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2019

Peng, J., Zhang, Z., Huang, J., Guo, P., Li, Y., Zhou, W., & Wu, Y. (2019). The effect of the inhomogeneous microstructure and texture on the mechanical properties of AZ31 Mg alloys processed by friction stir processing. *Journal of Alloys and Compounds*, 792, 16-24. doi:10.1016/j.jallcom.2019.04.014

<https://hdl.handle.net/10356/144440>

<https://doi.org/10.1016/j.jallcom.2019.04.014>

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The effect of the inhomogeneous microstructure and texture on the mechanical properties of AZ31 Mg alloys processed by friction stir processing

Jinhua Peng ^{a,c}, Zhen Zhang ^{a,b,*}, Ji'an Huang^a, Peng Guo^a, Yaozu Li^a, Wei Zhou ^{c,d},
Yucheng Wu ^{a,b}

^a School of Materials Science and Engineering, Hefei University of Technology, Hefei 230009, China

^b National-Local Joint Research Center of Non-ferrous Metal Materials and Processing Technology, Hefei 230009, China

^c Institute of Industry and Equipment Technology, Hefei University of Technology, Hefei 230009, China

^d School of Mechanical and Aerospace Engineering, Nanyang Technological University, 50 Nanyang Avenue, Singapore

Abstract

Friction stir processing caused obvious grain refinement and texture modification for the experimental AZ31 magnesium alloy. The microstructure, texture as well as mechanical properties were tested for samples cut at different depths from the processed surface of the work-pieces. The micro-hardness in the stir zone was improved to varying extent at different depth. The top layer showed the highest value due to the high density of dislocations. The tensile properties were improved significantly due to the microstructure and texture modification. The texture was weakened in the stir zone, which led to a lower yielding strength and a higher elongation due to the easier activation for basal slip. The true stress-strain curves could be divided into five stages depending on different strain hardening mechanisms, including twinning and dislocation slip. Conventional Hollomon equation was used to describe dislocation related plastic deformation stage while a corrected Arrhenius equation was successfully applied to describe twinning dominated stage.

Keywords: Friction stir processing; Texture; Strength; Strain hardening

1. Introduction

Magnesium (Mg) and its alloys show good properties such as weight saving, electromagnetic shielding and high specific strength, etc. [1,2]. However, they show limited formability, especially at ambient temperature [3,4]. This seriously restricts their widespread application in industry. Grain refinement and texture weakening were widely believed to be efficient ways to improve the ductility for magnesium [5,6]. In recent decades, friction stir processing (FSP), which was categorized as a severe plastic deformation (SPD) method, attracted more attention from people in magnesium field, because it could modify the microstructure and texture, and significantly improve the ductility of magnesium without loss in fracture strength [7-9].

Many previous studies investigated the microstructure and texture evolution of Mg alloys during FSP [10-12]. It was reported that significant grain refinement was realized by dynamic recrystallization (DRX); and in some cases, even nano-grains could be obtained by super cooling routes [13-15]. Meanwhile, strong B-fiber texture usually formed where the (0002) planes tended to be aligned with an ellipsoidal surface in the nugget-shaped stir zone [12,16,17.] The effect of the modified microstructure and texture on the mechanical properties has been studied [18,19]. The elongation was improved significantly due to the texture softening effect [7,9]. However, the previous studies paid more attention onto the stir zone. Outside the stir zone, the relationship between the microstructure, texture and mechanical properties was scarcely studied. The deformation mechanism at different deformation stages also remains unclear.

In the present work, FSP was conducted on hot-extruded AZ31 plates. Optical microscope, electron back-scattered diffraction (EBSD) and X-ray diffraction were used to reveal the microstructure and texture evolution at different depths. Micro-hardness was measured in a scanning way on different layers to reveal the micro-hardness distribution at different zones. Tensile properties were also tested with the tensile sample cut from different depths. The deformation mechanism at different deformation stages was also discussed in details.

2. Experiment procedures

The material used in this work was commercial AZ31 hot-extruded plates with a composition of 3%Al, 1%Zn and balance Mg. The as received plate, 5mm in thickness, showed inhomogeneous grain structure with the grain size ranging from 16 μ m to 300 μ m and an irregular (0002) basal texture, as shown in Figure 1. FSP was conducted using a stirring tool consisted of a Φ 15mm shoulder and a 3.5mm-conical pin. A constant tool rotating speed of 400rpm was adopted with an advancing velocity of 50mm/min.

Optical microscope (OM) was used to observe the microstructure evolution during FSP routines. The samples were polished and etched by a solution of picric-acetic. Electron back-scattered diffraction (EBSD) was conducted to reveal the variation of micro-texture in different locations. Meanwhile, macro-texture was measured on the PD-TD plane (Processing direction-Transverse direction) at different regions using Panalytical X-ray diffractometer equipped with a closed Eulerian cradle in Schulz reflection geometry. Moreover, the dislocation morphology was observed by transmission electron microscope (TEM) on PD-TD plane from the top, middle and bottom layers. The samples were prepared by double-jet electron-polishing with an ethanol-methanol-Mg(ClO)₄-LiCl electrolyte.

Vickers micro-hardness tests were also conducted on the PD-TD plane from AD (Advancing side) to RD (Retreating side), using a 300gf load for 10s. Dog-bone shaped tensile specimens were prepared with a dimension of 16mm in gage length, 4mm in width and 1.5mm in thickness by electrical discharge machining. Tensile specimens were cut from different regions

(Sample1-Sample6) along PD, as shown in Figure2. Tensile tests were conducted at room temperature with a crosshead speed of 1mm/min.

3. Results and discussion

3.1 Microstructure evolution during FSP

The morphology of the stir zone (SZ) from the ND-TD plane was provided in Figure 2d. Clearly, the SZ showed a typical conical shape from ND-TD view, which was consistent with the previous studies [12, 20]. The morphology on PD-TD plane from the top, middle and bottom layer was also examined, as shown in Figure 2a-c. It was clearly seen that the morphologies at different layers showed great differences. At the top layer, the SZ showed the largest width due to the direct contact between the tool shoulder and the work piece. It was interesting to find that the morphology of the bottom layer also changed although the stir pin was not long enough to reach this region. On the other hand, there was a distinctive boundary (as indicated by the yellow arrows) between the base material (BM) and SZ at the retreating side (RD) while the boundary seems to be blurred on the advancing side (AD). This phenomenon indicated that the material flow at the AD was more severe than RD [21,22]. In order to reveal the difference in microstructure at different locations, the microstructure at Point1-Point9 (P1-P9) on different layers was examined by optical microscope, as shown in Figure 3.

It could be clearly seen that the microstructure showed evident difference at different locations (P1-P9). Depending on the microstructure observation, these specific locations could be categorized into two groups: the stir zone-SZ (P2, P5 and P8) and the thermal-mechanical affected zone-TMAZ (P1, P3, P4, P6, P7 and P9). In the SZ (P2, P5 and P8), the grains were refined significantly compared with that of the BM. As shown in Figure 4, the grains size was measured by the linear intercept method where the line was parallel with TD. Clearly, the grain size in P2, P5 and P8 was only 3.6, 6.6 and 3.5 μm , respectively. Such was expected from the broad dynamic recrystallization (DRX) that occurred in stirring region due to severe plastic deformation at an evaluated temperature caused by fraction effect between the stirring pin and work pieces. In addition, it should be noted that the grain size in P2 and P8 was slightly smaller than in P5. The higher cooling rate at the top and bottom surface was responsible for the smaller grain size in P2 and P8.

On the other hand, in the TMAZ (P1, P3, P4, P6, P7 and P9), both small DRX grains and abundant twins were found, and the morphology of these twins were consistent with the $\{10\bar{1}2\}$ twins [23,24]. The original grains were refined due to the twinning and DRX. The grain size in these zones ranged from 15 μm to 19 μm which was higher than that in the SZ, as shown in Figure 4. During FSP, high pressure was imposed on the experimental plate. Since the critical resolved shear stress (CRSS) of $\{10\bar{1}2\}$ twinning was quite low, such twins were easily activated due to the complicated stress state near stirring pin during FSP [25]. Especially for P1, P3, P4 and P6,

plenty of $\{10\bar{1}2\}$ twins formed due to the squeeze force from the stir pin. However, the P7 and P9 were far from the SZ where the squeeze force was lower than that in the other points. Thus, both the density and size of $\{10\bar{1}2\}$ twins in P7 and P9 were smaller than that in the other zones. Meanwhile, DRX also occurred in the TMAZ due to the heat conduction from the SZ. Some small DRX grains were found at the grain boundary or twin boundary in these zones and gave rise to the grain refinement.

3.2 Texture distribution at different zones

For Mg and its alloys, (0002) basal texture often formed after the plastic processing [26,27]. In order to investigate the effect of FSP on the micro-texture distribution, EBSD was conducted on different layers (P2, P5, P8). The EBSD orientation maps and (0002) pole figures were showed in Figure 5, the grains were colored according to their crystal orientation relative to ND. As stated in section2-Experiment procedures, the BM had an irregular (0002) basal texture. After FSP, the (0002) basal texture of each layer was relatively uniform, however they show their own characteristics on different layers.

The top layer showed homogeneous fine grains, as well as an inclined (0002) basal texture. The (0002) basal planes inclined from ND to PD with an angle distributing from 20° to 60° . Similarly, the middle layer also had an inclined (0002) basal texture and some twins were observed. As shown in Figure 2, these two zones (P2 and P5) located in the SZ which underwent severe plastic deformation due to stirring effect of the tool. Thus, during FSP, the (0002) planes tended to be aligned with the conical stir pin and resulted in an inclined basal texture. Interestingly, at the bottom layer, the (0002) basal texture also inclined from ND to PD with a high angle (60° - 90°). Generally speaking, the (0002) planes would align with the interface between the BM and stir pin similar to the condition for the top and middle layers. However, the stir pin was not long enough to reach this zone. During the FSP, the material ahead the stir pin had suffered an onward pressure from the stir pin, a (0002)//ND texture would form immediately whether in top, middle or bottom layer due to $\{10\bar{1}2\}$ twinning. Whereas, the texture at the top and middle layers was modified again due to the rotation of stir pin while the (0002)//ND texture at the bottom layer was retained. Thus, the basal planes at the bottom layer were mostly parallel to ND when the top and middle layers showed the typical inclined basal texture.

On the other hand, in order to understand the distribution of macro-texture, X-ray diffraction was conducted on different areas locating at P2, P3, P5, P8 and P9. The results were shown in Figure 6. It was clearly seen that an inclined (0002) basal texture was formed on the top and middle layers of the SZ (P2 and P5), which was consistent with the EBSD results. The main difference was that the inclined angle from ND to PD was lower than that in EBSD result for the bottom layer. It indicated that the crystal orientation was complex due to the complicated stress state at the bottom layer. Furthermore, the macro-texture distribution located at P3 and P9 was also measured. Compared with the weak basal texture of the BM, the (0002) basal texture was

stronger, especially for P9 where the (0002) planes were mostly parallel to the PD-TD plane. The twinning and DRX would be responsible for the texture variation in these two regions.

3.3 The effect of FSP on the mechanical properties

As described above, both the microstructure and texture were modified after FSP, and different locations showed different characteristics. In order to investigate the effect of microstructure and texture on mechanical properties, firstly, the micro-hardness was measured from AD to RD on top, middle and bottom layers, respectively, as shown in Figure 7. The results invariably showed that the micro-hardness in SZ was higher than that in BM, which might be attributed to the grain refinement effect caused by DRX (well known as Hall-Petch effect). However, the Hall-Petch induced hardness increment on each layer was different. For example, the micro-hardness in SZ at the middle layer was about 50HV which had only 5HV improvement, while the micro-hardness was about 58HV at the top layer which had a 13HV improvement. As shown in Figure 5 and Figure 6, these two layers had similar texture distribution which meant that the crystal orientation was not the main reason for the difference. On the other hand, the variation in grain size at these two zones was not either expected to cause such evident difference in micro-hardness. Therefore, it was reasonable to think that the difference in crystal defects, such as dislocation, might be a most possible reason.

As shown in Figure 8, the dislocations on different layers were observed by TEM. It was clearly seen that the top layer had a high density of dislocation, while on the middle layer only a few dislocations could be seen. During FSP, the top layer always experienced higher plastic strain due to the direct friction from the shoulder, and the heat dissipation rate was usually higher than the central part of work-pieces. Thus, dislocations were expected to retain on the top layer. However, the temperature was high and the heat dissipation rate was low for the middle layer, where the dislocations were devoured by the DRX grains. As a result, the top layer showed a higher micro-hardness than the middle layer due to the higher dislocation density. Similarly, on the bottom layer, the temperature was lower and the heat dissipation rate was also higher than the middle layer, then the dislocation density was also higher than the middle layer. Thus, the micro-hardness at the bottom layer was higher than the middle layer.

In order to further reveal the effect of FSP on mechanical properties, uniaxial tensile tests were conducted on samples at different locations, which were marked in Figure 2 (Sample1-Sample6, S1-S6). The true stress-strain curves were plotted in Figure 9a and the variation of strength and elongation were measured depending on the engineering stress-strain curves, as shown in Figure 9b. It was clearly seen that the BM showed the lowest strength and elongation due to the inhomogeneous grains and the irregular texture. After FSP, the mechanical properties of all samples were improved significantly, but still showed great difference. These samples could be divided into two groups: Group1 (S1, S3, S4 and S6) and Group2 (S2 and S5) where the samples were locating in and besides the SZ respectively (Figure 2).

For the samples in Group1, all of the yield strength (YS), ultimate tensile strength (UTS) and elongation (EI) were improved compared with the BM, especially for the elongation which was about two times of the BM. As shown in Figure 3 and Figure 6, the microstructure in P1, P3, P4, P6, P7 and P9 was refined due to the twinning and DRX, and a stronger (0002) basal texture came into being at these zones. Thus, the YS was improved due to the grain refinement and texture strengthening effect. Meanwhile, the fine grains could accommodate more plastic deformation and resulted in higher elongation.

On the other hand, it was interesting to find that, for the samples in Group2, the YS showed limited difference compared with the BM though the grains were refined to several microns. The YS of S2 was even smaller than that of the BM. **Fortunately, the UTS and EI had evident improvement compared with BM, and the S2 had the highest EI (44.1%) while the S5 had the highest UTS (293.5MPa).** As stated above, an inclined (0002) basal texture formed in the SZ (top, middle and bottom layers) due to the shear deformation around the stir pin (Figure 5 and Figure 6). Thus, the Schmid factor of the basal slip increased significantly. In this way, basal slip was easier to be activated due to the high Schmid factor and low CRSS (critical resolved shear stress) [28]. Then, the yielding occurred at a low flow stress for the S2 and S5. And the YS of S2 was even lower than the BM with coarse grains. Furthermore, basal slip could accommodate more plastic deformation during tensile test, and other deformation mechanisms, such as $\{10\bar{1}2\}$ twinning, $\{10\bar{1}1\}$ twinning and $\bar{a} + \bar{c}$ slip, would be also activated due to grain refinement and texture modification [29]. The combination of these deformation mechanisms caused a quite high elongation for S2 and S5.

3.4 The deformation mechanism during tensile test

As shown in Figure 7 and Figure 9, the mechanical properties showed evident heterogeneity at different locations. The underlying reasons would be the inhomogeneous microstructure and texture introduced by FSP. Then, different deformation mechanisms were activated during the tensile tests. In order to reveal the deformation mechanisms and study the strain hardening behavior, the strain hardening rate (SHR) and strain hardening exponent (SHE) of each sample were calculated and discussed. As shown in Figure 10, the strain hardening rate-true strain curves were plotted separately for samples in Group1 and Group2. Clearly, the SHR of the BM and the samples in Group 1 decreased gradually which was consistent well with the traditional strain hardening law. However, for the S2 and S5, an abnormal trend was found in the evolution of SHR. The SHR firstly decreased to a valley and then it increased to a peak value. After that, it decreased to fracture as usual. It has been reported in the previous studies that, $\{10\bar{1}2\}$ twinning related refinement strengthening and texture strengthening effect were responsible for the abnormal increase in SHR [24,30].

In this study, the texture was modified, and an inclined (0002) basal texture formed in the SZ. The Schmid factors for basal slip, prismatic slip, pyramidal slip and $\{10\bar{1}2\}$ twinning were

calculated according to the EBSD data, as shown in Figure 11. It was clearly seen that plenty of grains owned a high Schmid factor for $\{10\bar{1}2\}$ twinning, especially for the bottom layer. Thus, the $\{10\bar{1}2\}$ twinning also played an important role during the tensile test. The nucleation of the $\{10\bar{1}2\}$ twinning accompanied with stress relaxation brought about the decrease in SHR and led to a valley value. With the extension of the $\{10\bar{1}2\}$ twinning, the grains were divided into many sub-grains and the crystal orientations were rotated from soft orientation to hard orientation. Then, the refinement strengthening and texture strengthening gave rise to the increase of SHR [23].

Depending on the Hollomon equation $\sigma = k\sigma^n$, the $\ln\sigma - \ln\varepsilon$ curves of the BM and S1-S6 were plotted in Figure 12, where the slope was the SHE-n. The $\ln\sigma - \ln\varepsilon$ curves could be divided into different stages due to the specific deformation mechanism. Three stages were found for the BM and the samples in Group1. Generally speaking, tensile curves could be divided two stages: elastic deformation and plastic deformation. For the BM and the samples in Group1, the deformation in Stage1 and Stage3 corresponded to the elastic deformation and plastic deformation respectively, while the Stage2 was the transitional stage between the elastic and plastic deformation. Moreover, it should be noted that, at the Stage3, the slope of the samples in Group1 was lower than the BM which meant that the samples in Group1 had a lower SHE. As shown in Figure 3 and Figure 6, the basal texture was stronger and the grains were refined in the zones where the samples in Group1 came from. Thus, both the basal slip and $\{10\bar{1}2\}$ twinning were suppressed which induced a lower SHE for samples in Group1.

In particular, five stages were found for the samples in Group2. In general, the Stage1 corresponded to the elastic deformation which could not be described by the Hollomon equation. Besides, the lines were almost straight at each stage except for the Stage4 which meant that the deformation at stage4 could not be described by the Hollomon equation either. The strain ranges of stage4 were 4%-19% and 2.5%-9% for S2 and S5, respectively. Interestingly, the strain range at the stage4 was corresponding to the true strain range where the SHR increased gradually, as shown in Figure 10b. As proposed in above, the extension of $\{10\bar{1}2\}$ twinning was the dominated deformation mechanism in this stage. Thus, it seemed that the Hollomon relationship was not suitable for the $\{10\bar{1}2\}$ twinning dominated deformation.

The SHE of each stage was calculated and showed in Table 1. It should be noted that the SHE of the Stage1 and Stage4 were not calculated because the deformation in these stages could not be described by the Hollomon equation. In the Stage2 which corresponded to the yielding stage, the SHE of S2 and S5 was only 0.15 and 0.34, respectively. As reported, basal slip was the dominated deformation mechanism at the very beginning of yielding [23]. Thus, the strain hardening was at a low level and resulted in a low SHE. Entering in Stage3, basal slip was still the dominated deformation mechanism and accompanied with the nucleation of $\{10\bar{1}2\}$ twinning. Thus, the SHE of this stage retained in a low value. However, in the Stage5, the SHE increased significantly and the values for S2 and S5 were 0.71 and 0.59, respectively. It was interesting to find that the

value of SHE in the Stage5 was beyond the general range of n (0.1-0.5). Concerning the specific texture distribution and deformation process, two points were summarized for the abnormally high SHE. Firstly, in the Stage4, $\{10\bar{1}2\}$ twinning rotated the grain orientation from soft orientation to hard orientation which induced a texture strengthening effect. Secondly, as shown in Figure 11, the frequency of grains with high Schmid factor for pyramidal slip was also high at each layer. Meanwhile, after Stage4, the stress increased to a sufficiently high level to activate pyramidal slip, as well as $\{10\bar{1}1\}$ twinning, probably. As a result, the texture strengthening effect and the activation of the multiple deformation mechanisms led to a high SHE at the Stage5. Moreover, the SHE of S2 and S5 also showed some differences due to the inhomogeneous microstructure and texture.

Apparently, the tensile deformation in Stage2-3 and Stage5, which had a constant SHE, could be described by the Hollomon equation. The specific equation for each stage was calculated and plotted, as shown in Table1 and Figure 13. It was clearly that the fitting lines coincided well with the true stress-strain curves for S2 and S5. However, the deformation dominated by the $\{10\bar{1}2\}$ twinning (Stage4) could not be expressed by the Hollomon equation due to the special deformation process. In order to describe the deformation in this stage, the corrected Arrhenius equation $y = y_0 + Ae^{Rx}$ (the y_0 , A and R were the constants associated with the material) was used to fit the strain at stage4. The fitting equations were obtained as follows:

$$\text{Sample2: } \sigma_{IV} = -398 + 489.16e^{1.52\varepsilon} \quad (1)$$

$$\text{Sample5: } \sigma_{IV} = -93.8 + 207.85e^{4.59\varepsilon} \quad (2)$$

Then the fitting lines were plotted in Figure 13 which also coincided well with the true stress-strain curves for S2 and S5 at the Stage4. It indicated that the Arrhenius equation was effective to describe the $\{10\bar{1}2\}$ dominated plastic deformation. The constants y_0 , A and R have to be a strong relation with the microstructure and crystal orientation which would be studied in future.

4. Conclusion

In this study, FSP was conducted on an AZ31 hot-extruded plate. The microstructure, texture, mechanical properties were studied systematically. The results could be summarized as follows:

1. In the SZ, the grains were refined to several microns due to severe plastic deformation. Besides the SZ, plenty of $\{10\bar{1}2\}$ twins and DRX grains were found due to the complex stress state and heat dissipation. The irregular basal texture of the BM was modified, and an inclined basal texture where the (0002) planes inclined from ND to PD was obtained in the SZ due to the shear deformation around the stir pin.

2. The micro-hardness in the SZ was improved to different values at the top, middle and bottom layer. The top and bottom layers had a higher micro-hardness than the middle layer due to

higher dislocation density. The UTS, YS and EI for the samples besides the SZ were all improved due to the grain refinement and texture strengthening. For the samples in SZ, the YS was even slightly lower than BM due to texture weakening effect, **while the widespread basal slip and $\{10\bar{1}2\}$ twinning lead to evident improvement in UTS and EI.**

3. Depending on the strain hardening law, the true stress-strain curves of samples in SZ could be divided into five stages due to the inhomogeneous microstructure and texture. Basal slip was the dominated deformation mechanism at stage2 and stage3 while the deformation at the stage4 was dominated by $\{10\bar{1}2\}$ twinning. Conventional Hollomon equation was used to describe dislocation related strain stage while a corrected Arrhenius equation was successfully applied to describe twinning dominated stage.

Acknowledgements

The authors thank the support by National Natural Science Foundation of China (No. 51871084 and No. 51401072).

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1. Microstructure, texture and mechanical properties in different zone were studied.
2. Inhomogeneous microstructure and texture induced great difference in tensile test.
3. Samples in the stir zone showed five stages during the tensile test.
4. Hollomon and Arrhenius equations were used to describe the tensile deformation.

Table 1 The strain hardening exponent (n), strain hardening coefficient (k) and the Hollomon equation at each stage of Sample2 (S2) and Sample5 (S5).

	Strain of each stage	n	k	Hollomon equation
S2	I : 0%-0.4%	--	--	--
	II : 0.4%-1.14%	0.15	178	$\sigma_{II} = 178 \times \varepsilon^{0.15}$
	III: 1.14%-4%	0.23	252	$\sigma_{III} = 252 \times \varepsilon^{0.23}$
	IV : 4%-19%	--	--	--
	V : 19%-36.1%	0.71	824	$\sigma_V = 824 \times \varepsilon^{0.71}$
S5	I : 0%-0.45%	--	--	--
	II : 0.45%-1.03%	0.34	550	$\sigma_{II} = 550 \times \varepsilon^{0.34}$
	III: 1.03%-2.5%	0.19	281	$\sigma_{III} = 281 \times \varepsilon^{0.19}$
	IV : 2.5%-9%	--	--	--
	V : 9%-21.5%	0.59	916	$\sigma_V = 916 \times \varepsilon^{0.59}$

Figure1

Figure 1 (a) The microstructure and (b) (0002) pole figure of the base material.

Figure2

Figure 2 The macro morphology of the SZ on PD-TD plane ((a) top layer, (b) middle layer and (c) bottom layer) and (d) ND-TD plane.

Figure3

Figure 3 The microstructure in different zones (P1-P9) which was marked in Figure 2.

Figure4

Figure 4 The variation of grain size at different zones (P1-P9).

Figure5

Figure 5 EBSD orientation maps and 0002 pole figures illustrating microstructure and texture at the (a) top layer-P2, (b) middle layer-P5 and (c) bottom layer-P8.

Figure6

Figure 6 (0002) pole figures showed the macro-texture distribution at different zones. The diffraction centers of X-ray were marked in Figure 2.

Figure7

Figure 7 The micro-hardness distribution from advancing side (AD) to retreating side (RD) at different layers.

Figure8

Figure 8 Bright filed TEM images showing the dislocation density at the (a) top layer, (b) middle layer and (c) bottom layer.

Figure9

Figure 9 (a) The true stress-strain curves of the BM and Sample1-Sample6 (S1-S6) and (b) the variation of ultimate tensile strength, yield strength and elongation.

Figure10

Figure 10 The strain hardening rate-true strain curves of (a) the BM and samples in Group1 (S1, S3, S4, S6) and (b) Group2 (S2, S5).

Figure 11 The Schmid factor of the basal slip, prismatic slip, pyramidal slip and $\{10\bar{1}2\}$ twinning depending on the EBSD data on (a) top layer (P2), (b) middle layer (P5) and (c) bottom layer (P8).

Figure12

Figure 12 The $\ln \sigma - \ln \epsilon$ curves of (a) the BM and the samples in Group1 (S1, S3, S4, S6) and (b) Group2 (S2, S5).

Figure13

Figure 13 The true stress-strain curves of sample2 and sample5 and the fitting lines at each stage depending on Hollomon equation and Arrhenius equation.