The effect of aging time on microstructure and hardness value of AZ80 Mg Alloy

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Abstract. AZ 80 Magnesium (Mg) alloy (AZ80) is the lightest structural metallic materials with good mechanical properties. However, Mg AZ80 has drawbacks which result in poor ductility and low strength where applications of Mg alloy have been restricted. The AZ80 has high aluminium content can cause the precipitation of β-phase which is Mg17Al12 in Mg-Al alloy. It can affect the mechanical properties such as poor strengthening. This paper was discussed the effect of aging time on microstructure and hardness value of AZ80. The AZ80 samples were cut to 1cm × 1cm. Samples heat treated at 360 °C for one-hour quenching in water. Then, samples aged at 170 °C with different aging times (2 to 8 hours) with same quench. Optical Microscope (OM), Scanning Electron Microscopy (SEM), X-ray Diffraction (XRD) and Vickers Microhardness machine were used to analyse the samples. As the results showed β-Mg17Al12 phase was discontinuously distributed along the grain boundary throughout solid solution treatment. The β-Mg17Al12 phase did not fully dissolve into the α-Mg phase and distributed along the grain boundary. The results showed that sample after 2 hours aging time with highest hardness value 62.5 HV is the optimum sample.

1.0 Introduction

Magnesium (Mg) is the lightest metallic material used in structural applications with a density of 1.738 g/cm³ compared to aluminium (Al) (2.70 g/cm³) and iron (Fe) (7.86 g/cm³). Mg alloys offer an excellent combination of properties that include an excellent strength-to-weight ratio, good impact strengths and fatigue, fairly large thermal and electrical conductivity and excellent biocompatibility [1].

Mg alloy can be classified into two different alloys which are wrought Mg alloy and cast Mg alloy. Mg-Al-Zn series alloys are widely used wrought Mg alloys for AZ31, AZ91, including AZ80 alloy. The wrought Mg alloy performs good mechanical properties compared to cast Mg alloy due to the deformation process that can eliminate pores,
homogenize composition and refine grain [2]. The AZ80 is a light magnesium alloy with excellent corrosion resistance and better forging ability [3]. Besides, a high content of aluminium in alloy such as AZ80 is associated with an appreciable amount of β-phase, which is $\text{Mg}_17\text{Al}_{12}$. The $\text{Mg}_17\text{Al}_{12}$ is harmful to the matrix's corrosion behaviour. Thus, by reducing the micro-galvanic effects, corrosion resistance can be enhanced [4]. However, AZ80 have drawbacks which result in poor ductility and low strength where application of Mg alloy has been restricted [2]. As AZ80 has high Al content caused the precipitation of β-phase in Mg-Al alloy thus effect the mechanical properties such as poor strengthening [3].

Based on the earlier research by Ju-Mei et al., [5], the elongation and tensile strength of the alloy increased significantly. After solid solution treatment, the hardness and yield strength decreased. The durability and strength of the alloy increased with the prolongation of the aging time meanwhile the ductility declined. The Al is the most essential alloying element for magnesium. Commercial AZ80 alloys contain less than 10 wt.% Al (maximum solubility 12.7 wt.% at eutectic temperature 437 °C), and zinc and manganese as secondary alloying elements. The Al contents increases castability, tensile strength, compressive strength, and fatigue resistance. In Mg–Al alloys, Al is normally partly in a solid solution and partly precipitated in the form of β-phase. The $\text{Mg}_17\text{Al}_{12}$, distributed along grain boundaries as a continuous phase as well as part of a lamellar structure [6]. According to Ben-Haroush [4], an appreciable amount of β-phase, related with a high Al content in alloy such as AZ80. Cast AZ80 alloy typically includes a large fraction of precipitate $\text{Mg}_17\text{Al}_{12}$ along the grain boundaries. The $\text{Mg}_17\text{Al}_{12}$ is cathodic with respect to the matrix and it reveals a passive behaviour over a wider pH range than either of its components of Mg or Al.

Pardo et al., [7] stated that an increase in the relative size of the β-phase at the expense of the Al rich-area will be increased the cathode to anode area ratio. Consequences, it causes greater localized corrosion. There is no effect on the nucleation of the primary Mg phase by add small amounts of Zn. Most of the Zn is segregated into a secondary phase close to the end of solidification. The morphology of eutectic phase resulting from non-equilibrium solidification of hypoeutectic Mg-Al alloys depends on the Al content of the alloy [8].

The samples will undergo an artificial aging process after past solution heat treated. Due to Ju-Mei et al., [5], the parameter of the samples was artificial aged at 200 °C for 4 to 32 hours. The continuous precipitates start and grow when the aging treatment prolongs to time which limits the nucleation and growth of discontinuous precipitate. Thus, the $\beta\text{Mg}_17\text{Al}_{12}$ phase precipitates continuous and discontinuous precipitates, which can be identified from their morphologies, in two different ways after aging treatment at 200 °C. Hence, the growth rate of the β phase is fast at the early stage of nucleation, but the growth will stop as it reaches equilibrium. Reported from Huang et al., [3] the nucleation of the continuous precipitation (CP) is a homogeneous process with the β-phase forming directly from the matrix. Therefore, after the aging treatment, there will no CP formed without solution treatment [3].

The effect of aging time was carried out on the AZ80. Due to some deficiencies in AZ80 such as the inadequate strength and poor ductility thus, solution and aging treatment is the recommended treatment to improve the mechanical properties of AZ80.

2.0 Methodology

The commercial AZ80 magnesium alloy ingot has composition of Mg - 8.2 wt.% Al and 0.51 wt.% Zn. This ingot cut to be 1 cm x 1 cm x 1 cm. The samples ground using silicon carbide (SiC) paper and polished to achieve mirror like surface.
Those samples heat treated at 360 °C for 1 hour in a furnace, followed by water quenching. The samples subjected to aging treatment at 170 °C with different aging times (2 hours, 4 hours, 6 hours and 8 hours), followed by water quenching.

The microstructure of the samples observed using optical microscope (OM) and scanning electron microscope (SEM), meanwhile phase analysis characterized using X-ray diffraction (XRD) and hardness using the Vicker’s microhardness.

3.0 Results and Discussions

3.1 Microstructure Analysis

The sample revealed two phases which were α-phase and β-phase. The presence of Al in AZ80 resulted in the formation of the phase β-Mg17Al12 (BCC) structure, forming an incoherent interface with the structure of the α-phase (HCP). The β-phase is a compositional intermetallic stoichiometric compound. There are two distinct and competitive precipitation modes for continuous and discontinuous precipitation of BCC β-Mg17Al12 from the parent HCP phase [9].

Fig. 1 showed the microstructure of untreated AZ80 and those aged at 170 °C for different aging times. Under all conditions, the AZ80 microstructures consist of which is the light background and β-Mg17Al12 that had the black regions or lines. Most of β-Mg17Al12 formed networks in Fig.1 (a) of untreated sample at the grain boundaries and are discontinuously broken and dispersed after the heat treatment. Few β phases can be seen in the untreated sample of AZ80. However, the grains of heat treated samples grew bigger.

![Fig.1. Optical images of AZ80 by different aging time (a) untreated sample; (b) 2 hours; (c) 4 hours; (d) 6 hours and (e) 8 hours.](image-url)
The main $\beta$-$\text{Mg}_{17}\text{Al}_{12}$ phase was discontinuously distributed along the grain boundary throughout solid solution treatment and did not fully dissolve into the $\alpha$-Mg phase. Then, after the aging treatment, the residual $\beta$-$\text{Mg}_{17}\text{Al}_{12}$ phase distributed along the grain boundary. Fig.1 (b) and (c) showed the amount $\beta$ precipitate of each specimen also increases when aging time increases from 2 hours to 4 hours. However, amount of $\beta$ precipitate decrease after 4 hours aging as the prolonging holding time. Due to the heat treatment, the amount and distribution of $\beta$ phase was significantly changed.

During aging treatment, aluminum (Al) atoms diffuse into grain boundaries to form $\beta$ precipitate, reducing the aluminum concentration in the $\alpha$-phase during aging treatment. As the aging time increasing, the proportions of $\beta$ phase $\text{Mg}_{17}\text{Al}_{12}$ grains also increasing. In spite of that, the $\beta$-$\text{Mg}_{17}\text{Al}_{12}$ phase as the temperature elevates and the holding time extended. These results were attributable to the solid solution effect of gains in $\beta$-phase $\text{Mg}_{17}\text{Al}_{12}$, resulting in a reduction in the diffusion phenomenon of $\beta$-$\text{Mg}_{17}\text{Al}_{12}$ phase grain into the $\alpha$-phase Mg matrix. This finding also supported by Chye at al.,[10].

Fig.2 represents the SEM images of the untreated and treated AZ80 with different aging time. Few $\beta$ phases can be seen in the both untreated and treated samples of AZ80 and also there are large particles along grain boundaries. Fig. 2 (b-e) shows the SEM micrographs of AZ80 aged at 170 °C for different aging time, from which the coarse white $\beta$-$\text{Mg}_{17}\text{Al}_{12}$ particle can be observed in the specimens where the phase $\beta$-$\text{Mg}_{17}\text{Al}_{12}$ was distributing discontinuously along the grain boundary after being treated for one hour at 360°C solution treatment followed by water quenching. Due to solution and secondary precipitate, the size of $\beta$-$\text{Mg}_{17}\text{Al}_{12}$ precipitates in the untreated sample alloys is smaller compared to that in the treated sample alloy.

![Fig.2 SEM images of AZ80 by different aging time (a) untreated sample; (b) 2 hours; (c) 4 hours; (d) 6 hours and (e) 8 hours; (f) Magnification of blue arrow at Fig. 3 (b), showing a large amount of lamellar discontinuous precipitation (D.P.) at grain boundaries and granular-shaped continuous precipitates (C.P.) in the matrix](image-url)
It could be observed that high amount of precipitates existed following aging. Fig.2 (f) shows the high magnification SEM micrograph of discontinuous precipitation at grain boundaries and continuous precipitation within grain boundaries as directed by the blue arrow in Fig.2 (b). The figure shows that the $\alpha$-phase was precipitated into two forms from the super-saturated $\alpha$-Mg solid solution, which is the discontinuous precipitate (DP) and continuous precipitate (CP). Furthermore, DP began to grow into the grains with lamellar growth from the grain boundaries, which is a typical morphology for DP of $\beta$-Mg$_{17}$Al$_{12}$. In the remaining areas, CP was found to form within the unoccupied grains of DP and the interface in both DP and CP areas was clear. The discontinuous precipitation stops with the rising aging time, and the continuous precipitates formed in the residual grain regions that are not occupied by discontinuous precipitates are continuing to increase.

3.2 X-Rays Diffraction Analysis

Fig.3 shows the XRD pattern for samples aged at 170 °C with different times (2 hours, 4 hours, 6 hours and 8 hours). The XRD result shows that the AZ80 consists of $\alpha$-Mg, $\beta$-Mg$_{17}$Al$_{12}$, Mg$_{32}$(AlZn)$_{49}$, Al$_3$Mg$_2$, AlMg and Mg$_7$Zn$_3$. However, the $\alpha$-Mg and $\beta$-Mg$_{17}$Al$_{12}$ were mainly present. After the aging treatment, the XRD result in Fig.3 (b) revealed that the $\beta$-Mg$_{17}$Al$_{12}$ phase has the highest diffraction peak between 32°, 34° and 37°. The intensity of $\beta$-Mg$_{17}$Al$_{12}$ phase becomes higher than $\alpha$-Mg. Moreover, by comparing the Fig. 3 (a) and (b), the diffraction peak of $\beta$-Mg$_{17}$Al$_{12}$ phase in Fig.3 (b) is higher than Fig.3 (a). Thus, the higher intensity result in the more phase presents. This result shows positive hardening effect towards the AZ80 alloy that aged at 2 hours.

Fig. 3(c) shows the XRD analysis for AZ80 aged at 170 °C for 4 hours of aging. The $\alpha$-Mg matrix appears and has the highest intensity at the peak 37°. It’s shown five peaks formation of $\alpha$-Mg at the peak 33°, 37°, 48°, 58° and 69°. While the formation of $\beta$-phase decrease where only three peak formed at 40°, 58° and 63°, this analysis can be conclude that aging for four hours will decreased the hardness compared to aging for two hours. In addition, Fig. 3 (d) and (e) shows the decreasing number of $\beta$-phase on both six and eight hours of aging, the number of $\beta$-Mg$_{17}$Al$_{12}$ were deceased. Other peaks have full with formation of $\alpha$-Mg, Mg$_{32}$(AlZn)$_{49}$, Al$_3$Mg$_2$, AlMg and Mg$_7$Zn$_3$. Therefore, from the both Fig. 3(d) and (e), hardening effect of AZ80 alloy that aged at 170 °C for six hours and eight hours are poorer than four hours aging.

Ju-mei et al., [5] found that the diffraction peak of the $\beta$-Mg$_{17}$Al$_{12}$ vanished after solution treatment, but then the $\beta$-phase appeared after the aging treatment. The volume fraction of the $\beta$Mg$_{17}$Al$_{12}$ phase increases very rapidly during the initial aging treatment period, however when the aging time has been prolonged, the volume fraction increase obviously slows down.
3.3 Hardness Analysis

Fig. 4 is the effect of the aging time on the hardness. After two hours aging process, the hardness of samples was slightly dropped. The average hardness value of the untreated sample is 61.4 HV. During aging treatment, the hardness of all samples increased with aging time until it reached the maximal hardness 62.5 HV at two hours of aging time. However, after the two hours aging time, all treated samples were decreased with the increasing of aging time. The hardness values decrease due to the increasing of aging time. Due to the Chye et al., [10] aluminium (Al) atoms diffuse into grain boundaries during aging treatment to form β-precipitate and this will reduce Al concentration during aging treatment in a phase. Moreover, the shape and amount of β precipitate are affecting the hardness. The more the formation of β precipitate will lead to the higher the aging hardness. The reason is caused by the resistance from the β-phase formation where the β-phase movement increased significantly and the structure became more difficult to move. But the prolonging of aging time will cause the material to fracture and reduce the hardness value.

![Image](image-url)
The hardness after heat treatment affected by the amount of secondary phase, the formation of super saturated grains and the change in average grain size. Reduced amount of secondary phase and grain growth will decrease the hardness while when the formation of super saturated grains decreases, the hardness will increases.

The phenomenon of decreasing the hardness value with increasing sensitization temperature and time can be attributed in part to decreasing the dislocation density during sample recrystallization. The density of dislocations would reduce as other grains are framed in the heat treatment of sensitization. It may also attribute precipitation of the Mg-rich stage and isolation of Mg molecules at grain boundaries, which extract solute Mg particles from the α-matrix. Thus, the remaining Mg particles left behind in a solid solution become ineffective barriers to dislocation movement [11].

4.0 Conclusion

1. The β-Mg₃Al₂ phase was getting to distribute along the grain boundary discontinuously via aging treatment at 170 °C after solution treatment at 360 °C for 1 hour followed by water quenching.
2. Aging treatment was found as a method of improving mechanical properties at 170 °C after solution treatment. Therefore, the optimal aging time of AZ80 is 2 hours as the highest hardness value is 62.5 HV.

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