Dynamic recrystallization behaviors of Mg-4Y-2Nd-0.2Zn-0.5Zr alloy and the resultant mechanical properties after hot compression

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Abstract The development behaviors of ultrafine grains (UFGs) due to continuous dynamic recrystallization (cDRX) were investigated in hot compression of a Mg-4Y-2Nd-0.2Zn-0.5Zr alloy pretreated in solution and subsequently peak-aging. In the aging sample containing statically precipitated particles (SPPs), the occurrence of cDRX starts to take place at medium to high strains, and finally a stable size of UFGs are fully developed in a whole volume. In the as-solution sample with no SPPs, by contrast, the size of UFGs evolved increases rapidly at lower strains, slowly at medium strains and then finally shows a bimodal distribution in high strain. In the latter, smaller grains accompanying with an incomplete formation of UFGs are developed by any effect of dynamically precipitated particles (DPPs). The microtexture evolved is effectively randomized in the regions of UFGs, leading to the formation of a weaker texture. The tensile elongation of the aging sample, with SPPs and a fully developed UFGs, was around 17.4 %. This was much higher than that of the as-solution one, with no SPPs and an incompletely developed UFGs, that was 11.8 %, which might result from the more randomized texture due to fully developed UFGs.

Keywords *precipitation particle; hot compression; microstructure; dynamic recrystallization; microtexture; mechanical properties.*

1. Introduction

Magnesium (Mg) alloys, known as the lightest structural materials, have been applied in modern

aerospace, transportation and electronics industries, etc. [1, 2], while the poor mechanical properties such as low formability limit their practical application as structure components. The poor mechanical properties as well as the development of strong basal texture in Mg alloys can be resulted from hexagonal close-pack structure and then limited number of slip systems [3]. The development of any texture weakening and some improvement of the formability of Mg alloys have attracted much interest of engineers and scholars by grain refinement taking place during various thermo-mechanical processes including mechanical alloying and severe plastic deformation [4, 5, 6]. It is known [2] that grain refinement in hot deformed magnesium alloys can be attained to reduce the basal texture. These SPD processes such as equal channel angular pressing (ECAP), accumulative roll bonding (ARB), high pressure torsion (HPT), however, are not simple enough for any application in practical production.

Recent demand for the development of higher strength Mg alloys has revived research interest on the strengthening effect of precipitation in Mg alloy [7, 8, 9]. Due to the high solubility of RE elements in Mg alloys at high temperature and their rapidly decrease with lowering temperature, Mg-RE alloys are expected to show remarkable age-hardening response during low temperature aging. For example, Su et al. [10] determined that high volume fraction of fine β' precipitates can result in high mechanical properties.

But if we are to use Mg-RE alloys widely, a combined properties of high strength and good formability are desired. Specifically, except the precipitates hardening effect, refined grains with weak basal texture are needed. Yu et al. [11] found that Mg₂RE precipitation particles lead to grain refinement as a result of particle pinning in a Mg-11Gd-1Nd-1.5Zn-0.5Zr. Besides, it has been reported in WE43 alloys [12] that dynamic precipitation occurred during hot deformation leads to the formation of β_1 precipitation particles. Some recent works on common Mg alloys have shown that precipitation particles have effective potential to refine the grain size and modify the texture through dynamic recrystallization affected by the characteristics of precipitation particles, such as the size, volume fraction, distributions, etc. [13, 14, 15]. However, for the Mg-RE alloys with Zn addition, the effects of precipitation particles on grain refinement and texture modification due to dynamic recrystallization behaviors have not been systematically investigated yet.

Therefore, a simple and practical processing, namely high-speed hot compression, was applied to refine the grains of a Mg-4Y-2Nd-0.2Zn-0.5Zr alloy in this study. The selected alloy has great potential to be used in aerospace industry due to its high strength and great creep resistant at room and elevated

temperatures [16]. Mg-4Y-2Nd-0.2Zn-0.5Zr alloy contains large amounts of two kinds of precipitation particles, i.e. β'' (Mg_3RE) and β' (Mg_{12}YNd), which exist stably at temperature as high as 520 °C [17]. By adopting suitable thermo-mechanical processes including aging treatment, these precipitates were fully developed before hot deformation. The present study was particularly intended to clarify any effect of precipitated conditions on grain refinement and the microtexture development in the present Mg alloy during high-speed hot compression.

2. Experimental procedure

The composition of the Mg alloy tested in the present study was melted in an electric resistance furnace under a protective atmosphere of CO_2 and SF_6 in ratio of 100 : 1. Alloying elements of Y, Nd, Zn and Zr were prepared from 99.9% Mg, 99.99% Zn and some master alloys of Mg-25%Y, Mg-25%Zr and Mg-30%Nd (wt.%), respectively. After these alloying elements were completely dissolved, the melt was refined by flux and hold for 15 min at 720 °C to homogenize and cast into cylindrical ingots of 60mm in diameter and 100 mm in height. The chemical composition of the ingot was finally determined as Mg-4Y-2Nd-0.2Zn-0.5Zr (wt.%). The alloy ingots were cut into 12 mm $\times$ 10 mm $\times$ 10 mm rectangular samples and then solution treated at 525 °C for 10 h, and then quenched in water. These samples are hereafter named as the as-solution (AS) one. A part of them were aged at 220 °C in oil bath for various periods of time from 5 to 25 h in order to investigate the age hardening vs. time behavior (see Fig. 1). The samples preheated at a peak aging time of 20 h were denoted here as the as-aged (AA) one. These samples were compressed at temperatures ranging from 420 °C to 510 °C with a constant strain rate of 0.3 s^{-1} using an Inston-type mechanical testing machine. They were held for 5 min at each compression temperature and then compressed. In order to avoid any precipitation and static restoration process taking place during cooling, they were immediately quenched into water just after hot compression.

The microstructures evolved were observed by optical microscopy (OM), scanning electron microscopy (SEM) equipped with electron backscatter diffraction (EBSD) and transmission electron microscopy (TEM). For OM observation, the compressed samples were cut to thin plates with a thickness of 1 mm parallel to along the compression axis at the center and then mechanically polished and subsequently etched in an acetic picral solution, i.e. 2.1 g picric acid + 5 mL water + 5 mL acetic acid + 35 mL ethanol [18]. For SEM and TEM analysis, the samples compressed were punched into 3

mm in diameter and prepared with an ion milling method after mechanically grinded. The step size used for the EBSD scan was 0.8 μm . It should be mentioned that selected OM and EBSD images, though only small part of the sample, they generally reveal the main characterization of the whole sample and thus can be trusted and compared between different conditions. The constituent of precipitations were identified by Rigaku D/max 2550 X-ray diffraction with Cu K α radiation. The Vickers hardness tests were carried out on each plate at room temperature under a load of 3 N with a 10 s duration, and 20 measurements were performed for each condition. Tensile samples with a gauge size of 6 mm $\times$ 1.5 mm $\times$ 1 mm were machined from deformed samples. Tensile test was performed at room temperature with a strain rate of $3 \times 10^{-3} \text{ s}^{-1}$ and three specimens were used for each testing condition in order to confirm the repeatability.

3. Results and discussion

3.1. Age-hardening behavior.

Fig. 1 shows a hardness vs. aging time (MPa - t) curve and the microstructures in as-solution and after aging at 220 °C for the present Mg alloy. The hardness (MPa) increases continuously with aging time and approaches a peak value of about 840 MPa at $t \approx 20 \text{ h}$, followed by a rapid decrease due to over aging. The average grain sizes of the AS and AA samples were about 134 μm and 140 μm , respectively.

Fig. 2 shows a bright field TEM image and XRD pattern(s) for the peak-aged (i.e. AA) sample(s). It is seen in Fig. 2 (a) that platelet-shaped precipitations, with a dimension of 30 - 50 nm in length and 2 - 5 nm in width, are densely dispersed. In addition, there are a few spherical precipitations with a diameter of about 8 nm. It is suggested by the XRD data and also some previous works [11, 19] that the platelet-shaped precipitates may be DO19- β'' and the spherical ones be bco- β' , which have been indexed in Fig. 2 (b). These two kinds of precipitates are considered to be (the) most effective strengthening ones in WE alloys [20].

3.2. Flow stress behaviors

Fig. 3 shows two series of true stress-true strain curves at various temperatures for both the as-solution (AS) and as-aging (AA) samples of the present Mg alloy. These samples were compressed to $\varepsilon \approx 1.5$ at a strain rate of 0.3 s^{-1} and at temperatures ranging from 420 °C to 510 °C. The flow stress reaches a peak value at a relatively low strain, followed by a gradual softening and a steady state flow

in high strain. The flow stresses drop clearly with increasing temperature. The flow stresses for the AS sample are always lower than those for the AA one at each temperature. This suggests that precipitated particles can play an effective role as a barrier for dislocation motion during hot compression. It is also seen in Fig. 3 that the difference of flow stresses for the AA and AS samples decreases with increasing temperature.

3.3. Microstructure changes in the AA sample.

Typical optical microstructures of the AA sample, deformed to $\varepsilon = 1.5$ at temperatures of 420 °C to 510 °C, are represented in Fig. 4. Equiaxial initial grains are pancaked by compression to $\varepsilon = 1.5$ accompanying with the development of deformation bands crossing some grain interiors. Ultrafine grains (UFGs) are concurrently evolved at along original grain boundaries and the regions of UFGs evolved in grain interiors increase with increase in temperature. The average size and the volume fraction of UFGs increase concurrently with elevated temperature, and a stable UFG structure is developed almost fully in a whole volume at 510 °C (Fig. 4 (d)).

Fig. 5 shows changes in the microstructures evolved in the AA sample with deformation at 510 °C. Several deformation bands are introduced in original coarse grains containing high density precipitates at $\varepsilon = 0.2$, while there are almost no new UFGs developed. With further straining, UFGs are developed at the boundaries of deformation bands introduced at low strains and in necklace at along grain boundaries at $\varepsilon = 0.5$ to 0.9. UFGs are almost fully evolved in a whole area at $\varepsilon = 1.5$ and then the volume fraction of UFGs approaches almost 1 (Fig. 5 (d)). It is interesting to note also under OM observations that high density precipitates existed at $\varepsilon = 0.2$ (Fig. 5 (a)) gradually disappear and the average size of UFGs evolved in necklace increases with deformation to high strains.

Fig. 6 shows SEM microstructures of the AA sample deformed to various strains at 510 °C. It is clearly seen in Fig. 6 (a) that precipitation particles are densely dispersed at along grain boundaries as well as in grain interiors at $\varepsilon = 0.2$. The length of long platelet-shaped precipitates is about 1.5 μm in Fig. 6 (a). Such platelet-shaped precipitates, recognized as β'' being a good thermal stability [12], are hardly dissolved into matrix, but gradually broken into fine ones during hot compression (Fig. 6). Then this process may be identified as crack and spheroidization [10]. The cracks of β'' precipitates occur continuously during deformation, then the β'' precipitates gradually disappear during hot deformation instead of dispersed spherical particles.

3.4. Microstructure changes in the AS sample

The microstructural results for the AS sample deformed at 510 °C are represented in Figs. 7 and 8. A few UFGs are evolved at some grain boundaries even at $\varepsilon = 0.2$ (Fig. 7 (a)). This is in contrast with that for the AA sample (Fig. 5 (a)). Then UFGs are gradually developed in necklace at along original grain boundaries and the regions of UFGs expand from the boundaries into grain interiors during further deformation (Figs. 7 (b) to (d)), while the volume fraction of UFGs attains about 0.6 even at $\varepsilon = 1.5$ (Fig. 7 (d)). On the other hand, Fig. 8 shows typical SEM microstructures of the AS sample deformed at 510 °C. When these SEM images in Fig. 8 are compared to those for the AA sample in Fig. 6, it is clearly seen that rather low density and finer precipitates are heterogeneously developed at grain boundaries as well as grain interiors. These finer precipitates are considered to be dynamically developed during hot deformation.

It is interesting to note also in Fig. 7 that there are two kinds of UFGs, i.e. rather coarser and finer ones, developed in original grain interiors. Finer UFGs are evolved surrounding with coarser ones formed at low strains, then leading to the development of a bimodal-like UFG structure at medium to high strains, i.e. $\varepsilon = 0.9 - 1.5$. It is seen in Fig. 8 that dynamic precipitation occurs at along both the boundaries of origin grains and of UFGs. This suggests any operation of an interaction between formation of UFGs and dynamic precipitates, then resulting in the formation of finer UFGs at medium to high strains in the AS sample [15, 21]. This phenomenon was also observed in Mg-Gd-Y-Nd-Zr alloys by Xia et al. [22].

3.5. Dynamic recrystallization during hot compression

Fig. 9 shows the typical OIM maps of the AA sample deformed to different strains. Different colors indicate different crystallographic orientations, as defined in the inverse pole figure, and lines indicate various boundaries, i.e. white and thin black lines represent low angle boundaries with a misorientation angle of $3^\circ \sim 5^\circ$ and $5^\circ \sim 15^\circ$, respectively, while bold black ones represent high angle boundaries with a misorientation angle of over 15° .

As shown in Fig. 9 (a), the AA sample before deformation is composed of an equiaxial grain structure and the texture distribution is rather random. In Fig. 9 (b), new UFGs are developed at along original grain boundaries at low strains and progressively in grain interiors through grain fragmentation by kink bands (KBs), which are one kind of deformation bands. The frequent development of such KBs

is confirmed in the retained coarse grains (see Fig. 9 (b)). Thus the formation of UFGs in Mg alloys can take place in-situ in segments fragmented by KBs and then be controlled by a series of strain-induced continuous reaction, i.e. continuous DRX (cDRX) [4, 23, 24]. In addition, UFGs are developed frequently at along the grain boundaries of coarse grains by the operation of necklace cDRX, as seen in Figs. 9 (b) and (c). During necklace cDRX, new grains with nonequilibrium low angle boundaries are developed at low strains and gradually transformed into UFGs with medium to high angle boundaries accompanying with an increase of the volume fraction with further deformation [4]. The point-to-point misorientation and the point-to-origin misorientation are measured along the line in Fig. 9 (b). It is clearly seen that the distortions occur inside initial grains, a continuous increase of the misorientation angle from a boundary to opposite side is revealed. The original coarse grains are gradually divided into some substructures, which indicate that the dynamic recrystallization originates in the vicinity of grain boundaries. It is interesting to note that the point-to-point misorientation degrees ranges from 2° to 3° , but exceeds 5° at several places for occurrence of KBs. With increasing strain from $\epsilon = 0.9$ to 1.5, the fraction of new grains increases rapidly and new grains are developed almost homogeneously at $\epsilon = 1.5$ (Fig. 9 (b) and (c)). Crystal orientations of new grains distribute randomly with the strain of 1.5.

The distribution of misorientation angles during hot compression of AA sample is shown in Fig. 10. It is obvious that the grain boundaries misorientation distribution changes with the strain. Compared with the AA sample before compression, the sample compressed to $\epsilon = 0.9$ present a high fraction of ($2\text{-}10^{\circ}$) low angle grain boundaries. When the compressive strain increases from $\epsilon = 0.9$ to $\epsilon = 1.5$, the fraction of the low angle grain boundaries decreases while the fractions of the high angle grain boundaries increases. It is suggested that new grains with non-equilibrium low angle boundaries are developed at low strains and gradually transformed into UFGs with medium to high angle boundaries accompanying with an increase of the volume fraction with further deformation, which is corresponded with the cDRX process discussed previous [4].

This is clearly in contrast with conventional necklace discontinuous DRX (dDRX), where new grains with equilibrium high angle boundaries are formed during early stages of hot deformation. It is well known in [4, 25, 26] that dDRX takes place generally during hot deformation of coarse-grained face-centered cubic materials with low to medium stacking fault energy. The dDRX takes place through two steps of nucleation and long-range growth controlled by lattice diffusion frequently operating at high temperature, and then the nature of dDRX is considered to be thermal. By contrast, necklace

cDRX takes place through a single step strain-induced process and so the nature of cDRX is considered to be athermal. Thus cDRX occurs not only during hot deformation, but also under warm and even cold severe plastic deformation [4].

3.6. Microtextures developed during cDRX.

The microtexture developed in the present Mg alloy was examined by EBSD techniques. Typical OIM maps for both the AS and AA samples deformed to $\varepsilon = 1.5$ at 510 °C which have average confidence indexes of 3.9 and 3.6, respectively, are represented in Fig. 11. The following results can be obtained from Fig. 11. Namely,

- 1) Equiaxial UFGs are developed partially in the AS sample (Fig. 11 (a)) and almost fully in a whole area of the AA one (see Fig. 11 (b)).
- 2) Newly developed UFGs have rather random lattice orientations irrespective of both the AS and AA samples.
- 3) The average size of UFGs developed in the AS sample is smaller than that for the AA one.

Next, in order to analyze separately the textures developed in the two regions, i.e. newly developed UFGs and retained coarse grains, OIM maps of the AS sample deformed to $\varepsilon \approx 1.5$ at 510 °C with the inverse pole figures and their corresponding {0001} pole figures are represented in Fig. 10. It is seen in Figs. 12 (a) and (b) that remained coarse grains rotate nearly perpendicular to the compression axis and the relative peak intensity is as high as over 30. The development of such {0001} basal plane texture is frequently reported in many Mg alloys deformed by compression [23, 27] and rolling [28, 29]. By contrast, the basal peaks in the regions of UFGs present a rather dispersed distribution and the relative peak intensity is only around 4 (Figs. 12 (c) and (d)). This suggests that the development of random lattice orientations in the Mg alloy can be strongly resulted by the formation of UFGs. It is concluded from Figs. 11 and 12, therefore, that UFGs evolved by cDRX result in a more random and weak texture irrespective of aging treatment, leading to the development of weak basal texture in the present Mg alloy.

3.7. Quantitative analysis of UFGs developed during cDRX.

The volume fraction and the grain size of UFGs developed by cDRX (V_{UFG} and D_{UFG}) were measured by using Figs. 5 and 7. Changes in V_{UFG} and D_{UFG} with deformation at 510 °C are represented by solid line for the AA sample and by broken line for the AS one in Fig. 13. The V_{UFG}

starts to increase at $\varepsilon \approx 0.4$ for both the samples, and then approaches roughly 100% at $\varepsilon \approx 1.5$ in the AA sample, while the V_{UFG} for the AS one attains to only about 60%. It is suggested by Figs. 7 and 8 that a pinning effect of dynamic precipitation may operate effectively at newly formed UFGs and so retards the growth of UFGs in the AS sample [23].

On the other hand, the D_{UFG} for the AA sample increases continuously with strain accompanying with a full formation of UFGs at $\varepsilon \approx 1.5$. This result is roughly similar to those for many Mg alloys [4, 12]. The average value of D_{UFG} for the AS sample, by contrast, increases rather rapidly at low strains and becomes slowdown at medium to high strains, finally leading to an incomplete formation of a bimodal structure composed of relatively coarse and finer UFGs (Figs. 7 and 11). It is interesting to note in Fig. 13 (b) that the average D_{UFG} in the regions of coarser UFGs developed in the AS sample is almost similar to that of UFGs evolved in the AA samples. This suggests that these coarse and finer UFGs in AS samples may develop during early deformation and finer UFGs may occur subsequently at higher strains due to an interaction between dynamic strain aging and cDRX as mentioned in section 3.4. While in AA samples, because of the full aging treatment, no dynamic precipitates would occur during deformation and only coarse UFGs were obtained. Specifically, dynamic precipitates occurs during the hot compression of AS samples and could stop the growth of UFGs, resulting in the evolution of a finer grained structure as compared with that of AA samples.

3.8. Resultant mechanical properties after hot compression

Fig. 14 shows typical engineering tensile stress - strain curves tested at room temperature for both the AS and AA samples deformed to $\varepsilon \approx 1.5$ at temperatures ranging from 420 °C to 510 °C. The flow stress at $\varepsilon = 0.05$ ($\sigma_{\varepsilon=0.05}$) and the total elongation to fracture measured from Fig. 14 are plotted against temperature of prior hot deformation in Fig. 15. It is seen in Fig. 15 (a) that the flow stresses at $\varepsilon = 0.05$ for the AA samples are higher than those for the AS one at lower deformation temperature because of statically precipitated particles (SPPs), while those for the AA one decrease rapidly and attains to lower values at higher temperatures. By contrast, the flow stresses for the AS sample hardly change at the temperatures tested, because dynamically precipitated particles (DPPs) are evolved frequently at higher temperatures. It is clearly seen in Fig. 15 (b) that the elongation to fracture for the AA sample is always larger than that for the AS one, and increases rapidly at higher temperature comparing with that for the AS sample. The largest elongation of the AA sample is 17.8%, which is excellent compared with 11.4%

for the AS one: namely, an enhancement of 50% in total elongation is attained by the full development of UFGs following aging treatment before hot deformation.

Fig. 16 shows the relationships between the elongation to fracture and the volume fraction of UFGs (V_{UFG}) evolved at $\varepsilon \approx 1.5$ for both the AA and AS samples. It is clearly seen that the elongation to fracture increases with increase in V_{UFG} for both the samples. Tensile elongation may be controlled by the development of UFGs themselves and also the volume fraction of UFGs evolved by cDRX. Because the latters have a more random and weak texture (Figs. 11 and 12). Next, the elongation to fracture for the AS sample appears to be always smaller than that for the AA one. This is because dynamic precipitation taking place in the AS sample can result in the incomplete formation of smaller UFGs even in high strain (Figs. 7, 11, 12 and 13). This may bring about smaller elongations to fracture as well as rather higher proof stresses at higher prior deformation temperature (Fig. 15).

4. Conclusion

Hot deformation and microstructural behaviors of a Mg-4Y-2Nd-0.2Zn-0.5Zr alloy were studied in compression up to strains of 1.5 at temperatures from 420 °C to 510 °C. The alloy was heat-treated in as-solution (the AS sample) and subsequent a peak aging (the AA one). The main results are summarized as follows:

(1) In the AA samples, statically precipitates particles (SPPs) are gradually cracked and spheroidized during hot compression. In addition, SPPs in AA samples have a positive effect on grain refinement resulting in about 100% volume fraction of UFGs in $\varepsilon = 1.5$. By contrast, in the AS samples containing no SPPs, the average size of UFGs evolved increases more slowly and the development of UFGs does not occur completely even at high strain, i.e. $V_{\text{UFG}} \approx 60\%$, at $\varepsilon \approx 1.5$, which would be due to the growth retarding effect on the UFGs by the dynamic precipitates (DPPs) formed during deformation in AS samples.

(2) DRX occurs along the boundaries of original grains and KBs, resulting in a high volume fraction of UFGs and then weaker textures. Such new grain evolution can be controlled by a series of strain-induced continuous reaction, i.e. continuous DRX (cDRX).

(3) The tensile elongations to fracture for the present Mg samples with a partial and a full formation of UFGs were 11.4% to 17.8%, respectively; namely, it was enhanced by 50% through hot deformation to $\varepsilon \approx 1.5$ following aging treatment. The latter can results in the formation of UFG themselves and

further the volume fraction developed.

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Figure captions

Fig. 1. Vickers microhardness curves and typical microstructures of Mg-4Y-2Nd-0.2Zn-0.5Zr alloy treated in solid solution and peak aging at 220 °C.

Fig. 2. (a) TEM image and (b) XRD patterns of the AA sample heated at 510 °C for 5 min before hot deformation.

Fig. 3. True stress - true strain curves at $\dot{\epsilon} = 0.3 \text{ s}^{-1}$ and at various temperature from 420 °C to 510 °C for Mg-4Y-2Nd-0.2Zn-0.5Zr alloy treated in solid solution (broken line) and subsequent aging (solid line).

Fig. 4. Microstructures of the AA sample deformed to a strain of 1.5 at various temperatures (a) 420 °C, (b) 450 °C, (c) 480 °C and (d) 510 °C.

Fig. 5. Microstructures developed at strains of (a) $\epsilon = 0.2$, (b) $\epsilon = 0.5$, (c) $\epsilon = 0.9$ and (d) $\epsilon = 1.5$ for the AA sample deformed at 510 °C.

Fig. 6. The SEM images of the AA samples (a) $\epsilon = 0.2$, (b) $\epsilon = 0.5$, (c) $\epsilon = 0.9$ and (d) $\epsilon = 1.5$ deformed to various strains at 510 °C.

Fig. 7. Microstructures developed in the AS sample deformed at 510 °C (a) $\epsilon = 0.2$, (b) $\epsilon = 0.5$, (c) $\epsilon = 0.9$ and (d) $\epsilon = 1.5$.

Fig. 8. The SEM images of the AS samples deformed at 510 °C (a) $\epsilon = 0.5$ and (b) $\epsilon = 1.5$.

Fig. 9. Typical OIM maps of the AA sample deformed to (a) $\epsilon = 0$, (b) $\epsilon = 0.9$ and (c) $\epsilon = 1.5$, and (d)

misorientation profiles along the indicated directions marked in (b).

Fig. 10. The misorientation of the AA sample deformed to (a) $\epsilon = 0$, (b) $\epsilon = 0.9$ and (c) $\epsilon = 1.5$.

Fig. 11. Typical OIM maps of Mg-4Y-2Nd-0.2Zn-0.5Zr alloy deformed to $\epsilon = 1.5$ at 510 °C for both the AS (a) and AA (b) samples.

Fig. 12. Inverse pole figures and their corresponding {0001} pole figures in the two regions of remained original grains ((a) and (b)) and DRX grains evolved in the AS sample ((c) and (d)) deformed to $\epsilon = 1.5$ at 510 °C (See Fig. 10(a)). Note the relative intensity number attached in (b) and (d).

Fig. 13. Changes in (a) the volume fraction and (b) the average grain size of UFG developed with strain at 510 °C for both the AA and AS sample.

Fig. 14. Effect of aging treatment on engineering tensile stress - strain curves of Mg-4Y-2Nd-0.2Zn-0.5Zr alloy after hot deformation to $\epsilon = 1.5$ at various temperatures.

Fig. 15. Changes in tensile mechanical properties at room temperature for Mg-4Y-2Nd-0.2Zn-0.5Zr alloy after hot deformation to $\epsilon = 1.5$ with previous compression temperatures.

Fig. 16. Relationship between tensile elongation to fracture at room temperature and the volume fraction of UFGs developed by cDRX for Mg-4Y-2Nd-0.2Zn-0.5Zr alloy deformed to $\epsilon = 1.5$ at various compression temperatures.

Fig. 1. Hardness curves and typical microstructures of Mg-4Y-2Nd-0.2Zn-0.5Zr alloy treated in solid solution and peak aging at 220 °C.

Fig. 2. (a) TEM image and (b) XRD patterns of the AA sample heated at 510 °C for 5 min before hot deformation.

Fig. 3. True stress - true strain curves at $\dot{\epsilon} = 0.3 \text{ s}^{-1}$ and at various temperature from 420 °C to 510 °C for Mg-4Y-2Nd-0.2Zn-0.5Zr alloy treated in solid solution (broken line) and subsequent aging (solid line).

Fig. 4. Microstructures of the AA sample deformed to a strain of 1.5 at various temperatures (a) 420 °C, (b) 450 °C, (c) 480 °C and (d) 510 °C.

Fig. 5. Microstructures developed at strains of (a) $\epsilon = 0.2$, (b) $\epsilon = 0.5$, (c) $\epsilon = 0.9$ and (d) $\epsilon = 1.5$ for the AA sample deformed at 510 °C.

Fig. 6. The SEM images of the AA samples (a) $\epsilon = 0.2$, (b) $\epsilon = 0.5$, (c) $\epsilon = 0.9$ and (d) $\epsilon = 1.5$ deformed to various strains at 510 °C.

Fig. 7. Microstructures developed in the AS sample deformed at 510 °C (a) $\epsilon = 0.2$, (b) $\epsilon = 0.5$, (c) $\epsilon = 0.9$ and (d) $\epsilon = 1.5$.

Fig. 8. The SEM images of the AS samples deformed at 510 °C (a) $\epsilon = 0.5$ and (b) $\epsilon = 1.5$.

Fig. 9. Typical OIM maps of the AA sample deformed to (a) $\varepsilon = 0$, (b) $\varepsilon = 0.9$ and (c) $\varepsilon = 1.5$, and (d) misorientation profiles along the indicated directions marked in (b).

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