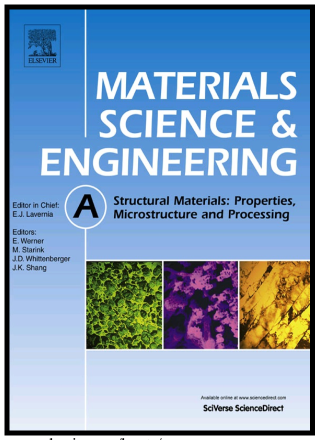
Author's Accepted Manuscript



www.elsevier.com/locate/msea

Negligible effect of twin-slip interaction on hardening in deformation of a Mg-3Al-1Zn alloy

Peng Chen, Bin Li, Duke Culbertson, Yanyao Jiang

PII: S0921-5093(18)30726-3
DOI: https://doi.org/10.1016/j.msea.2018.05.067
Reference: MSA36505

To appear in: *Materials Science & Engineering A*

Received date: 2 April 2018
Revised date: 17 May 2018
Accepted date: 18 May 2018

Cite this article as: Peng Chen, Bin Li, Duke Culbertson and Yanyao Jiang, Negligible effect of twin-slip interaction on hardening in deformation of a Mg-3Al-1Zn alloy, *Materials Science & Engineering A*, https://doi.org/10.1016/j.msea.2018.05.067

Negligible effect of twin-slip interaction on hardening in deformation of a Mg-3Al-1Zn alloy

Peng Chen[a,b], Bin Li[a,b]*, Duke Culbertson[c], Yanyao Jiang[c]

[a] Department of Chemical and Materials Engineering, University of Nevada, Reno, USA

[b] Nevada Institute for Sustainability, University of Nevada, Reno, USA

[c] Department of Mechanical Engineering, University of Nevada, Reno, USA

* Corresponding author. Tel: 775-7840-4507; e-mail: binl@unr.edu

Abstract:

Extruded magnesium alloys with a strong basal texture present a strong tension-compression asymmetry in deformation due to the fact that dislocation slips dominate plastic deformation in tension along the extrusion direction (ED) whereas twinning dominates plastic deformation in compression along the ED. The deformation characteristics allows for a study of the twin-slip interaction by introducing dislocations in the material before activating twinning. In the current work, dislocations of different densities are first introduced by prestraining in tension an extruded Mg-3Al-1Zn alloy along the ED to 5% and 10% of total strain, respectively. Then the loading direction is reversed to compression such that deformation twinning is activated. The results show that the hardening rate after different prestrains remains nearly unchanged, indicating that twin-slip interaction produces negligible effect on strain hardening. The yield stress in the reversed compression increases only slightly with increasing prestrain: ~20 MPa at every 5% increase in prestrain. Electron backscatter diffraction results show that prestraining can retard twin nucleation at the very early stage of plastic deformation, but it has little effect on twin growth.

Keywords: Magnesium alloy; prestrain; twin-slip interaction; strain hardening.

1. Introduction

As the lightest structural metal, magnesium (Mg) alloys are attractive for engineering applications in which lightweight is desired [1]. However, over 80% of Mg components currently in use are cast [2]. Application of wrought Mg is limited due to the poor room temperature formality and the asymmetry of yield behavior associated with texture generated in processing. Because of the low symmetry of the hexagonal-close-packed (hcp) structure, the active deformation modes during plastic deformation of magnesium alloys are highly dependent on the crystal orientation and the loading direction [3]. For instance, dislocation slips dominate plastic deformation in tension along the extrusion direction (ED) of extruded Mg alloys, whereas $\{10\bar{1}2\}\langle 10\bar{1}1\rangle$ twinning is the dominant mechanism in compression [4–7]. Recent measurement of contribution of $\{10\bar{1}2\}$ twinning to plastic strain in an extruded Mg alloy indicated that ~90% of the plastic strain comes from $\{10\bar{1}2\}$ twining during low stress stage deformation [8]. Yield stress and strain hardening behavior are largely dependent on the dominant mechanism of plastic deformation. Hence, extruded Mg alloys with a strong texture exhibit tension-compression asymmetry, i.e. the yield strength under ED compression is lower than that under ED tension. Also, strain hardening rate at the low stress stage is generally lower during ED compression due to the activation of extension twinning.

The influence of prestrain on the flow curve and work hardening behavior has been studied [9–13]. In general, slip-slip [14–17], twin-twin [18–20], and twin-slip interactions [21–23] are contributors to the hardening. Slip-slip interaction plays an important role especially when dislocations glide on multiple slip systems. The contribution to work hardening mainly comes from the increased interactions which give rise to a reduction in dislocation mobility [24]. The effect of twinning on hardening differs from that of slip, showing a dynamic Hall-Petch effect [25] as a result of grain refinement by twin boundaries. Culbertson et al. [26] investigated the effect of pre-compression on microstructure evolution during reversed tension load of an extruded pure magnesium. Irregular shaped secondary twins were observed during reversed tension due to the competition between detwinning of primary twin and propagation of secondary twin. It was found that the amount of sub-grains resulted from secondary twinning was proportional to the amount of prestrain and may be responsible for the increased tensile strength via Hall-Petch effect. Recent work [27] shows that twin-twin interactions also provide an additional strengthening mechanism.

Twin-slip interactions have been investigated both experimentally and theoretically [28–34], but the nature of the interaction is rather complex and is still unclear. Based on lattice correspondence in classical twinning theory, the corresponding slip modes for the major twinning modes in hcp metals were calculated by Niewczas [35] who showed that the dislocation in parent can be incorporated into twins by transformation of planes and directions. The interaction between dislocation slips and twinning can be summarized in two scenarios: (1) dislocations glide and penetrate into the twin lattice and; (2) twin boundaries (TBs) migrate and encompass the dislocations. For the first scenario, Serra et al. [34] suggested that TBs may act as obstacles to subsequent slip, resulting in an increase of hardening. For the second scenario, Proust et al. [36] proposed that dislocations introduced by preloading may act as barriers to twin nucleation or/and twin propagation. El Kadiri et al. [28] suggested that glissile matrix dislocations can be transmuted by twinning and become sessile inside twins. It was also suggested that when lattice dislocations meet with a TB, dislocation dissociation may occur, and a unit twinning dislocation is left behind at the TB, resulting in enlarging or shrinking the twin domain [37].

The purpose of this work is to investigate twin-slip interaction and its effect on strain hardening at macroscopic level based on deformation characteristics of an extruded Mg that slip and twinning can be activated separately by controlling the loading direction with respect to the extrusion direction. The results provide a new insight on the deformation behavior of Mg alloys.

2. Experiments

The material used in this work is a commercially available extruded AZ31B Mg alloy. The initial grain structure and texture of the material are shown in Figure 1. The average grain size of the as-extruded material is approximately 15 μm. The material presents a typical rod-texture in which the (0002) basal pole is nearly perpendicular to the ED, and the $\{10\bar{1}0\}$ prismatic pole presents a strong intensity along the ED (Figure 1). Intensity variation of the basal pole can be seen due to the likely inhomogeneity in the as-extruded material.

Dog-bone shaped specimens with a gage length of 13.0 mm and gage diameter of 9.0 mm were machined from the extruded Mg bar with the specimen axis along the extrusion direction.

An Instron load frame with a loading capacity of ±25 kN was used to conduct the experiments. The specimens were carefully aligned with the loading axis to avoid buckling.

It has been widely observed that the strong texture in the extruded Mg alloys results in tension-compression asymmetry. In compression along the ED, deformation twinning dominates the low stress stage plastic deformation; whereas in tension along the ED, dislocation slips dominate the plastic deformation. Such asymmetric deformation behavior allows us to study the influence of dislocations on twinning experimentally. The extruded dog-bone AZ31B specimens were prestrained by tension along the ED to 5% and 10% total strain, respectively, such that different densities of dislocations were introduced into the specimens. The prestrained specimens were compressed along the ED such that $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ twinning was activated. As such, twin nucleation and growth can be investigated in matrix that has different dislocation densities, and twin-slip interaction can be studied as well. The prestrains of 5% and 10% were chosen based upon our previous testing data [38]. For each prestrain condition, five companion specimens were used for the microstructure observations. Each companion specimen was terminated at a given strain after pre-tension. The mechanical experiments were conducted in a strain-controlled mode at a strain rate of 8×10^{-4} s^{-1} at ambient temperature.

Samples for electron backscatter diffraction (EBSD) scans were prepared after mechanical testing. All the EBSD scans were performed on cross-section surfaces perpendicular rather than parallel to the ED. This generates inverse pole figure (IPF) maps that allows better measurements of twin volume fraction (TVF) in image processing because the parent is reoriented by ~90° after twinning. The sample preparation procedure is as follows: the samples were mechanically ground down to 1200/P4000 grit number on SiC sand papers, followed by electrochemical polishing with a solution of 5% nitric acid, 0.5% perchloric acid and 94.5% ethanol at 20 V for about 20 seconds. EBSD scans were conducted on a JEOL 7100F field emission scanning electron microscope (SEM) with an Oxford HKL Channel 5 instrument. An acceleration voltage of 20 kV, a working distance of 25 mm, and a step size of 1.0 μm were used in the scans.

3. Results

Figure 2 shows the stress-strain curves with 0% pre-tension strain (monotonic compression), 5% pre-tension strain and 10% pre-tension strain. Yield asymmetry can be observed for all three loading conditions. In tension, the flow stresses present a typical dislocation slip dominated deformation, whereas in compression, a low stress stage is observed which is typical of twinning dominated deformation. However, no significant difference can be seen in the yield stress under compression (~200 MPa) irrespective of the amount of tension prestrains.

To better resolve the influence of tension prestrain on the strain-hardening behavior, the stress-strain curves of Figure 2 were re-plotted in Figure 3 such that the compression part is overlapped for all the three deformation conditions. A comparison of stress-strain curves for 0%, 5% and 10% prestrains displays two important features. First, the yield stresses in compression of the prestrained specimens are only slightly higher than the specimen without prestrain. As the prestrain increases, the yield stress in compression slightly increases (by about 20 MPa for every 5% prestrain increase). A recent measurement of twin volume fraction at low stress stage compression of the extruded AZ31 specimens [8] reveals that the main contributor to plastic strain at the very early stage of compression (ε_p < 0.25%) is basal slip. After the compressive plastic strain exceeds 0.25%, the twin volume fraction rapidly increases as the strain increases. During the whole low stress stage deformation, twinning accounts for 80~90% of the plastic strain. Therefore, the slight increase in yield stress in Figure 3 is most likely due to the hardening effect of the preexisting dislocations generated during prestraining. To summarize, a higher tensile prestrain leads to a larger pre-existing dislocation density, which in turn results in a larger yield stress in compression.

Second, and most interestingly, the hardening behavior during the low stress stage of compression across all three loading curves remains almost identical, despite the large difference in prestrains. Although the yield stress slightly increases with increasing prestrain, the trend of the flow stresses remains almost unchanged as the three stress-strain curves nearly fall on top of each other. This unexpected behavior raises a question as to what a role twin-slip interaction plays in the hardening and in the deformation behavior of the extruded AZ31 Mg alloy.

4. Analysis and discussion

The results presented in Figures 2-3 unambiguously show that the interaction between the pre-existing dislocations produced in prestraining to 5% and 10% total strain and the growing extension twins only has a negligible contribution to the hardening at the low stress stage in which $\{10\bar{1}2\}\langle 10\bar{1}\bar{1}\rangle$ extension twinning dominates the plastic deformation. This surprising observation contrasts the literature reports in which finite contributions from twin-slip interaction to hardening were extensively discussed [28,34,36]. The results from the current study suggests that migrating twin boundaries (TBs) do interact with the dislocations, although the detailed twin-slip interactions are unclear. In the following, we characterize the evolution of twin volume fraction (TVF) at selected strain levels of all three types of specimens based on the EBSD results, and discuss in detail possible twin-slip interaction mechanisms.

4.1 Evolution of twin volume fraction

First we examine the effect of prestrains on twin nucleation and growth as a function of plastic strain in compression in all three loading cases. The details on measuring the TVF and calculation of contribution of twinning to plastic deformation were described in [8], and we only provide a brief description here.

Figure 4 shows the evolution of extension twins at six selected plastic strains, 0.25%, 0.71%, 1.37%, 2.47%, 3.72% and 6.65%, for the specimen without prestrain (monotonic compression). Similar results were reported in [8]. For the sake of comparison, the IPF maps were presented to serve as the baseline to reveal the effect of prestrains on twinning behavior. Note that at the plastic strain as low as 0.25%, i.e. shortly after yielding, extension twinning was already activated. The TVF increases as the plastic strain increases. At small strains (Figure 4b-d), the twins appear to be thin plates, but as the plastic strain increases, the twins thicken and coalesce to form large twins such that grains can be totally twinned (Figure 4e-f).

The twinning behavior in the specimens with 5% tension prestrain is shown in Figure 5.

Interestingly, although a 5% prestrain is significant in dislocation-dominated plastic deformation, the twinning behavior displays a similar trend to that without prestrain: the density of twins increases rapidly with increasing strain. Twin nucleation also occurred at very low plastic strains, for example, ε_p = 0.33% (Figure 5a). The twin growth behavior is similar to that of the zero prestrain. Twinning almost saturated at ε_p = 6.70% (Figure 5f).

In stark contrast, for the specimens with 10% prestrain, a substantial difference in the twinning behavior can be observed, which is shown in the IPF maps in Figure 6. At ε_p = 0.31% (Figure 6a), almost no extension twins can be seen. When the plastic strain is increased to ε_p = 0.74% (Figure 6b), few slender twins at the early stage can be observed. Compared to the scenarios in 0% and 5% prestrains, the number of twins is much fewer. Even after the plastic strain is increased to ε_p = 1.22% (Figure 6c), the number of twins only slightly increases but still remains very low. As the plastic strain is further increased to ε_p = 2.18% (Figure 6d), the density of the extension twins increases but lower than that of the specimens of 0% and 5% prestrains. At ε_p = 3.58% (Figure 6e), twin growth and coalescence become dominant. It is clear that twin nucleation is retarded by the much higher density of dislocations generated by the 10% prestrain.

Quantitatively measured evolution of TVF (f_{twin}) as a function of plastic strain is shown in Figure 7. In this graph, at least three measurements were performed at each plastic strain levels and the averaged value was plotted. The effect of prestrains on the TVF can now be better observed. The most significant influence comes at the low plastic strain levels. Although the evolution trend of TVF of the specimens with 0% prestrain is close to that of the specimens with

5% prestrain, at strains $\varepsilon_p < 2.34\%$, it can be seen that the TVF is slightly lower for the specimens with 5% prestrain. When the prestrain is increased to 10%, the curve is further pushed to the right side. It can be seen that the TVFs at $\varepsilon_p = 0.31\%$, 0.74% and 1.22% for the specimens with 10% prestrain are much lower than the specimens with 0% and 5% prestrains. At these low strains the overall TVF is below 5%, which strongly suggests that twinning activity especially twin nucleation is retarded at the low strains. However, for both 5% and 10% prestrains, the effect of prestrain on TVF becomes insignificant at higher strain levels.

Following the quantitative measurement of the TVFs at various strain levels for all three prestrains, we calculated the contribution of extension twinning to the plastic strain. Details of the calculation method were described in [8], in which the Schmid law was not used because $\{10\bar{1}2\}\langle10\bar{1}\bar{1}\rangle$ twinning actually does not obey the Schmid law [8,39] due to the breakdown of the invariant plane strain condition [40]. Our calculation is based on the fact that for single crystal Mg, if a tensile load is applied along the *c*-axis, an elongation of 6.7% is produced by twinning after the crystal is totally twinned, i.e. the parent lattice is reoriented by nearly 90° around the zone axis [41]. Thus, a misfit strain $\varepsilon_{misfit} = 6.7\%$ is produced. For the highly textured Mg specimens, the contribution of extension twinning to the plastic strain can be expressed as $\varepsilon_{twin} = f_{twin} \times \varepsilon_{misfit}$ (f_{twin} denotes the TVF) [8].

The contribution of extension twinning to the plastic strain, expressed as $\varepsilon_{twin}/\varepsilon_p$, for the three types of prestrained specimens is plotted with respect to the plastic strain, as shown in Figure 8. Compared with the specimens without prestrain, the curves for 5% and 10% prestrains are shifted to the right. A general trend can be observed on these curves. At the beginning of the plastic deformation when the plastic strains are low (ε_p < 1.0%), the contribution of twinning is insignificant. Thus, dislocation slip dominates plastic deformation. However, after the plastic strain is increased to ε_p > 1.0%, for the 0% and 5% prestrains, the contribution of extension twinning rapidly increases and becomes the dominant mechanism in the low stress stage deformation. But in sharp contrast, for the 10% prestrain, the contribution of extension twinning to the plastic strain remains very low ($\varepsilon_{twin}/\varepsilon_p$< 20%) at ε_p < 1.22%. At ε_p = 2.18%, the contribution increases to about 35%. As the plastic strain further increases, the contribution from extension twinning to the plastic strain rapidly increases to the level comparable to the 0% and 5% prestrains.

4.2 Retardation of twin nucleation by dislocations

The IPF maps (Figures 4-6) and the quantitative analyses of TVF evolution and its contribution to the plastic strain (Figures 7-8) clearly reveal that the pre-existing dislocations generated by the prestrains only influence nucleation of extension twins. There is almost no influence on twin growth and hence work hardening.

Mahajan [42] investigated how prestraining affects twin formation in iron prior to shock-loading. The annealed iron specimens were first compressed at room temperature to various strains up to 20%, no twin was observed in these prestrained specimens. Then the prestrained specimens were shock loaded. It was found that: (1) the number of twins decreases with increasing amount of prestrain; (2) a twin, once nucleated, is able to propagate through a grain, even in heavily prestrained specimen. The influence of prestrain on twinning behavior in niobium (Nb) single crystals was investigated by Boucher and Christian [43]. Two groups of specimens were prestrained at room temperature and 158 K to introduce dislocations with different distribution and density. It was found that prestrains at both room temperature and 158 K were effective in suppressing twinning at temperatures down to 77 K. They speculated that the prestrain may increase twinning stress, and homogeneously distributed dislocations can effectively suppress twin nucleation. They also pointed out that the suppression in twinning was highly dependent on dislocation density. Similar study on the effect of dislocation density on twinning behavior in tantalum (Ta) was conducted by Florando et al. [44]. The annealed polycrystalline Ta specimens were first prestrained by compression to different strain levels at room temperature with a strain rate of $10^{-3}/s$ to introduce dislocations of different density, then the specimens were reloaded at 77 K and $1/s$ to activate twinning. It was found that the specimens with higher amounts of prestrain showed a higher dislocation density and a higher yield stress. Twins were only observed in specimens prestrained less than 1%, and almost no twins appeared in specimens prestrained to ≥ 2%. This indicates that an increase in dislocation density suppresses twinning, resulting in the increase of the stress for twin nucleation.

These previous results imply that prestrain only suppresses twin nucleation and its effect on twin propagation is relatively insignificant. Despite the differences in crystal structure and possibly in twinning mechanism, our observations of the effect of prestrains on twinning behavior in AZ31 appear to be consistent with previous reports. When the specimens are prestrained by tension along the ED, basal and non-basal dislocations are expected to be generated. In subsequent compression, these pre-existing dislocations retard twin nucleation. Prestrain also increases the stress for twin nucleation, although only slightly (Figure 3).

4.3 Possible twin-slip interaction mechanisms

(1) Dynamic Hall-Petch effect: Twin boundaries act as new grain boundaries that reduce the effective grain size and the mean free path of dislocations. Twin-slip interaction has been considered as an important contributor to the increase in hardening rate during twinning and after exhaustion of twinning [28,29,34,45]. Dynamic Hall-Petch effect has long been considered an important factor that contributes to the superior mechanical properties of TWIP (twinning induced plasticity) steels. Bouaziz and Guelton [46] proposed a work hardening model that takes into account the dynamic Hall-Petch effect introduced by twinning. It is assumed that twin boundaries act as impenetrable obstacles for gliding dislocations, decrease the mean free path of dislocations, and hence increase the hardening rate.

However, Liang et al. [47] argued that the contribution of twinning to hardening in TWIP steel is overestimated in the literature. They quantitatively calculated the contribution of twins and dislocations to work hardening rate with synchrotron X-ray diffraction (XRD). According to their results, dislocations account for ~60% of total flow stress and ~90% of the flow stress after yielding, whereas twins contribute only ~10% to the flow stress after yielding, indicating that the direct contribution of twinning to hardening should be insignificant. More recently, Luo and

Huang [48] deformed a TWIP steel at 373 K, 473 K and room temperature. They observed that the stress-strain curves are almost parallel and the hardening rates are comparable at these temperatures. However, deformation twins are significantly suppressed at 373 and 473 K as compared to room temperature, and the corresponding twin volume fractions are less than 1% at 10% true strain. This indicates that the formation of deformation twins actually has little contribution to the high hardening rate in TWIP steels, hence dynamic Hall–Petch effect is ineffective.

Kalidindi et al. [49] investigated the strain hardening behavior and microstructure evolution in compression of high purity titanium at room temperature. Three distinct hardening regimes were observed. Particularly, an increase in hardening rate at stage B is attributed to deformation twinning which gives rise to dynamic Hall–Petch mechanism with reduced effective slip distance. A crystal plasticity model [50] was developed by incorporating this effect. In more recent work, Kalidindi et al. [51] quantitatively measured the contribution of extension twins to strain hardening in AZ31 Mg. They showed that the hardening caused by extension twinning is only about 0.03G (G is elastic modulus), which only accounts for a very small portion of peak strain hardening rate (~0.3G). Comparison of hardening rates between strong textured AZ31 and α-Ti shows that the strain hardening rate in AZ31 is much higher than that in α-Ti. Consequently, extension twins should not be responsible for the high hardening rate in AZ31. They also concluded that the dynamic Hall-Petch effect caused by extension twins in Mg alloys is ineffective in hardening, because extension twins grow very quickly and consume the entire grain, hence extension twins did not effectively refine the grain size. Instead, contraction twins might play a significant role in hardening. Recent work by Wu et al. [52] also shows that elastic deformation mostly accounts for the sharp increase in hardening after the exhaustion of twining in a Mg alloy.

These reports suggest that the dynamic Hall-Petch effect induced by twinning, particularly $\{10\bar{1}2\}$ twinning which can grow and entirely consume parent grains, may not be a significant contributor to hardening. This is consistent with our observations as well.

(2) The Basinski effect and dislocation transmutation: As an alternative mechanism for strain hardening resulting from deformation twinning, Basinski et al. [53] proposed that glissile dislocations in parent would be transformed to sessile dislocations by twinning, and result in increase in strength as measured in microhardness of twins. In twinning of fcc Cu, the twinning plane is $(1\bar{1}1)$ twinning plane and the direction of twinning shear is along $\frac{1}{6}[112]$. The lattice correspondence for slip systems shows that slip dislocations with a Burgers vector of $1/2\,[101]$ lying on the $(1\bar{1}\bar{1})$ and $(11\bar{1})$ planes in parent will transform to sessile dislocations on $\{200\}$ planes. As a result, the twinned lattice is expected to be harder than the parent and contributes to the overall hardening.

In light of the Basinski mechanism, El Kadiri et al. [28,33] proposed that dislocation transmutation may take place during twin-slip interaction. As a result of transmutations, the number and type of dislocations in twins are multiplied, resulting in increased latent hardening [28]. Wang et al. [29,33] observed dislocations in the vicinity of $\{10\bar{1}2\}$ twin boundaries in ex-situ TEM characterization. These dislocations lie on the basal plane of the twin, but with a $\langle c \rangle$ component. To explain how these dislocations were generated, they suggested that transmutation of the basal $\langle a \rangle$ dislocations in the matrix took place. However, according to classical twinning theory [4] and lattice correspondence analysis [35], a basal dislocation in matrix must be transformed to a prismatic dislocation in a $\{10\bar{1}2\}$ twin. Thus, the conclusion that the dislocations in the twin were generated by transmutation is questionable.

In contrast to these reports, Hiura et al. [54] reported that the Basinski hardening mechanism is actually ineffective in $\{10\bar{1}2\}$ twinning in Mg. They conducted nano-indentation measurements and found that the hardness difference between a twin and the surrounding matrix is insignificantly small. It seems that more investigations are need to resolve the controversies.

(3) Dislocation-TB interaction: In cubic metals, TBs remain highly coherent during twinning and the boundary plane is one of the slip planes of the matrix. Twinning dislocations can be well-defined on the twinning plane. When a matrix dislocation impinges on a TB, the dislocation may transmit through the TB and cross slip in the twin, or interact with the TB via dislocation reaction, or be hindered by the TB [30,55,56]. However, interaction between matrix dislocations and $\{10\bar{1}2\}$ TBs can be distinctively different.

Tomsett and Bevis [57] reported that in polycrystalline zinc, the $1/3\,[11\bar{2}0]$ dislocations can penetrate $\{10\bar{1}2\}$ TBs and decompose into two Shockley partial dislocations, and result in the formation of stacking faults between the leading and trailing partial dislocations in twins. Capolungo et al. [58] investigated the interaction between prismatic dislocations and twins in hcp

Zr by controlling the deformation temperature. In some of their experiments, the specimens were prestrained at various strain levels at room temperature to introduce prismatic dislocations, then deformed in liquid nitrogen to introduce twins. It was found that the pre-induced prismatic dislocations have almost no impeding effect on the propagation of $\{10\bar{1}2\}$ deformation twinning, but increase the resistance for $\{11\bar{2}2\}$ twin growth. However, they do have negative effect on $\{10\bar{1}2\}$ twin nucleation, but do not change the onset of $\{11\bar{2}2\}$ twinning. It is worth noting that the twinning mechanisms of $\{10\bar{1}2\}$ and $\{11\bar{2}2\}$ are very different [59].

Xin et al. [10] investigated the effect of dislocation-twin interaction on deformation behavior by conducting pre-compression along the TD to introduce $\{10\bar{1}2\}$ twinning followed by tension along the RD to introduce dislocations in a hot-rolled AZ31 alloy. Their results show that interaction between dislocations and TBs can reduce the peak hardening rate, indicating that the impedance of the TBs on dislocations is weak. “Basal slip transmission” across twin boundaries in Mg was reported by using in-situ SEM [60].

Recently, Li and Ma [61] showed that $\{10\bar{1}2\}$ twinning in Mg alloy is actually mediated by atomic shuffling. A large deviation between the actual twin boundaries and the theoretical $\{10\bar{1}2\}$ twinning plane has been frequently observed [40] in HRTEM. More recently, Li and Zhang [62] showed that the twinning shear of $\{10\bar{1}2\}\langle 10\bar{1}1\rangle$ mode can only be zero because the twinning plane is not an invariant plane and thus no twinning dislocations should be involved. In fact, most $\{10\bar{1}2\}$ TBs in HRTEM and atomistic simulations are highly incoherent. Thus, the mechanistic difference in twinning may give rise to very different twin-slip interaction in hcp metals from cubic metals. Despite the extensive discussions of possible twin-slip interaction mechanisms, key questions remain unanswered: (1) given the fact that $\{10\bar{1}2\}$ TBs are mostly incoherent, as observed in extensive experiments and simulations [40,63], it can be envisaged that these incoherent TBs are hard obstacles for most matrix dislocations to penetrate through. Additionally, due to the lack of twinning dislocations on the TBs [62,64,65], can $\{10\bar{1}2\}$ TBs actually act as a dislocation sink? If this is the case, then the negligible contribution of twin-slip interaction to hardening can be reasonably well explained. (2) It should not be expected that all types of Burgers vectors of matrix dislocations can be transformed/transmuted by migrating TBs. Such transformations may be dependent on the actual directions of the Burgers vectors with respect to the zone axis between parent and twin. The question is how such transformations exactly occur? Obviously, post-mortem observations can only provide very limited insight on such important interaction which takes place on the atomic scale. To answer these critically important questions and fully understand twin-slip interaction, in-situ transmission electron microscopy combined with atomistic simulations may be required.

5. Conclusions

In this work, we take the advantage that twinning and dislocation slip can be separately introduced into an extruded AZ31B Mg alloy by controlling the loading direction and study the effect of twin-slip interaction on the strain hardening behavior. The following conclusions can be reached:

(1) Twin-slip interaction produces negligible effect on hardening. Even with large tension

prestrains, the hardening rate at the low stress stage during compression along the ED remains almost unchanged, indicating that the contribution of twin-slip interaction to hardening is negligible in deformation of Mg alloys. However, the mechanism remains unclear at this time. More research on the atomic scale is required.

(2) The yield stress in compression along the ED only slightly increases with increasing prestrain. The magnitude of increase is only ~20 MPa at every 5% increase in prestrain.

(3) Twin volume fraction measurements show that, after yielding, twin nucleation is suppressed by the pre-existing dislocations. However, the twin volume fraction catches up as the strain increases, indicating that twin growth is hardly affected by the prestrains.

(4) Due to the retardation of twin nucleation, the contribution of twinning to the plastic strain at the early stage after yielding is reduced as the prestrain increases, but catches up at higher strains as twin growth re-dominates the plastic deformation.

Acknowledgements

Bin Li gratefully thanks support from the U.S. National Science Foundation (NSF) (CMMI-1635088). Yanyao Jiang acknowledges support from the NSF (CMMI-1462885).

References:

[1] T.M. Pollock, Weight Loss with Magnesium Alloys, Science. 328 (2010) 986–987. doi:10.1126/science.1182848.

[2] D. Griffiths, Explaining texture weakening and improved formability in magnesium rare earth alloys, Mater. Sci. Technol. 31 (2015) 10–24. doi:10.1179/1743284714Y.0000000632.

[3] E. Kelley, W. Hosford, Plane-strain compression of magnesium and magnesium alloy crystals, Trans Met Soc AIME. 242 (1968) 5–13.

[4] J.W. Christian, S. Mahajan, Deformation twinning, Prog. Mater. Sci. 39 (1995) 1–157.

[5] M.H. Yoo, J.K. Lee, Deformation twinning in hcp metals and alloys, Philos. Mag. A. 63 (1991) 987–1000.

[6] M.H. Yoo, Slip, twinning, and fracture in hexagonal close-packed metals, Metall. Trans. A. 12 (1981) 409–418.

[7] M.R. Barnett, Twinning and the ductility of magnesium alloys, Mater. Sci. Eng. A. 464 (2007) 8–16. doi:10.1016/j.msea.2007.02.109.

[8] P. Chen, B. Li, D. Culbertson, Y. Jiang, Contribution of extension twinning to plastic strain at low stress stage deformation of a Mg-3Al-1Zn alloy, Mater. Sci. Eng. A. 709 (2018) 40–45. doi:10.1016/j.msea.2017.10.038.

[9] S. Zhao, E. Guo, L. Wang, T. Wu, Y. Feng, Effect of pre-compressive strain on microstructure and mechanical properties of Mg–2.7Nd–0.4Zn–0.5Zr alloy, Mater. Sci. Eng. A. 647 (2015) 28–33. doi:10.1016/j.msea.2015.08.092.

[10] Y. Xin, L. Lv, H. Chen, C. He, H. Yu, Q. Liu, Effect of dislocation-twin boundary interaction on deformation by twin boundary migration, Mater. Sci. Eng. A. 662 (2016) 95–99. doi:10.1016/j.msea.2016.03.061.

[11] T. Hama, H. Nagao, Y. Kuchinomachi, H. Takuda, Effect of prestrain on work-hardening behavior of magnesium alloy sheets upon cyclic loading, Mater. Sci. Eng. A. 591 (2014) 69–77. doi:10.1016/j.msea.2013.10.083.

[12] T. Hama, Y. Kariyazaki, N. Hosokawa, H. Fujimoto, H. Takuda, Work-hardening behaviors of magnesium alloy sheet during in-plane cyclic loading, Mater. Sci. Eng. A. 551 (2012) 209–217. doi:10.1016/j.msea.2012.05.009.

[13] J. Zhang, G. Xi, X. Wan, C. Fang, The dislocation-twin interaction and evolution of twin boundary in AZ31 Mg alloy, Acta Mater. 133 (2017) 208–216. doi:10.1016/j.actamat.2017.05.034.

[14] T. Nogaret, W.A. Curtin, J.A. Yasi, L.G. Hector Jr, D.R. Trinkle, Atomistic study of edge and screw 〈c + a〉 dislocations in magnesium, Acta Mater. 58 (2010) 4332–4343. doi:10.1016/j.actamat.2010.04.022.

[15] B. Li, Q.W. Zhang, S.N. Mathaudhu, Basal-pyramidal dislocation lock in deformed magnesium, Scr. Mater. 134 (2017) 37–41. doi:10.1016/j.scriptamat.2017.02.040.

[16] J. Geng, M.F. Chisholm, R.K. Mishra, K.S. Kumar, The structure of 〈c + a〉 type

dislocation loops in magnesium, Philos. Mag. Lett. 94 (2014) 377–386. doi:10.1080/09500839.2014.916423.

[17] F.F. Lavrentev, The type of dislocation interaction as the factor determining work hardening, Mater. Sci. Eng. 46 (1980) 191–208. doi:10.1016/0025-5416(80)90175-5.

[18] Q. Yu, J. Wang, Y. Jiang, R.J. McCabe, C.N. Tomé, Co-zone 10-12 Twin Interaction in Magnesium Single Crystal, Mater. Res. Lett. 2 (2014) 82–88. doi:10.1080/21663831.2013.867291.

[19] H. El Kadiri, J. Kapil, A.L. Oppedal, L.G. Hector, S.R. Agnew, M. Cherkaoui, S.C. Vogel, The effect of twin–twin interactions on the nucleation and propagation of {10-12} twinning in magnesium, Acta Mater. 61 (2013) 3549–3563. doi:10.1016/j.actamat.2013.02.030.

[20] Q. Yu, J. Wang, Y. Jiang, R.J. McCabe, N. Li, C.N. Tomé, Twin–twin interactions in magnesium, Acta Mater. 77 (2014) 28–42. doi:10.1016/j.actamat.2014.05.030.

[21] L. Lv, Y. Xin, H. Yu, R. Hong, Q. Liu, The role of dislocations in strain hardening of an extension twinning predominant deformation, Mater. Sci. Eng. A. 636 (2015) 389–395. doi:10.1016/j.msea.2015.04.007.

[22] C.M. Cepeda-Jiménez, J.M. Molina-Aldareguia, M.T. Pérez-Prado, Origin of the twinning to slip transition with grain size refinement, with decreasing strain rate and with increasing temperature in magnesium, Acta Mater. 88 (2015) 232–244. doi:10.1016/j.actamat.2015.01.032.

[23] T. Ezaz, M.D. Sangid, H. Sehitoglu, Energy barriers associated with slip–twin interactions, Philos. Mag. 91 (2011) 1464–1488. doi:10.1080/14786435.2010.541166.

[24] R. Madec, B. Devincre, L. Kubin, T. Hoc, D. Rodney, The Role of Collinear Interaction in Dislocation-Induced Hardening, Science. 301 (2003) 1879–1882. doi:10.1126/science.1085477.

[25] S. r. Kalidindi, A. a. Salem, R. d. Doherty, Role of Deformation Twinning on Strain Hardening in Cubic and Hexagonal Polycrystalline Metals, Adv. Eng. Mater. 5 (2003) 229–232. doi:10.1002/adem.200300320.

[26] D. Culbertson, Q. Yu, J. Wang, Y. Jiang, Pre-compression effect on microstructure evolution of extruded pure polycrystalline magnesium during reversed tension load, Mater. Charact. 134 (2017) 41–48. doi:10.1016/j.matchar.2017.10.003.

[27] P. Chen, J. Ombogo, B. Li, Interaction between primary {10-12} twins and formation of mobile 60°〈01-10〉 boundaries in deformation of a magnesium alloy, (2018) (Submitted for publication).

[28] H. El Kadiri, A.L. Oppedal, A crystal plasticity theory for latent hardening by glide twinning through dislocation transmutation and twin accommodation effects, J. Mech. Phys. Solids. 58 (2010) 613–624. doi:10.1016/j.jmps.2009.12.004.

[29] F. Wang, S.R. Agnew, Dislocation transmutation by tension twinning in magnesium alloy AZ31, Int. J. Plast. 81 (2016) 63–86. doi:10.1016/j.ijplas.2016.01.012.

[30] Y.T. Zhu, X.L. Wu, X.Z. Liao, J. Narayan, L.J. Kecskés, S.N. Mathaudhu, Dislocation–twin

interactions in nanocrystalline fcc metals, Acta Mater. 59 (2011) 812–821. doi:10.1016/j.actamat.2010.10.028.

[31] Y. Xin, L. Lv, H. Chen, C. He, H. Yu, Q. Liu, Effect of dislocation-twin boundary interaction on deformation by twin boundary migration, Mater. Sci. Eng. A. 662 (2016) 95–99. doi:10.1016/j.msea.2016.03.061.

[32] K.D. Molodov, T. Al-Samman, D.A. Molodov, Profuse slip transmission across twin boundaries in magnesium, Acta Mater. 124 (2017) 397–409. doi:10.1016/j.actamat.2016.11.022.

[33] F. Wang, K. Hazeli, K.D. Molodov, C.D. Barrett, T. Al-Samman, D.A. Molodov, A. Kontsos, K.T. Ramesh, H. El Kadiri, S.R. Agnew, Characteristic dislocation substructure in {10-12} twins in hexagonal metals, Scr. Mater. 143 (2018) 81–85. doi:10.1016/j.scriptamat.2017.09.015.

[34] A. Serra, D.J. Bacon, R.C. Pond, Twins as barriers to basal slip in hexagonal-close-packed metals, Metall. Mater. Trans. A. 33 (2002) 809–812. doi:10.1007/s11661-002-0149-7.

[35] M. Niewczas, Lattice correspondence during twinning in hexagonal close-packed crystals, Acta Mater. 58 (2010) 5848–5857. doi:10.1016/j.actamat.2010.06.059.

[36] G. Proust, C.N. Tomé, A. Jain, S.R. Agnew, Modeling the effect of twinning and detwinning during strain-path changes of magnesium alloy AZ31, Int. J. Plast. 25 (2009) 861–880. doi:10.1016/j.ijplas.2008.05.005.

[37] S. Lay, G. Nouet, HREM study of the (0112) twin interface in zinc, Philos. Mag. A. 70 (1994) 261–275. doi:10.1080/01418619408243184.

[38] P. Chen, B. Li, D. Culbertson, Y. Jiang, Twin-Slip Interaction at Low Stress Stage Deformation in an AZ31 Mg Alloy, in: Magnes. Technol. 2018, Springer, Cham, 2018: pp. 193–198. doi:10.1007/978-3-319-72332-7_30.

[39] M.Z. Bian, K.S. Shin, {10-12} twinning behavior in magnesium single crystal, Met. Mater. Int. 19 (2013) 999–1004. doi:10.1007/s12540-013-5012-4.

[40] X.Y. Zhang, B. Li, X.L. Wu, Y.T. Zhu, Q. Ma, Q. Liu, P.T. Wang, M.F. Horstemeyer, Twin boundaries showing very large deviations from the twinning plane, Scr. Mater. 67 (2012) 862–865. doi:10.1016/j.scriptamat.2012.08.012.

[41] B. Li, X.Y. Zhang, Global strain generated by shuffling-dominated {101-2}<10-1-1> twinning, Scr. Mater. 71 (2014) 45–48. doi:10.1016/j.scriptamat.2013.10.002.

[42] S. Mahajan, Effects of existing substructure on shock-twinning behaviour of iron, Phys. Status Solidi A. 2 (1970) 217–223. doi:10.1002/pssa.19700020205.

[43] N.A. Boucher, J.W. Christian, The influence of prestrain on deformation twinning in niobium single crystals, Acta Metall. 20 (1972) 581–591. doi:10.1016/0001-6160(72)90013-2.

[44] J.N. Florando, B.S. El-Dasher, C. Chen, D.C. Swift, N.R. Barton, J.M. McNaney, K.T. Ramesh, K.J. Hemker, M. Kumar, Effect of strain rate and dislocation density on the twinning behavior in tantalum, AIP Adv. 6 (2016) 045120. doi:10.1063/1.4948528.

[45] H. El Kadiri, C.D. Barrett, J. Wang, C.N. Tomé, Why are twins profuse in magnesium?, Acta Mater. 85 (2015) 354–361. doi:10.1016/j.actamat.2014.11.033.

[46] O. Bouaziz, N. Guelton, Modelling of TWIP effect on work-hardening, Mater. Sci. Eng. A. 319 (2001) 246–249.

[47] Z.Y. Liang, Y.Z. Li, M.X. Huang, The respective hardening contributions of dislocations and twins to the flow stress of a twinning-induced plasticity steel, Scr. Mater. 112 (2016) 28–31. doi:10.1016/j.scriptamat.2015.09.003.

[48] Z.C. Luo, M.X. Huang, Revisit the role of deformation twins on the work-hardening behaviour of twinning-induced plasticity steels, Scr. Mater. 142 (2018) 28–31. doi:10.1016/j.scriptamat.2017.08.017.

[49] A.A. Salem, S.R. Kalidindi, R.D. Doherty, Strain hardening regimes and microstructure evolution during large strain compression of high purity titanium, Scr. Mater. 46 (2002) 419–423.

[50] A.A. Salem, S.R. Kalidindi, S.L. Semiatin, Strain hardening due to deformation twinning in α-titanium: Constitutive relations and crystal-plasticity modeling, Acta Mater. 53 (2005) 3495–3502. doi:10.1016/j.actamat.2005.04.014.

[51] M. Knezevic, A. Levinson, R. Harris, R.K. Mishra, R.D. Doherty, S.R. Kalidindi, Deformation twinning in AZ31: Influence on strain hardening and texture evolution, Acta Mater. 58 (2010) 6230–6242. doi:10.1016/j.actamat.2010.07.041.

[52] P.D. Wu, X.Q. Guo, H. Qiao, S.R. Agnew, D.J. Lloyd, J.D. Embury, On the rapid hardening and exhaustion of twinning in magnesium alloy, Acta Mater. 122 (2017) 369–377. doi:10.1016/j.actamat.2016.10.016.

[53] Z.S. Basinski, M.S. Szczerba, M. Niewczas, J.D. Embury, S.J. Basinski, The transformation of slip dislocations during twinning of copper-aluminum alloy crystals, Rev. Métallurgie. 94 (1997) 1037–1044. doi:10.1051/metal/199794091037.

[54] F. Hiura, R.K. Mishra, M. Lukitsch, M. Niewczas, Nano-Indentation Studies of Twinned Magnesium Single Crystals, in: Magnes. Technol. 2012, Springer, Cham, 2012: pp. 117–119. doi:10.1007/978-3-319-48203-3_22.

[55] Z.X. Wu, Y.W. Zhang, D.J. Srolovitz, Dislocation–twin interaction mechanisms for ultrahigh strength and ductility in nanotwinned metals, Acta Mater. 57 (2009) 4508–4518. doi:10.1016/j.actamat.2009.06.015.

[56] B. Skrotzki, Crystallographic aspects of deformation twinning and consequences for plastic deformation processes in γ-TiAl, Acta Mater. 48 (2000) 851–862. doi:10.1016/S1359-6454(99)00385-7.

[57] D.I. Tomsett, M. Bevis, The formation of stacking faults in 1012 twins in zinc as a result of slip dislocation-deformation twin interactions, Philos. Mag. 19 (1969) 533–537. doi:10.1080/14786436908216310.

[58] L. Capolungo, I.J. Beyerlein, G.C. Kaschner, C.N. Tomé, On the interaction between slip dislocations and twins in HCP Zr, Mater. Sci. Eng. A. 513–514 (2009) 42–51.

doi:10.1016/j.msea.2009.01.035.
[59] B. Li, H.E. Kadiri, M.F. Horstemeyer, Extended zonal dislocations mediating twinning in titanium, Philos. Mag. 92 (2012) 1006–1022. doi:10.1080/14786435.2011.637985.
[60] K.D. Molodov, T. Al-Samman, D.A. Molodov, Profuse slip transmission across twin boundaries in magnesium, Acta Mater. 124 (2017) 397–409. doi:10.1016/j.actamat.2016.11.022.
[61] B. Li, E. Ma, Atomic Shuffling Dominated Mechanism for Deformation Twinning in Magnesium, Phys. Rev. Lett. 103 (2009) 035503. doi:10.1103/PhysRevLett.103.035503.
[62] B. Li, X.Y. Zhang, Twinning with zero twinning shear, Scr. Mater. 125 (2016) 73–79. doi:10.1016/j.scriptamat.2016.07.004.
[63] X.Y. Zhang, B. Li, Q. Liu, Non-equilibrium basal stacking faults in hexagonal close-packed metals, Acta Mater. 90 (2015) 140–150. doi:10.1016/j.actamat.2015.02.036.
[64] S.G. Song, G.T. Gray, Structural interpretation of the nucleation and growth of deformation twins in Zr and Ti—I. Application of the coincidence site lattice (CSL) theory to twinning problems in h.c.p. structures, Acta Metall. Mater. 43 (1995) 2325–2337. doi:10.1016/0956-7151(94)00433-1.
[65] S.G. Song, G.T. Gray, Structural interpretation of the nucleation and growth of deformation twins in Zr and Ti—II. Tem study of twin morphology and defect reactions during twinning, Acta Metall. Mater. 43 (1995) 2339–2350. doi:10.1016/0956-7151(94)00434-X.

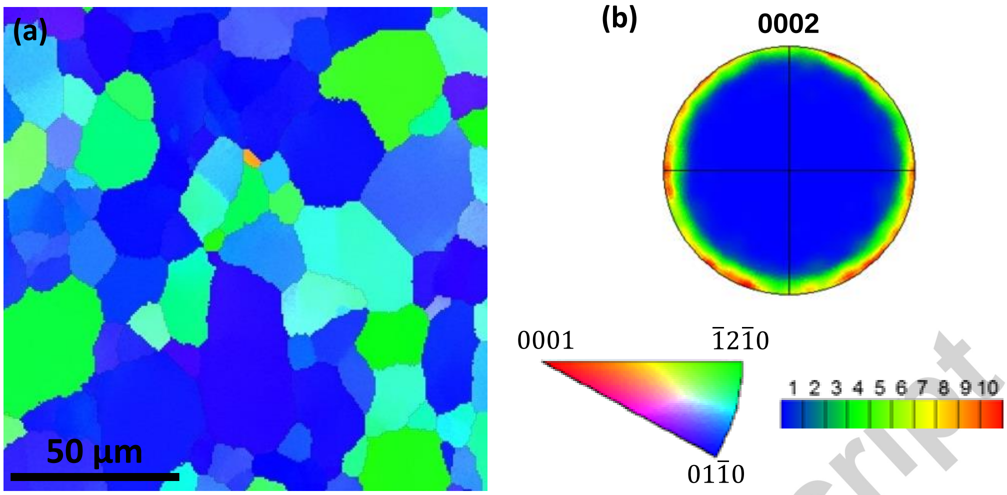


Figure 1. Initial grain structure and texture of the as-extruded AZ31 alloy. The scanned surface is perpendicular to the extrusion direction (ED). (a) Inverse pole figure (IPF) map. (b) Pole figure. The basal planes are mostly parallel to the ED, which is a typical of extruded Mg alloys.

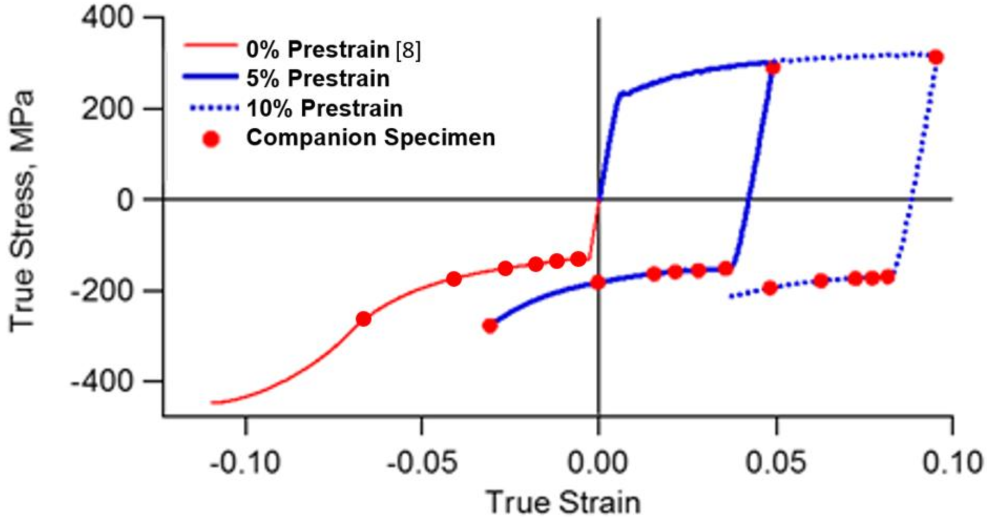


Figure 2. Stress-strain curves with 0% pre-tension strain (red line), 5% pre-tension strain (blue solid line), and 10% pre-tension strain (blue dashed line). The red dots denote strain levels at which companion specimens were terminated for EBSD analysis.

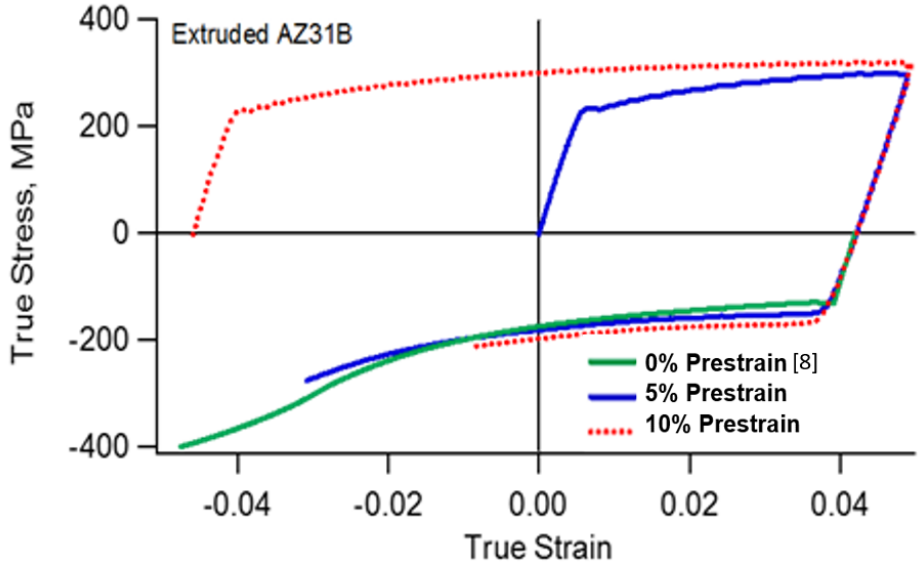


Figure 3. Comparison of stress-strain curves for 0%, 5%, and 10% prestrains by overlapping the flow stages of three specimens. The yield stress under compression only slightly increases with increasing prestrain. However, the hardening rate at the low stress stage during compression remains practically unchanged.

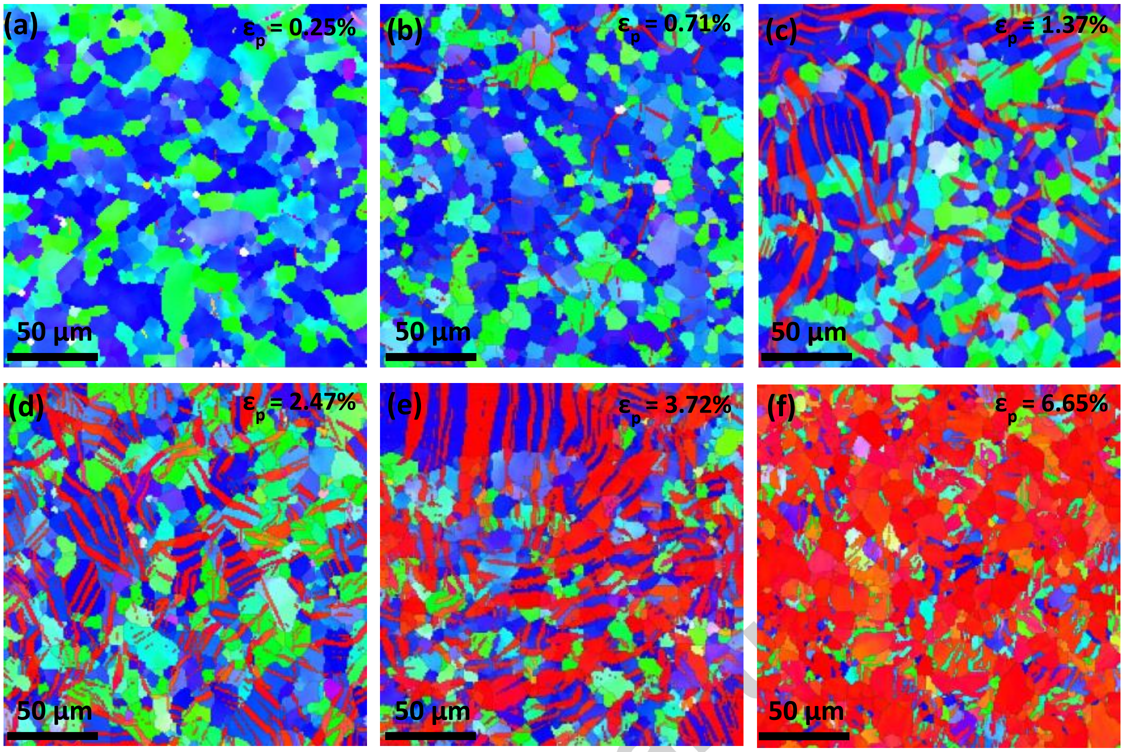


Figure 4. IPF maps at selected plastic strain levels for 0% prestrain (monotonic compression).

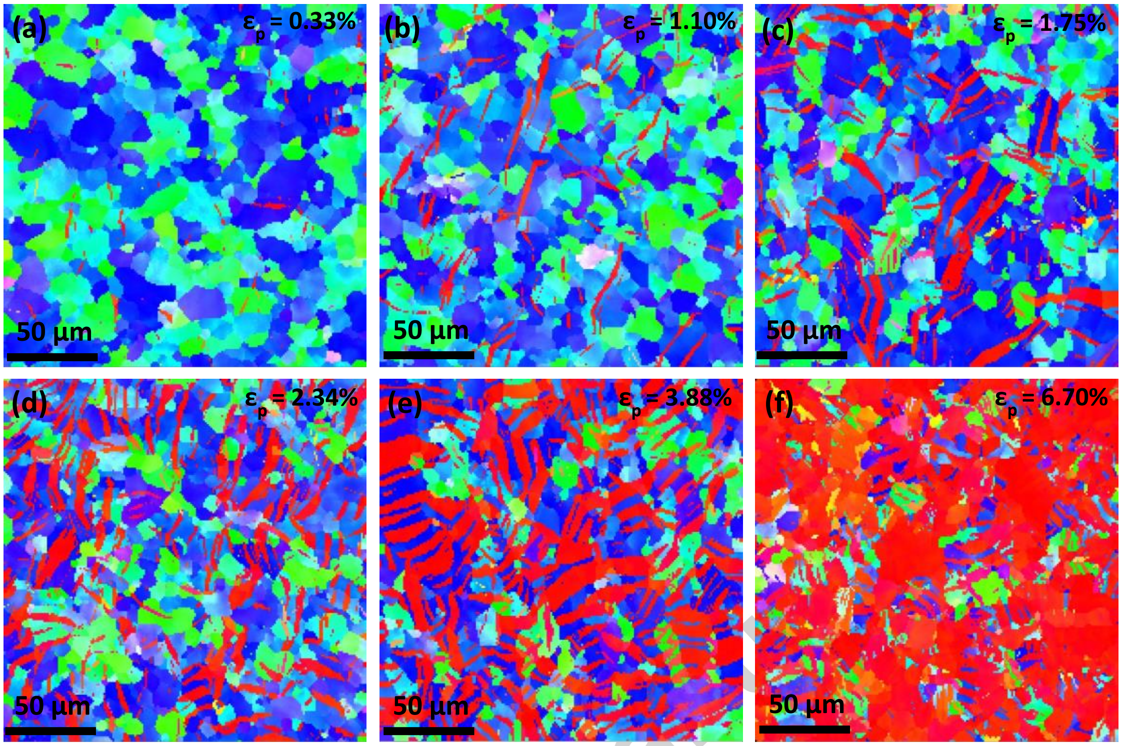


Figure 5. IPF maps at selected plastic strain levels for the specimen with 5% prestrain. The twinning behavior is similar to the specimen without prestrain.

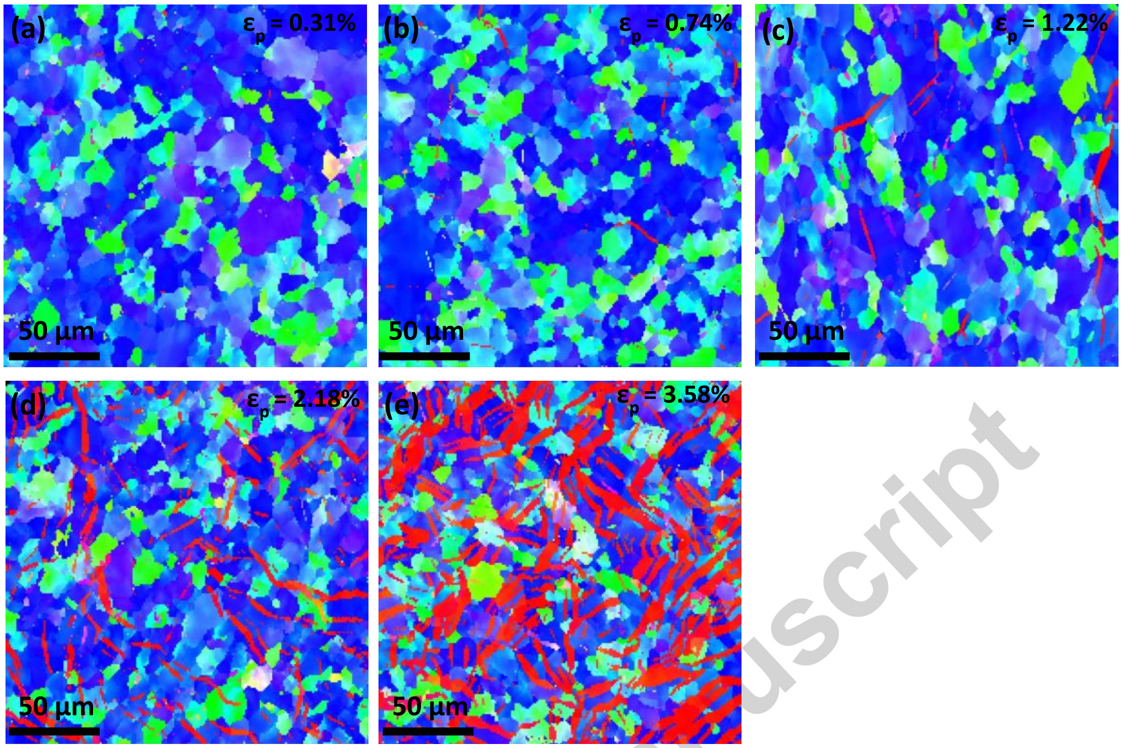


Figure 6. IPF maps at selected plastic strain levels for the specimen with 10% prestrain. Compared with 0% and 5% prestrains, the number of twins is much lower at the early stage when $\varepsilon_p < 2.18\%$. Obviously, twin nucleation is retarded by prestrain.

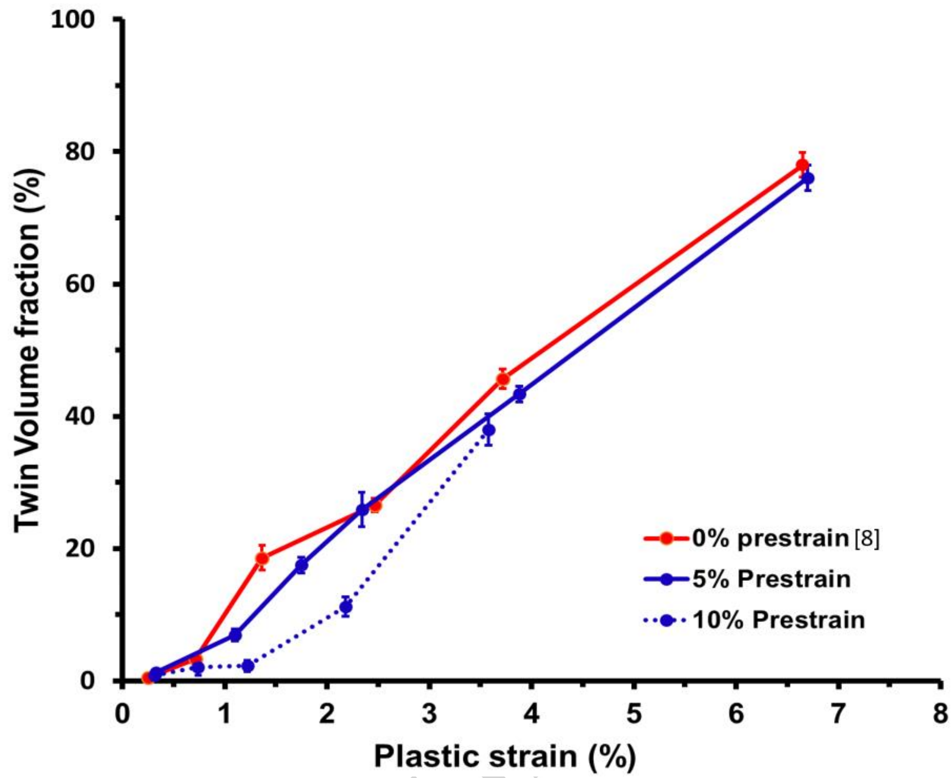


Figure 7. Evolution of twin volume fraction (TVF, f_{twin}) as a function of plastic strain. The TVFs for the different prestrains were measured from the IPF maps in figure 4, 5 and 6, respectively. At least three measurements were performed at each plastic strain level and the averaged value was plotted. The TVF for 5% prestrain is close to the TVF without prestrain. However, at strains $<3.5\%$, the TVF for 10% prestrain is lower than that for 0% and 5% prestrain.

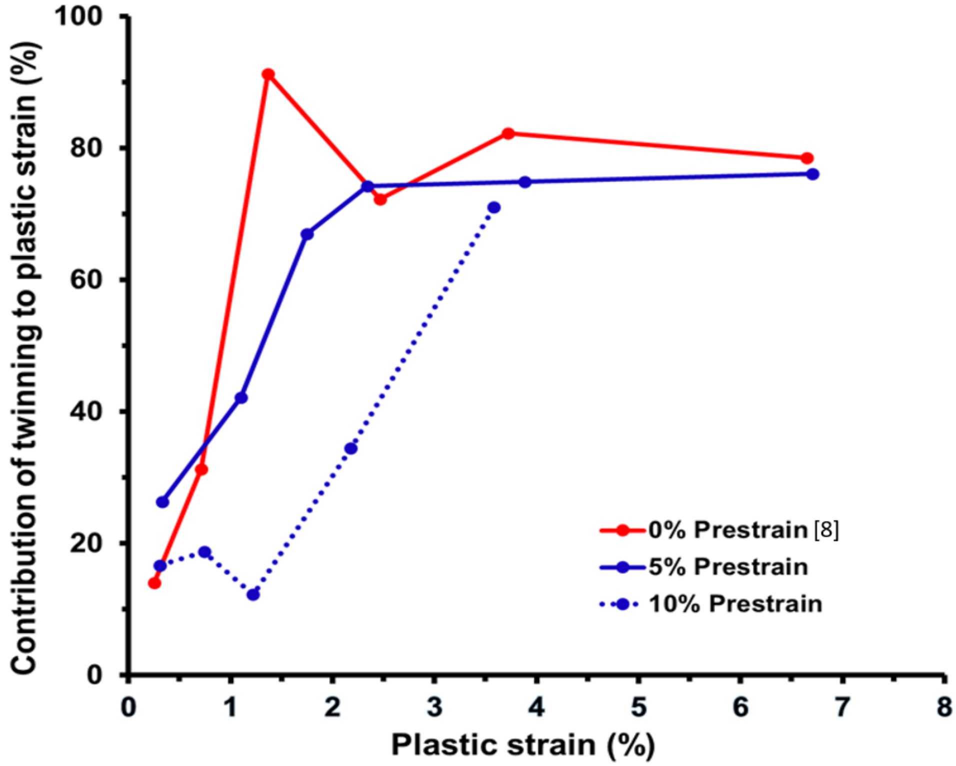


Figure 8. Contribution of $\{10\bar{1}2\}$ extension twinning to plastic strain at the low stress stage for 0%, 5% and 10% prestrain. At plastic strains <1.0%, the contribution of twinning is relatively insignificant, especially for the 10% prestrained specimen in which a plateau is present when $\varepsilon_p<1.2\%$. In general, prestrains push the curves to the right. 5% prestrain slightly reduces the contribution of twinning to plastic strain at low strains. However, 10% prestrain significantly reduces the contribution when $\varepsilon_p<3.5\%$, indicating retardation of twinning at low strains.