Effects of yttrium on microstructure and mechanical properties of Mg-6Al magnesium alloy

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Abstract: Since Y has a great solid solubility in magnesium alloys, it helps enhancing the heat-resistant property of magnesium alloys. The effects of Y on microstructures and mechanical properties of Mg-6Al alloy have been studied in this work. The results show that Y addition refines grains of Mg-6Al alloy, and reduces the amount of the Mg₁₇Al₁₂ phase. At the same time, the high melting-point Al₂Y phase particles are formed. According to the mathematical model of the two-dimensional lattice misfit proposed by Braffit, it is believed that the Al₂Y particles can serve as the nucleation sites for α-Mg. After T6 treatment, both elongation and ultimate tensile strength of Mg-6Al alloy at the room temperature and high-temperature increased firstly and then decreased, with increasing Y addition. The peak mechanical properties were achieved in the Mg-6Al-1.2Y alloy system. Y addition appears to change the fracture characteristic of Mg-6Al alloy. With 1.2wt%Y, the fracture surface of the alloy showed a lot of dimples and tearing ridges which connected the microscopic dimples and the fracture is mixed fracture of quasi-cleavage and ductile fracture.

Key words: yttrium; Mg-6Al alloy; microstructure; mechanical properties

Table 1: Chemical compositions of the tested alloys (wt.%)

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Al</th>
<th>Y</th>
<th>Mg</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mg-6Al</td>
<td>6.0</td>
<td>0</td>
<td>Bal.</td>
</tr>
<tr>
<td>AY606</td>
<td>6.0</td>
<td>0.6</td>
<td>Bal.</td>
</tr>
<tr>
<td>AY612</td>
<td>6.0</td>
<td>1.2</td>
<td>Bal.</td>
</tr>
<tr>
<td>AY618</td>
<td>6.0</td>
<td>1.8</td>
<td>Bal.</td>
</tr>
<tr>
<td>AY624</td>
<td>6.0</td>
<td>2.4</td>
<td>Bal.</td>
</tr>
</tbody>
</table>

Magnesium alloys are the lightest engineering structural material [1]. Magnesium alloys are being increasingly used in the aerospace, communications, aircraft and automotive industries due to their higher specific strength, stiffness, higher dimensional stability and damping characteristics, good machining properties and easy recycle-ability. But so far, magnesium alloys have not been in the large-scale application because of its poor high-temperature properties/performances. When the working temperature is above 120°C, the high temperature strength and creep resistance of magnesium alloy are decreased significantly [2].

Through the analysis on the form and distribution of the precipitated phases as well as the tensile properties, the strengthening mechanism has been investigated.

1 Experimental procedures

Chemical compositions of the studied alloys are listed in Table 1. Alloy ingots were prepared from high purity Mg (99.95%), Al (99.98%), and Mg–24.54Y (wt.%) master alloy in an induction melting furnace under the shielding gas consisting of mixed CO₂ and SF₆ with the ratio of 100:1. The molten alloys were held for 15 min at 720°C, and then were poured into a metallic mold which was preheated to 250°C. The small samples cut out from the ingots and covered with MgO powders were solution heat treated at 420°C for 24 h and then water quenched. Artificial aging treatments were performed at 220°C for 14 h.

Tensile tests were carried out at a strain rate of 1 mm/min in an AG-I250kN precision universal material test machine at the room temperature (20°C) and elevated temperatures (150°C, 175°C). Microstructures, fracture surface morphologies and compositions of the alloys were analyzed by optical microscopy and scanning electron microscopy (SEM) (JSM-5610LV). Phase analyses were performed with an X’pert X-ray diffractometer (XRD). The grain size was determined using a linear intercept method from a large number of non-overlapping measurements.
2 Result and discussion

2.1 Microstructure

Figure 1 shows the micrographs of the studied alloys. The microstructure of Mg-6Al contains the matrix (α-Mg) and discontinuous reticular lath compounds Mg17Al12, which were mainly distributed at grain boundary and intragranular, as shown in Fig.1(a). With increasing Y content, the morphology and distribution of Mg17Al12 phases were changed. Lath Mg17Al12 phases were dissolved, leading to formation of larger circular/nodular precipitates and particulate phases, as shown in Fig.1(b). The electron-negativity differences of two elements were used for predicting the possibility to form metallic compound. The larger the electron-negativity differences, the stronger the familiarity and the higher the possibility to form metallic compound. Electron-negativity difference between Al (electron-negativity value is 1.5) and Y (electron-negativity value is 1.2) is larger than that between Mg (electron-negativity value is 1.3) and Y, so Al is much easier to form Al-Y compounds than Mg. According to XRD results and previous studies[4], the larger circular/nodular precipitates were Mg17Al12 phases and the tiny particle phases were rare-earth compounds Al2Y. It can be seen from Fig.2(b) that almost no Mg17Al12 phase is observed, suggesting the particulate phases should be the rare-earth compounds Al2Y [Fig.1(c)]. When the content of Y was at 1.8wt.%, the rare-earth phase Al2Y grew up, especially when the content of Y was at 2.4wt.%, the rare-earth phase Al2Y was segregated, as shown in Fig.1(d) and Fig.1(e). The above analysis shows that a small amount of Y addition to Mg-6Al can inhibit Mg17Al12 phase precipitates and promote the formation of the new phase Al2Y. With further increase of Y addition, Mg17Al12 phase disappeared and the Al2Y phase grew up and tended to form clusters. The reasons for this may be[5]: (1) Y has lower diffusion velocity in solid-state alloys, so solution and aging treatment cannot eliminate segregation and aggregation of Y which occur during the casting process, especially when the amount of Y addition to Mg-6Al is too high; and (2)
the higher the Y content, the more the rare-earth compound of Al\textsubscript{17}Y. Al\textsubscript{2}Y phases which are formed firstly during solidification are squeezed into the liquid-solid interface and form cluster.

With the increase in Y content, the average grain size of Mg-6Al, AY606, AY612, AY618 and AY624 are 333.3 μm, 219.8 μm, 170.9 μm, 206.8 μm, 192.3 μm, respectively. This may be explained by the following:

1) According to the work by Bramfitt\cite{6}, the mathematical model of the two-dimensional lattice misfit is:

\[ \delta = \sum_{\gamma} \left( \left[ d_{\gamma} \right] - \cos \theta \left[ d_{\gamma} \right] \right) / 3 \times 100\% \]

where \((hkl)\) is the low-index plane of the substrate, \([uvw]\) is the low-index direction in \((hkl)\); \((hkl)\) is the low-index plane in the nucleated solid, \([uvw]\) is the low-index direction in \((hkl)\); \(d[uvw]\) and \(d[uvw]\) are the atomic spacing along \([uvw]\) and \([uvw]\); \(\theta\) is the angle between \([uvw]\) and \([uvw]\). Bramfitt’s research showed that one criterion for heterogeneous nucleation is that the disregistry of nucleation plane should be less than 15%. Table 2 shows the matching parameters between the low-index plane of Al\textsubscript{2}Y and \(\alpha\)-Mg. Al\textsubscript{2}Y is cubic crystal structure, \(a = 0.786 \times 10^{-9}\) nm. By calculation, \(\delta = 10.94\%\) when the orientation relationship between Al\textsubscript{2}Y phase and \(\alpha\)-Mg is \((111)_{\text{Al}_2Y}//(1100)_{\alpha\text{-Mg}}, \) which is less than 15%. Therefore, Al\textsubscript{2}Y can probably cause the heterogeneous nucleation of the \(\alpha\)-Mg phase based on this orientation relationship.

2) Y atoms are squeezed into the solid-liquid interface during solidification, which leads to constitutional supercooling and promotes the formation of a nucleus. At the same time, the Al\textsubscript{2}Y forms among interdendritic \(\alpha\)-Mg phase and can mechanically hinder the growth of \(\alpha\)-Mg grains.

### Table 2: The matching parameter between the low-index plane of \(\alpha\)-Mg and Al\textsubscript{2}Y

<table>
<thead>
<tr>
<th>[hkl]\textsubscript{\text{Al}_{2}Y}</th>
<th>[101]</th>
<th>[011]</th>
<th>[2\overline{1}1]</th>
<th>[0\overline{1}0]</th>
<th>[\overline{1}0\overline{1}]</th>
<th>[\overline{1}0\overline{0}]</th>
<th>[\overline{1}1\overline{2}]</th>
<th>[\overline{1}2\overline{0}]</th>
<th>[1\overline{1}0]</th>
</tr>
</thead>
<tbody>
<tr>
<td>[hkl]\textsubscript{\alpha\text{-Mg}}</td>
<td>[2\overline{1}\overline{\overline{1}}0]</td>
<td>[\overline{1}\overline{2}0]</td>
<td>[1\overline{0}0]</td>
<td>[2\overline{2}0]</td>
<td>[2\overline{1}\overline{0}]</td>
<td>[1\overline{1}20]</td>
<td>[1\overline{1}23]</td>
<td>[\overline{1}\overline{2}3]</td>
<td></td>
</tr>
<tr>
<td>(\theta)</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>14.5</td>
<td>14.5</td>
<td>0</td>
<td>4</td>
<td>4</td>
</tr>
<tr>
<td>(\delta)</td>
<td>14%</td>
<td>18.36%</td>
<td>10.94%</td>
<td></td>
<td></td>
<td></td>
<td></td>
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<td></td>
</tr>
</tbody>
</table>

### 2.2 The mechanical properties at room and elevated temperatures

Figure 3 shows the results of mechanical properties of the studied alloys. As can be seen in Fig.3(a), with increasing Y content, the mechanical properties at room temperature and elevated temperatures (150°C, 175°C) firstly increased and then decreased. When the content of Y addition was 1.2wt.%, the values of the tensile strength (\(R_m\)) at room temperature and elevated temperature (150°C, 175°C) peaked simultaneously at 224 MPa, 221 MPa and 192 MPa, respectively. When the content of Y was above 1.2wt.%, the strength of the studied alloy decreased sharply. The relationship between elongation (\(\Delta\)) and RE content in Mg-6Al alloys is shown in Fig.3(b). The elongation of the studied alloys at room temperature and at elevated temperatures were also first up, and then down. When the content of Y was 1.2wt.%, the elongation values at room temperature and elevated temperatures (150°C, 175°C) were all up to their maximum of 7.0%, 9.0%, and 9.4% respectively. When the content of Y was above 1.2wt.%, the elongation of the alloys decreased quickly.

For Mg-6Al alloy, Mg\textsubscript{17}Al\textsubscript{12} phase is the main strengthening phase, but the thermal stability of the Mg\textsubscript{17}Al\textsubscript{12} phase (with a melting point of 437°C) is poor, which is easily softened and coarsened with increasing temperature. Since the Mg\textsubscript{17}Al\textsubscript{12} phase is mainly distributed at grain boundaries, the softening and coarsening of the Mg\textsubscript{17}Al\textsubscript{12} phase weaken the grain boundaries at elevated temperatures, becoming the key factors accounting for the low heat resistance of the Mg-6Al alloy. The
main mechanisms of enhancing the heat-resistant properties by Y addition to Mg-6Al alloys are solution strengthening, dispersion strengthening of Al₂Y and the reduction of Mg₁₇Al₁₂ phase. The high melting point of Al₂Y phase which distributed at grain boundaries has higher thermal stability than Mg₁₇Al₁₂ phase, which can in turn stabilize the surrounding grains and effectively resist dislocation movement and grain boundary sliding at elevated temperatures. With the more content of Y, the Mg₁₇Al₁₂ phase decreases. But when the Y content is too high, a lot of asymmetric conglomerations (Al₂Y) appear, weakening the effect of dispersion strengthening, causing segregation of composition and microstructure, increasing stress concentration, and leading to the decrease of the elevated temperature mechanical properties of the studied alloys.

2.3 Fracture behavior observation

SEM images of the tensile fracture surfaces are shown in Fig.4. When the content of Y addition is at 0.6wt.%, the main characters of the fracture surfaces of the studied alloys are cleavage fracture which was reflected by the cleavage steps and river pattern [Fig.4(a)]. When the content of Y addition is up to 1.2wt.%, a lot of dimples and tearing ridges which connected the microscopic dimples are observed, as shown in Fig.4(b). Therefore, the alloy exhibited good ductility. When the content of Y addition is above 1.2wt.%, the second phases start to have major segregation and aggregation, causing the decrease of the amount of the dimples. A great amount of particle phases were observed at the fracture surfaces of the studied alloys, as shown in Fig.4(c) and (d).

(a) AY606; (b) AY612; (c) AY618; (d) AY624

Fig.4: Fracture surface of AY alloys at room temperature
3 Conclusions

(1) With appropriate Y addition to Mg-6Al alloy, the grains of Mg-6Al alloys can be dramatically refined and the quantity of the Mg17Al12 phase can be reduced. Meanwhile, high melting-point Al2Y phases are formed.

(2) Y addition improves the tensile strength of Mg-6Al alloys at room and elevated temperatures. When the content of Y is 1.2 wt.%, the mechanical properties were optimum.

(3) Y addition changes the fracture characteristics of Mg-6Al alloys. When the content of Y is 1.2 wt.%, on the fracture surface of the alloy a lot of dimples and tearing ridges which connect the microscopic dimples are observed and the fracture surface shows mixed fracture of quasi-cleavage and ductile fracture morphology.

References


