Tensile Properties at Room Temperature to 823 K of Mg–4Y–3RE Alloy

Mamoru Mabuchi¹, Yasumasa Chino¹ and Hajime Iwasaki²

¹National Institute of Advanced Industrial Science and Technology (AIST), Institute for Structural and Engineering Materials, Nagoya 463-8560, Japan
²College of Engineering, Department of Materials Science and Engineering, Himeji Institute of Technology, Himeji 671-2201, Japan

This paper describes tensile properties of a peak-aged Mg–4Y–3RE alloy at room temperature to 823 K with \(10^{-5} - 10^{-1}\) s\(^{-1}\). The Mg alloy exhibited high strength (> 250 MPa) at room temperature to 473 K. However, the strength rapidly decreased at 573 K. It is suggested that a large decrease in strength at 573 K is attributed to grain boundary sliding. Also, elongation increased rapidly at 723–823 K. This is likely to arise from the relatively high strain rate sensitivity of about 0.3 due to the glide-controlled dislocation creep.

(Received February 12, 2002; Accepted June 3, 2002)

Keywords: magnesium, alloys, mechanical properties, tensile tests, grain boundary sliding, glide-controlled dislocation creep

1. Introduction

Magnesium alloys have high potential as light-materials for use in the structural components in automobile and aerospace applications due to their low density. It has been demonstrated that Mg alloys containing yttrium (Y) and/or rare earth metals (RE) show excellent heat-resistance. For example, a Mg–5Y–4RE alloy exhibits good creep resistance. It was shown that because the grain boundary diffusion coefficient of Mg is very large, the strengthening mechanisms reduced by diffusion are not effective to attain high strength at elevated temperatures for Mg based materials. This indicates that the strengthening mechanism such as the Orowan process contributes to improvement of elevated temperature strength in Mg based materials since the Orowan process is not easily reduced by diffusion. Recently, Nie and Muddle investigated the precipitates of a Mg–5Y–4RE alloy by TEM observation and they reported that there was little change in the distribution of the \(\beta\) plate precipitates even after annealing at 523 K for 8.64 \times 10^6\ s. Hence, the excellent creep resistance of Mg–Y–RE alloys may arise from the precipitation strengthening mechanism.

However, it has been reported that the strength of a Mg–4Y–3RE alloy decreased rapidly at 573 K. This fact suggests that the deformation mechanisms of Mg–Y–RE alloys are complicated. Hence, it is worthwhile to investigate deformation characteristics in a wide range of temperature for understanding of the complicated deformation mechanisms of Mg–Y–RE alloys. In the present paper, tensile properties of the peak-aged Mg–4Y–3RE alloy are investigated at room temperature ~ 823 K.

2. Experimental Procedure

The chemical composition of the Mg–Y–RE alloy used is listed in Table 1. The cast Mg alloy was annealed at 798 K for 1.8 \times 10^4\ s for solution treatment and then was aged at 473 K for 1.08 \times 10^4\ s for peak-aging treatment. The Mg\(_{12}\)NdY phase was precipitated by aging. Details of aging behavior and precipitates are described in the previous study. The grains of the aged alloy were almost equiaxed and the grain size was 61 \(\mu\)m.

Tensile tests were carried out at room temperature ~ 823 K with an initial strain rate range of \(10^{-5} - 10^{-1}\) s\(^{-1}\). The dimensions of tensile specimens were 10 mm in gauge length and 2.5 mm in gauge diameter, respectively. The samples required about 1.8 \times 10^3\ s to equilibrate at the given testing temperatures prior to initiation of straining. The precipitates in the specimens deformed to failure at room temperature and 573 K were observed by transmission electron microscope. Additional tensile tests with the specimens lined by a polishing paper were conducted to investigate grain boundary sliding, where the specimens were deformed to \(\varepsilon = 10\%\). The transverse offset was measured from shift of the lines at grain boundaries. The strain due to grain boundary sliding is given by

\[\varepsilon_{GBS} = \frac{\Phi \bar{w}}{\bar{L}} \tag{1}\]

where \(\varepsilon_{GBS}\) is the strain due to grain boundary sliding, \(\bar{w}\) is the transverse offset, \(\bar{L}\) is the average linear intercept grain size and \(\Phi\) is a constant (= 1.1\(^{12}\)).

The contribution of grain boundary sliding to total strain is given by

\[\xi = \frac{\varepsilon_{GBS}}{\varepsilon_{tot}} \tag{2}\]

where \(\xi\) is the contribution of grain boundary sliding to total strain and \(\varepsilon_{tot}\) is the total strain.

3. Results and Discussion

The nominal stress-nominal strain curves at room temperature, 373, 473 and 573 K with \(10^{-5} - 10^{-1}\) s\(^{-1}\) are shown in Fig. 1. Strain hardening behavior was apparently observed from the stress-strain curves at room temperature, 373 and 473 K, except the one at 573 K. It is of interest to note that the flow stress at 573 K was significantly low, compared to those at room ~ 473 K.

The variation in ultimate tensile strength at \(10^{-5} - 10^{-1}\) s\(^{-1}\) as a function of testing temperature is shown in Fig. 2. A
Table 1 The chemical composition of the Mg–4Y–3RE alloy.

<table>
<thead>
<tr>
<th></th>
<th>Mg</th>
<th>Y</th>
<th>RE</th>
<th>Zr</th>
<th>Li</th>
<th>Zn</th>
<th>Cu</th>
<th>Mn</th>
<th>Si</th>
<th>Ag</th>
<th>Ni</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>mass (%)</td>
<td>Bal.</td>
<td>4.0</td>
<td>3.2</td>
<td>0.41</td>
<td>0.15</td>
<td>&lt;0.03</td>
<td>&lt;0.03</td>
<td>&lt;0.03</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>&lt;0.005</td>
<td>0.004</td>
</tr>
</tbody>
</table>

decrease in strength from room temperature to 473 K was relatively low and high strength > 250 MPa was obtained at an elevated temperature of 473 K in the strain rate range investigated. However, it should be noted that the strength decreased rapidly at 573 K.

Kamado et al.7,8) revealed that Mg alloys containing Gd, Y and Zr showed high strength > 200 MPa at an elevated temperature of 573 K. The variation in ultimate tensile strength of Mg–RE alloys as a function of testing temperature is shown in Fig. 3. The aged Mg–Gd–Zr and Mg Gd–Y–Zr alloys7,8) showed a large decrease in strength at 573 K as well. The data of solution-treated Mg–Y–RE alloy11) are added in Fig. 3. It should be noted that the solution-treated Mg–Y–RE alloy also exhibited a large decrease in strength at 573 K as well as the peak-aged Mg–Y–RE alloy. The precipitates coarsened during the tests at 573 K (Fig. 4). However, the fact that both the peak-aged alloy and the solution-treated alloy exhibited a large decrease in strength at 573 K cannot be explained only by change of the precipitation state.

The variation in ln $\sigma$ as a function of $1/T$ for the aged Mg–Y–RE alloy ($\dot{\varepsilon} = 10^{-5}$ s$^{-1}$) and the solution-treated Mg–Y–RE alloy$^{11}$ is shown in Fig. 5, where $\sigma$ is the maximum flow stress and $T$ is the absolute temperature. For both alloys, the slopes in a temperature range blow 473 K were very low, however, the slope drastically changed around 573 K. This indicates that a change in deformation process occurred around 573 K. It is therefore suggested that a change in deformation

---

**Fig. 1** The nominal stress-nominal strain curves at room temperature, 373, 473 and 573 K with $10^{-5}$ s$^{-1}$.

**Fig. 2** The variation in ultimate tensile strength at $10^{-5}$–$10^{-1}$ s$^{-1}$ as a function of testing temperature.

**Fig. 3** The variation in ultimate tensile strength of Mg–RE alloys as a function of testing temperature.

**Fig. 4** Transmission electron micrographs of the precipitates in the specimens deformed to failure at $10^{-5}$ s$^{-1}$, where (a) room temperature and (b) 573 K.
process is responsible for the decrease in strength at 573 K.

The grain boundary diffusion coefficient of Mg is large.\textsuperscript{9} Hence, it is suggested that a drastic change in deformation process around 573 K for the Mg–Y–RE alloys is related to grain boundary sliding. The surfaces of the specimens deformed to $\varepsilon = 10\%$ at $10^{-5}$ s$^{-1}$ were observed by scanning electron micrographs to investigate grain boundary sliding. The results are shown in Fig. 6, where the testing temperature is (a) room temperature, (b) 373 K, (c) 473 K and (d) 573 K, respectively. Inspection of Fig. 6 revealed that grain boundary sliding occurred at 573 K. It should be noted that grain boundary sliding significantly occurred at 573 K, in spite of the relatively large grain size of 61 $\mu$m. On the other hand, there was no evidence for grain boundary sliding at room temperature $\sim 473$ K. Therefore, the active grain boundary sliding is likely to be responsible for the large decrease in strength at 573 K in the Mg–Y–RE alloy.

The surfaces of the specimens deformed to $\varepsilon = 10\%$ at $10^{-3}$ and $10^{-1}$ s$^{-1}$ with 573 K are shown in Fig. 7. The evidence for grain boundary sliding was observed even at a high strain rate of $10^{-1}$ s$^{-1}$.

The variation in contribution of grain boundary sliding to total strain at 573 K as a function of strain rate is shown in Fig. 8. The contribution of grain boundary sliding to total strain was 30$\%$ at $10^{-3}$ s$^{-1}$. The contribution of grain boundary sliding decreased with increasing strain rate. Anyway, it is clear that grain boundary sliding strongly contributed to the deformation at 573 K in the strain rate investigated.

Raj and Ashby\textsuperscript{14} developed the rate of grain boundary sliding caused by diffusion. According to their results, the rate of grain boundary sliding due to grain boundary diffusion is given by

$$\dot{U} = \frac{8\tau \Omega \delta}{kT_h^2} D_{GB} \tag{3}$$

where $\dot{U}$ is the rate of grain boundary sliding due to grain boundary diffusion, $\tau$ is the shear stress, $\Omega$ is the atomic volume, $\delta$ is the effective thickness of the grain boundary, $D_{GB}$ is the grain boundary diffusion coefficient, $k$ is the Boltzmann’s constant and $h$ is the total height of the boundary shape. From eq. (3), the rate of grain boundary sliding due to grain bound-
The surfaces of the specimens deformed to \( \varepsilon = 10\% \) at 573 K, where (a) \( 10^{-3} \) s\(^{-1} \) and (b) \( 10^{-1} \) s\(^{-1} \).

The variation in contribution of grain boundary sliding to total strain at 573 K as a function of strain rate.

The variation in \( \Omega \delta \bar{D}_{\text{GB}}/T \) as a function of \( T/T_m \) for Mg, Al and Fe, where \( \Omega \) is the atomic volume, \( \delta \) is the effective thickness of the grain boundary, \( \bar{D}_{\text{GB}} \) is the grain boundary diffusion coefficient, \( T \) is the absolute temperature and \( T_m \) is the melting point.

The variation in elongation to failure at \( 10^{-5} - 10^{-1} \) s\(^{-1} \) as a function of testing temperature is shown in Fig. 10. The elongation increased significantly in an elevated temperature range \( \geq 723 \) K. In particular, a large elongation \( > 250\% \) was obtained at 823 K with low strain rates of \( 10^{-4} \) and \( 10^{-5} \) s\(^{-1} \).

Recently, it has been reported that dynamic recrystallization occurred during hot deformation and grain refinement was attained in a Mg–9Al–1Zn alloy. \(^{18} \) In this case, large elongation of 600% was attained due to grain boundary sliding resulting from grain refinement. It is well known that for structural superplasticity, the dominant deformation process is grain boundary sliding whose contribution to total strain is very large (\( = 50–80\% \)). \(^{12,13,19,20} \) In the present investigation, however, no grain refinement due to dynamic recrystallization was observed for the Mg–Y–RE alloy. Furthermore, the contribution of grain boundary sliding to total strain at 723–823 K was about 30%. Therefore, the dominant deformation process at 723–823 K for the Mg–Y–RE alloy is not grain boundary sliding.

The variation in stress as a function of strain rate is shown in Fig. 11. The strain rate sensitivity of stress increased with the testing temperature and the high strain rate sensitivity of about 0.3 was attained at 723–823 K. The fact that the strain rate sensitivity is about 0.3 indicates that the deformation at 723–823 K is the dislocation creep whose rate controlling process is the viscous dislocation glide. \(^{21–24} \) Recently, it
Coarse-grained Al-high Mg alloys. The large elongation has been reported that elongation above 300% is obtained in Mg–4Y–3RE alloy were investigated at room temperature due to the glide-controlled dislocation creep as well as for the Al–Mg alloys. Therefore, it is likely that the large elongation at 723–823 K for the Mg–Y–RE alloy is attributed to the relatively high strain rate sensitivity due to the glide-controlled dislocation creep. Therefore, it is likely that the large elongation at 723–823 K for the Mg–Y–RE alloy is attributed to the relatively high strain rate sensitivity due to the glide-controlled dislocation creep.

In the present paper, tensile properties of the peak-aged Mg–4Y–3RE alloy were investigated at room temperature to 823 K. However, the strength rapidly decreased at 573 K.

Observation of the surfaces of the deformed specimens revealed that grain boundary sliding significantly occurred at 573 K, however, there was no evidence for grain boundary sliding at room temperature. Therefore, occurring of active grain boundary sliding is likely to be responsible for the large decrease in strength at 573 K.

The elongation increased rapidly at 723–823 K. In particular, large elongation > 250% was attained at 823 K. The strain rate sensitivity at 723–823 K was about 0.3, indicating that the deformation process is the glide-controlled dislocation creep. Therefore, it is likely that the large elongation at 723–823 K for the Mg–Y–RE alloy is attributed to the relatively high strain rate sensitivity due to the glide-controlled dislocation creep.

### Table 2

The data of $\Omega$, $\delta D_{GB}$ and $T_n$ of Mg, Al and Fe.

<table>
<thead>
<tr>
<th></th>
<th>Mg</th>
<th>Al</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\Omega$ (m$^3$)</td>
<td>$2.33 \times 10^{-28}$</td>
<td>$1.66 \times 10^{-29}$</td>
<td>$1.18 \times 10^{-29}$</td>
</tr>
<tr>
<td>$\delta D_{GB}$ (m$^3$/s)</td>
<td>$5 \times 10^{-12} \exp \left( \frac{-92}{RT} \right)$</td>
<td>$5 \times 10^{-14} \exp \left( \frac{-84}{RT} \right)$</td>
<td>$1.1 \times 10^{-12} \exp \left( \frac{-174}{RT} \right)$</td>
</tr>
<tr>
<td>$T_n$ (K)</td>
<td>924</td>
<td>933</td>
<td>1810</td>
</tr>
</tbody>
</table>

$R$ is a Gas constant ($= 8.314 \times 10^{-3}$ (kJ/mol·K))

$T$ is the absolute temperature.

### References


