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An intersecting-shear model for strain-induced martensitic transformation

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Abstract

Strain-induced martensitic transformation in an austenitic 18Cr–10Mn–0.4N steel was investigated using neutron diffraction and transmission electron microscopy (TEM). Based on experimental evidence from neutron diffraction and TEM indicating that a sequential $\gamma \rightarrow \varepsilon \rightarrow \alpha'$ transformation occurred and the ε intersection played a definite role in the α' formation, an intersecting-shear model for the strain-induced α' nucleation at the ε intersection is proposed. Apart from previous models for direct $\gamma \rightarrow \alpha'$ transformation, two-step transformation composed of $\gamma \rightarrow \varepsilon$ followed by $\varepsilon \rightarrow \alpha'$ is regarded as a main transformation path. In this model, two invariant-plane strains are required to complete the $\varepsilon \rightarrow \alpha'$ transformation: the first shear is of the $\{0001\}\langle 10\bar{1}0\rangle$ type and amounts to one-half the twinning shear of γ; the second shear whose magnitude is one-third the twinning shear of γ is consecutively introduced parallel to the $\langle 2\bar{1}\bar{1}0\rangle$ direction on the $\{01\bar{1}1\}$ plane. An indirect verification of the model was provided by careful analysis of the precise rotational relationship involved in the $\varepsilon \rightarrow \alpha'$ transformation. It was found that a partial dislocation ($[0\bar{1}10]$) in moving the ε variant interacted with a partial dislocation ($[10\bar{1}0]$) in the stationary ε variant, and this interaction resulted in the formation of a stair-rod dislocation ($[2\bar{1}\bar{1}0]$) which connects two ε variants.

Keywords: Martensitic transformation; Austenitic steels; Intersecting-shear model; TEM

1. Introduction

In face-centered cubic (fcc) austenitic steels various types of deformation mechanism that are responsible for accommodating plastic strains have been recognized; these are closely related to the change in microstructure under certain conditions of plastic deformation [1]. As part of ongoing efforts to improve their strength and elongation, considerable efforts and discussions have been directed to understanding two important deformation mechanisms, namely deformation-induced martensitic transformation (DIMT) [2–8] and deformation twinning (DT) [9,10]. Although strength and elongation are often mutually exclusive, i.e. a material may be strong or ductile but rarely both at the same time, DIMT and DT can provide a possibility to simultaneously increase the strength and elongation of austenitic steels. Hence DIMT and DT have been the subject of intensive research for development of high-performance austenitic steels such as transformation-induced plasticity (TRIP) [2,11] and twinning-induced plasticity (TWIP) steels [9,10].

For DIMT, metastable austenite (γ) transforms into martensite in the course of deformation, and two types of transformed products, ε martensite (hexagonal close-packed (hcp)) and α' martensite (body-centered cubic or tetragonal (bcc, bct)), have been identified depending on factors such as temperature, strain rate and chemical composition [2–4,11]. In addition, stress-assisted and strain-induced martensitic nucleation have been also

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recognized [3,5]. The former involves the same sites and embryos as does the spontaneous transformation, and the yielding is initiated by the onset of martensitic transformation. On the other hand, the latter (strain-induced martensitic transformation (SIMT)) depends on the creation of new sites and embryos by plastic deformation, and the applied stress must reach or exceed the yield stress of γ to initiate the martensitic transformation. In this regard, the strain-induced nucleation of α' martensite at the intersection of ε martensite has been known as an important example of SIMT in that nucleation site of martensite is produced by plastic deformation.

The formation of ε martensite in metastable austenitic steels with low stacking fault energy (SFE) has been well established [1,7,8,11–15]. It is known that 12 ε variants can be obtained from the 12 $\{1\,1\,1\}\langle 1\,1\,\bar{2}\rangle$ shear systems of γ. Once the variants form under thermal cycle or applied stress, their propagation results in a variety of mutual interactions, and the strains introduced during the intersection process should be accommodated in any energetically favorable way [12]. The intersection mechanisms underlying the strain accommodation can vary according to the relative configuration of individual strains associated with the intervariant relationships of the ε martensites (strain accommodation modes).

So far, there is much experimental evidence that the ε martensites provide favorable nucleation sites for the formation of α' martensite [2,7,8,11,13], since their interactions introduce new, highly potent sites for the α' nucleation. Notwithstanding many instances where the ε intersection played an important role for the purpose, the existence of ε martensite, however, has been overlooked in some cases, and it has been also argued that α' martensite can nucleate without aid of ε martensite [6,16,17]. As noted by Schumann [4], the transformation path, either $\gamma \rightarrow \varepsilon \rightarrow \alpha'$ or $\gamma \rightarrow \alpha'$, showed a strong dependence on the chemical composition of the alloys and the sequential $\gamma \rightarrow \varepsilon \rightarrow \alpha'$ transformation prevailed in the alloys containing higher amounts of γ-stabilizing elements such as Ni and Mn. The critical Mn content for the transition of transformation path from $\gamma \rightarrow \alpha'$ to $\gamma \rightarrow \varepsilon \rightarrow \alpha'$ was reported to be around 10 wt.%.

Our previous study on deformation behavior of 18Cr–10Mn–N or N + C steels showed that the deformation mode gradually changed from SIMT to DT as N + C content increased and a linear correlation between SFE and N + C content was obtained [7,8]. Along with Schumann's explanation [4], a sequential transformation path of $\gamma \rightarrow \varepsilon \rightarrow \alpha'$ can be expected in 18Cr–10Mn–N or N + C steels because the addition of N and C, which are strong γ stabilizers, into 18Cr–10Mn system may further shift towards the $\gamma \rightarrow \varepsilon \rightarrow \alpha'$ region. Given that the $\gamma \rightarrow \varepsilon \rightarrow \alpha'$ transformation is a main transformation path for SIMT in this alloy system, it is important to know the mechanism underlying strain-induced nucleation of α' martensite at the ε intersections. The present investigation is intended to give a comprehensive understanding of SIMT in relation to the ε intersection. Based on the experimental evidence in which a sequential $\gamma \rightarrow \varepsilon \rightarrow \alpha'$ transformation occurred and the ε intersection played a definite role in the α' formation, an intersecting-shear model for the strain-induced α' nucleation at the ε intersection is proposed in terms of hcp–hcp intersection and discussed in comparison with previous models [5,6].

2. Experimental

The investigated alloy was an austenitic stainless steel with the following composition (wt.%): 17.83Cr; 9.73Mn; 0.39N; 0.03C; balance Fe. After homogenization at 1250 °C for 2 h under argon atmosphere, the ingot was hot-rolled into sheets 4 mm thick, followed by air cooling. These sheets were then solution-annealed in a single-phase region of γ at 1050 °C for 1 h and then quenched into water. Interrupted tensile testing was performed using a universal testing machine (INSTRON 5882, Canton, USA) at a constant strain rate of $1.67 \times 10^{-3}\ s^{-1}$, and tensile tests were interrupted at true strains of 0.1, 0.2, 0.3 and 0.4, respectively.

After the interrupted tensile testing, deformation microstructures were characterized by means of neutron diffraction and transmission electron microscopy (TEM). Neutron diffraction experiments were conducted using a high-resolution powder diffractometer equipped with 32 detectors at HANARO, Korea Atomic Energy Research Institute, Republic of Korea. The neutron beam was monochromatized to a wavelength of 1.834 Å, and the instrumental resolution of the diffractometer was $\Delta d/d \cong 2.0 \times 10^{-3}$. The change in peak position (peak shift) was used to determine the stacking fault probability (SFP), and its sensitivity was less than 1/10th of the instrument resolution. Neutron diffraction profiles were collected at intervals of 0.05° between 30° and 140° in the 2θ range (for γ phase, (1 1 1), (2 0 0), (2 2 0), (3 1 1) and (2 2 2) peaks were obtained) with a sample rotation speed of ~30 rpm. Instrument correction was accomplished using standard reference material (SRM 640C, NIST, USA).

The procedure for line profile analysis consists of two independent schemes: (i) the Rietveld whole-profile fitting [18–22]; and (ii) the subsequent double-Voigt size–strain analysis [23]. In whole-profile fitting, a modified Thomson–Cox–Hasting (TCH) pseudo-Voigt function [21] was used as profile shape function, and in total 16 parameters for each phase including Gaussian and Lorentzian components of the full width at half maximum (FWHM) were refined using a commercial program, FULLPROF [22]. In subsequent double-Voigt size–strain analyses, both size and strain components were assumed to be modeled by independent Voigt functions. Separation of size and strain parameters could be accomplished from the difference in diffraction order dependence for the (1 1 1) and (2 2 2) reflections of γ [18–20] using the BREADTH program [24]. A detailed description on this analysis procedure was given in our previous paper [7].