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Aging damping associated with discontinuous transformation in Al–Zn and Al–Zn–Mg–Cu alloys

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Abstract. In this work, the aging damping associated with the discontinuous phase transformation was studied in two Al–Zn alloys using a forced-vibration pendulum. Correlating these results with an x-ray diffraction analysis and TEM observations, we suggested that the aging damping was directly proportional to the total interface area which was related to the nucleation and growth during the discontinuous transformation.

1. Introduction

Previous studies [1–4] have shown that there is an aging damping peak (denoted P_2) during the discontinuous transformation in an as-quenched Al–Zn eutectoid alloy and it was suggested that this peak was probably attributable to the variation in the area of the interface, i.e. the maximum area of the interface during the transformation. This was suggested because there was no minimum of the modulus corresponding to the peak P_2 and the tendency of change in the aging damping and modulus is independent of frequency, i.e. the local soft modulus associated with the mobility of the phase interface did not occur during the discontinuous transformation. However, as another aging damping peak (called P_1) associated with the appearance and disappearance of coherency during the spinodal pre-transition, which just preceded the discontinuous transformation, overlapped P_2, we were unable to obtain a complete plot of P_2, and therefore the detailed relationship of the variation in aging damping such as P_2 with the variation in interface area during the discontinuous transformation was not fully understood.

In this work, we have let the transformation of two as-quenched Al–Zn alloys take place at an appropriate temperature so that only the discontinuous phase transformation occurs during the whole aging process, and we have investigated in detail the aging damping associated with the discontinuous transformation. Based on the x-ray diffraction analysis and TEM observation, a clear schematic diagram has also been suggested to illustrate the relation between the aging damping variation and the total interface area variation during the transformation.

2. Experimental details

In order to study the effect of the alloying element, we have chosen two kinds of extruded Al–Zn alloy, namely Al–78 wt% Zn and Al–78 wt% Zn–0.03 wt% Mg–0.2 wt% Cu (for

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short, called specimens 1 and 2, respectively). They are annealed at 648 K for 1.5 h and then quenched into water at room temperature (303 K). The aging damping and x-ray diffraction measurements are conducted in the air at room temperature (303 K) at various times after the quenching.

The specimens for the aging damping experiment with dimensions 60 mm × 4 mm × 1 mm are cut from pieces of alloy 1 and alloy 2, respectively. The aging damping is measured using a forced-vibration pendulum of our own design. The damping Q^{-1} is calculated from $Q^{-1} = \tan \varphi$, where φ is the phase angle by which strain lags stress during the forced vibration. The measuring strain amplitude is less than 5×10^{-6}. A detailed description of the pendulum can be found in [5].

The specimens for the x-ray diffraction experiment are flat sheets 0.5 mm in thickness which are cut from pieces of alloy 1 and alloy 2, respectively. The x-ray diffraction is performed using a computer-controlled x-ray diffractometer with a Si filter. The Cu radiation is applied within the angle 2θ range from 30° to 100°. There are two kinds of scanning. One is continuous scanning which is carried out at a higher scanning speed within $2\theta = 30$–$100°$ at various times after quenching. This is done in order to know the intensity variation of various diffraction peaks so that a pronounced intensity variation of a peak can be detected. In doing so, we find that the intensity of the peak of $\beta(10\bar{1}1)$ ($2\theta = 43.3°$) shows the greatest variation. The other kind of scanning is localized and is only performed around the peak of $\beta(10\bar{1}1)$ so that an accurate measurement of the variation in the amount of β-phase with aging time can be obtained. The speed for continuous scanning is $0.1°\,(2\theta)\ s^{-1}$; the speed for the localized scanning is $0.02°\,(2\theta)\ s^{-1}$.

The foils for TEM observation were prepared from the specimens on which the damping and x-ray-diffraction had been measured and which had been completely decomposed after the quenching as described above. The foils were made by electropolishing in a solution containing 15% perchloric acid and 85% methanol (by volume) at about 240 K and 20–30 V. Those foils are examined using a JEM-200CX transmission electron microscope.

3. Results

3.1. Results for alloy 1

Figure 1 shows the results of the aging damping and x-ray diffraction in the as-quenched alloy 1. There is an obvious damping peak in figure 1(*a*) at about 10 min after quenching, which corresponds to the transition point (which is also at 10 min after quenching) of the increasing rate of β-phase x-ray diffraction intensity as shown in figure 1(*b*) (in fact, after this point, there is still a relatively slow increase in β-phase x-ray diffraction intensity [1]), and the damping at the beginning of transformation is extremely small. Figure 1(*a*) also reveals that the aging damping in the whole process decreases with increasing frequency. Figure 2 is the TEM micrograph of the microstructure of the completely decomposed alloy 1 which consists of equiaxial grains of α-phase (white) and β-phase (black) with quite a homogeneous grain size.

3.2. Results for alloy 2

Figures 3(*a*) and 3(*b*) illustrate the results of the aging damping and of x-ray diffraction, respectively, for the as-quenched alloy 2. Compared with alloy 1, the aging damping in

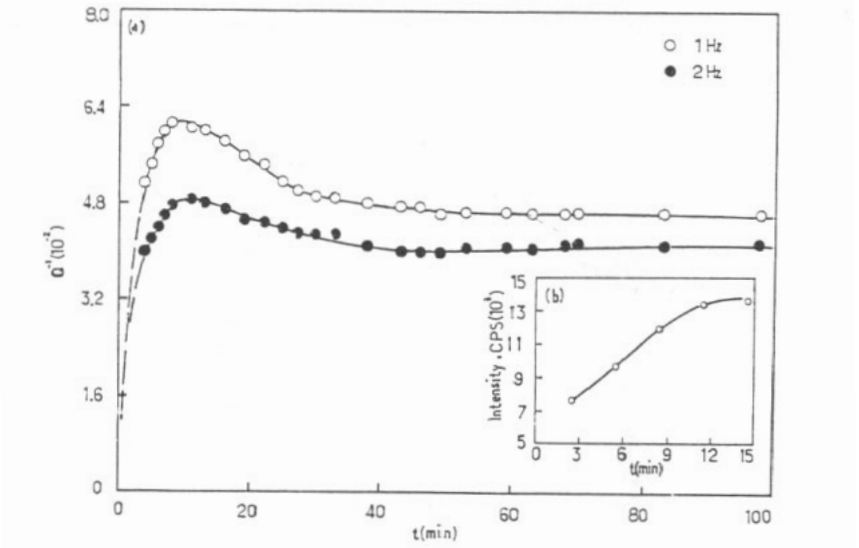

Figure 1. (*a*) The aging damping curve of Q^{-1} against t and (*b*) the variation in the x-ray diffraction intensity of $\beta(10\bar{1}1)$ during the room-temperature decomposition of as-quenched alloy 1.

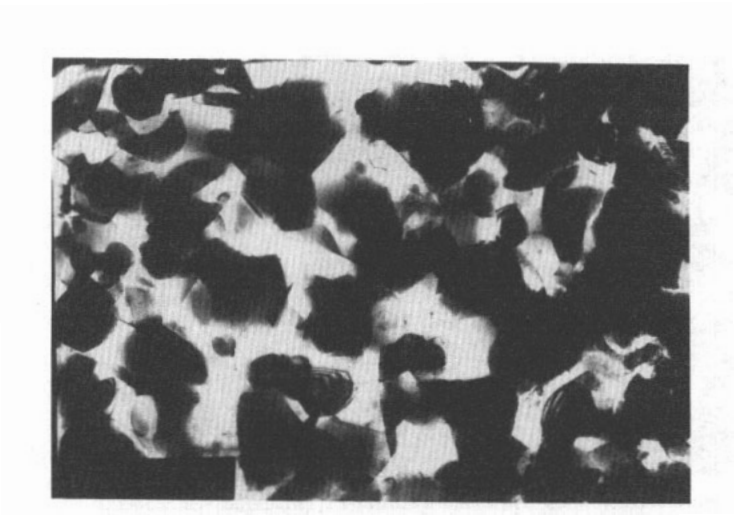

Figure 2. Decomposed product of alloy 1. (Magnification, 20000.)

alloy 2 increases only monotonically and there is no longer an obvious aging damping peak. In the process, the initial damping increases relatively quickly whereas, after about 100 min, the damping increases very slowly and finally reaches a stable value. It is also obvious that the whole aging damping of alloy 2 is lower than that of alloy 1 but is dependent on frequency which is similar to, but weaker than, that in alloy 1, i.e. in the whole process the aging damping decreases with increasing frequency. The x-ray diffraction analysis (see figure 3(*b*)) shows that the transformation of alloy 2 is intensively

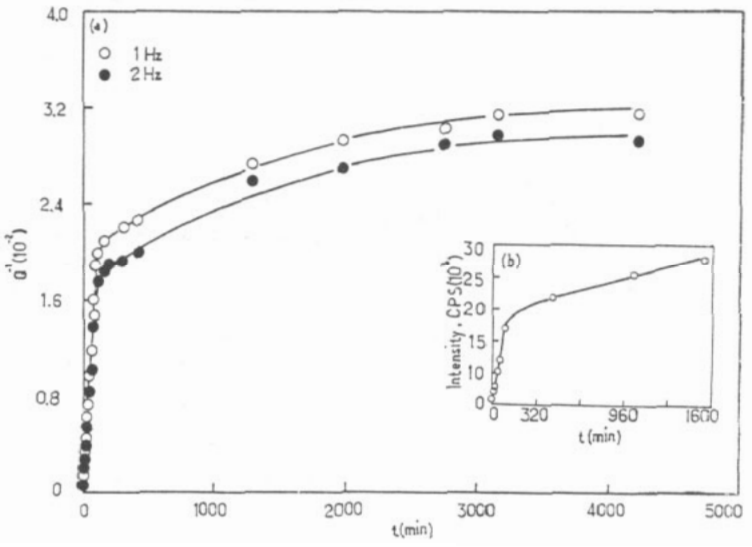

Figure 3. (*a*) The aging damping curve of Q^{-1} against t and (*b*) the variation in the x-ray diffraction intensity of $\beta(10\bar{1}1)$ during the room-temperature decomposition of as-quenched alloy 2.

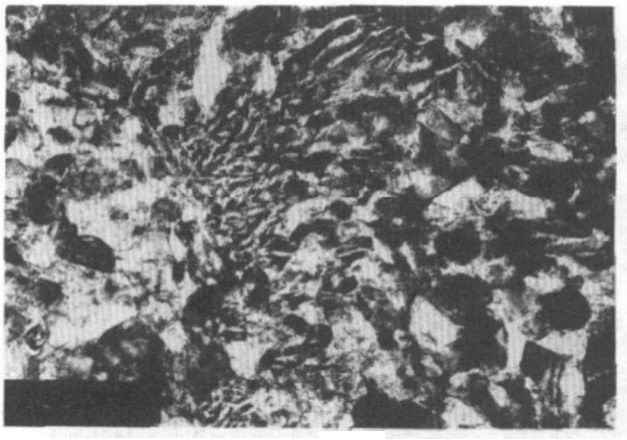

Figure 4. Decomposed product of the alloy 2. (Magnification, 20000.)

delayed, the increasing rate of β-phase is relatively quicker at the beginning but, after aging for about 100 min, it is obviously reduced. The completely decomposed product (figure 4) consists of a heterogeneous microstructure (with both equiaxial and lamellar grains) of α-phase and β-phase with a smaller average grain size than that of alloy 1.

4. Discussion

It has been proved [5] that the kind of interface, α–β, α–α' or α'–β, in the alloys has no obvious effect on Q^{-1} as long as they are all incoherent because the diffusivity of Zn in

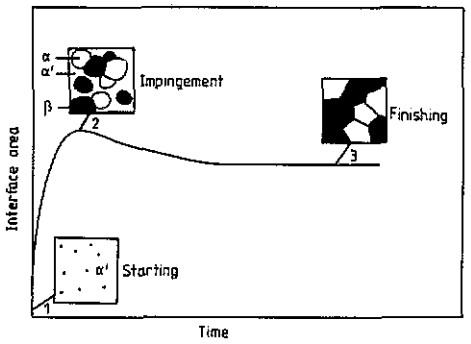

Figure 5. Schematic diagram for the decomposition process and accompanying variation in interface area for alloy 1.

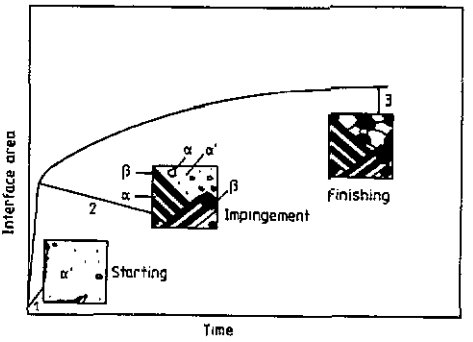

Figure 6. Schematic diagram for the decomposition process and accompanying variation in interface area for alloy 2.

Al and the diffusivity of Al in Zn are almost same, according to the viscous interface motion model [5], the Q^{-1} values for these interfaces are also identical. The fact that there is no damping peak as a precursor corresponding to the appearance and disappearance of the coherency during the spinodal transition [2, 3] indicates that the spinodal transition does not occur under the quenching condition as described in this paper. (The quenching surely avoids the metastable miscibility gap and a spinodal transition does not occur. This agrees with the results in [1].) In terms of the analysis in [2], the aging dampings both in alloy 1 and in alloy 2 are only caused by the variation in the interface area value during the discontinuous decomposition $\alpha \rightarrow \alpha + \beta$. Because the measurement strain amplitude is less than 5×10^{-6}, according to results in [1], the aging damping is amplitude independent or linear. The dependency of damping on frequency proves that the damping originates from a non-hysteresis motion of the phase interface.

The eutectoid decomposition process and accompanying variation in interface area in alloy 1 can be schematically described as shown in figure 5. On the basis that we have a decomposed product of the equiaxial grains of two phases with quite a homogeneous grain size as shown in figure 2, we can assume that all nuclei form homogeneously right at the beginning of the transformation (see stage 1 in figure 5), and thus the intensity of the β-phase diffraction (see figure 1(b)), the interface area and, therefore, aging damping (see figure 1(a)), increase rapidly at the beginning. As the parent phase α is consumed by the growing spherical new phase, the adjacent transformation phases will impinge on each other when most of the spherical new phases grow to a certain radius as shown at stage 2 in figure 5; thus the interface area will begin to decrease and a maximum value of the interface and, therefore, of the aging damping appear (see figure 1(a)). When the transformation is finished, the interface area and, therefore, the aging damping remain unchanged.

In alloy 2, because of the addition of alloying elements Mg and Cu, which can effectively destroy the quenched-in vacancies, retard the atom diffusion and therefore reduce greatly the transformation rate [6], the increasing rate of intensity of β-phase diffraction is evidently reduced (see figure 3(b)), compared with the result in figure 1(b). As a result of the difficulty in diffusion, heterogeneous nucleation will occur in some favoured sites, nucleation will take place throughout the transformation, and some nuclei will nucleate in a lamellar form in order to facilitate the atom diffusion and to

form a product with a wide range of grain sizes and shapes (equiaxial and lamellar) as shown in figure 4. Owing to this kind of nucleation and growth, there will be no specific grain size in most of the growing phases as shown at stage 2 in figure 5 corresponding to an obvious maximum of phase interface area or aging damping. So the aging damping increases relatively more quickly before stage 2 (figure 6) because, before stage 2, both the relatively faster lamellar nucleation and growth and the spherical nucleation and growth coexist (see stage 1). After stage 2, however, the lamellar growth is terminated by the impingement of adjacent transformed volume, and only the very slow spherical nucleation and growth remain; so the transformation rate (see figure 3(*b*)) is reduced greatly, and the interface area or the aging damping increases very slowly to a constant value at the end of transformation as shown in figure 6. The reason why Q^{-1} in alloy 2 is lower than that in alloy 1 is that the alloying elements impede the viscous motion of the interface [4] but, in the same alloy, Q^{-1} is always proportional to the interface area.

Acknowledgments

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