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Mechanisms for enhanced plasticity in magnesium alloys

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Abstract—We have recently reported that a high strain-rate rolling process is effective for producing strong and ductile Mg alloy sheets. Here we elucidate the fundamental mechanisms that are responsible for plastic deformation in this process via systematic physical thermomechanical plane-strain rolling simulations on a Mg–Zn–Zr alloy. The strain-rate sensitivities of the alloy's microstructure and flow curves were closely correlated to the onset of deformation twinning and dynamic recrystallization (DRX). Unlike deformation at low strain rates, deformation at the high strain rates applied here induced a high number density of twins, including a predominance of $\{10\bar{1}1\}$–$\{10\bar{1}2\}$ double twins in coarse grains, and a predominance of $\{10\bar{1}2\}$ twins in fine DRX grains. We also report a new observation of $\{10\bar{1}2\}$ nanotwins in ultrafine grains with high density in bulk Mg alloy when processed at high strain rates. We propose that the high propensity for twinning at high strain rates provides nucleation sites for DRX, resulting in a uniform, ultrafine-grained microstructure that exhibits a weakened basal texture and thus excellent mechanical properties.

Keywords: Mg alloy; Thermomechanical processing; Deformation twinning; Dynamic recrystallization; Ultrafine-grained materials

1. Introduction

As the lightest structural metallic material, magnesium (Mg) alloys have significant potential in aerospace, automotive, and electronics applications [1]. However, two major problems have severely limited the industrial application of Mg alloys to date. One is the poor formability of these alloys due to the difficulty of slip in their hexagonal close-packed (hcp) crystal structure [2,3], and the fact that this issue is exacerbated when strong (e.g.) basal textures form during plastic processing [4]. The other major problem is the relatively low strength and ductility of Mg alloys when compared with steels and aluminum alloys [1].

In order to improve the formability of Mg alloys, processing at elevated temperature is required, since this has been shown to activate multiple slip systems [5]. However, high temperature processing is expensive and leads to grain coarsening. Moreover, to reduce the tendency for strain-hardening and cracking in Mg and its alloys, low deformation strain rates and intermediate annealing steps are often required during processing.

Grain refinement and texture modification are important for improving the mechanical properties of Mg alloys [1]. The average grain sizes of conventionally deformed Mg alloys are usually larger than 10 μm [1]. The basal texture in Mg alloys can be intensified both by conventional plastic deformation [1,4] and by subsequent thermal annealing [6]. These tendencies reduce greatly the formability of these alloys. In contrast, severe plastic deformation (SPD) has been a major method of grain refinement and texture modification of Mg alloys. Some SPD processes, including equal-channel angular pressing [7–9], differential speed rolling [10–12], conshearing [13] and equal-channel angular rolling [4], have resulted in combined grain refinement and texture weakening. However, these SPD processes are expensive and difficult to scale for industrial production.

We have recently developed a high strain-rate rolling (HSRR) technique that enhances the formability of Mg alloys and improves the mechanical properties of resulting Mg sheets [14–16]. High strain rates facilitate high-density twinning and, subsequently, twinning induced dynamic recrystallization (DRX) such that the twinning and DRX release stress and consume strain energy induced by plastic deformation, resulting in remarkable improvement in the formability of the alloys, via formation of an ultrafine-grained (UFG) microstructure. Significantly, both high strength and high ductility are obtained in HSRR-processed Mg sheets. To date, the precise mechanisms operating in HSRR processed Mg alloys are unclear. This is

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because previous investigations have examined only the final HSRR sheets. In this study, we applied thermomechanical plane-strain simulations to systematically investigate the microstructural evolution induced by high strain-rate deformation. Analysis of flow curves and microstructure provides new insights into the deformation mechanisms during HSRR.

2. Experimental

A ZK60 (Mg–6Zn–0.5Zr, in wt.%) Mg alloy billet prepared by permanent mold casting and a two-step homogenization (330 °C for 24 h and 420 °C for 4 h) was used in this study. The as-homogenized billet had an average grain size of ~150 μm and random grain orientations. Samples for rolling simulations were cut from the as-homogenized billet. The sample size was 6 mm in the rolling direction (RD), 20 mm in the transverse direction (TD) and 10 mm in the normal direction (ND). The rolling process was simulated by a plain strain deformation carried out on a computer-servo Gleeble 1500 simulator. Fig. 1 shows the schematics of the rolling process and setup of plane strain deformation simulation. The stress, strain and heat flow of the plane strain deformation simulation are similar to those of rolling process [17]. The samples were compressed between two anvils in one pass from 10 to 2 mm in thickness, with a true strain of 1.61 at 300 °C. These parameters were the same as those in the HSRR process conducted in our previous investigation [14–16]. A mixture of graphite and engine oil was used for lubrication. The deformation along the TD during plane strain deformation was constrained by the elastic constraint from the undeformed ends of the sample beyond the anvils, and the larger friction at the interfaces between the sample and anvils along the TD than that along the RD [17]. It was observed that the metal flow was mostly along the RD rather than the TD, which is due to the principle of least resistance, and that the post simulation extension along the TD was negligible. These observations suggest that plane strain deformation was realized by this approach. The contact area of $6 \times {\sim}20\ \mathrm{mm}^2$ (RD × TD) between the anvils and sample remained almost constant throughout the process. Each experiment was repeated three to four times, and showed excellent reproducibility. Three typical flow curves and microstructure at strain rates of 1, 15 and 50 s^{-1} were chosen to analyze the deformation mechanisms, microstructural evolution and grain orientation. The low strain rate of 1 s^{-1} is similar to the strain rates used in conventional deformation, and the strain rate of 10 s^{-1} is close to the strain rates introduced in the HSRR carried out in our previous investigations. The simulation at the strain rate of 50 s^{-1} was applied to guide future design of the HSRR process. The samples were water-quenched immediately after compression to preserve microstructure.

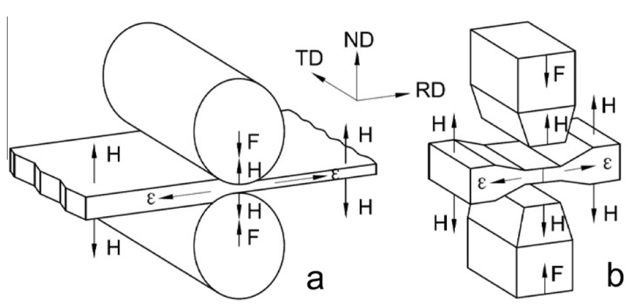

Fig. 1. Schematics of (a) rolling process and (b) setup of plane strain deformation simulation. *F*, *H* and ε indicate force, heat and strain, respectively. The figure is reproduced from Ref. [17].

The microstructure and grain orientation of the processed samples were examined on the RD–TD plane using optical microscopy (OM), electron backscatter diffraction (EBSD) and transmission Kikuchi diffraction (TKD) in a scanning electron microscope (SEM). TKD is a novel technique developed from EBSD very recently, which significantly improves the spatial resolution down to less than 10 nm and is ideal to observe ultrafine and nanosized features [18,19]. For the samples deformed at 15 s^{-1}, the microstructure was observed at the following levels of strain: $\varepsilon = 0.11$, 0.36, 0.69 and 1.61, so as to obtain a systematic elaboration of the microstructural evolution. The EBSD and TKD analyses were performed on a Zeiss Ultra field emission scanning electron microscope using AZtec and Channel 5 software to collect and analyze data. The EBSD/TKD data were used to characterize the twinning as follows: the thickness of each $\{10\bar{1}2\}$ twin, λ, was measured from the two-dimensional maps by defining it as the length of the minor axis of an ellipse fitted to each twin. The true twin thickness λ_{true} was calculated by multiplying λ with the cosine of the angle between the twin plane and the ND of the scanning plane [20].

3. Results

3.1. Flow curves

Fig. 2 provides the flow curves for the ZK60 alloy after deformation to a true strain, ε, of 1.61 at the three strain rates of 1, 15 and 50 s^{-1}. The curves in Fig. 2a were derived from the measured load–stroke curves and the stress values indicated the applied pressures from the anvils (called "applied stress" hereafter, σ_a). σ_a can be used to evaluate the required instrument capability for the future rolling process. σ_a not only contributes to the deformation but also overcomes the friction between the sample and anvils. The stress causing deformation can be represented by the Von Mises equivalent stress σ_e. σ_e can be extracted from σ_a using the following equation [21]:

$$\sigma_a = \frac{2}{\sqrt{3}}\sigma_e\left(1 + \frac{\mu a}{2h}\right) \quad (1)$$

where μ is the friction coefficient between the sample and the anvils, a is the instantaneous size along the RD and h is the instantaneous thickness of the sample. It has been reported that the mixture of graphite and engine oil is the most effective lubricant for Gleeble experiments and μ can be very low (close to zero) in plain strain compression [22]. In this study, since the compression was a one-pass deformation to a high strain and the samples could not be re-lubricated during deformation, μ should be higher. $\mu = 0.2$ is considered as a reasonable value for the calculation of σ_e [22,23]. Fig. 2b provides the σ_e–ε flow curves which were deduced from Fig. 2a using $\mu = 0.2$. It is seen that increasing strain leads to a more significant difference between σ_a and σ_e due to the increase of the a/h aspect ratio. This implies that the effect of friction on σ_a became more notable and the increase of σ_a was mostly caused by friction. To exclude the effect of friction on the flow curves,