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ScienceDirect

Scripta Materialia 69 (2013) 793–796

Scripta MATERIALIA

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Strain localization behavior in low-carbon martensitic steel during tensile deformation

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Received 13 August 2013; revised 30 August 2013; accepted 30 August 2013
Available online 6 September 2013

The microscopic strain distribution of lath martensitic steel during tensile deformation up to a strain of 10% has been measured in situ. Strain localization, which indicates grain interaction, is clearly observed in the vicinity of subblock boundaries, and it affects the inhomogeneous crystal rotation behavior within the block (or subblock), and hence the resultant rapid grain subdivision of the lath martensitic steel during elongation.

Keywords: Martensitic steels; Plastic deformation; Grain refining; Digital image correlation; Electron backscattering diffraction (EBSD)

Grain refinement in steel is a unique method for achieving high strength and high toughness simultaneously [1–3]. Recently, Tsuji et al. have reported that ultrafine-grained steel can be easily fabricated using a low-carbon martensite steel (Fe–0.13 wt.% C) as a starting material by a conventional cold-rolling process to reduce thickness by only 50% and a subsequent heat treatment at approximately 723 K [2]. They suggested that the deformation during the cold-rolling process results in the formation of an ultrafine-grained structure owing to the inhomogeneous deformation associated with its complex starting microstructures, such as packets, blocks, subblocks, and laths. In fact, Shanthraj and Zikry suggested from a numerical study that complex interactions between dislocations and crystallographic boundaries, such as lath and block boundaries, lead to an inhomogeneous deformation behavior in the vicinity of those interfaces during deformation [4]. In contrast, Ohmura et al. demonstrated by in situ observation during nanoindentation inside a transmission electron microscope (TEM) that the effect of interactions in a lath boundary is small [5]. In our previous study [6], we demonstrated using an in situ tensile experiment in a field emission scanning electron microscope (FESEM) equipped with an electron backscattered diffraction (EBSD) analyzer that the inhomogeneity in the activation of the slip system within a single block actually starts to develop with a relatively small strain and causes rapid grain refinement. However, it was not clarified which boundary was responsible for the inhomogeneity. In our present study, we aim to clarify the grain interaction behavior in lath martensite by investigating the changes in the strain distribution and activation behavior of the slip system using digital image correlation (DIC) [7] and EBSD analysis coupled with a crystal plasticity model, respectively.

A multilayered structure [8–11] was employed as a tool to acquire sufficient plastic deformation in martensitic steel under uniform tensile loading. The employed multilayered structure is composed of two 0.2 wt.% C martensitic steel layers and one Type 316L austenitic steel layer. First, the 20 mm thick Type 316L steel layer was sandwiched by the two 4 mm-thick 0.2 wt.% C steel layers. Then, the thickness of the as-sandwiched steel composite was reduced by hot rolling and subsequent cold rolling, so that the final thickness of the 0.2 wt.% C steel layer was approximately 130 μm. The chemical compositions of the components are shown in Table 1. The specimens were austenized at 1373 K for 1200 s and subsequently quenched by Ar and 3% H_2 gas to obtain a full lath martensitic structure. The surface of the specimen was mechanically polished, and then lightly etched using nital (2% solution of nitric acid in ethanol) to reveal a high-contrast microstructural pattern, which enables the effective tracking of the movement of the specimen surface by

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1359-6462/$ - see front matter
http://dx.doi.org/10.1016/j.scriptamat.2013.08.030

Table 1. Chemical compositions of constituent materials (wt. %).

Type of steel	C	Si	Mn	Ni	Cr	Mo	Fe
Martensitic steel layer	0.2	0.25	0.25	14	–	–	Bal.
Austenitic steel layer	0.02	0.63	0.84	12.09	17.76	2.12	Bal.

DIC during the subsequent in situ tensile experiments. The multilayered steel plate (30 mm $\times$ 5 mm $\times$ 1 mm) was uniformly elongated using a microtensile tester with an engineering strain rate of $1.0 \times 10^{-3}\ s^{-1}$. During tensile testing, the microstructure on the surface of the specimen was filmed using the FESEM (JSM7001F, JEOL). The distance between two indentation markers was measured on the surface of the martensitic steel layer, such that macroscopic tensile strains of 0, 3, 6 and 10% were applied in a stepwise manner to a single sample. Crystallographic features were analyzed by EBSD (TexSEM Laboratory (TSL)) using a detector attached to the FESEM, with acceleration voltage 15 kV, beam spot diameter 20 nm and step size 0.5 μm. Scanning areas were approximately 150×200, 180×200, 200×250 and $250 \times 300\ \mu m^2$ for the 0, 3, 6 and 10% elongations, respectively. The rolling (RD), transverse (TD) and normal (ND) directions of the specimen were used as the standard coordinate system in the crystallographic orientation analysis. The tensile direction corresponds to RD in this study. Accordingly, all the inverse pole figures and color coding in the inverse pole figure maps in this study plot the crystal orientation parallel to the RD. Strain fields were derived using the DIC software based on Image J [12] developed in house on the basis of a classical algorithm [13], keeping the average distance between the points of interest at 0.5 μm.

Figure 1a and b shows the microstructure of the martensitic steel before and after 10% elongation, respectively, as revealed by EBSD analysis. The black lines in Figure 1 indicate the boundaries where the misorientation between adjacent points is >5°, and it is demonstrated that, after 10% elongation, many new boundaries with misorientations of >5° are formed. Figure 2 shows the change in distribution of the maximum shear strain in the martensitic steel (highlighted area in Fig. 1) during deformation obtained by DIC analysis. The grain structure derived by EBSD analysis is superimposed on the figure, and the black lines indicate boundaries with misorientations of >15°. The increment in maximum shear strain for macroscopic tensile strain intervals of 0–3%, 3–6% and 6–10% are shown in Figure 2a–c, respectively. As shown in Figure 2a, the deformation behavior of the martensitic steel is inhomogeneous even at a strain of <3%. From 3% to 6% elongation (see Fig. 2b), a significant strain localization is found in some of the blocks. Subsequently, the intensity of strain accumulation in the strain localized area steadily increases up to 10% elongation, as shown in Figure 2c. The increment in maximum shear strain for macroscopic tensile strain intervals of 6–10% and the grain boundary map before deformation within a selected block of the martensitic steel are presented in Figure 3a and b, respectively. The profiles of the increment of maximum shear strain along line A (depicted in Fig. 3a) and the misorientation angles along line B (depicted in Fig. 3b) are shown in Figure 3c and d, respectively. A total of four subblock boundaries, indicated in Figure 3d by B1, B2, B3 and B4, are found within the block. The misorientation angles at the subblock boundaries are 4.9°, 8.2°, 7.4°, and 5.6°, respectively, which are in good agreement with those obtained in the previous study by Kitahara et al. [14]. As shown in Figure 3c, the peak positions of the increment of maximum shear strain indicated as P1, P2, P3 and P4, correspond well to the positions of the subblock boundaries. This result indicates that a subblock boundary acts as a barrier that interacts with mobile dislocations during deformation [15]. Noticeable strain localization at the P4 that corresponds to the B4 subblock boundary seems to be due to grain interaction in the vicinity of not only the B4 subblock boundary but also pre-existing (such as inner

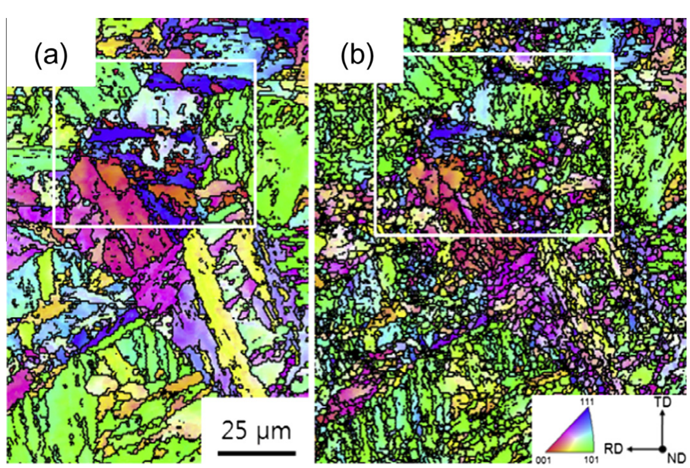

Figure 1. Crystal orientation maps of 0.2 wt.% C martensitic steel (a) before and (b) after 10% elongation.