Effect of extension $\{10\bar{1}2\}$ twins on texture evolution at elevated temperature deformation accompanied by dynamic recrystallization

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Abstract

High temperature deformation processing of magnesium and its alloys is often accompanied by dynamic recrystallization (DRX). Deformation twinning is one of the main deformation mechanisms in HCP metals, but very few works are available in literature (experimental or modeling) which investigate the effect of the deformation twinning on dynamic recrystallization. The current study provides insights regarding the deformation mechanisms of magnesium alloy during compression test at elevated temperatures. The role of the extension twins in DRX is examined using a previously developed dynamic recrystallization model based on the crystal plasticity finite element approach. In order to consider extension twins, an EBSD IPF map with reoriented twinning elements is used as input. An entire element is reoriented into twin orientation and DRX proceeds within the new microstructure. This way, twins can serve as a nucleation site because they provide high angle grain boundaries. The results of compression simulations for extruded AZ31 Mg alloy are presented and validated with the experimental data available in the literature. **The results show twinning when present affects the resultant texture as most of the new nucleating grains are close to the twin matrix**

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interface and have twin orientation.

Keywords: dynamic recrystallization, deformation twinning, crystal plasticity, magnesium alloys

1. Introduction

The limited formability of magnesium and its alloys at room temperature leads to the need in high temperature deformation processing. High temperature forming of cast alloys is inevitably accompanied by dynamic recrystallization (DRX) driven by the stored energy associated with dislocations (Ion et al. (1982)). Magnesium has a hexagonal close-packed crystal structure, and limited slip systems available at room temperature. Deformation twinning is one of the main deformation mechanisms at room temperature and beyond. Extension twins appear when grains' c-axes are placed under tensile loading. While at room temperature extension twinning is a major deformation mechanism, elevated temperature tests reveal very few twinning (Sitdikov and Kaibyshev (2001); Kuc et al. (2008); Al-Samman et al. (2010)). This can be explained by thermal activation of non-basal slip systems, namely, pyramidal $\langle c+a \rangle$ slip system (Barnett (2003); Jain and Agnew (2007); Al-Samman and Gottstein (2008)). Since dynamic recrystallization in magnesium alloys was reported at temperatures as low as 150 °C (Ion et al. (1982)), it is important to study the effect of the deformation twinning on DRX.

There are limited works available on deformation twinning behaviour and its effect on texture evolution during DRX. Martin et al. (2011) assumed that extension twin boundaries are highly mobile, and therefore, that they thicken rather than serve as a recrystallization site. The same assumption was made in the work by Ma et al. (2011), where EBSD analysis showed the presence of twins during an extrusion of Mg AZ61 alloy at 450 °C, emphasizing the possibility of twins at high temperatures. The major suggestion was that at the elevated temperature extrusion, $\langle c+a \rangle$ type dislocations are generated, which can also be favourable for deformation twinning (Morris et al. (2000)). In the work by Al-Samman and Gottstein (2008) it was assumed that the deformation twinning can be a potential nucleation site for DRX, since the stored energy of twinning is higher than in the matrix. However, recrystallization that occurs inside twin lamellae was restricted within twin area, which was observed in the experimental studies of the single crystal, as in the

work by Al-Samman et al. (2012) for example. This observation was also confirmed by the other studies of single crystals under compression and tension at different temperatures (Wonsiewicz and Backofen (1967); Kelley and Hosford (1968a,b)). Although in the same work by Al-Samman and Gottstein (2008) it was also observed extension twins without recrystallization at the end of the deformation, and the reason why twins were left unrecrystallized was unknown, however, the authors suggested that they might have appeared during unloading of the sample.

The strain rate has bigger effect on the activation of twinning during high temperature deformation. Thus, at the strain rates $\dot{\epsilon} < 10^{-1} s^{-1}$, the effect of twinning was negligible at high temperature deformation (Ma et al. (2014)), and some elevated temperature tests also revealed very few twins (Sitdikov and Kaibyshev (2001); Al-Samman and Gottstein (2008); Kuc et al. (2008)). However, Dudamell et al. (2012) reported high amount of twins during uni-axial tension and compression at dynamic strain rates, and it was shown in the work of Li et al. (2013a) that the texture evolution is more sensitive to strain rate at 400 °C in AZ31 Mg alloy. Galiyev et al. (2001) in their work reported active twinning, basal and $\langle c + a \rangle$ slip systems during deformation at < 200 °C in ZK60 Mg alloy. Twins were observed in TEM analysis after hot torsion at temperatures between 180-360 °C and strain rates $10^{-2} s^{-1}$ and $1.0 s^{-1}$ (Myshlyaev et al. (2002)). As the temperature decreased, the twin boundaries became less sharp and resulted in serrated boundaries, which are often the precursors of DRX. Spigarelli et al. (2013) studied high temperature formability of AZ31 between 200 °C and 400 °C. They found that at lower temperatures there was very little DRX and twinning is observed. While at higher temperatures that there was recrystallization but little twinning was observed.

Commercial magnesium alloy AZ31, the main material discussed in this work, has a strong basal texture, and therefore, different slip and twinning systems are active depending on the loading direction. The DRX behaviour of AZ31 Mg alloy has been studied at different temperatures and strain rates via experiments (Myshlyaev et al. (2002); Al-Samman and Gottstein (2008); Srinivasarao et al. (2012)). The experimental results showed that the initial texture has a significant effect on the stress-strain response as well as on the twinning behaviour at different temperatures and strain rates (Barnett (2001); Wang et al. (2011); Dudamell et al. (2012); Li et al. (2013a)). In the work by Srinivasarao et al. (2013), it was concluded that $\{10\bar{1}2\}$ twins appeared at the initial stage of the deformation before DRX started, and

twinning was not a dominant deformation mechanism above 200 °C. Tension and compression tests at the temperature range of 200 °C and 400 °C and strain rates $10^{-2}s^{-1}$ and $10^{-4}s^{-1}$ of AZ31 showed that during 200 °C compression test along extrusion direction (ED), most of the grains undergo twinning (Al-Samman et al. (2010)). The c-axis of HCP crystal was perpendicular to the compressive loading direction, which is favourable for the activation of twinning. A sign of active deformation twinning in the flow curves (plateau-type behaviour and rapid hardening) was present during the channel die compression of pure Mg and AZ31 alloy at 200 °C and 300 °C and the strain rate $10^{-4}s^{-1}$ (Gottstein and Al-Samman (2005); Gehrmann et al. (2005)). Li et al. (2013b) presented experimental analysis that showed the relationship between twinning and DRX. They showed that the $\langle c+a \rangle$ slip enhanced the DRX when twinning is active.

Based upon the review of the aforementioned works on the effect of the deformation twinning on dynamic recrystallization in magnesium alloys, it is concluded that there is a need in a research study of the effect of twinning during high temperature deformation to provide insights of the role of twinning during DRX.

Crystal plasticity models already have been an established tool to model grain level behaviour of the material. Since dynamic recrystallization is a grain scale problem, crystal plasticity is an ideal model to be utilized. The DRX-CPFE model presented in Popova et al. (2014) showed a great potential in capturing the flow curve as well as texture evolution in magnesium alloy AZ31. The present work provides an extended model to include extension twinning, and the effect of twins on the evolution of DRX is investigated at quasi static rates.

The current paper consists of four parts. The DRX-CPFE model formulation was published before and can be found in the previous paper (Popova et al. (2014)), therefore, it is not presented here. However, the important equations modified for incorporation of twinning along with the strategy to account for deformation twinning is presented in the next section. Then, the results of the simulations of extruded Mg alloy AZ31 are compared with experimental data from literature. Finally, observations with respect to the effect of twinning are made and outstanding for future modeling efforts are highlighted.

2. Modelling procedure

The incorporation of the twins within the current model is accomplished by considering the effect of the twin orientation on the texture evolution during DRX. Contraction twins are not taken into account due to their thin size and inability to grow Martin et al. (2011). When extension twins are active, they are known to induce sigmoidal shaped hardening curve (Lou et al. (2007); Khan et al. (2011); Bettles and Barnett (2012)). However, capturing that behaviour of the stress-strain curve is beyond the scope of this study and will not be discussed here. Rather, modelling the softening due to DRX as well as the effect of twinning orientation on microstructure evolution during DRX are the main objectives of the proposed model.

The main idea of the proposed approach is a geometrical incorporation of twins into the discretized microstructure to study the effect of the twin orientations on DRX nuclei.

- The deformation twinning is active and the twin volume fraction (f) evolves until the first DRX step is taken.
- Once the total volume fraction of twins in an element (integration point) reaches the chosen threshold f_{th} , the current element is reoriented into the dominant twin orientation. The volume fractions of the matrix as well as the other twin orientations (non-dominant) are set to zero (Fig. 1).
- DRX proceeds with a new microstructure with twin orientations. Further twin evolution is suppressed.

The strategy of reorientation is similar to the predominant twin reorientation scheme (PTR) introduced by Tome' et al. (1991). Such reorientation leads to the loss of already accumulated other twin orientations and the matrix orientation, which can lead to inaccurate predictions of the final texture. It is not rare that more than one twin modes can be active in magnesium alloys. If one twin mode is much more dominant than the others, the effect from other twin orientations will be small. While if two or more twin modes have comparably similar volume fractions the dominant one will be chosen and the effect of the other orientations will be not considered. Also, such reorientation will lead to abrupt changes in mechanical response. However, the goal of the current work is to capture the orientation of nuclei that would

lead to the correct final DRX texture as well as softening behaviour due to DRX.

The basic assumption is that the nucleated twins do not grow during further deformation and they can act as the potential nucleation sites. Therefore, the grain (consisting of a few elements) can be divided into two grains as presented in Fig. 1b, where M and TW denote matrix and twin respectively, or an entire grain can be reoriented to the twin orientation based on the volume fraction of twins. Another assumption is made regarding the dislocation content of the twins. A rotated element inherits the dislocation content accumulated by that element before the reorientation, since the dislocation density inside twins is still not fully understood in literature, though it is a subject of active research Wang and Agnew (2016).

Figure 1: Schematic representation of the twin incorporation into the mesh

The evolution of the deformation twinning is presented by the power-law equation and it is set to zero once DRX step is taken:

$$\dot{f}_{(\beta)} = \begin{cases} \frac{\dot{f}_{(0)}}{\gamma^{tw}} \text{sgn} \tau_{(\beta)} \left| \frac{\tau_{(\beta)}}{g_{(\beta)}} \right|^{1/m} & \text{if } t < t_{DRX} \\ 0 & \text{if } t \geq t_{DRX} \end{cases} \quad (1)$$

where $\tau_{(\beta)}$ is the resolved shear stress on the slip system β , and it is calculated as $\tau_{(\beta)} = P_{(\beta)} : \sigma$, $g_{(\beta)}$ is the hardness of the slip system.

The Cauchy stress is calculated as a volume average of the twinned part and the matrix Kalidindi (1998):

$$\sigma = \begin{cases} \left(1 - \sum_{\beta=1}^{N^{tw}} f_{(\beta)}\right) \sigma^{mt} + \sum_{\beta=1}^{N^{tw}} f_{(\beta)} \sigma^{tw\beta} & \text{if } t < t_{DRX} \\ \sigma^{mt} \text{ or } \sigma^{tw\beta} & \text{if } t \geq t_{DRX} \end{cases} \quad (2)$$

Once the element is reoriented to a twin orientation, the volume fraction of the matrix is set to zero.

To demonstrate the proposed approach, the simulation was performed on a small specimen and the approach is illustrated in Fig. 3. The initial microstructure, marked initial in Fig. 3, has very similar texture as of the full sample shown in Fig. 4. A representative volume element with the size of $450\mu m \times 450\mu m$ is used in the simulations with 50 elements in each direction (step size of $9\mu m$). Uniaxial compression is applied along the X-axis (radial direction) shown in Fig. 3. This promotes the extension of the c-axis, aligned with extrusion direction (ED) in Fig. 3. The temperature was $200^\circ C$ and the simulation was stopped when the total strain in the specimen reaches about 20%, i.e. just before the DRX begins. Misorientation distribution after reorientation procedures should demonstrate pronounced 86 degrees, which indicates the reoriented twins Fig. 2. During the simulation, the total volume fraction of twins in the simulated microstructure reaches about 50%, and the misorientation angle distribution reveals a high peak of 86 degrees compared to the initial distribution. Notably, the misorientation distribution presented in Fig. 2 mimics that observed experimentally for Mg alloys undergoing twinning Al-Samman and Gottstein (2008).

$$T = \begin{cases} T_{matrix} & \text{if } f_{total} < f_{th} \\ T_{twin\beta} & \text{if } f_{total} \geq f_{th} \end{cases} \quad (3)$$

where T is a rotation matrix of the current element, subscripts *matrix* and *twin β* denote matrix orientation and β twin mode orientation respectively.

The illustration of reorientation procedure is presented in Fig. 3. The EBSD maps show the initial microstructure and after the reorientation is completed as well as the partitioning of the twinned and matrix elements. The partitioning shows which elements were reoriented to the twin orientations. It can be seen that some of the grains were entirely reoriented, while in some only a few elements within the grains were reoriented. The threshold value of the volume fraction of twins for reorientation used in this test simulation is $f_{th} = 60\%$, and DRX proceeds with a "new" microstructure. The value of threshold is chosen such that the twin is not only the most dominant with respect to the other active twins but also has a higher volume fraction than the matrix in that particular element. As outlined earlier this reorientation takes place till the first DRX step and after this point further evolution of the twins is suppressed. The first DRX step (nucleation) is chosen based

Figure 2: Misorientation angle distribution: initial and after reorientation (degrees < 10 are excluded)

on the criteria based on ρ_{cr} , $d\rho_{cr}$ and M . This ensures that there is enough stored energy in the system, represented by ρ , for DRX to happen. $d\rho$ and M restricts the subgrains chosen as a nuclei to a subset that constitutes viable nuclei. The conditions on $d\rho$ and M force a viable nucleus to have driving force and a mobile boundary respectively. The length of the DRX step is linked to the growth of the nucleus and is determined by the grain boundary velocity (M).

The coupled crystal plasticity–probabilistic cellular automata model developed by the authors Popova et al. (2014) is also employed in this work. It has number of parameters that need to be calibrated based on the deformation conditions. The temperature dependent parameters of the model are as follows:

- CRSS values for slip systems, which are calibrated depending on the yield point of the stress-strain curve.
- Critical condition for DRX initiation (Roberts and Ahlblom (1978)):

$$\rho_{cr} = \left(\frac{20S\dot{\epsilon}}{3bLM\tau^2} \right)^{1/3} \quad (4)$$

where S is the grain boundary energy per unit area, M is the grain boundary mobility, $\tau (= c\mu b^2)$ is the dislocation line energy and L is the dislocation mean free path, which is defined as $L = K/c_2\sqrt{\rho}$, the constants K, c are of the order 10 and 0.5 respectively. Since at higher

Figure 3: Example of the reoriented structure: EBSD IPF maps showing initial and reoriented structure and partitioning of the twins and the matrix.

temperatures DRX starts earlier, the value of ρ_{cr} decreased to allow more grains to recrystallize.

- For the DRX nucleus to grow it also has to satisfy a critical value for the stored energy given by:

$$d\rho_{ij} \geq d\rho_{cr} \quad (5)$$

where $d\rho_{ij} = \rho_i - \rho_j$ is the dislocation density mismatch between the elements i and j . The magnitude of the $d\rho_{cr}$ is lowered with increasing temperature providing more potential nucleation sites.

- Grain boundary velocity that depends upon the mobility curve. The mobility follows an Arrhenius relation ship. At higher temperatures, the mobility of the grain boundary is higher, hence, it has higher velocity to travel the size of the element. Consequently, another parameter of the model, Δt , the DRX time step, becomes smaller.

$$M = M_0 \exp \left(-\frac{H_m}{kT} \right) \quad (6)$$

where H_m is the activation enthalpy and M_0 is the pre-exponential factor.

3. Experimental and input data

Al-Samman and coworkers (Al-Samman (2013); Al-Samman and Gottstein (2008, 2007); Al-Samman et al. (2010, 2012)) presented a series tests including uniaxial compression, channel die compression etc. on extruded commercial magnesium alloy at different temperatures. This provided a complete data set for comparison. In the present work, in order to analyze the behaviour of the extension twins during DRX, the data for the orientation labeled CD0ED (c-axis is perpendicular to the extrusion axis), from their work, is chosen. An EBSD map of an extruded AZ31 magnesium alloy is used as input to the DRX-CPFE model, pole figure shows extruded texture (Fig. 4). The original EBSD map is re-sampled, and the step size is $9\mu\text{m}$ in both x and y directions.

Figure 4: Initial texture of the extruded AZ31 Mg alloy (a) Pole figure; (b) EBSD IPF map.

4. Results and Discussions

Compressive load is applied along the extrusion direction (ED), so that c-axes are under a tensile load to activate extension twinning. The parameters for the model are obtained in two steps (similar to Popova et al. (2014)). First, the crystal plasticity model parameters such as CRSS values for slip and twinning systems are calibrated for each simulation temperature. *The calibration of the flow curves (CRSS values) is performed only to fit the yielding point, not the rest of the flow curve as the first peak is*

controlled in this case mostly by the twinning, and subsequent softening behaviour is controlled by the DRX evolution i.e. $\rho_{cr}, d\rho_{cr}$ and M . The correlation between the critical resolved shear stress (CRSS) values for different slip systems and temperatures was investigated in many works, however, the evidence for this relationship is still inconclusive. For the various crystal plasticity models the obtained CRSS values have a broad spread (Lou et al. (2007)), for example, in the work by Barnett (2003), the ratio of CRSS values for intermediate hot working temperatures is suggested to be $\tau_{basal} : \tau_{prismatic} : \tau_{\langle c+a \rangle} : \tau_{twinning} \sim 1 : 6 : 5 : 6$. It is accepted in literature that the CRSS value for deformation twinning is assumed to be temperature insensitive (Barnett (2003); Li et al. (2013b)), therefore, in the current simulations, the CRSS is kept the same for all the temperatures. The CRSS values for the slip systems were varied to get the best match to the experimental data and are shown in fig. 5. The magnitudes of CRSS for different slip systems and the twin system used in the simulations are presented in Table 1.

Table 1: CRSS values for different temperature simulations (in MPa). The strain rate is the same $\dot{\epsilon} = 10^{-4} s^{-1}$

Temperature (°C)	basal	prismatic	pyramidal $\langle a \rangle$	pyramidal $\langle c + a \rangle$	twinning
200	18.0	30.0	45.0	38.0	30
300	6.0	9.0	14.0	14.0	30
400	2.5	4.2	4.2	5.0	30

Table 2: DRX parameters for different temperature simulations. The strain rate is the same $\dot{\epsilon} = 10^{-4} s^{-1}$

Temperature (°C)	K	C	C_1 in ($S = C_1 \theta (1 - \log(\theta))$)	C_0 in ($d\rho = C_0 \rho_{max}$)
200	10.0	1E-4	0.3	0.33
300	10.0	1E-3	0.3	0.4
400	10.0	1E-2	0.3	0.33

The experimental and simulated stress – strain curves for compression test at 200 °C are presented in Fig. 6. The strain rate is taken to be the same for all the simulations ($\dot{\epsilon} = 10^{-4} s^{-1}$). The predicted stress-strain curve

Figure 5: CRSS values for slip systems used in the simulations for different temperatures

shows a good match with experimental data. The peak stress is controlled in this case mostly by the twinning, and the softening behaviour is well captured by DRX-CPFE model.

During the earlier stages of the deformation, the reorientation of the elements is performed once the volume fraction of twins in the element reaches a threshold $f_{th} = 60\%$. The histogram of the number of the reoriented elements at each incremental time step is given in Fig. 7 along with the evolution of the total volume fraction of twins. The total number of the reoriented elements during 200 °C compression simulation is about 70%. In the work of Al-Samman and Gottstein (2007), the texture obtained from the experiments at 200 °C indicated that most of the grains were twinned or twinned grains grew and consumed the other grains. Also, experimental measurements of twins volume fraction in Jain and Agnew (2007) showed that about 50% of the sample was twinned during the compression of AZ31B sheet metal. **The work of Jain and Agnew was for a fine grained material while in this work the grain size is larger. This should result in higher twin volume fraction. This is inline with the observation made in this work.**

The final texture obtained from the simulation at 200 °C shows that the orientations between 0001 and 10 $\bar{1}$ 0 poles have been reoriented during the deformation (Fig. 8). The initial extruded sample had c-axes lying perpen-

Figure 6: Experimental (Al-Samman and Gottstein (2007)) and simulated true stress vs. true strain curves for 200 °C

Figure 7: Histogram of the number of the reoriented elements and the volume fraction of the reoriented elements vs. incremental time step

dicular to extrusion direction (ED) with basal plane normals pointing around the radial direction (see Fig. 4). The texture strength after the simulation did not change significantly, which might be because the nuclei in the model inherit those initially strong orientations at 0001 and $10\bar{1}0$. The slip systems activity plot shows that the deformation is dominated by prismatic and basal slip systems. Pyramidal $\langle c + a \rangle$ slip becomes active due to the twin reorientation (Fig. 9a), since the deformation twinning causes 86 degrees rotation of the basal 0001 planes towards the extrusion direction, and it becomes favourable for pyramidal $\langle c + a \rangle$ slip. Fig. 9b presents the evolution of the volume fraction of the recrystallized elements throughout the deformation. At 200 °C deformation, less than 40% of the elements were recrystallized, which is a good agreement with experimental observations.

Figure 8: Final {0001} pole figure after DRX at 200 °C. (a) From this work (b) is from the work of al Samman (2008).

The experimental and simulated stress-strain curves of the compression along ED at 300 °C and 400 °C are presented in Fig. 10. The predictions, using the parameters listed in tables 1 and 2 obtained from the CPFE-DRX model show a very good match with experimental flow curves.

In the higher temperature simulations, deformation twinning is not observed due to lower CRSS values for non-basal slip systems, compared to the twinning CRSS. During uniaxial compression along ED, prismatic slip is highly favourable as well as twinning and pyramidal $\langle c + a \rangle$ slip. At 300 °C, prismatic slip is the most dominant followed by pyramidal $\langle c + a \rangle$ as shown in Fig. 11a. The CRSS for pyramidal $\langle c + a \rangle$ is lower than CRSS for

Figure 9: Simulation results of 200 °C compression (a) Slip systems activity (pyram1 and pyram2 denote pyramidal a and $\langle c + a \rangle$ systems respectively) (b) Evolution of the volume fraction of DRX.

Figure 10: Experimental (Al-Samman and Gottstein (2007)) and simulated true stress vs. true strain curves for ED compression at 300 °C and 400 °C

twinning, hence most of the strain along the c axis is accommodated by the pyramidal $\langle c + a \rangle$ system. This results in only about 2% of grains undergo twinning. Basal, prismatic and pyramidal $\langle c + a \rangle$ slip systems, during DRX, were reported as being the most active by Srinivasarao et al. (2013). At 400 °C deformation, pyramidal $\langle c + a \rangle$ slip is the most dominant, and the first order pyramidal a is active along with prismatic slip system (Fig. 11b). The simulation results suggest that $\langle c + a \rangle$ is a dominant mechanism at the highest temperatures, but also other dislocation climb based mechanisms may be operative, which are not considered within the CPFEE-DRX model.

Figure 11: Slip systems activity during ED compression simulations at (a) 300 °C (b) 400 °C.

Fig. 12 illustrates the evolution of the volume fraction of recrystallized elements during the compression simulations at 300 °C and 400 °C. At 400 °C the sample was fully recrystallized by the end of the simulation, while at 300 °C 80% of the elements were recrystallized. These results are consistent with those of the other studies by Al-Samman and Gottstein (2008) and suggest that the lower temperature DRX leads to the less fraction of recrystallized grains.

The grain size distribution histograms at different temperatures after DRX are illustrated in Fig. 13. Initial data has bimodal distribution, while at 200 °C the recrystallized grains (35% see Fig. 9b) appear in the first peak of the histogram. However, the rest of the distribution remains the same with lower fraction. At 300 °C the data shows a single peak. After 400 °C it can be seen that the first peak at the position ($\sim 15\mu m$) is smaller. A new peak is developed at a larger grain size of $40\mu m$. This implies that the viable nuclei from the earlier steps have grown into larger size grains, and the grains

Figure 12: Evolution of the volume fraction of recrystallized elements (a) 300 °C (b) 400 °C.

recrystallized later contribute to the first high peak of the histogram plot.

Figure 13: The histogram of the grain size distribution: initial and final after DRX at three simulated temperatures.

In the present model, since the nucleation sites consist of the subgrains, they have the same or similar orientations as the parent grains. The rotation of crystal due to slip alone in magnesium alloys reaches up to ~ 30 degrees, and texture in magnesium alloys is known not to change significantly after DRX compared to FCC metals (al Samman (2008)). The reason is the activation of basal and prismatic slip systems, which accommodate most of the strain during the deformation. Therefore, the final textures are not much different from the initial one after 300 °C and 400 °C compression tests (Fig. 14). The 400 °C pole figure shows the splitting poles that form pyramidal texture, which also was reported in previous studies by Al-Samman and Gottstein (2007).

Figure 14: Final pole figures after DRX at 400 °C (a) from this work (b) from al Samman (2008).

4.1. Effect of twinning on the final microstructure

The main difference in the simulations at 200 °C and those at 400 °C is the presence of twins in the microstructure before DRX step is taken. At 200 °C there is about 70% twinning while at 400 °C there is almost no twinning. Figures 8a and 14a show the polefigures after the deformation. They show that at 200 °C the overall texture is dominated by twin orientation while 400 °C the matrix texture is still the most dominant. The fact that twinning will have a major impact on the texture is obvious, given the strong crystallographic reorientation which accompanies it. What is of greater interest here is the impact of twinning upon the microstructure evolution during dynamic recrystallization. The microstructures at these temperatures are shown as inverse pole figure (IPF) maps in Fig. 15. The subgrain boundaries are shown in white. Most places that show subgrain boundaries are the new nucleated grains growing into the surrounding matrix. These show that at 200 °C the new nucleating grains in the microstructure are close to the twin interfaces and most have twin orientations. While the ones at 400 °C have the same orientation as the matrix grains.

Figure 16 shows the microstructure after 22.5% strain and 33% strain. A comparison clearly shows the location and the orientations of the recrystallized nuclei. The results of the current simulation work, which is based upon sound physical reasoning suggest that twins play a key role in nucleation. It

Figure 15: Final inverse pole figure maps. Grain boundaries ($\theta > 15$) are shown in black and subgrain boundaries ($5 < \theta < 15$) are shown in white. The colors represent the texture.

can be seen that all of the new grains marked in the figures are all close to twin interfaces and preferential have the orientation of the twins. This helps to explain why the cases involving twinning have a texture characteristic of twinning, even though dynamic recrystallization is also active.

There is little doubt that twinning played a very minor role during quasi-static deformation at 400 °C. However, the final texture results obtained at 400 °C call into question the notion that non-basal slip of $\langle a \rangle$ dislocations is so soft relative to basal slip and twinning at high temperatures, despite the suggestion of prior literature (Barnett (2003)) and despite the fact that the predicted flow curves match the experimental well. During compressive deformation, basal slip and twinning will rotate the grains c-axes toward the center of the pole figure, as observed experimentally (note the region of moderate intensity towards to the center of the experimental pole (fig. 14b.)) On the other hand, prismatic (and pyramidal I) slip will stabilize grains with their c-axes at the periphery, as observed in our modeling (figs. 14a and 11b.). The sort of donut in the basal pole figure with low intensity precisely along the compression axis and higher intensity slightly tilted away has previously been explained in terms of pyramidal II (a.k.a. $\langle c+a \rangle$) slip (Agnew et al. (2001)). The current results (Fig. 11b) stress the activity of $\langle c+a \rangle$ slip, yet there is no donut shaped feature in the predicted pole figure. This is because basal slip has not driven c-axes close to the center

Figure 16: Microstructure for the 200 °C simulation showing the location and orientation of the nuclei.

because basal slip is predicted to be essentially inactive (Fig. 11b). This raises questions about how well-suited current crystal plasticity models are for handling problems of high temperature deformation, since there is no explicit accounting for dynamic recovery processes of climb and cross-glide which may alleviate the need for so much non-basal glide of $\langle a \rangle$ dislocations and permit basal glide to be a more active participant. This could represent an attractive direction for future research.

5. Conclusions

The effect of the extension twinning on dynamic recrystallization in magnesium alloy AZ31 was investigated using DRX-CPFE model. The compression test along extrusion direction (ED) is simulated at various temperatures with the $10^{-4} s^{-1}$ strain rate. The following conclusions have been made:

- Since the CRSS values for slip systems are much lower at the elevated temperatures, the extension twinning is not active at temperatures higher than 300 °C at the low strain rates. However, the experimental observations show that at the higher strain rates, twinning can persist even up to 450 °C (Dudamell et al. (2012)).
- The study of slip systems activity shows that during compression along ED of the extruded AZ31 alloy, the majority of plastic deformation is

accommodated by prismatic, basal and pyramidal $\langle c + a \rangle$ slip systems. Simulation results suggest that as temperature rises, the CRSS values for pyramidal $\langle c + a \rangle$ slip decreases and it increases its share of the strain accommodation.

- At 200 °C twinning still has a big impact on the texture evolution. Most of the grains were reoriented into twin orientation, which can be seen from the final pole figures. The experimental data suggests that most of the grains underwent twinning, and the simulations predict that 70% of the grains undergo twinning.
- 300 °C and 400 °C deformation simulations have not revealed active twinning, and the deformation was predicted to have been accommodated by prismatic and pyramidal $\langle c + a \rangle$ slip systems.
- Twinning has a big impact on the texture even after DRX. The resultant texture is closer to the twin orientation.
- When twinning occurs, the model predicts that new nuclei are preferentially nucleated at the twin interfaces and mostly adopt the orientation of twins.

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