

Shape memory effect in Fe₃Al single crystals with D0₃ structure

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Received 3 July 2006; revised 27 July 2006; accepted 31 July 2006

Available online 1 September 2006

The shape memory effect of D0₃-ordered Fe₃Al single crystals based on the motion of 1/4⟨111⟩ superpartial dislocations was investigated. The strain recovery during heating was quantitatively discussed, focusing on the stress acting on the superpartials. © 2006 Acta Materialia Inc. Published by Elsevier Ltd. All rights reserved.

Keywords: Iron aluminides; Transmission electron microscopy; Order–disorder phenomena; Shape memory alloys; Dislocation

In general, pseudoelasticity (PE) and the shape memory effect (SME) in several crystals appear to be based on a thermoelastic martensitic transformation [1]. For PE, strain recovery takes place during unloading when the crystals are deformed above the austenite-finish temperature (A_f). On the other hand, a residual strain of the crystals deformed below the martensite-start temperature (M_s) is recovered by SME during heating above A_f . Fe₃Al single crystals with the D0₃ structure, however, were found to exhibit PE [2–11] and SME [2–4], though the martensitic transformation never occurred in them. In the D0₃ phase, a ⟨111⟩ superdislocation is generally dissociated into four 1/4⟨111⟩ superpartial dislocations bound by the nearest-neighbour (NN) and next-nearest-neighbour (NNN) anti-phase boundaries (APBs) [12]. However, in Fe₃Al with Al content between 22.0 and 25.0 at.%, 1/4⟨111⟩ superpartials moved independently, dragging the NNAPB [4–11]. At room temperature, the NNAPB pulled back the superpartials during unloading resulting in the PE of which maximum recoverable strain is approximately 5%. Moreover, an ordered domain structure of the D0₃ phase played an important role in the individual motion of 1/4⟨111⟩ superpartials; the amount of strain recovery showed a maximum at 23.0 at.% Al where the fine domain structure developed [7]. On the other hand, at 77 K, an applied strain was never recovered immediately after unloading, while strain recovery occurred during heating to room temperature resulting in the SME [2–4]. Recently, the PE of

Fe₃Al single crystals has been extensively examined to understand the mechanism [7–11] and to improve the PE properties [9], while the mechanism of the SME is not yet deeply understood. In the present study, we quantitatively analysed the SME in D0₃-ordered Fe₃Al single crystals, focusing on the stress acting on 1/4⟨111⟩ superpartials, and the possible mechanism was discussed.

Master ingots of Fe₃Al containing 23.0 and 24.7 at.% Al were prepared by melting high purity Fe and Al in a plasma arc furnace. Then, the single crystals were grown from the ingots by the floating zone method at a rate of 5 mm/h. These crystals are described as Fe–23.0Al and Fe–24.7Al throughout this paper. After homogenisation at 1373 K for 48 h, these crystals were slowly cooled to room temperature at a cooling rate of 80 K/h for D0₃ ordering. Compression specimens with gauge dimensions of 2.5 × 2.5 × 5.5 mm³ were cut from the single crystals by spark machining. The loading axis of these specimens was selected to be $[\bar{1}49]$ where the Schmid factor for the primary $(\bar{1}01)[111]$ slip is 0.50. It is noted that Fe–23Al single crystals showed a tensile elongation of 45% at the orientation, though the elongation of the polycrystals was limited to 5%. After mechanical and electrolytic polishing, a gauge mark was introduced by scratching the samples to measure a residual strain. Compression tests were performed at and below room temperature at a constant cross-head speed corresponding to an initial strain rate of 1.67×10^{-4} /s. After loading to a maximum plastic strain (ϵ_p) of 5.0%, the specimens were unloaded at the same cross-head speed. During loading and unloading, the compression samples were kept in a Dewar vessel partly filled with liquid nitrogen. After

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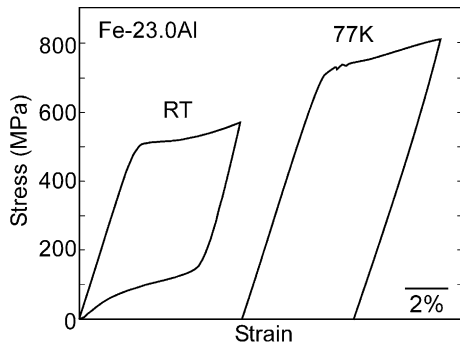


Figure 1. Stress–strain curves of Fe–23.0Al deformed to $\varepsilon_p = 5.0\%$ at room temperature and 77 K.

the deformation, the samples were taken from the vessel and heated to room temperature or 473 K. The amount of strain recovery in the crystals was evaluated using the recovery ratio (r) defined as follows:

$$r = \frac{\varepsilon_p - \varepsilon_r}{\varepsilon_p} \times 100 \quad (1)$$

where ε_r is the residual strain. Slip trace analysis was done using an optical microscope equipped with Nomarski interference contrast. Deformation substructure developed in the crystals was observed by a transmission electron microscope (TEM) operated at 300 kV.

Figure 1 shows stress–strain curves of Fe–23.0Al deformed at room temperature and 77 K. Nearly perfect pseudoelasticity is obtained at room temperature while little strain recovery occurs at 77 K immediately after unloading. It is also noted that a small serration accompanied with a deformation twin is observed after yielding in the stress–strain curve at 77 K. In addition, a clicking sound could be heard during the formation of the deformation twin because of its large shear. In contrast, there was no deformation twin in Fe–24.7Al even at 77 K. Figure 2 shows the recovery ratio of Fe–23.0Al and Fe–24.7Al as a function of deformation temperature. Recovery ratios of both the crystals just after unloading decrease with decreasing temperature and finally become less than 10%. However, heating to room temperature leads to an increase in recovery ratio, especially in the crystals deformed below 173 K. In Fe–23.0Al, further heating to 473 K results in higher recovery ratios of more than 80% (Fig. 2(a)). It is also noted that the recovery ratio of Fe–23.0Al at room temperature is slightly higher than that of Fe–24.7Al. Since the ordered domains in Fe–23.0Al were smaller than those in Fe–24.7Al, an individual motion of $1/4[111]$ superpartials was accelerated in Fe–23.0Al, resulting in a higher recovery ratio [10]. In contrast, Fe–24.7Al showed a higher recovery ratio than Fe–23.0Al at lower temperatures immediately after unloading and by heating, particularly at 173 K.

The slip trace and dislocation structure in Fe–23.0Al deformed at room temperature or 77 K are shown in Figure 3. After deformation at room temperature, a faint contrast of $(\bar{1}01)$ slip traces are observed (Fig. 3(a)). In comparison, the fine traces parallel to $(\bar{1}01)$ and $(\bar{2}11)$ slip planes can be clearly seen at 77 K as shown in Figure 3(b). $(\bar{1}01)$ slip markings smoothly connect to $(\bar{2}11)$ traces; thus, cross-slip from

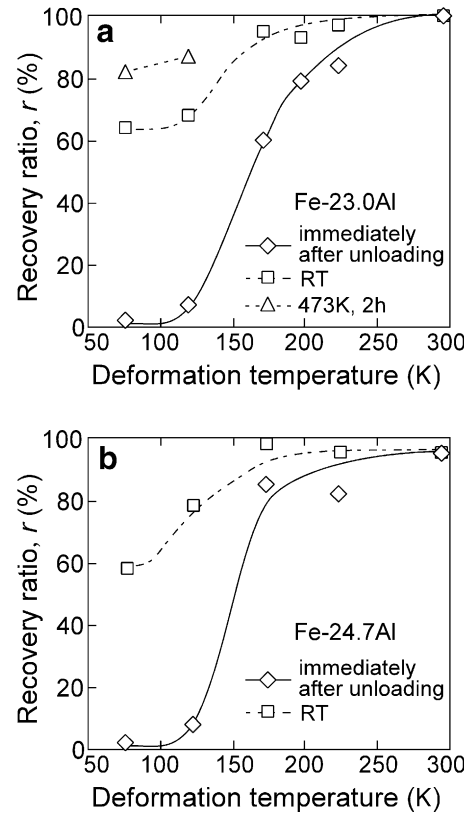


Figure 2. Variation in recovery ratio with deformation temperature in Fe–23.0Al (a) and Fe–24.7Al (b) at $\varepsilon_p = 5.0\%$. The recovery ratios immediately after unloading and those after heating to room temperature or 473 K are plotted.

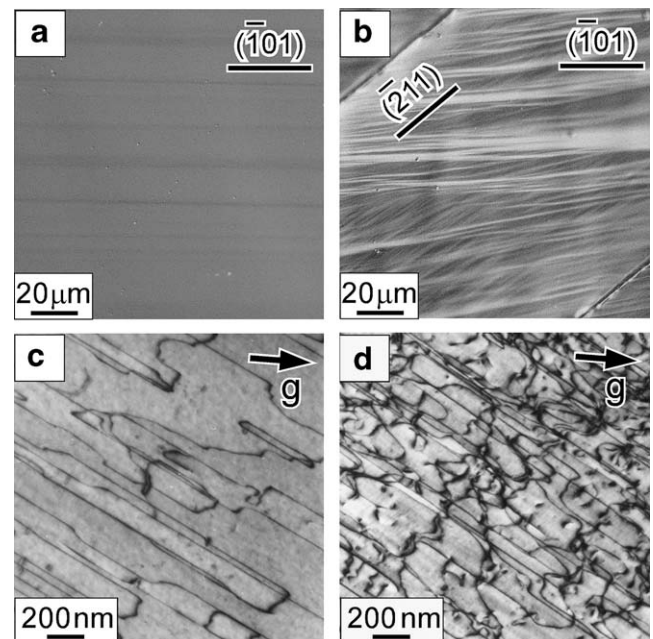


Figure 3. Slip trace and dislocation structure in Fe–23.0Al deformed to $\varepsilon_p = 5.0\%$ at room temperature (a), (c) and 77 K (b), (d). (a), (b) slip traces; (c), (d) dislocation structure; $B \approx [101]$, $g = 202$.

$(\bar{1}01)$ to $(\bar{2}11)$ plane frequently occurs. In body-centered cubic (bcc) metals, $\{211\}$ slip becomes more

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