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Effect of strain rate on deformation mechanism for ultrafine-grained interstitial-free steel

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ABSTRACT

Ultrafine-grained (UFG) interstitial-free steels with grain sizes of 0.39, 0.42, and 0.51 μm were used to ascertain effects of the strain rate ($\dot{\varepsilon}$) on the primary deformation mechanism at room temperature. Tensile tests were performed to obtain the strain-rate sensitivity exponent of 0.2% proof stress. The value was evaluated as 0.02 at high strain rates but as −0.01 at low strain rates. The transition was observed at $\dot{\varepsilon}$ of 10^{-3} s^{-1} for each sample. Although the negative m value might result from strain aging, the influence of grain boundary sliding (GBS) increased remarkably at a low strain rate because it reached 76% of plastic strain and became about six times as much as that at a high strain rate. Therefore, it is claimed that the dominant deformation mechanism was changed by the strain rate from dislocation motion to GBS with decreasing strain rate across $\dot{\varepsilon} \approx 10^{-3}$ s^{-1}.

1. Introduction

The mechanical properties of ultrafine-grained (UFG) metals with grain size (d) of <1 μm are well known to be better than those of conventional coarse-grained (CG) metals [1,2]. To produce UFG metals, several severe plastic deformation methods are applied, such as equal-channel angular process (ECAP), high-pressure torsion, and accumulative roll bonding (ARB) [2,3]. Among those, several papers have emphasized mechanical properties for ARBed materials because ARB process, as a prospective method to manufacture structural materials for engineering, is useful for fabricating bulk UFG metal plate continuously [2,4–7].

Based on those circumstances, several deformation models have been proposed to predict the deformation behavior for UFG materials. Although the models suggest that the role of the grain boundary (GB) is important for deformation, the descriptions are based on dislocation motion [8] or GB deformation [9,10]. The former described that mechanical properties are influenced by depinning processes during dislocation bow-out from GB [8]. The model explained that dislocation motion is controlled by the activation volume of the process, and showed good agreement with experimental data at the share strain rate of 10^{-4} s^{-1}. However, the latter of the two models described above focused on dislocation annihilation at GB. In this model, diffusion along GB affects mechanical properties [9,10]. Hahn et al. claimed that the occurrence of GB migration or sliding activates easily because diffusion occurs in GB that is wider than that in coarse-grained materials [9]. Blum and Zeng showed that their annihilation model reconstructs experimental data at strain rates ($\dot{\varepsilon}$) of 10^{-8}–10^{-3} s^{-1} [10].

According to earlier papers [8–10], the dominant deformation mechanism might change at $\dot{\varepsilon}$ of about 10^{-4} s^{-1} between GB deformation and dislocation motion for UFG materials. However, those papers emphasized a study of face-centered cubic (FCC) materials [8,10–12]. For body-centered cubic (BCC) ones, only that at high strain rates such as $\dot{\varepsilon} > 10^{-4}$ s^{-1} has been examined [13–16] because the machining performance of UFG steels is attractive. Therefore, the present study addressed deformation mechanisms in the low strain rate region, i.e. $\dot{\varepsilon} < 10^{-4}$ s^{-1}; mechanical tests and some microscopy were performed using interstitial-free (IF) steel.

2. Experimental procedures

IF steel with $d = 20$ μm was used as a virgin material in this study. Its chemical composition was Fe–0.13Mn–0.05Cr–0.03Ti–$<$0.004Si–0.01P–0.01S–0.003C–0.002O–0.002N [weight %]. UFG steels were manufactured by five-cycled ARB process at 823 K, which introduced the equivalent strain of about four, following annealing at 723 K for 0.5 h (sample 1 and 2) and 823 K and 1 h (sample 3). To evaluate the grain size, electron backscatter diffraction pattern analyses were performed after the

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pretreatments. The grain boundary maps were taken in the ND–RD plane, where ND is the nominal direction and RD is the rolling direction (Fig. 1). The grains of these samples were pancake-like [17] with grain thickness, width, and length of 0.39, 0.75, and 1.1 μm for sample 1, 0.42, 0.93, and 1.5 μm for sample 2 and 0.51, 1.1, and 1.8 μmfor sample 3, respectively. Although the grain sizes differed for samples 1 and 2 after the same treatment, that might derive from the difference of the plate batches.

To ascertain the strain rate effect on the deformation mechanism, tensile tests were performed to obtain the strain-rate sensitivity exponent (m) of 0.2% proof stress ($\sigma_{0.2}$) with constant cross-head speeds corresponding to $\dot{\varepsilon}=10^{-6}$–$10^{\circ}$ s^{-1} at room temperature. The loading direction corresponded to RD. In the tests, the strain was measured using strain gauges mounted directly on the specimen surfaces. The specimens were prepared with 6-mm gauge length using an electric discharge machine.

Next, transmission electron microscopy (TEM) was performed with sample 2 to observe the dislocation structure after deformation. The observations were operated with voltage of 200 kV with LaB_6 cascade. Samples were prepared by twin-jet electro-polishing with a solution of 5% perchloric acid and 95% acetic acid at 288 K at a voltage of 50 V.

Moreover, atomic force microscopy (AFM) was conducted using sample 2 to evaluate the amount of grain-boundary sliding (GBS) perpendicular to a specimen surface (v). At least 350 grain-boundary steps were measured before and after tensile tests stopped around $\sigma_{0.2}$ and tensile strength (σ_B). To obtain the contribution of GBS on plastic strain, the strain generated by it, i.e. ε_{gb}, was commuted from v using $\varepsilon_{gb}=fv/d_l$ [18], where f is the geometrical factor of 1.1 [19] and d_l is the grain length parallel to the loading direction.

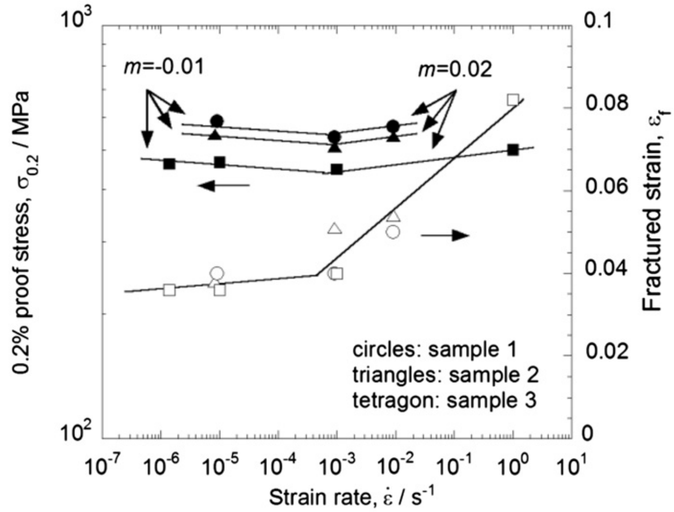

Fig. 2. Double logarithmic plot of $\sigma_{0.2}$ and $\dot{\varepsilon}$ for UFG IF steels. The m values were evaluated as 0.02 and −0.01 at high and low strain rates, respectively. It also shows that $\dot{\varepsilon}_f$ functioned by $\dot{\varepsilon}$. ε_f decreased rapidly from 0.08 to 0.04 with decreasing $\dot{\varepsilon}$.

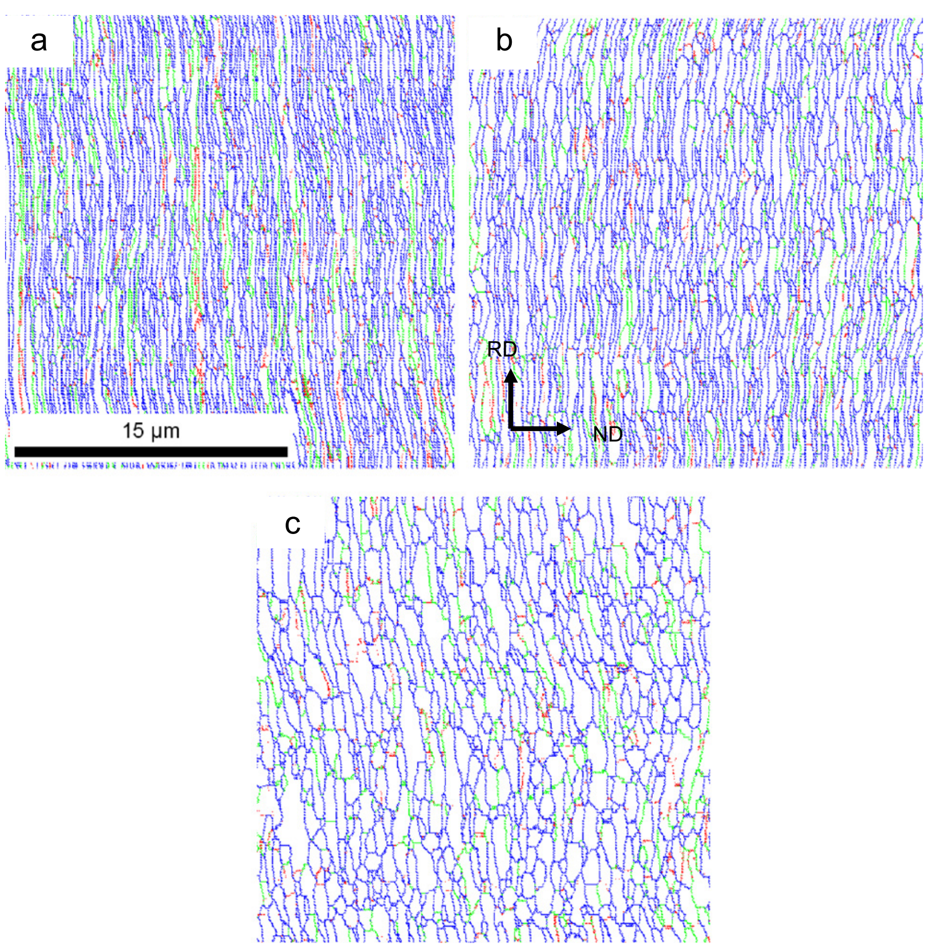

Fig. 1. Grain boundary maps of (a) sample 1, (b) sample 2 and (c) sample 3 in the ND–RD plane. The blue line shows high angle grain boundaries with misorientation of $>15^{\circ}$. Green and red lines represent low angle grain boundaries with misorientation of 5–15 and 2–5°, respectively. (For interpretation of the references to color in this figure, the reader is referred to the web version of this article.)