Influence of aging modes on microstructure and mechanical properties of AZ80 magnesium alloy

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Abstract: The microstructure and mechanical properties of AZ80 magnesium alloy after solid solution and aging treatments were studied by using optical microscope (OM), X-ray diffraction (XRD), scanning electron microscopy (SEM) as well as tensile testing. The results indicated that $\beta$-Mg$_{17}$Al$_{12}$ phase was getting to distribute discontinuously along the grain boundary after treated at 395°C ageing for 12 h followed by water-cooling, but it did not dissolve into $\alpha$-Mg completely. The residual $\beta$-Mg$_{17}$Al$_{12}$ phase distributed along the grain boundary and had block-like or island shapes. The size of $\alpha$-Mg was getting to be coarsening but not significantly. The $\beta$-Mg$_{17}$Al$_{12}$precipitates appeared in discontinuous and continuous patterns from supersaturated $\alpha$-Mg solid solution after aged at 200°C. The precipitation patterns were associated with the aging time essentially. The tensile strength and elongation of the alloy increased significantly but the hardness and yield strength decreased after solid solution treatment. However, with the prolonging of aging time, the hardness and strength of alloy increased while the ductility decreased.

Keywords: magnesium alloy; solid solution; aging treatment; microstructure; mechanical property

1 Experimental procedures

The development in light metal field is continuously pushing magnesium alloys into an important position in industrial applications for its high specific strength and specific stiffness, good castability, damping characteristic, machining and excellent electromagnetic shielding. So, more and more magnesium alloy structural components have been used in aviation, automobile and electron industries [1-4]. Due to the formation of large amount of brittle intermetallic $\beta$-Mg$_{17}$Al$_{12}$ phases [5] that are mainly distributing along the $\alpha$-Mg grain boundary continuously, the cast magnesium alloy with high Al content, such as AZ80 and AZ91, has the decreasing toughness and ductility [6]. In previous researches, many researchers have focused their efforts on improving the performance of magnesium alloys through heat treatment, which are widely applied in industries. Many literature reviews about homogenizing or solution and aging heat treatment of Mg-Al alloys [7-9] have been published.

Therefore, the purpose of this present study was to discuss the influence of aging modes on the microstructure and mechanical properties of AZ80 magnesium alloy after solution treatment.

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2 Results and discussion

2.1 Influence of solid solution treatment on microstructure of AZ80 alloy

Figure 1 shows X-ray diffraction patterns of as-cast and as-solution AZ80 alloy. The XRD results reveal that the AZ80 alloy is composed of $\alpha$-Mg matrix and $\beta$-Mg$_{17}$Al$_{12}$ phases. The $\alpha$-Mg matrix also appears in AZ80 alloy after solid solution treatment at 395°C for 12 h followed by water-cooling, while the peaks of $\beta$-Mg$_{17}$Al$_{12}$ phase disappear significantly. The typical SEM morphology of as-cast AZ80 alloy, as shown in Fig.2, indicates that most of the $\beta$-Mg$_{17}$Al$_{12}$ eutectic phases, owing to the non-equilibrium solidification, distributed along the grain boundary with block-like or island shapes, only the small amount of them formed within $\alpha$-Mg matrix. The eutectic structure occurs in the grain boundary as the divorced eutectic. Figure 3 shows the OM images of AZ80 alloy treated at 395°C for 12 h, water-cooling. Compared with the as-cast microstructure, the heat-treated grains grew bigger, but not enough. The main $\beta$-Mg$_{17}$Al$_{12}$ phase discontinuously distributed along the grain boundary and did not completely dissolve into $\alpha$-Mg phase during solid solution treatment. The residual $\beta$-Mg$_{17}$Al$_{12}$ phase distributed along the grain boundary and had block-like or island shapes. Therefore, the results indicated that the $\beta$-Mg$_{17}$Al$_{12}$ phase, discontinuously distributed along the grain boundary, did not dissolve into $\alpha$-Mg phase completely at 395°C for 12 h. It could be possible to obtain the homogeneous microstructure of $\alpha$-Mg solid solution only by either increasing solution temperature or prolonging holding time.

2.2 Influence of aging treatment on microstructure of AZ80 alloy

XRD analysis was performed on selected specimens to identify the phases via alloy aged at 200°C after solution treatment at 395°C for 12 h, as shown in Fig. 4. The diffraction peak of $\beta$-Mg$_{17}$Al$_{12}$ phase that had vanished after solution treatment reappeared after the aging treatment. It illuminates that the $\beta$-Mg$_{17}$Al$_{12}$ phase precipitated from oversaturated solid solution owing to aging treatment. The longer the aging time and the stronger the diffraction peak of $\beta$-Mg$_{17}$Al$_{12}$, that is to say, the amount of $\beta$-Mg$_{17}$Al$_{12}$ phase is prominent. The volume fraction of $\beta$-Mg$_{17}$Al$_{12}$ phase as a function of aging time is shown in Fig.5, in which the volume fraction at the starting point 0 belongs to the alloy after solution treatment. It can be seen from Fig.5 that the volume fraction of $\beta$-Mg$_{17}$Al$_{12}$ phase increases very quickly at the initial period of the aging treatment, but when the aging time prolonged to ~30 h, the increasing of the volume fraction evidently slows down. It is attributed to the fact that the degree of saturation of the as-solution alloy is big at the beginning of precipitating, and the activating force of precipitation is big too. So, the free energy for phase transformation gradually releases.

| Table 1 Chemical composition of as-cast AZ80 magnesium alloy, wt-% |
|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|
|     Al          |     Zn          |     Mn          |     Si          |     Fe          |     Cu          |     Ni          |     Mg          |
| 7.8-9.2        | 0.2-0.8        | 0.15-0.5       | 0.01          | 0.01          | 0.01          | 0.01          | 0.05          | Bal.           |

Fig.1 X-ray diffraction patterns of as-cast and as-solution AZ80 alloy

Fig.2 SEM morphology of as-cast AZ80 alloy

Fig.3 OM image of as-solution AZ80 alloy

Fig.4 X-ray patterns of AZ80 alloys aged at 200°C
exhaustedly, which causes the precipitation rate to slow down, and the amount of precipitate $\beta$-$\text{Mg}_17\text{Al}_{12}$ phase during the late stage of aging has no significant change. The microstructural transformation in Fig.6 also testifies the results.

Many researchers [10] have indicated that the precipitation of $\beta$-$\text{Mg}_17\text{Al}_{12}$ phase occurred as two forms, discontinuous and continuous. Normally, the two forms coexist, but usually the discontinuous precipitation separates out as the forerunner, and then carries on the continuous precipitation.

Figure 6 shows the SEM micrographs of AZ80 alloy aged at 200°C for various periods, from which the coarse white $\beta$-$\text{Mg}_17\text{Al}_{12}$ particle can be still observed in the specimens after solution treatment (395°C, 12 h, water-cooling). After the aging treatment for 4 h, discontinuous precipitates preferentially form at the grain boundaries, as shown in Fig. 6(a). With the increasing of aging time, the discontinuous precipitation ceases, and the continuous precipitates formed in the remaining regions of the grain that are not occupied by discontinuous precipitates increases, as shown in Fig. 6(b, c). The size of $\beta$-$\text{Mg}_17\text{Al}_{12}$ precipitates in the aging alloys is smaller in comparison with that in the as-cast alloy, due to solution and secondary precipitate.
2.3 Effects of aging treatment on mechanical properties of alloy

Table 2 listed the average values of Brinell hardness (HB), tensile strength ($R_m$), yield strength ($R_{0.2}$) and elongation (A) for all of the specimens tested in the study. It can be seen that solid solution and aging treatments of the AZ80 alloy resulted in a significant influence on mechanical properties. Compared with as-cast AZ80 alloy, yield strength of the alloys with solid solution treatment remarkably decreased, while the tensile strength and ductility increased significantly. Namely, the yield strength of solid solution AZ80 is lower than that of as-cast AZ80, which is ascribed to the coarsening of β-Mg₁₇Al₁₂ phase distributed along the grain boundary and dissolved into α-Mg. The β-Mg₁₇Al₁₂ phase dissolved into α-Mg phase can cause a remarkable weakening of the alloy for dislocation motion. However, the dissolution of brittle β-Mg₁₇Al₁₂ phase also resulted in the weakening of disseverment for α-Mg matrix. Therefore, the elongation and tensile strength obviously enhanced, even more, the average value of elongation and the ratio of tensile to yield strength ($R_m/R_{0.2}$) increase to 7.22% from 4.96%, 1.84 from 1.43 of as-cast, respectively. In addition, the solid solution treatment not only made the secondly strengthening β-Mg₁₇Al₁₂ phase disappeared, but also drove the solute atoms to dissolve into the matrix, resulting in a slightly decrease in the hardness of the alloy. When the as-solution specimens of the alloys were aging-treated at 200 °C for various period, the β-Mg₁₇Al₁₂ phase precipitated from the oversaturated solid solution of α-Mg. With prolonging the aging time, the amounts of β-Mg₁₇Al₁₂ phase discontinuously precipitated at the grain boundaries and continuously precipitated within the grain increased gradually. Moreover, the grain size of β-Mg₁₇Al₁₂ precipitates is relatively finer, compared with the β-Mg₁₇Al₁₂ phase of as-cast alloy, and morphology exhibits a biggish transformation. Accordingly, the hardness and tensile strength of the alloy increased obviously and their average values increased from 62.5 HB and 162.5 MPa of as-cast to 77.1 HB and 199.41 MPa aged at 200 °C for 32 h, respectively. However, the ductility index A of alloys decreased remarkably with the prolonging of aging time, even the average value of elongation at 200 °C for 32 h is lower than that of as-cast AZ80 alloy. The reason can be attributed to the increasing restriction of dislocation motion from precipitate β-Mg₁₇Al₁₂ phase with the aging process, in which the dislocations are packed, leading to the stress accumulated large enough to cause micro-crack, as a result, the ductility of alloy decreases gradually.

Comparison with mechanical property data between various heat treatment modes, it can be suggested that the discontinuous precipitation of β-Mg₁₇Al₁₂ phase has more contribution to promote the mechanical properties of AZ80 magnesium alloy.

3 Conclusions

(1) The β-Mg₁₇Al₁₂ phase was getting to distribute along the grain boundary discontinuously after treated at 395 °C ageing for 12 h followed by water-cooling. However, it did not completely dissolve into α-Mg phase. The residual β-Mg₁₇Al₁₂ phase distributed along the grain boundary and had block like or island shapes. The size of α-Mg was getting to be coarser but not significantly.

(2) The processes of β-Mg₁₇Al₁₂ phase precipitation aged at 200 °C include two patterns, continuous and discontinuous, which were associated with the aging time essentially.

(3) The tensile strength and elongation of the alloy increased remarkably but the hardness and yield strength decreased after solid solution treatment. However, with the prolonging of aging time, the hardness and strength of alloy increased while the ductility decreased.

References


Table 2 Mechanical properties of alloys for age treatment at 200°C

<table>
<thead>
<tr>
<th>Conditions</th>
<th>HB</th>
<th>$R_m$</th>
<th>$R_{0.2}$</th>
<th>A (%)</th>
</tr>
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<tbody>
<tr>
<td>as-cast</td>
<td>62.5</td>
<td>162.05</td>
<td>113.66</td>
<td>4.96</td>
</tr>
<tr>
<td>395 °C, 12 h</td>
<td>58.1</td>
<td>172.17</td>
<td>93.85</td>
<td>7.22</td>
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<td>200 °C, 4 h</td>
<td>72.8</td>
<td>184.46</td>
<td>102.29</td>
<td>6.03</td>
</tr>
<tr>
<td>200 °C, 24 h</td>
<td>76.3</td>
<td>196.29</td>
<td>128.81</td>
<td>4.42</td>
</tr>
<tr>
<td>200 °C, 32 h</td>
<td>77.1</td>
<td>199.41</td>
<td>126.32</td>
<td>3.66</td>
</tr>
</tbody>
</table>