Effect of Al Addition on the Precipitation Behavior of a Binary Mg-Zn Alloy

Ye-Lim Kim[†], Hiroyasu Tezuka*, Equo Kobayashi* and Tatsuo Sato*

Graduate student, Department of Metallurgy and Ceramics Science, Tokyo Institute of Technology,
2-12-1, O-okayama, Meguro-ku, Tokyo 152-8552, Japan

*Department of Metallurgy and Ceramics Science, Tokyo Institute of Technology, 2-12-1,
O-okayama, Meguro-ku, Tokyo 152-8552, Japan

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Abstract The effect of Al addition on the precipitation behavior of a binary Mg-Zn alloy was investigated based on the changes in the morphology, distribution and element concentration of precipitates formed during aging treatment. The as-cast Mg-6.0 mass%Zn (Mg-6Zn) and Mg-6.0 mass%Zn-3.0 mass%Al (Al-added) were homogenized at 613 K for 48 h and at 673 K for 12 h; they were then solid solution treated at 673 K for 0.5 h and 1 h, respectively. The Mg-6Zn and Al-added alloys were aged at 403 K and 433 K. The peak hardness of the Al-added alloy was higher than that of the Mg-6Zn alloy at each aging temperature. Rod-like, plate-like, blocky, and lath-like precipitates were observed in the Al-added alloy aged at 433 K for 230.4 ks, although the rod-like and plate-like precipitates were observed in the TEM microstructure of the Mg-6Zn alloy aged at 433 K for 360 ks. Moreover, the precipitates in the Al-added alloy were refined and densely distributed compared with those in the Mg-6Zn alloy. The Cliff-Lorimer plots obtained by the EDS analysis of the rod-like β_1' and plate-like β_2' phases in the Al-added alloy peak aged at 433 K for 230.4 ks were examined. It was confirmed that the β_2' phases had higher concentration of solute Al atom than was present in the β_1' phases, indicating that the properties of precipitates can be changed by Al addition.

Key words Mg-Zn-Al alloy, precipitation behavior, transmission electron microscopy (TEM), Cliff-Lorimer plot.

1. Introduction

The Mg-Zn alloys have a great potential to improve the strength by various heat treatments and micro-alloying as compared with the Mg-Al alloys which do not generally undergo thermal treatments such as an aging treatment, whereas their strength is industrially achieved through the grain refinement. However, the predominant strengthening β_1' precipitates in the binary Mg-Zn alloy are widely reported to exhibit a coarse and inhomogeneous distribution.¹⁻⁴⁾ It is generally accepted that, for Mg-Zn alloys containing 4-9 mass% Zn and aged isothermally at 393-533 K, the strengthening precipitate phases are β_1' and β_2' . The metastable β_1' phase forms as rods with its long axis parallel to the $[0001]_{\text{Mg}}$ direction of the α -Mg matrix, while the metastable β_2' phase forms as plates on $(0001)_{\text{Mg}}$. Historically, it is determined that both β_1' and β_2' phases have a hexagonal structure ($a = 0.520$ nm, $c = 0.857$ nm) that is identical to that of the Laves phase MgZn_2 . In a newly published paper, it is found that the β_1' phase is mainly formed as $[0001]_{\text{Mg}}$ rods, unlike the MgZn_2 structure, with a base-centered monoclinic structure ($a =$

2.596 nm, $b = 1.428$ nm, $c = 0.524$ nm, $\gamma = 102.5^\circ$) similar to that of the Mg_4Zn_7 phase.^{1,5-10)} In the development of the equilibrium β phase during the precipitation heat treatment, several transition phases are formed in a specific sequence as follows:^{1,11-12)}

Supersaturated solid solution(SSSS) $\rightarrow$ solute clusters $\rightarrow$ GP zones $\rightarrow \beta_1'$ (rods and blocky precipitates $\perp \{0001\}_{\text{Mg}}$; MgZn_2 or possible Mg_4Zn_7) $\rightarrow \beta_2'$ (coarse plates $\parallel \{0001\}_{\text{Mg}}$ and laths $\perp [0001]_{\text{Mg}}$; MgZn_2) $\rightarrow$ equilibrium β phase (MgZn or Mg_2Zn_3)

The decomposition of the SSSS initially occurs through clustering of solute atoms. Depending on the aging temperature mainly below 373 K and alloy composition, three types of coherent GP zones may form: GP1 zones as plates on $\{11\bar{2}0\}_{\text{Mg}}$, GP2 zones as oblate spheroids on $\{0001\}_{\text{Mg}}$ and GP3 zones as discs on $\{0001\}_{\text{Mg}}$.¹²⁾ Recently, spherical GP zones containing a high concentration of Zn were discovered in the ZM61 alloy aged at 343 K for 48 h using TEM and atom probe. GP zones can act as heterogeneous nucleation sites for the transition phases β_1' and β_2' during high temperature aging, resulting in finer microstructure as well as enhanced age hardening response.^{8-9,13-15)}

If the effective control of the size and distribution of these precipitates is achieved, the expanded applications of

[†]Corresponding author
E-Mail : forever-yl@nate.com (Y. -L. Kim)

magnesium alloys can be expected due to the ultimately improved precipitation hardening. Microalloying is one of the useful methods to produce the finely and uniformly distributed precipitates because the additives are expected to affect the formation of small clusters and provide preferential nucleation sites for precipitates.^{10,12,16)} Recently, it has been reported that not only precipitation hardening responds but also microstructure and mechanical properties of Mg-Zn-Mn alloys were improved due to the refined precipitates by the Al addition.^{8,17)} The Mg-Zn-Al alloys have achieved improved elevated-temperature properties in the as-cast state comparable to those of commercial Mg-Al-Zn alloys, e.g. AZ91, AM60 and AZ31 alloys.¹⁸⁻¹⁹⁾ The Mg-Zn-Al alloys have an advantage to show a sound heat resistance due to the high stability of the $\text{Mg}_{32}(\text{Zn},\text{Al})_{49}$ phase and the low interfacial energy between the $\text{Mg}_{32}(\text{Zn},\text{Al})_{49}$ phase and matrix.²⁰⁾ Furthermore, Al containing alloys give good mechanical properties at room temperature due to the solid-solution hardening, excellent tensile properties due to the refined precipitates and good castability.

However, the effect of Al addition and the characteristics of various precipitates in the Mg-Zn-Al alloys are rarely carried out. Therefore, in this study, the effect of Al addition on enhancing the precipitation hardening was estimated by investigating the aging behaviors and the characteristics of various precipitates in the binary Mg-Zn and modified Mg-Zn-Al alloys.

2. Experimental Procedure

The as-cast Mg-6.0 mass%Zn (Mg-6Zn) and Mg-6.0 mass%Zn-3.0 mass%Al (Al-added) alloys were prepared by a conventional casting method. The chemical compositions of the Mg-6Zn and Al-added alloys are presented in Table 1. The temperatures for the homogenization and solid-solution treatment were determined using the differential scanning calorimetry (DSC). The DSC was operated ranging from room temperature to 773 K with 30 mg of Mg (99.99%) as a reference under an argon atmosphere with the heating rate of 10 K/min. The Mg-6Zn and Al-added alloys were homogenized at 613 K for 48 h and at 673 K for 12 h, respectively. Then the homogenized specimens were solid solution treated at 673 K for 0.5 h and 1 h of the Mg-6Zn and Al-added alloys, respectively. The

specimens were subsequently aged in oil baths at 403 K and 433 K. Prior to the experiments, the aged specimens with $\sim 10 \times 10 \times 5 \text{ mm}^3$ dimension were mechanically polished using 120–4000 grit emery paper. The precipitation hardening response was examined by the micro Vickers hardness measurement determined from five data out of 7 individual indentations with a load of 200 gf for 15 s. The element distribution and backscattered electron images were obtained using the electron probe X-ray micro-analyzer (EPMA). The existing phase analysis of the alloys were analyzed using the X-ray diffraction analysis (XRD) instrument with the $\text{CuK}\alpha$ radiation at a grazing angle 1° . The XRD patterns were identified by the JCPDS X-ray diffraction databank. The precipitates formed during the aging treatment were observed and characterized using a transmission electron microscope (TEM) equipped with an energy dispersive X-ray spectroscopy (EDS). Thin foil specimens for TEM were prepared by punching 3 mm diameter discs and twin jet electro-polishing using a solution of 8.3 g LiCl, 18.6 g $\text{Mg}(\text{ClO}_4)_2$, 833.4 ml methanol, and 166.7 ml 2-butoxy-ethanol at approximately 243 K with 17 V.

3. Results

3.1 Changes in microstructure

The DSC measurement was carried out under a controlled heating program in order to understand the characteristic temperatures of the phase transformations evolved during heating. Fig. 1 shows the DSC curves of the as-cast Mg-6Zn and Al-added alloys with a heating rate of 10 K/min. There are endothermic peaks at 614 K and 630 K emerging in the curves of the as-cast Mg-6Zn and Al-added alloys, respectively, corresponding to the second phase transformation. It can be inferred that the monotectoid reaction between α -Mg and $\text{Mg}_{32}(\text{Zn},\text{Al})_{49}$ phases,

Fig. 1. DSC curve of the as-cast Mg-6Zn and Al-added alloys recorded under a heating program at a controlled speed of 10 K/min.

Table 1. Chemical compositions of the Mg-6Zn and Al-added alloys.

Alloys (mass%)	Designation		Zn	Al	Mg
Mg-6Zn	Mg-6Zn	mass%	6.39	-	Bal.
		at.%	2.48	-	
Mg-6Zn-3Al	Al-added	mass%	6.12	2.90	Bal.
		at.%	2.37	2.72

Fig. 2. DSC curve of the solid-solution treated Mg-6Zn and Al-added alloys at 673 K for 0.5 h and 1 h, respectively, recorded under a heating program at a controlled speed of 10 K/min.

Fig. 3. XRD patterns of the as-cast and solid-solution treated binary Mg-6Zn and ternary Al-added alloys solid-solution treated at 673 K for 0.5 h and 1 h, respectively.

which is identified by the XRD analysis in Fig. 3. The endothermic peak at 630 K of the monotectoid reaction in the Al-added alloy is much higher than that in the Mg-6Zn. The Al addition affects the formation of thermally more stable phase in the Mg-6Zn alloy. The details are discussed in the explanation parts of Fig. 3 and Fig. 4.

The DSC results of the solid-solution treated Mg-6Zn and Al-added alloys at 673 K for 0.5 h and 1 h, respectively, with a heating rate of 10 K/min are shown in Fig. 2. The DSC curve for the solid-solution treated Al-added alloy is characterized by two exothermic peaks at 534 K and 612 K (tops of the peaks), on the other hand, the DSC curve plotted from the solid-solution treated Mg-6Zn alloy shows the only exothermic peak at 541 K. The exothermic peaks are attributed to the formations of precipitates. The peak 1 is broaden and shifted from 541 K to 534 K by the Al addition. Furthermore, there is a new peak 2 at 612 K in the Al-added alloy which is not found in the Mg-6Zn alloy. This may be because of the facts that

Fig. 4. EPMA results of the solid-solution treated Al-added alloy solid-solution treated at 673 K for 1h; (a) BSE image and elemental maps for (b) Mg, (c) Zn and (d) Al.

either the detected heat representing the peak 2 in the Mg-6Zn alloy is too small to be detected by the DSC experiments due to the very small volume fraction of the precipitate corresponding to the peak 2 in the Mg-6Zn alloy or the precipitate corresponding to the peak 2 is not formed in the Mg-6Zn alloy. It is noted that the Al addition caused the different aspect of formation of precipitates compared to that in the Mg-6Zn alloy.

The XRD patterns of the as-cast and solid-solution treated Mg-6Zn and Al-added alloys are shown in Fig. 3. The as-cast Mg-6Zn alloy is consisted of α -Mg, MgZn and MgZn₂ phases, while the as-cast Al-added alloy is composed of α -Mg and Mg₃₂(Zn,Al)₄₉ phases. Almost of the MgZn, MgZn₂ in the Mg-6Zn alloy and Mg₃₂(Zn,Al)₄₉ phase in the Al-added alloy were completely dissolved after the solid solution treatment. The EPMA result of the solid-solution treated Al-added alloy shown in Fig. 4 supports the XRD result after solid-solution treatment. The elements in the solid-solution treated Al-added alloy were distributed homogeneously in the matrix without any segregation and/or remained intermetallic compounds. It indicates that the supersaturated solid solution in the investigated alloys is successfully obtained and the Zn and Al mainly act as solute atoms. It is in good agreement with the Mg-Zn binary and Mg-Zn-Al ternary phase diagrams.²¹⁻²²⁾ However, there is no XRD peak corresponding to the Mg₁₇Al₁₂ phase which causes poor mechanical properties at higher temperatures above 393 K.²⁰⁾ It is understood that the Zn addition suppressed the formation of the Mg₁₇Al₁₂ phase even though 3 mass%Al was added. It is confirmed that the Mg₃₂(Al,Zn)₄₉ phase is produced before the formation of the Mg₁₇Al₁₂ phase during solidification based on the phase diagram of the Mg-Zn-Al ternary alloy. The Mg₃₂(Al,Zn)₄₉ phase is formed at $\sim$ 773 K, while the Mg₁₇Al₁₂ phase exists at temperatures below $\sim$ 723 K.

Fig. 5. Micro Vickers hardness changes in the Mg-6Zn and Al-added alloys during aging at 403 K and 433 K.

3.2 Changes in hardness during aging treatment

The age-hardening profiles of the Mg-6Zn and Al-added alloys are shown in Fig. 5. In the initial as-quenched condition, the hardness of the Al-added alloy is higher than that of the Mg-6Zn alloy due to the solid-solution hardening by the Al addition. A precipitation-hardening behavior is exhibited when the Mg-6Zn and Al-added alloys are exposed to the subsequent aging treatment. Apparently, the precipitation hardening responses of the Al-added alloy were significantly enhanced at 433 K, with the peak hardness of ~85 HV as compared to that of ~76 HV for the Mg-6Zn alloy. However, the difference of peak hardness obtained at 403 K is smaller as ~82 HV of Mg-6Zn and ~85 HV of Al-added alloys. The Mg-6Zn alloy needs longer incubation time during aging treatment at 403 K than the Al-added alloy which shows higher increment of hardness after 4 h of aging time at 403 K. Moreover, the Mg-6Zn and Al-added alloys have almost similar peak aging time as ~128 h at 403 K and ~64 h at 433 K. Note that the addition of Al enhances the precipitation hardening at 433 K, although the Al addition is less effective on the improvement of precipitation hardening at 403 K compared with the Mg-6Zn alloy.

3.3 TEM microstructures of precipitates

The TEM microstructures of precipitates in the Mg-6Zn and Al-added alloys are compared. The TEM bright field images of the Mg-6Zn and Al-added alloys aged at 433 K for 360 ks and 230.4 ks are shown in Fig. 6 and Fig. 7, respectively. The images were observed along the $\langle 1\bar{1}00 \rangle_{Mg}$ direction. There are two types of precipitates in the Mg-6Zn alloy which can be characterized as a rod-like precipitate parallel to the c-axis of the matrix and a plate-like precipitate with $[10\bar{1}0]_{Mg}$ direction. As described in the introduction, the precipitates have the certain

Fig. 6. TEM micrograph of the Mg-6Zn alloy peak aged at 433 K for 360 ks; (a) bright field image and (b) the diffraction pattern of $\langle 1\bar{1}00 \rangle_{Mg}$ beam direction. Electron beam is approximately parallel to $\langle 1\bar{1}00 \rangle_{Mg}$.

Fig. 7. TEM micrograph of the Al-added alloy peak aged at 433 K for 230.4 ks; (a) bright field image and (b) the diffraction pattern of $\langle 11\bar{2}0 \rangle_{Mg}$ beam direction. Electron beam is approximately parallel to $\langle 11\bar{2}0 \rangle_{Mg}$.

orientation relationships with the α -Mg matrix.¹¹⁻¹²⁾ The details of the orientation relationships are confirmed to be as follows:

$$\cdot \beta'_1 \text{ phase (similar structure to } Mg_4Zn_7) \\ [0001]_{\beta'_1} \parallel [11\bar{2}0]_{\alpha} \text{ and } (11\bar{2}0)_{\beta'_1} \parallel (0001)_{\alpha}$$

$$\cdot \beta'_2 \text{ phase (MgZn}_2) \\ [11\bar{2}0]_{\beta'_2} \parallel [10\bar{1}0]_{\alpha} \text{ and } (0001)_{\beta'_2} \parallel (0001)_{\alpha}$$

Precipitates in the Al-added alloy can be distinguished with four different morphologies, on the other hand only the rod-like and plate-like precipitates observed in the Mg-6Zn alloy. The TEM microstructure of the Al-added alloy consists of not only the rod-like β'_1 precipitates and plate-like β'_2 precipitates but also a number of blocky β'_1 and small volume fraction of lath-shape β'_2 precipitates. It is obvious that the Al-added alloy exhibits much higher number density of the blocky and plate-like precipitates compared with the Mg-6Zn alloy. This suggests that Al affects the nucleation of this morphological and orientational variant of the β'_1 and β'_2 phases. It is found that the Al addition to the binary Mg-Zn alloy produces

Fig. 8. TEM bright field images of the Mg-6Zn alloy peak aged at 433 K for 360 ks with (a) low magnification and (b), (c) high magnification, and (d) the diffraction pattern of $[0001]_{\text{Mg}}$ beam direction. Electron beam is approximately parallel to $[0001]_{\text{Mg}}$.

Fig. 9. TEM bright field images of the Al-added alloy peak aged at 433 K for 230.4 ks with (a) low magnification and (b), (c) high magnification and (d) the diffraction pattern of $[0001]_{\text{Mg}}$ beam direction. Electron beam is approximately parallel to $[0001]_{\text{Mg}}$.

blocky β_1' and some of lath-shape β_2' precipitates, on the other hand, the microstructures of the aged binary Mg-Zn alloy consist of two types of precipitates showing the morphologies of rod-like and plate-like shapes.^{6,12)} Fig. 8 and Fig. 9 are the TEM bright field images of the Mg-

Fig. 10. Cliff-Lorimer plots for the (a) rod-like β_1' and (b) plate-like β_2' phases in the Al-added alloy aged at 433 K for 230.4 ks.

6Zn and Al-added alloys aged at 433 K for 360 ks and 230.4 ks, respectively, observed along the matrix $[0001]_{\text{Mg}}$ direction. The precipitates in the Al-added alloy are more finely and densely distributed compared to those in the Mg-6Zn alloy. In the observation along the $[0001]_{\text{Mg}}$ direction, the rod-like precipitates are aligned along the $\langle 10\bar{1}0 \rangle_{\text{Mg}}$ as well as the $\{11\bar{2}0\}$ prismatic plane direction and the aligned precipitates may contribute to the precipitation hardening which is produced by the plate-like precipitates.

4. Discussion

4.1 Effect of Al addition on the precipitation hardening of the Mg-Zn alloy

Fig. 10 shows the Cliff-Lorimer plots for the rod-like β_1' and plate-like β_2' phases in the Al-added alloy aged at 433 K for 230.4 ks. The Cliff-Lorimer plots were obtained from the EDS concentration analysis representing solute atoms as a function of the Mg concentration. The atomic percentages of solute Zn and Al atoms were plotted versus concentration of Mg. Then, the least square method was used for calculating the gradient of each straight line of the EDS results as described in Fig. 10. The negative slope of each straight line indicates that the Zn and Al atoms are incorporated in the precipitates. The concentration of Al in the β_2' phase is slightly higher than that in the β_1' phase. It indicates that the Al addition accelerates the nucleation of the plate-like β_2' precipitates and suppresses their growth during aging. The enrichment of Al in the precipitates is mainly caused by the strong attractive interaction with Zn. The atomic radii of Al and Zn are almost similar as 0.182 and 0.153 nm, respectively, which means that Al and Zn have almost the similar atomic size; the atomic radius of Mg is 0.172 nm. It is certain that the Al addition provides a direct influence on the nucleation and growth of precipitates.

The hardness increases very slightly in the initial stage of aging due to the formation of GP zone. If the precipitate has the same crystal structure and a similar lattice parameter to the α -matrix phase, the two phases can form low-energy coherent interfaces on all sides. This situation arises in the early stage of aging. This kind of precipitate is named as a fully coherent precipitate or a GP zone. Even the two crystal structures match more or less perfectly across all interfacial planes, the zone can have any morphology and fully coherent. There are several reports on the formation of the GP zones in Mg-Zn alloys. Mima and Tanaka²³⁻²⁴⁾ examined age-hardening and precipitation sequence in the Mg-Zn alloys by electrical resistivity, hardness measurements, and TEM observations, and reported that the microstructures after the aging at room temperature for 270 h as well as at 383 K for 100 h exhibited a mottled structure implying the presence of GP zones. A X-ray diffraction experiment has reported that two types of GP zones are formed as a plate along $\{11\bar{2}0\}$ planes and an ellipsoidal shape on the basal planes during aging at temperatures below 333 and 353 K, respectively.²⁵⁾ The present TEM and atom probe results have shown that the GP zones are spherical containing a high concentration of Zn. This could be the first report demonstrating the spherical morphology of the GP zones in the Mg-Zn alloy system, since there is no article which has ever indicated the clear image of GP

zones in Mg-Zn alloys. Also, it seems that the zones do not have an ordered structure. In the early stage of aging of Mg-RE-Zn-Zr and Mg-Ca-Zn alloys, microstructure analyses using atom probe and high resolution TEM have revealed that the atoms of RE (mainly containing Ce and Nd) and Ca along with Zn aggregate on the basal planes, *i.e.*, the formation of monolayer plate-like GP zones.^{13,16)} This is because that the atomic radius of the rare earth elements or Ca is substantially larger than that of Mg, so the plate-like morphology is energetically favorable. There are no intermetallic compounds between the Mg and MgZn at low temperature in a Mg-Zn binary phase diagram, and so the existence of a large metastable miscibility gap is expected in the α -Mg and MgZn two phase region. Therefore the formation of the Zn-enriched GP zones before the precipitation of the stable MgZn phase is reasonable. Since the atomic radii of Mg and Al are more similar than the atomic radii of Mg and Zn, the Al-rich GP zones may be easily formed from the Mg-rich matrix; the atomic radii of Mg, Zn and Al are 0.172, 0.153 and 0.182 nm, respectively. After an incubation time, *i.e.* midterm stage of aging, the GP zone has a role of nucleation site for the subsequent precipitates. Therefore, higher number density of GP zones in the Al-added alloy than that in the Mg-6Zn alloy can produce more precipitates which enhance the hardness during aging.

4.2 Effect of Al addition on the morphology and distribution of the precipitates in the Mg-Zn alloy

From the TEM microstructures (Fig. 6 and Fig 7), it is obvious that the Al-added alloy exhibits much higher number density of the plate-like and blocky precipitates compared with the Mg-6Zn alloy. Buha¹²⁾ reported that the Zn-enrich GP zones are formed on the $\{0001\}_{\text{Mg}}$ planes in the Mg-Zn alloy after aging at 433 K for 90 min. Moreover, Saito et al.²⁶⁾ found that the planar GP zones in the Mg-Gd-Zn alloy are existed perpendicularly to the c-axis of Mg matrix and parallel to a-axis of the Mg matrix after aging at 473 K for 100 h. Furthermore, the heavier elements, *i.e.* Zn and Gd are enriched into the GP zones while the light element of Mg is depleted on the locations of the GP zones.

The GP zones in the Mg-Zn-Al alloys investigated in the present study will be formed on the $\{0001\}_{\text{Mg}}$ planes and Al will be enriched into the GP zones due to the strong interaction energy between Zn and Al. The lattice misfit between the α -Mg matrix and precipitate phases can be changed by the Al addition. Shin et al.²⁷⁾ discovered that the large solutes, such as Ca, K, Sr and Ba, increase the lattice parameters of Mg in both a-axis and c-axis significantly, and interestingly Ba increases a-axis far more than c-axis. From this result, it can be assumed that the Al addition changes more a-axis than c-axis. The misfit across

the a-axis results large coherency strains parallel to the $(0001)_{\text{Mg}}$, but no coherency strains will exist across the c-axis. Therefore, the growth of not only plate-like precipitate but also blocky precipitate in the c-axis of the matrix will be suppressed differently from the rod-like precipitate.

From Fig. 9 observed along the matrix $[0001]_{\text{Mg}}$ direction, the precipitates in the Al-added alloy are more finely and densely distributed compared with those in the Mg-6Zn alloy. In observation along the $[0001]_{\text{Mg}}$ direction, the rod-like precipitates are aligned along the $\langle 10\bar{1}0 \rangle_{\text{Mg}}$. The aligned precipitates may also contribute to the dispersion hardening. It is found that the precipitates distributed on the prismatic planes of the Mg matrix were most effective in the dispersion strengthening.¹³⁾ In general, the stacking fault energy of an alloy decreases with increasing solute elements. It has been reported that the stacking fault energy of pure Mg is reduced to about one third by the addition of 3 mass% Al.¹⁴⁾ The decrease of the energy can cause the decomposition of dislocations into partial ones with the formation of stacking fault, $1/3\langle 11\bar{2}0 \rangle \rightarrow 1/3\langle 10\bar{1}0 \rangle + 1/3\langle 01\bar{1}0 \rangle$. It is considered that the precipitation takes place to the $\langle 10\bar{1}0 \rangle$ directions, and extends to the prismatic planes. It is found that the precipitates distributed on the prismatic planes of the Mg matrix were most effective in the dispersion strengthening.¹³⁾ Therefore, refinement and changed morphology and distribution of precipitates could provide improved precipitation strengthening in the Al containing Mg-Zn alloy.

5. Conclusion

In this study, the changes of microstructure and aging behavior of a Mg-6.0 mass%Zn alloy by the additions of 3.0 mass%Al was investigated. The following results were obtained.

1) The as-cast Mg-6Zn alloy is mainly composed of the α -Mg, MgZn and MgZn₂ phases, while the as-cast Al-added alloy is composed of the α -Mg and Mg₃₂(Zn,Al)₄₉ phases. The MgZn, MgZn₂ and Mg₃₂(Zn,Al)₄₉ phases are dissolved into the α -Mg matrix after solid-solution treatment.

2) The peak hardness, obtained at aging temperatures of 403 and 433 K, in the Al-added alloy was higher than that in the Mg-6Zn.

3) The TEM microstructure of the Mg-6Zn alloy aged at 433 K for 360 ks shows rod-like β_1' and plate-like β_2' precipitates, whereas four types of precipitates are observed in the Al-added alloy at 433 K for 230.4 ks: rod-like β_1' precipitates parallel to the $[0001]_{\text{Mg}}$ direction, a number of blocky β_1' , plate-like β_2' precipitates and small volume fraction of lath-shape β_2' precipitates.

4) The Cliff-Lorimer plots obtained by the EDS analysis of the β_1' and β_2' phases in the Al-added alloy aged at

433 K for 230.4 ks indicate that those phases contain solute Zn and Al atoms which might affect the nucleation and growth behavior of precipitates.

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