Damping Properties of Homogenized and Cold-Rolled Mg–14.3Li–0.8Zn
β-Phase Magnesium Alloy

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As-homogenized β-phase Mg–14.3Li–0.8Zn (LZ141) alloy exhibits a P1 peak and a conspicuous high-temperature damping background (HTDB) in the heating internal friction Q–¹ curves. Cold-rolled LZ141 alloy with 80% thickness reduction exhibits a significant P2 peak in the Q–¹ curves due to the recrystallization process during heating. The activation energy (H) of the HTDB for as-homogenized LZ141 alloy is 1.44 eV, decreasing to 0.91 eV after 80% cold-rolling. These H values for β-phase LZ141 alloy are lower than those for α-phase AZ80 alloy. Cold-rolled LZ141 alloy exhibits extremely high damping capacity at temperatures around the P2 peak.

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1. Introduction

Magnesium, with a density of 1.74 g/cm³, is one of the lightest structural metals. This characteristic is the most important factor in the selection of magnesium for engineering designs, especially for automotive and aerospace components. Magnesium alloys also possess the advantages of acceptable mechanical and electrical properties, high heat dissipation and good electromagnetic shielding, and are widely used for the manufacture of cell phones, laptop computers, cameras, and other hand-held electronic components.¹,² However, magnesium has a hexagonal close-packed (HCP) structure and little deformation and strain hardening can be accomplished at room temperature. Several studies have reported that the low processability and ductility of magnesium can be improved by adding lithium to form Mg–Li alloys.³–⁶ These Mg–Li alloys are exceptionally lightweight since lithium has a density of only 0.53 g/cm³. The Mg–Li binary phase diagram indicates that HCP-structured α-phase of Mg solid solution changes to BCC-structured β-phase of Li solid solution when the Li content is above 11.5 mass%.⁷ β-phase Mg–Li alloys have the advantages of ultra-light weight with a specific gravity ≤ 1.5 g/cm³ and good formability at room temperature, but their relatively low strength and thermal resistance restrict their practical engineering applications.⁸,⁹ In order to overcome these shortcomings, Zn is usually added to β-phase Mg–Li alloys.⁹ The α-phase Mg and Mg alloys can exhibit good damping capacity (Q–¹) due to the ease with which they absorb energy elastically.¹⁰–¹² The damping properties of (α + β) dual-phase LZ60, LZ100 and LZ110 alloys have also been studied in detail using dynamic mechanical analysis (DMA) tests.¹³ Nevertheless, the damping characteristics of β-phase Mg–Li alloys are less well understood. The aim of this study is, therefore, to investigate the low-frequency damping properties of β-phase Mg–14.3Li–0.8Zn (mass%), i.e., ASTM

LZ141 alloy, using DMA tests. The preliminary experiments suggest that the cold-rolling process can significantly improve the mechanical properties of LZ 141 alloy. However, the way in which the cold-rolling process affects this alloy’s damping capacity is still unknown. Thus, this study also investigates the effect of cold-rolling on the damping properties of LZ141 alloy.

2. Experimental Procedures

LZ141 alloy used in this study was prepared from the pure raw materials: high purity magnesium (purity 99.95%), lithium (purity 99.9%) and zinc (purity 99.99%). These raw materials were induction-melted and protected by argon gas, and then cast in a steel mold followed by homogenization at 350°C for 12 h. The homogenized ingot was sliced into 6–10 mm thickness plates. The precise chemical composition of the as-homogenized LZ141 alloy was measured, with an inductivity coupled plasma-optical emission spectrometer (ICP-OES), as Mg–14.3Li–0.8Zn (mass%) with trace amounts of Al, Mn and Fe elements below 0.02%.

Some of the homogenized plates were cold-rolled at room temperature to achieve an 80% reduction in thickness, using a BDR 2HI-MILL rolling machine manufactured by Daito Seiki Co., Japan. Specimens for DMA tests, with the dimension 35.0 × 5.0 × 1.2 mm³, were cut from the as-homogenized and cold-rolled plates. The internal friction, Q–¹, of the as-homogenized and cold-rolled LZ141 specimens was determined using a TA 2980 DMA equipment configured with a single cantilever and liquid nitrogen cooling apparatus. Each DMA specimen was heated from 0 to 300°C at a constant heating rate of 3°C/min, and the testing frequency and strain amplitude were set at 0.1–10 Hz and 1 × 10⁻³–1 × 10⁻⁵, respectively. The specimens of the LZ141 alloy were prepared for microstructural examination using a standard metallographic procedure with an etching solution of 7 ml alcohol + 1 ml picric acid + 1 ml acetic acid + 1 ml deionized water. Microstructural observations were performed with a Nikon FX-35DX optical microscope (OM).
3. Results and Discussion

3.1 The dependence of damping capacity on strain amplitude

Figure 1 plots the dependence of damping capacities \( Q^{-1} \) on strain amplitude \( (\varepsilon) \) for as-homogenized and cold-rolled LZ141 alloys, respectively, measured at a constant frequency (1 Hz) and constant temperature (25°C). From Fig. 1, both the \( Q^{-1} \) values of as-homogenized and cold-rolled LZ141 alloys are almost constant when the applied \( \varepsilon \) is below 3.5 \( \times 10^{-3} \%), but gradually increase as \( \varepsilon \) increases further. Accordingly, the damping capacities can be divided into two components:\(^{14,15}\)

\[
Q^{-1} (\varepsilon) = Q_0^{-1} + Q_{\text{H}}^{-1} (\varepsilon)
\]

where \( Q_0^{-1} \) represents the damping capacity that is independent or only weakly dependent on the strain amplitude, and \( Q_{\text{H}}^{-1} (\varepsilon) \) represents the damping capacity that increases with the strain amplitude. It is difficult to determine if \( Q_0^{-1} \) or \( Q_{\text{H}}^{-1} (\varepsilon) \) is the more important indicative parameter of the damping capacity of \( Q^{-1} (\varepsilon) \), because \( Q_0^{-1} \) is dominant when the applied strain amplitude is small, while \( Q_{\text{H}}^{-1} (\varepsilon) \) becomes dominant when the applied strain amplitude is large. Figure 1 shows that the \( Q_0^{-1} \) value of as-homogenized LZ141 alloy is approximately 1.1 \( \times 10^{-2} \), but this value increases to about 1.5 \( \times 10^{-2} \) after cold-rolling. These values are much higher than that for LZ100 alloy \( (Q_0^{-1} = 6 \times 10^{-3}) \) measured at the same temperature.\(^{13}\) This indicates that \( \beta \)-phase LZ141 alloy exhibits better damping capacity than \( (\alpha + \beta) \) dual-phase LZ100 and LZ110 alloys at room temperature. Figure 1 also shows that the damping capacity of \( Q_{\text{H}}^{-1} (\varepsilon) \) for LZ141 alloy increases after cold-rolling when it is measured at the same \( \varepsilon \) value. The cold-rolled LZ141 alloy exhibits higher \( Q_0^{-1} \) and \( Q_{\text{H}}^{-1} (\varepsilon) \) values, because abundant defects and dislocations are introduced during the cold-rolling process. This can be explained by the Granato–Lücke theory,\(^{16,17}\) which states that the pinned dislocation loop oscillating under an applied stress leads to the energy loss during damping.

3.2 Damping characteristics of as-homogenized and cold-rolled LZ141 alloys

Figure 2(a) shows the heating internal friction curve, \( Q^{-1} (T) \) versus temperature, for as-homogenized LZ141 alloy measured at a constant strain amplitude (1.7 \( \times 10^{-4} \)) and heating rate (3°C/min) but different frequencies from 0.1 to 10 Hz. From Fig. 2(a), the damping capacity of each heating \( Q^{-1} (T) \) curve only increases gently at temperatures below 200°C, and then rises significantly faster until 300°C. The heating \( Q^{-1} (T) \) curve for temperatures below 200°C is termed the athermal damping background (ADB) and that for temperatures above 200°C, which exhibits an exponential damping background, is termed as high-temperature damping background (HTDB).\(^{13,18-20}\)

Typically, HTDB possesses a much higher damping capacity than that of ADB but this is only apparent at an elevated temperature, above one-half of the absolute melting temperature of the specimen. Therefore, in view of the engineering applications of Mg alloys, \( Q_{\text{HT}}^{-1} \) is the more important indicative parameter of the damping capacity at temperatures around room temperature, while HTDB becomes dominant at elevated temperatures.

Figure 2(a) also shows that an internal friction peak appears at around 30 to 60°C, as indicated by an arrow. This internal friction peak is also observed in pure Mg of 99.96 mass% purity and Mg–Ni alloy of 6.2–22.6 mass% Ni.\(^{15}\) It is termed the P\(_1\) peak and is caused by the movement of dislocations on the basal planes of \( \alpha \)-phase Mg and Mg alloys. Figure 2(b) shows the heating \( Q^{-1} (T) \) curves versus temperature for cold-rolled LZ141 alloy measured by experimental parameters identical to those of Fig. 2(a). Compared with Fig. 2(a), the heating \( Q^{-1} (T) \) curves for
cold-rolled LZ141 alloy exhibit a much more significant internal friction peak at about 140 to 170°C, while the P1 peak becomes inconspicuous. This significant internal friction peak is termed the P2 peak and is caused by the sliding of grain boundaries.15) For α-phase Mg alloys, a P2 peak with such an extremely high damping capacity (Q⁻¹ value is above 0.3 when the specimen is measured at 0.1 Hz) has never been observed.

Figure 3(a) shows the OM micrograph for as-homogenized LZ141 alloy. As can be seen, the grains of as-homogenized LZ141 alloy are isotropic. Their average grain size is calculated as 264 μm, following ASTM E112-88. Figure 3(b) shows the OM micrograph of cold-rolled LZ141 alloy after heating to 150°C in the DMA furnace, and then in-situ cooling in a load-cell to room temperature using liquid nitrogen cooling apparatus. As shown in Fig. 3(b), the original coarse grains no longer exist. Instead, abundant fine grains with a much smaller grain size (about 18 μm in average) appear in the specimen. This implies that recrystallization takes place in the cold-rolled LZ141 alloy during the heating process. Accordingly, the significant P2 peak can be ascribed to the grain refinement that occurs during the recrystallization process, which leads to an increase in the number of movable grain boundaries within the specimen.

Figures 2(a) and 2(b) also reveal that the peak temperatures of the P1 and P2 peaks both shift to higher temperatures as frequency increases. This indicates that the P1 and P2 peaks exhibit a thermal activated relaxation characteristic, where their relaxation times and activation energy obey the Arrhenius equation:21)

$$\tau = \tau_0 \exp\left(\frac{H}{kT}\right)$$  \hspace{1cm} (2)

where $\tau$ and $\tau_0$ are the relaxation time and relaxation constant, respectively, $H$ is the activation energy, $k$ is the Boltzmann constant and $T$ is the absolute temperature. At the peak temperature, relaxation time $\tau$ and angular frequency $\omega$ satisfy the relationship: $\tau_0 \omega_p = 1$. Here, the angular frequency is defined as $\omega = 2\pi f$, where $f$ is the applied frequency. Therefore, eq. (2) can be rewritten as

$$\omega_p^{-1} = \tau_0 \exp\left(\frac{H}{kT_p}\right)$$  \hspace{1cm} (3)

where $\omega_p$ and $T_p$ are the angular frequency and the absolute temperature at peak temperature, respectively.

Figures 4(a) and 4(b) plot the Arrhenius relation of $\ln(\omega)$ versus $1000/T_p$ for (a) the P1 peak of as-homogenized LZ141 alloy and the P2 peak of cold-rolled LZ141 alloy, respectively. From the slopes of the fitting lines in Figs. 4(a) and 4(b), the activation energy ($H$) of the P1 and P2 peaks can be calculated as $H_{P1} = 1.02$ eV and $H_{P2} = 2.07$ eV, respectively. Table 1 lists the activation energy ($H$) of the P1 and P2 peaks for as-homogenized and cold-rolled LZ141 specimens, as well as those of α and (α + β) phase(s) Mg alloys for the purposes of comparison. As shown in Table 1, the $H_{P1}$ value of as-homogenized LZ141 alloy ($H_{P1} = 1.02$ eV) is higher than that of as-homogenized LZ100 alloy ($H_{P1} = 0.87$ eV).13) This can be explained by the fact that (α + β) dual-phase LZ100 alloy possesses abundant α/β phase interfaces, which facilitate diffusion-assisted dislocation climbing.13)
are constants. For analysis of the HTDB, it is convenient to
say a

$$\frac{Q^{-1}(T)}{C_0} = \frac{K}{\omega \exp(H/kT)}$$

(4)

where $Q^{-1}_{at}$ is the damping capacity of the ADB, and $n$ and $K$
are constants. For analysis of the HTDB, it is convenient to
use the logarithmic representation of eq. (4):

$$\ln(Q^{-1}(T) - Q^{-1}_{at}) = \ln K - n \ln \omega - n \frac{H}{kT}.$$ (5)

Figures 5(a) and 5(b) plot $\ln(Q^{-1}(T) - Q^{-1}_{at})|_{\ln\omega=0}$ versus $1/T$ for as-homogenized and cold-rolled LZ141 specimens measured at different temperatures above 250°C. The activation energy of the HTDBs ($H_{HTDB}$) for as-homogenized and cold-rolled specimens can be determined from the slope of the fitting lines as $H = 1.44$ and 0.91 eV, respectively.

Table 1 lists the calculated $H_{HTDB}$ values for as-homogenized and cold-rolled LZ141 specimens and other Mg alloys. It shows that the $H_{HTDB}$ value for as-homogenized LZ141 alloy ($H_{HTDB} = 1.44$ eV) is similar to those for cold-rolled AZ80 alloy ($H_{HTDB} = 1.37$ eV)\(^{20}\) and as-cast Mg-1Ca (mass%) alloy ($H_{HTDB} = 1.35$ eV),\(^{24}\) but is higher than that for as-annealed LZ100 alloy ($H_{HTDB} = 0.59$ eV).\(^{13}\) Weller et al.\(^{15}\) reported that the HTDB in TiAl and NiAl intermetallics can be assigned to the diffusion-assisted climb of dislocations, since the activation energy of TiAl intermetallics measured by high temperature damping experiments and by creep experiments are very similar.

In Figs. (a) and (b), both as-homogenized and cold-rolled LZ141 alloys show HTDBs at temperature above 250°C (523 K). This temperature exceeds the high-temperature creep region of LZ141 alloy, i.e., above one-half of its absolute melting temperature (861 K). Therefore, it is believed that the HTDB of LZ141 alloy is also caused by the mechanism similar to that for TiAl alloy. Accordingly, the smaller $H_{HTDB}$ value for LZ100 alloy can be explained by the
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