The influence of heat treatment on damping response of AZ91D magnesium alloy

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Abstract

The effect of parameters such as strain amplitude, frequency and temperature on damping capacity of magnesium alloy AZ91D as-cast, solution and aged was investigated. Granato–Lücke model was employed to explain the influences of parameters on damping capacity of magnesium alloy. It is shown that solution treatment decreases the amount of second-phase particles distributing inside grains, weakens the strong pinning on dislocations, and increases the internal friction of AZ91D, while aging reduces the value due to precipitation of second-phase particles. The influence of solution treatment and aging on damping capacity of AZ91D at room temperature is the same as at 100 °C. The damping at 100 °C is greater than that at room temperature because of thermal activation.

Keywords: Magnesium alloy; Damping; Internal friction; Granato–Lücke theory; Strain amplitude; Frequency

1. Introduction

In order to reduce the vibration and noise in industry fields, applications of damping alloys are considered. Among the structural metallic materials, pure magnesium has the lowest specific gravity (1.738 g/cm³) [1] and the highest damping capacity ($Q^{-1} = 0.11$ for a strain of $\varepsilon = 10^{-4}$). However, mechanical properties of pure magnesium are low, e.g. the tensile strength is only of 100 MPa [2] and the elastic modulus 45 GPa [3]. The strength can be improved by alloying [4–6], while the damping capacity decreasing. Therefore, the crux lies in resolving the contradiction between damping capacity and mechanical properties, and to exploit magnesium-based damping alloys having high damping and sufficient mechanical properties. Huang Liangyu et al. [7] developed a Mg-based damping alloy ZMJD-1S having high damping and high strength, which has a specific damping capacity of 48% ($Q^{-1} = 0.076$) and tensile strength of 165 MPa.

In the present work, the effect of heat treatment, i.e. solution treatment and aging, on the damping capacity and microstructure of AZ91D and their relations is investigated by a combination of dynamic mechanical analyzer (DMA), optical microscope (OM), and scanning electron microscope (SEM).

2. Experimental

AZ91D (basic composition (in wt.%): Al, 8.3 ~ 9.7; Mn, \( \leq 0.15 \); Zn, 0.35 ~ 1.0; Si, \( \leq 0.10 \); Fe, \( \leq 0.005 \); Cu, \( \leq 0.030 \); Ni, \( \leq 0.002 \); other impurities, \( \leq 0.02 \); Mg, the rest) was cast in a permanent mould by using a gas shield (SF6 + CO2). Castings were dimensioned and machined into 30 mm x 3 mm x 1 mm rectangle lamellar specimens, and subsequently heat-treated in accordance with conditions summarized in Table 1.
Table 1
Heat treatment conditions for damping specimen

<table>
<thead>
<tr>
<th>No.</th>
<th>Parameters of heat treatment</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>As-cast</td>
</tr>
<tr>
<td>2</td>
<td>Solutionized for 16 h at 413°C, quenched in water at 20°C</td>
</tr>
<tr>
<td>3</td>
<td>Solutionized for 16 h at 413°C, quenched in water at 20°C, aged 4 h at 168°C</td>
</tr>
<tr>
<td>4</td>
<td>Solutionized for 16 h at 413°C, quenched in water at 20°C, aged 8 h at 168°C</td>
</tr>
<tr>
<td>5</td>
<td>Solutionized for 16 h at 413°C, quenched in water at 20°C, aged 12 h at 168°C</td>
</tr>
</tbody>
</table>

Room temperature. OM and SEM were used to investigate the microstructures of test specimens. Specimens for OM were prepared by standard techniques and subsequently etched in 4% HNO₃ in ethanol.

3. Results and discussion

3.1. Microstructure

The microstructures of the specimens are shown in Fig. 1. It is seen that microstructure of as-cast alloy consists of the primary α-Mg phase, white γ-phase (Mg₃TiAl₂) and black lamellar eutectic phase (α plus γ). After solution treatment for 16 h, the γ-phase and eutectic phase disappear and the matrix transforms to supersaturated solid solution. The precipitates, which are distributed at and along grain boundaries, appear after aging (c ~ e) and increase with an increase in aging time.

A study of the literature [8–10] reveals that when aged at low temperatures, the supersaturated solid solution will decompose and the non-coherent equilibrium phase (Mg₃TiAl₂) will precipitate without either pre-precipitation or metastable phases. The two precipitation patterns are: (a) continuous and (b) non-continuous precipitation. In most cases, non-continuous precipitation starts at the grain boundaries and dislocations; whereby, the Mg₃TiAl₂ phase grows as patches into grains following a definite orientation. Continuous precipitation was evident within the grains; whereby, the Mg₃TiAl₂ phase grows as fine patches along the basal plane (0001). Fig. 1 shows the non-continuous precipitates at grain boundaries increase with an increase in aging time. Fig. 2 shows the SEM micrograph of AZ91D aged for 12 h. As can be seen from Fig. 2, a large amount of lamellar non-continuously precipitated Mg₃TiAl₂ at grain boundaries.

3.2. Damping capacity

3.2.1. The influence of heat treatment on damping capacity of AZ91D at room temperature

The strain dependence of damping capacity was investigated by DMA at room temperature at a vibration frequency...
Fig. 2. SEM micrograph of T6, 12 h: (A) discontinuously precipitated particles and (B) continuously precipitated particles.

Fig. 3. tan $\phi$–$\varepsilon$ relation of AZ91D as-cast, T4 and T6 at room temperature. The results are shown in Fig. 3, where x-axis corresponds to strain amplitude ($\varepsilon$) and y-axis the Loss Factor (tan $\phi$). Each curve in Fig. 3 is nearly horizontal at a strain of $5 \times 10^{-4}$–$1 \times 10^{-2}$%, which indicates that the Loss Factor does not change with strain. The Loss Factor increases with an increase in strain for strain above $1 \times 10^{-2}$%. Therefore, there is a critical strain amplitude $\varepsilon_{cr}$ above which, the Loss Factor (tan $\phi$) increases remarkably with strain increasing. At the same strain, the higher value of tan $\phi$ implies higher damping capacity.

Above results can be interpreted by using Granato–Lücke model [11] (Fig. 4a). For large enough concentration of impurity atoms, the length of loop determined by the intersection of dislocation network loops is further pinned down by the impurity particles through the Cottrell mechanism. There are, therefore, two characteristic lengths in the model: the network length $L_N$, and the length $L_C$, determined by the impurities. If an external stress is applied, there will be, in addition to the elastic strain, an additional strain due to the dislocations strain. For zero applied stress, the length $L_N$ is pinned down by the impurity particles (A). For a very small stress (B), the loops ($L_C$) bow out and continue to bow out until the breakaway stress is reached. At the breakaway stress, a large increase in the dislocation strain occurs without stress increasing (C–D). Now, for further increases in the stress, the loop length ($L_N$) bows out (D–E), until the stress required to activate the Frank–Read source [12], ($L_{FR}$) is reached. Further increases in the applied stress lead to creation and expansion of new closed dislocation loops (F–G). The stress-dislocation strain law corresponding to the model is shown in Fig. 4(b). The curve has a hysteresis loop due to the different dislocation movement, which leads to the internal friction.

The loss caused by forced vibration of dislocation segment ($L_C$) is called as resonance internal friction and is strain–amplitude independent and frequency dependent. It can be calculated using the following equation [11]:

$$\delta_I = \frac{2\Omega(1 - \nu)}{\pi^3} \times \frac{A L_C^2 B_0}{G b^2}$$  

where $\Omega$ is the orientation factor, $\nu$ the Poisson’s ratio, $A$ the total length of movable dislocation line, $L_C$ an effective loop length which should be given by 3.3 times $L_C$, $B$ the damping force per unit length, $\omega$ the angular frequency, $G$ the shear modulus and $b$ is the Burgers vector.

The loss caused by dislocation segment pinned by strong pinning points is named as static hysteresis internal friction and is strain–amplitude dependent and frequency independent. Then strain–amplitude dependence of the logarithmic decrement can be calculated using the relationship [11,13]:

$$\delta_H = C_1 \exp \left( - \frac{C_2}{\varepsilon} \right)$$  

Fig. 4. Granato-Lücke model (see Ref. [11]): (a) the bowing out, breakaway and multiplication of dislocation by an increasing applied stress and (b) the stress-dislocation strain law.
where

\[ C_1 = \frac{E_0 L_1}{6bE_L C} \quad C_2 = \frac{F_b}{bEL_C} \quad (3) \]

\( F_b \) is the binding force between a dislocation and a solute atom (weak pinning point), \( E \) is the unrelaxed modulus, \( \varepsilon \) is the strain amplitude, \( L_S \) is the mean length of dislocation segments between nodes of the dislocation network and \( \rho \) is the density of dislocations participated in the breakaway process. \( C_1 \) and \( C_2 \) are constants that can be experimentally estimated.

Because of the two different internal friction mechanisms, the damping curves of the specimens do not vary with strain initially. However, they rise rapidly with an increase in strain above a critical strain.

When the strain is above \( \varepsilon_{cr} \), the curve of solutionized specimen shows the highest internal friction value, while the as-cast specimen exhibits the lowest value. As to the other three aged specimens, their internal friction values are between the as-cast and solutionized specimens. Furthermore, the value of internal friction decreases with an increase in aging time.

Eq. (2) can be alternated \([13]\) as

\[ \ln(\delta_{H\varepsilon}) = \ln(C_1 - C_2) \quad (4) \]

and Granato–Lücke (G–L) plots (i.e. \( \ln(\delta_{H\varepsilon}) \) versus \( \varepsilon^{-1} \)) should be straight lines, whose intercept and slope are the values of \( \ln(C_1) \) and \( C_2 \) respectively.

The expression for calculating the specific damping capacity (S.D.C. or \( \Phi \)) \([14]\):

\[ 2\pi \tan \phi = \Psi = \begin{cases} \frac{2\delta}{1 - \exp(2\delta)}, & \Psi < 40\% \\ \frac{1}{\ln(1 - 2\pi \tan \phi)}, & \Psi \geq 40\% \end{cases} \quad (5) \]

can be rewritten as the \( \delta - \tan \phi \) relation:

\[ \delta = \begin{cases} \pi \tan \phi, & \tan \phi < 0.06 \\ \frac{\pi}{2\pi} \ln(1 - 2\pi \tan \phi), & \tan \phi \geq 0.06 \end{cases} \quad (6) \]

then, the \( \tan \phi - \varepsilon \) relation shown in Fig. 3 can be transformed to \( \ln(\delta_{H\varepsilon}) - \varepsilon^{-1} \) relation, leading to the G–L plots (Fig. 5). Values of \( C_1 \) and \( C_2 \) obtained according to the G–L plots (Fig. 5) are given in Table 2.

It can be observed that all the specimens almost have equal \( C_2 \) values, and that the \( C_1 \) value of the solutionized specimen is highest while that of the as-cast is lowest. Also, the three aged specimens show moderate \( C_1 \) values, which decrease with increasing aging time. It can be explained using Eq. (3) as follows: since the compositions of specimens are same, the parameters in equations except \( L_C \) and \( L_S \) can be treated as constants. As heat treatment has little effect on \( L_C \), the \( C_2 \) values are approximately equal. For the volume of grain is constant during heat treatment, \( L_S \) of as-cast specimen is minimal due to the great quantity of \( \gamma \)-phase (Mg17Al12), which is larger than precipitated particles, resulting in the lowest \( C_1 \). Most \( \gamma \)-phase and eutectic phase in grains are dissolved and the matrix transforms to supersaturated solid solution after solution treatment, so \( L_N \) reaches the maximum which is approximately equal to the diameter of grain, accordingly, the \( C_1 \) value is highest. The \( C_1 \) values of aged specimens are higher than the value of as-cast one due to the prior solution treatment. And, the more aging time, the more and larger precipitated particles appear forming strong pinning points, causing \( L_N \) to shorten, and the \( C_1 \) values decrease with increasing aging time consequently.

The temperature dependence of damping capacity for as-cast specimen is shown in Fig. 6. It can be seen that the curve rises with increasing temperature. The thermal activation has great influence on the formation of point defects, with the temperature increasing, equilibrium concentration of point defects increases, leading to the enhancement of concentration of line defects and dislocations, which results in

![Fig. 5. G–L plots of AZ91D at room temperature.](image)

![Fig. 6. tan–temperature graph of AZ91D as-cast.](image)
the greater effect of pinning points, so the internal friction value increases with increasing temperature. At the temperature of 350–400 °C, the slope of the curve descends, thus a peak can be predicted to appear at higher temperature, which is in accord with literature [2].

The frequency dependence of damping capacity for as-cast specimen is shown in Fig. 7. The curve falls at the elevated frequencies but to a little extent, i.e. at experimental strain (above the critical strain), damping capacity of the alloy is almost unrelated to frequency, which verifies the dislocation theory that the high damping capacity of Mg alloys is due essentially to the static hysteresis internal friction.

3.2.2. The influence of heat treatment on damping capacity of AZ91D at high temperature

The strain dependence of damping capacity at 100 °C, vibration frequency of 1 Hz is shown in Fig. 8. The values of $C_1$ and $C_2$ obtained according to the G–L plots (Fig. 9) are given in Table 3.

The values indicate the same conclusions as those of the previous study, which is explained using the same mechanisms. However, the values of $C_1$ and $C_2$ of specimen aged for 8 h are higher than those of specimen aged for 12 h. This is because of its less concentration of impurities causing $L_C$ to be enlarged.

In comparison with the data in Table 2, data in Table 3 indicate that the $C_1$ value at 100 °C is higher than that at room temperature for the same specimen, which is due to the influence of thermal activation.

4. Conclusions

1. The solid solution treatment makes Mg$_{17}$Al$_{12}$ phase formed during solidification dissolved and the alloy transform to supersaturated solid solution. The strong pinning points of dislocations are decreased considerably while the static hysteresis internal friction of alloy increases.

2. The aging process causes the precipitation of Mg$_{17}$Al$_{12}$ that leads to formation of the strong pinning points to dislocations. The pinning effect increases with an increase in aging time, and so the damping capacity descends with aging.

3. The damping capacity at 100 °C is greater than that at room temperature, primarily because the thermal activation causes an increase in point defects and line defects. This promotes greater interaction between dislocations and pinning points, and the reason for the internal friction value of AZ91D as-cast to increase at elevated temperatures and at strain levels above the critical strain.

4. Solution and aging treatment show the same influence on damping capacity of AZ91D even if the testing temperature is increased from room temperature to 100 °C.
References