficial to the improvement of the mechanical properties. According to Hall–Petch theory, the yield strength of joint after PWHR is remarkably improved due to grain boundaries strengthening.

Many previous studies indicate that the basal texture can be weakened and modified by tilting an angle from the normal direction (ND) in SZ of the joint by FSW [32]. When the rolling reduction is 14%, the basal texture near the SZ-side region is relatively strong due to the *c* axis of majority grains nearly parallel to the ND direction (Fig. 11), which is positive for tensile strength due to the suppression of extension twinning and basal slips. Combined the effects of above reasons, the strength of the joint obtains a certain degree of improvement. During hardness value tests (compression) along the transverse direction, the strong basal texture in BM is beneficial to activate extension twinning, which hint both grain refinement and texture modification, hence are the positive factors for compression strengthening in SZ, leading to the hardness value in SZ zone is higher than BM.

However, the volume fraction of the β -Mg₁₇Al₁₂ phase decreases because the phase is dissolved into the matrix with the increase in the rolling reduction. This is one reason for the strength of joint reduction. On the other hand, dislocation density decreases due to the occurrence of DRX, which causes a reduction in the strength of joint.

The improvement in the mechanical properties of joint and joint coefficient is mainly attributed to grain refinement and texture strengthening. The decreases of β -Mg₁₇Al₁₂ phase and dislocation density have a little negative effect. Combining the effects of above reasons, the transverse tensile properties and the joint coefficient are significantly improved.

Due to the occurrence of recrystallization, the metal internal dislocation and other defects are eliminated and the grain is refined, resulting in the increase in elongation of the joints during PWHR.

As shown in Fig. 7, the grain sizes of the joints L₀ and H₀ are different. The effect of grain size on yield strength can be assessed by Hall–Petch:

$$\sigma_y = \sigma_0 + kd^{-\frac{1}{2}}, \quad (2)$$

where σ_y is the yield strength of the material, d is the grain size, σ_0 is the total resistance of the lattice dislocation movement, k is a constant which is regarded as 160 MPa $\mu\text{m}^{-1/2}$ [33]. In this study, the yield strength gap of the joints L₀ and H₀ is estimated to be about 4.5 MPa, corresponding to the stress–strain curve in Fig. 9.

3.3 Microhardness

Figure 12 presents the Vickers microhardness profiles measured along the mid-thickness line of the cross section of the joints under different rolling reduction. It can be observed that the hardness value of BM is 55–65 HV under different conditions. As shown in Fig. 12, the hardness value of sample H₀ and sample L₀ in SZ zone is slightly higher than BM before rolling. After rolling, the hardness value of joint ranges from 46 to 72 HV. The result shows conformity with former study [27]. With increasing rolling

Fig. 11 Pole figures near SZ-side region of the sample L_1

Fig. 12 Hardness profiles along the mid-thickness of joints welded in different conditions: **a** sample H, **b** sample L

reduction, the hardness value in SZ zone decreases firstly and then increases.

The change of hardness is mainly related to the grain size, crystallographic texture, dislocation density and the distribution of intermetallic compounds with small particles [34–36]. As AZ31 does not contain particle reinforced phase, the grain refinement plays an important role in improving strength according to the Hall–Petch formula [27]. The hardness values of sample H_0 and sample L_0 in SZ zone are higher than BM. Compared to the BM, the increase in the hardness of the SZ is mainly attributed to the grain refinement and the texture variation of the plate after FSW, which had been discussed former.

On one hand, it can be seen from Fig. 8 that the grain size of the SZ increases when the rolling reduction is 14%. Thus, the hardness of the SZ decreases according to the Hall–Petch formula. On the other hand, as the rolling reduction increases, the basal texture of SZ region gradually becomes stronger which is negative factor for hardness

due to the activation of extension twinning and basal slips. According to the above analysis, the hardness value of the SZ decreases (Fig. 11).

When the rolling reduction is further increased to 32%, the grain size of the SZ reduces. Therefore, according to the Hall–Petch relationship, the hardness values of the SZ increases. However, the hardness value of the sample L_2 in SZ decreases. This result may be related to other strengthening factors. One negative factor is that DRX occurs in the SZ when the rolling reduction is from 14 to 32%, which results in the dislocation density and its hardness value also decreases. Another negative factor is that the basal texture of SZ region becomes stronger. Meanwhile, the amount of $\beta\text{-Mg}_{17}\text{Al}_{12}$ phase in the SZ decreases. For the sample H_2 , the hardness value of the sample H_2 in the SZ increases. The main factor is that the grain size of sample H_2 in the SZ reduces greatly.

According to the above analysis, when the rolling reduction is 50%, the grain sizes of the sample H_3 and

sample L_3 reduce. Meanwhile, the dislocation density and the amount of $\beta\text{-Mg}_{17}\text{Al}_{12}$ phase in the SZ are further reduced, and the former is the main factor, which leads to the hardness values of the sample H_3 and sample L_3 in the SZ increase.

3.4 Fracture Analysis

According to the macroscopic characteristics of the fractured samples, it can be found that all the fracture positions are in the SZ side on AS, which is corresponding to the previous reports [1, 32]. Fracture surfaces of each specimen of sample H are shown in Fig. 13. Figure 13a shows the fracture of the sample H_0 . A large number of tearing ridges, quasi-cleavage planes and few dimples are observed on the fracture surface. It implies that the fracture mechanism of the H_0 specimen is a mixed-type consisting of quasi-cleavage planes and dimples.

Fracture morphology of each specimen of PWHR is shown in Fig. 13b–d. It is found that the number and size

of dimples are obviously increased on the tensile fracture surfaces of the specimens. It means that these samples have good plasticity. The fracture morphology can be inferred that PWHR is regarded as positive effects to enhance the elongation of welded joints. This is consistent with the results of tensile test specimens.

4 Conclusions

The effects of welding speed and post-weld hot rolling on the microstructure and mechanical properties of friction stir-welded AZ31 magnesium alloy were studied. The conclusions are as follows:

1. When the rotational speed is 880 rpm, the welding speed increases from 60 to 80 mm/min, the average grain size of the SZ is decreased from about 6.18 μm to about 5.4 μm . The microstructures of BM, HAZ, TMAZ and SZ of the welded AZ31 alloy changed after each pass. With increasing rolling reduction, the

Fig. 13 SEM images of fracture surfaces of sample H: a H_0 , b H_1 , c H_2 , d H_3

microstructures of different regions become more and more consistent. Ultimately, the average grain size in each region after three passes of hot rolling is almost the same.

2. The mechanical properties of the FSW joints and the joint coefficient can be effectively improved by PWHR. When the rolling reduction is 50%, the gap of tensile strength between FSW joint and corresponding base metal reaches the minimum. The sample L₃ achieves the best comprehensive properties, with the yield strength, ultimate tensile strength and elongation of 162, 287.9 MPa, and 17.9%, respectively.
3. Compared to the BM, the hardness value of the SZ zone is slightly higher before rolling. With increasing rolling reduction, the hardness value in the SZ decreases firstly and then increases.
4. The fracture surfaces of sample H₀ contain a large number of tearing ridges, quasi-cleavage planes and the dimples on the fracture surface indicating that it is a mixed-type fracture mode. With increasing rolling reduction, the number and size of dimples in the tensile fracture surface of specimens increase obviously, and the elongation of welded joints can be effectively improved.

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