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Revisit the role of deformation twins on the work-hardening behaviour of twinning-induced plasticity steels

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ABSTRACT

The present study found that the work-hardening rate and dislocation density in a twinning-induced plasticity (TWIP) steel deformed at 373 K and 473 K are comparable to that deformed at 298 K, but deformation twins are considerably prohibited at 373 and 473 K. High dislocation density induced by dynamic strain aging (DSA) is the dominant mechanism responsible for the high work-hardening rate of TWIP steels at 373 and 473 K. It indicates that TWIP steels can also achieve high working-hardening rate without the formation of deformation twins.

Over the past two decades, high manganese twinning-induced plasticity steels (TWIP) with an excellent combination of strength and ductility have attracted intensive research interest [1–6]. Tremendous efforts have been made to explore the role of deformation twins on their work-hardening behaviour [2,7–9]. It has been reported that the high work-hardening rate of TWIP steels is related to the continuous formation of deformation twins. Their boundaries could act as strong obstacles for dislocation gliding, resulting in a “dynamic Hall-Petch” grain refinement effect [7,10,11]. It has also been accepted that the twin boundaries (TBs) could enhance the multiplication of dislocations by efficiently reducing the mean free path of dislocations [7,12]. However, Liang et al. [13] recently demonstrated that the contribution of deformation twins to the flow stress is insignificant. It can be seen that the mechanisms determining the high work-hardening rate of TWIP steels are still under debate. The aim of the present work is to further clarify these mechanisms by investigating the work-hardening behaviour of a typical TWIP steel at various temperatures and strain rates. The evolutions of dislocation density and twin volume fraction with strain were evaluated quantitatively.

The starting material has a chemical composition of Fe-18Mn-0.75C-1.7Al-0.5Si (wt.%) and a fully recrystallized microstructure with an average grain size of 4.4 μm. Tensile specimens with gauge dimensions of 10 × 4 × 1.8 mm were cut from the plate. Tensile tests were carried out on a universal machine equipped with a heating furnace and high-temperature extensometer. The specimens were tested at a strain rate of 0.00625 s^{-1} at eight temperatures ranging from 253 K to 573 K. A K-type thermocouple was welded on the center of the sample to measure the temperature. The tests were interrupted at different true strains (0.1, 0.2, 0.3, 0.4). These interrupted samples are named S10, S20, S30 and S40 hereafter. The interrupted samples were prepared for scan electron microscope (SEM) and synchrotron X-rate diffraction (XRD) experiments. The SEM samples were ground to 50–100 μm in thickness followed by twin-jet electropolishing with an ethanol solution of 5% perchloric acid at 243 K. In order to get the best contrast of the twins, the Inlens mode was applied under a working voltage of 5 kV in a LEO 1530 SEM machine. During image collection, a line integration method was used to reduce the noise, and the brightness was set to 0 while the contrast was about 44.9. A typical SEM image is shown in the supplementary material (Fig. S1). The synchrotron XRD experiments were carried out at the of Shanghai synchrotron facility (beamline No. 14B). The energy of the synchrotron X-ray was 18 keV.

Fig. 1a shows the engineering stress-strain curves deformed at eight temperatures ranging from 253 K to 573 K at a strain rate of 0.00625 s^{-1}. The temperature-dependent tensile properties are summarized in Fig. 1b. It reveals that the yield strength decreases monotonically with temperature, which is consistent with results reported in [14]. However, the uniform elongation (UE) decreases firstly and then increases and finally decreases again with temperature. According to the values of UE, we subdivided Fig. 1b into three regions. It is interesting to note that the ultimate tensile strength (UTS) keeps almost unchanged within region II (from 198 K to 473 K), and the Portevin–Le Chatelier (PLC) effect is also observed in this region. Fig. 1c reveals the work-hardening rate

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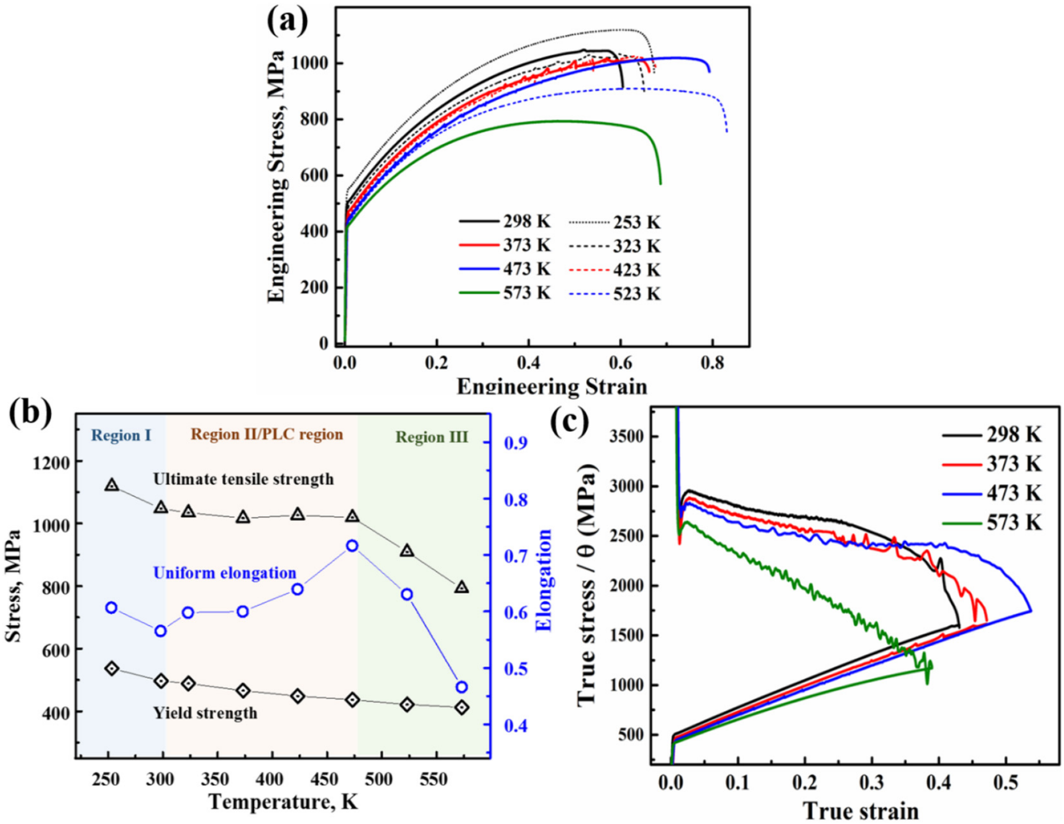

Fig. 1. Tensile properties at various temperature at a strain rate of 0.00625 s^{-1}: (a) the engineering stress-strain curves; (b) the summarized tensile properties showing a temperature region with PLC effect (Region II); (c) the work hardening rates and true stress-strain curves at four selected temperatures. For the color images, please refer to the online version of this article.

and the true stress-strain curves at four selected temperatures. It shows that the work-hardening rate at 298 K, 373 K and 473 K are all very high compared to that at 574 K.

Fig. 2 shows SEM images of samples interrupted at a true strain of 0.3 at various temperatures. More SEM pictures are shown in Fig. S2 in the supplementary material. Intensive deformation twins can be found in the samples deformed at 298 K (Fig. 2a). By comparing among Fig. 2 a-c, it demonstrates clearly that the twin density decreases with temperature. Fig. 2d presents the evolution of twin volume fraction with true strain at various temperatures. The method used to evaluate the volume fraction of twins is described in the supplemental material (Fig. S3). The results indicate that the volume fraction of twins decreases significantly with temperature as shown in Fig. 2d, which is consistent with the results reported in [15,16].

By combining Fig. 1 and Fig. 2, one can find that the mechanical properties and work-hardening rate of TWIP steel deformed at 373 K and 473 K are comparable to that deformed at 298 K, although the deformation twins are prohibited significantly at 373 K and 473 K. At a true strain of 0.3, the respective volume fractions of twin in the sample deformed at 298 K and 473 K are 7.8% and 0.6%, while the corresponding flow stress increment (true stress at 0.3 true strain minus yield stress) are 810 MPa and 770 MPa. The difference in volume fractions of twin at 298 K and 473 K is significant (7.2%), but the difference in flow stress increment is very small (40 MPa). It indicates that the deformation twins may not be necessary for the high flow stress and high work-hardening rate of TWIP steels. In fact, it has also been reported that the work-hardening rate of Hadfield steel at 498 K (without deformation twins) was higher than that at 223 K (with intensive deformation twins) [17].

Dislocation densities are determined by using the convolution multiple whole profile (CMWP) method developed by T. Ungár et al. [18,19]. The parameters used for fitting were carefully determined based on [20]. Fig. 3a shows the {110} Bragg peaks of the deformed samples (S40) at four temperatures and the as-received sample. The large plastic deformation contributes to a significant peak broadening as shown in Fig. 3a. Fig. 3b reveal the calculated dislocation densities of samples deformed at different temperature. One can find that the evolution of dislocation densities at 298 K, 373 K and 473 K are similar. As the true strain exceeds 0.2, the dislocation densities at 298 K, 373 K and 473 K are much higher than that at 573 K.

At room temperature, a large amount of deformation twins can be observed as shown in Fig. 2a. Since the twin boundaries could be obstacles for dislocation glide, Bouaziz et al. [7] proposed that these twins increase the dislocation multiplication rate by decreasing the dislocation mean free path. However, in the present study, the respective dislocation density of S30 samples deformed at 298 K and 373 K are 290×10^{14} m^{-2} and 292×10^{14} m^{-2}, while the corresponding twin volume fraction are 7.8% and 1.0%. The large difference in twin volume fraction does not result in different dislocation density. It suggests that the twins may not be the dominant mechanism for the high dislocation multiplication rate in TWIP steels, especially at 373 K and 473 K.

It has also been accepted that the dislocation accumulation of TWIP steels is controlled by the stacking fault energy (SFE) [21]. Owing to the low SFE (20–40 mJ m^{-1}), perfect dislocations in TWIP steels tend to dissociate into two Shockley partial dislocations. The cross-slip is strongly inhibited by such extended dislocation core unless this extended core is constricted by external stress with the assistance of thermal activation [22]. Therefore, the dislocation multiplication can be enhanced with suppressed dynamic recovery caused by low SFE. Nevertheless, in this paper, the dislocation multiplication within the temperature ranging from 298 K to 473 K are comparable as proved in Fig. 2c, while the SFE increases from 25 mJ m^{-2} to 58 mJ m^{-2} as calculated through Dumay's thermodynamic model [23]. It suggests that the SFE