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Q1 Enhanced damping capacity of magnesium alloys by tensile twin boundaries

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We propose a novel method to fabricate Mg alloys with enhanced damping capacity without sacrificing mechanical properties by introducing {10−12} twin boundaries. We find that the twin boundaries and their reciprocating movements play a significant role in remarkably enhancing the damping capacity of Mg alloys at high vibration strain amplitude. Such enhancement is especially pronounced in a slightly hot-forged sample with a subsequent annealing at 250 °C for a short period, where incoherent twinning boundaries can be reorganized into coherent ones.

Keywords: Mg alloys; Twin boundary; Damping capacity; Plastic deformation

Damping capacity of magnesium (Mg) is generally dominated by basal slip, which absorbs the external vibration energy through reciprocating motion of dislocations. The high damping capacity of pure Mg can usually be ascribed to the extremely low critical resolved shear stress (CRSS) of basal slip (∼0.6 MPa) as compared to that of non-basal slip systems (>38 MPa) [1,2]. However, damping capacity of Mg is significantly reduced once it is alloyed in order to improve its mechanical properties [3]. The reduction in damping capacity has been ascribed to the inhibition of dislocation motion in basal planes and the lowered sweeping activity by solution atoms and precipitates [4, 5], leading to an increased CRSS for basal slip (e.g. to 2 MPa for AZ31 alloy) [6].

Q3 Several techniques have been applied to modify the damping capacity of Mg alloys by adding Ni [4], Cu [7], or heat treatment [8]. However, these methods generally either deteriorate the mechanical properties or lead to other problems, such as complications in manufacturing. In addition, the influence of cold or hot working on damping capacity of Mg alloys has also been studied by Fan et al. [9] via equal-channel angular pressing and Kamado et al. [10] by cold-rolling. The results indicated that plastic deformation to large strain level is unfavourable for enhancing the damping capacity of Mg alloys because it leads to an extremely high dislocation density that lowers the mobility of dislocations through their mutual interaction. This poses a further considerable challenge for the development of Mg alloys with a high damping capacity, the origin of which is twofold. On the one hand, it is difficult to improve the density of mobile dislocations in basal planes on a large scale, because a high density of dislocations generally leads to low mobility due to their mutual interactions. On the other hand, although the new interfaces formed via doping other elements or phases may play a role in increasing the damping capacity of Mg alloys, such effect is small because the intrinsic defects such as the non-coherent interface greatly inhibit the interface motion [11]. Novel methods to substantially enhance the damping capacity of Mg alloys, without compromising mechanical properties, would be highly valuable for engineering applications as well as in fundamental scientific research.

It is well known that {10−12} tensile twin boundaries in Mg are movable and can shrink and grow even at a stress much lower than the nucleation stress [11–14]. CRSS for the growth of {10−12} twin was found to be ∼2–3 MPa for AZ31 alloys [6], which is slightly higher but very close to that of the basal slip system (approximately 2 MPa) [15,16]. However, no clear evidence has been provided, and it remains unknown how twins can affect the damping capacity of Mg alloys, and also to what extent the damping capacity can be improved, although Tsai et al. [17] indicated that twins may be crucial for increasing the damping capacity of Mg−1 wt.% Zr alloys. In this work, we introduce a large number of {10−12} twins in AZ31Mg alloy through hot compression, and demonstrate, for the first

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time, that twin boundaries enhance the damping capacity of this alloy. This opens a novel route for obtaining a high damping capacity in Mg alloys and also in a wide range of other alloys with similar crystalline structures.

AZ31B extruded rods used were obtained from Osaka Fuji Corp., Japan. Cylindrical samples with dimensions of 15 mm in diameter and 40 mm in height were cut from the extruded rods. Prior to hot compression, the alloy was subjected to solution treatment (ST) at 400 °C for 2 h followed by water quenching to room temperature. Hot compressions were carried out at 250 °C along the extrusion direction (ED) to an engineering strain of 4% under strain rates of 0.001 and 20 s^{-1} in vacuum using a computer-aided hot-forging simulator. Since the duration of the compression to reach a strain level of 4% at a strain rate of 0.001 s^{-1} was 100 s, while it was only 5×10^{-3} s at 20 s^{-1}, further annealing treatment at 250 °C for 100 s was carried out on sample compressed to 4% at 20 s^{-1}, in order to eliminate the time difference. To evaluate the mechanical property of these samples, hardness of each sample was measured 10 times using a Vickers hardness tester (Shimazu, HMV) with a load of 2 N.

Microstructures were characterized by examining the centres of the samples by EBSD equipped with data acquisition software (TSL-OIM 5.0) and by transmission electron microscopy (TEM, JEOLJEM2010, JEOL Ltd.). Damping capacity was measured in terms of internal friction with an internal friction measuring system (EG-HT, Nihon Techno-Plus Co. Ltd., Japan) at room temperature, with vibration strain amplitudes ranging from 0.01% to ~0.15%. To determine the influence of texture on the internal friction of Mg alloys, we also evaluated the internal friction of ST samples cut from the extruded rods, with the longitudinal direction of test samples either parallel (0°) or perpendicular (90°) to the ED.

Figure 1(a) shows the inverse pole figure (IPF) of the initial microstructure free of twins. The average grain size is about 20 μm. Figure 1(b) and (c) shows typical microstructures of the alloy compressed to 4% at a strain rate of 0.001 and 20 s^{-1}. As expected, no obvious twinning is visible in sample compressed at 0.001 s^{-1}, indicating the dominant plastic deformation by dislocation slipping. In addition, no variation in grain size is observed, owing to the low strain level during hot compression. In contrast, a large number of $\{10\bar{1}2\}$ twins are formed in the sample compressed at 20 s^{-1} (c). After annealing at 250 °C for 100 s, no significant change is observed in either twin number or twin profile (Fig. 1(d)).

To shed light on the details of variation in the microstructure of the sample compressed at 20 s^{-1} after annealing, we further analysed the local regions denoted by rectangles A in Figure 1(c) and B in Figure 1(d). These areas include grains enriched with large numbers of $\{10\bar{1}2\}$ twins, and both areas share identical orientations. Figure 2(a) and (b) shows the grain average misorientation (GAM) maps of these two areas. Although the microstructures before and after the annealing undergo no significant change, the misorientation generated during the prior plastic deformation is remarkably reduced after annealing. The decrease in GAM after short annealing is more obvious as shown in distribution of different misorientations (Fig. 2(c)).

Figure 3 shows the internal friction of samples under various conditions as a function of strain amplitude. For comparison, the results for ST and ST-90° samples are also given. At low vibration strain amplitude (0.025%), the sample compressed at a low strain rate (0.001 s^{-1}) shows no twins, and exhibits a slightly lower internal friction than ST sample. For the unannealed sample compressed at 20 s^{-1}, which contains large number of $\{10\bar{1}2\}$ twins, the internal friction at strain amplitude of 0.025% is comparable to that of ST, implying that the presence of twin boundaries is not detrimental to internal friction. The internal friction is obviously observed to increase with the increase of the strain amplitude, in all samples. However, compared to the twin-free samples (ST, ST-90°, 0.001 s^{-1}), both twin-containing samples (20 s^{-1} and 20 s^{-1} + annealing) exhibit significantly higher internal friction. This is more obvious in the sample after annealing, implying that annealing a twin-containing sample for a short period helps to enhance the internal friction. ST

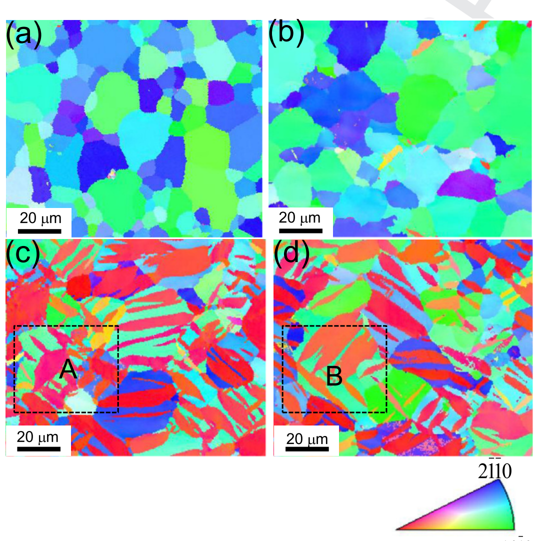

Figure 1. IPF maps of (a) ST sample and those compressed at strain rate of (b) 0.01 s^{-1}, (c) 20 s^{-1}, and (d) 20 s^{-1} with a subsequent annealing.

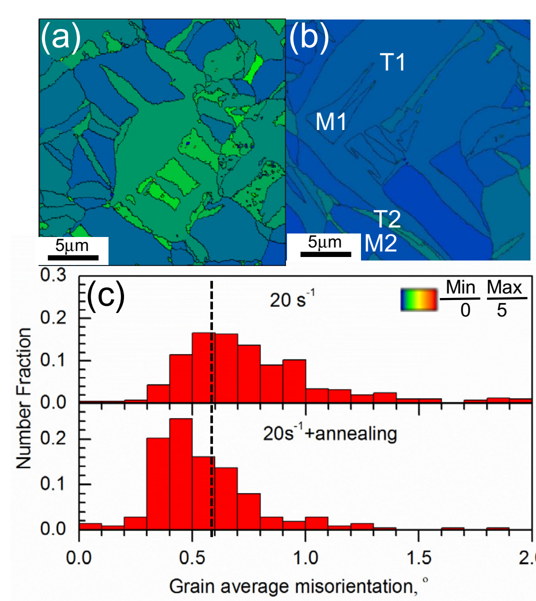

Figure 2. GAM maps of the samples compressed to 4% (a) at 20 s^{-1} and (b) 20 s^{-1} with subsequent annealing. (c) Corresponding number fractions of GAMs in (a) and (b).