Acta Materialia 132 (2017) 264–275

Contents lists available at ScienceDirect

Acta Materialia

journal homepage: www.elsevier.com/locate/actamat

Full length article

Simultaneous twinning nucleation mechanisms in an Fe–Mn–Si–Al twinning induced plasticity steel

B. Mahato [a, b], T. Sahu [c], S.K. Shee [d], P. Sahu [a, *], T. Sawaguchi [e], J. Kömi [f], L.P. Karjalainen [f]

[a] Department of Physics, Jadavpur University, Kolkata, 700 032, India
[b] Materials Science and Technology Division, National Metallurgical Laboratory, Jamshedpur, 831 007, India
[c] Department of Physics, S.A. Jaipuria College, Kolkata, 700 005, India
[d] Department of Physics, Midnapore College, Midnapore, 721 101, West Bengal, India
[e] Vibration Control Materials Group, National Institute for Materials Science, 1-2-1, Sengen, Tsukuba, Ibaraki, 3050047, Japan
[f] Centre for Advanced Steels Research, Box 4200, FIN–90014, University of Oulu, Finland

ARTICLE INFO

Article history:
Received 30 January 2017
Received in revised form
18 April 2017
Accepted 21 April 2017
Available online 21 April 2017

Keywords:
High-Mn TWIP steels
Transmission electron microscopy
Deformation twinning
Stacking faults
Stacking fault energy

ABSTRACT

Two simultaneous twinning nucleation mechanisms were observed in transmission electron microscopy (TEM) during low strain rate tensile deformation of a Fe–27Mn–2.5Si–3.5Al twinning induced plasticity (TWIP) steel at −40 °C. Deformation twinning took place on the conjugate $(1\bar{1}1)$ plane through the activation of pole and three–layer mechanisms producing Shockley twinning dislocations having the same Burgers vector. Secondary twinning mechanisms were not identified. The active twinning mechanisms were influenced by the stacking fault energy (SFE) of the deformed austenite, which as estimated using electron and X-ray diffraction was ~15 mJ/m^2 and 20 mJ/m^2 for 2% and 5% strains, respectively. This low SFE is believed to hinder cross–slip of twinning Shockley partials so that the stair–rod cross–slip twinning mechanism was not favored. The selection of twinning mechanisms by the microstructure is explained through critical twinning stress, the incidence of intrinsic and extrinsic stacking faults, as well as proper $\frac{a}{2}[1\bar{1}0]$ dislocations having a screw component that serve as suitable pole dislocations. The active twinning mechanism is plausibly not contributing markedly to the strain hardening of TWIP steels but the SFE and other hardening mechanisms are seemingly more prevalent.

1. Introduction

The past several years have seen considerable efforts to interpret the deformation microstructure of twinning induced plasticity (TWIP) steels since they reveal excellent strain hardening due either to dislocation glide and/or deformation twinning [1,2]. The stacking fault energy (SFE) of austenite is a very crucial factor among other parameters in controlling the deformation behavior and the occurrence of deformation twinning in such steels [1,2]. Twinning in low SFE fcc metals/alloys depends on a number of internal and external parameters, namely, the chemical composition, SFE, temperature, strain rate, grain size and orientation, etc. [3].

Mechanical twinning starts when the twinning stress becomes lower than the slip stress and only if the critical twinning stress $(\tau_c)_{twin}$ of the metal is surpassed. The critical event in twinning is its nucleation and growth can occur at stresses that are a fraction of the nucleating stress [4]. Various nucleation and growth mechanisms for twinning in fcc metals have been established [5–11], which are also admissible for the TWIP steels and is best observed using Transmission Electron Microscopy (TEM). However, the infinitesimal length and time scales of in- or ex-situ experiments restrain observations on the twin nucleation. High–resolution TEM studies in TWIP steels involving the characterization of perfect and/or partial dislocations for identification of the twinning mechanism under different loading conditions are scarce and quite divergent views on the twinning mechanisms in these steels are reported in the literature [12–15].

Huang and coworkers [16–18] while studying deformation twinning in micron and submicron TWIP steel pillars have demonstrated that twinning nucleates through glide of Shockley partials from the surface of the pillars and is strongly correlated with the 'size-effect'. The common perception for twinning in TWIP steel is that one mechanism is responsible for nucleation and that

* Corresponding author.
E-mail addresses: psahu74@gmail.com, psahu@phys.jdvu.ac.in (P. Sahu).

http://dx.doi.org/10.1016/j.actamat.2017.04.046
1359-6454/

the growth takes place by secondary mechanisms [13,14]. Mahato et al. [15] recently reported that both nucleation and growth of deformation twinning in a Fe–27Mn–2.5Si–3.5Al TWIP steel is according to the three–layer mechanism [7] and no secondary mechanisms were identified. However, despite some previous efforts to study the twinning nucleation in TWIP steels [13–15], several issues still remain unclear. Namely, for an Fe–Mn–Si–Al TWIP steel, the $(\tau_c)_{twin}$ values at room temperature (RT) for the occurrence of pole and three–layer mechanisms are very much comparable ~ 200 MPa [15], in spite of that, no evidence could be obtained as to why only the tree-layer mechanism was discriminated by the microstructure. Further, it is not irrefutably known on which closed packed plane twinning first nucleates and the nature of dislocations involved in it.

The critical SFE to obtain the TWIP effect is in the range of 20–30 mJ/m^2. However, such a SFE value alone is not sufficient to predict the twin tendency of such steels but the stress state and the difficulty for perfect dislocation glide must also be taken into consideration [1]. Thus, the observation of twin nucleation mechanisms and the interrelation between SFE and plasticity modes are fundamental to interpret the strain hardening behavior of such steels. However, in such low SFE steels, the extension of stacking fault (SF) should be deformation dependent because of the dislocations' interaction on the SF in real crystals that introduce lattice distortions in it. This lead to the concept of apparent or effective SFE [19], wherein, an elastic coherency strain energy (E_{str}) is added up to the ideal SFE corresponding to infinitely wide SFs. Nevertheless, the SFE of austenite has a very important role in controlling the twinning mechanism in these steels [1,2]. A sub-zero test temperature should lower the SFE further, which in turn would increase the incidence of twinning.

Therefore, we chose to investigate the twinning mechanism(s) in an Fe–27Mn–2.5Si–3.5Al steel during low strain rate interrupted uniaxial tensile tests at −40 °C. This steel at RT had a deformation dependent SFE ~20 mJ/m^2 and it revealed profuse twinning [15]. The deformation dependent SFE of austenite at some low terminal strains related to emergence of twinning were estimated using Weak Beam Dark Field (WBDF) imaging in TEM and also complemented by X–ray diffraction (XRD) analyses. Further, the dislocation activity pertaining to twinning nucleation was studied and identified according to the two-beam dynamical theory of electron diffraction [20,21]. In view of the diverging perceptions about the twinning mechanism in TWIP type steels, the present work is envisaged to unambiguously resolve the nebulous concepts related to the onset of twinning with a greater emphasis on the nucleation stage and to introspect potential effect of the observed twinning mechanism(s) on the strain hardening behavior of the steel, as reported in the literature [13,14]. This is important from a fundamental point of understanding the SFE mediated twinning phenomenon in such steels.

2. Experimental

The steel used in the present work had the nominal composition Fe–27Mn–2.3Si–3.5Al (all compositions are in wt%) and it was prepared by induction furnace melting followed by hot rolling at 1100 °C. It was then homogenized at 1100 °C and water quenched to RT. The steel had an average grain size of ~48 μm. Uniaxial tensile tests were carried out until failure in a Zwick Z 100 tensile testing machine (Zwick Roell, GmbH) at −40 °C with a low strain rate of 10^{-4} s^{-1}. Here, we primarily concentrate microstructures corresponding to 2% and 5% true strains, because they revealed the nucleation of twinning and are thus most relevant. The uniformly strained gauge region of deformed samples were mechanically ground to about 100 μm thickness and few 3 mm disks were punched, which were subsequently electropolished with a TenuPol–5 using a solution of 90% methanol and 10% perchloric acid. TEM observations were carried out on a JEOL 2200FS TEM equipped with a GATAN 994UltraScan® 1000XP imaging system. XRD data acquisition using monochromatic $Cu-K_{\alpha 1}$ radiation ($\lambda = 1.5418$ Å) was carried out with a Bruker D8 Advance X-ray powder diffractometer operating in the Bragg–Brentano geometry.

3. Methodologies for stacking fault energy measurements

The SFE can experimentally be determined adopting the WBDF technique in TEM either by measuring the radii of isolated extended nodes [22] and/or by measuring the extension widths of $\frac{a}{2}\langle 110\rangle$ glide dislocations [23]. Another method is in the use of X-ray analyses following Schramm and Reed method [24], later modified by Dey et al. [25], which considers the effect on dislocation parameters on SFE. It is trivial that both the TEM and X-ray methods have their own merits and demerits and should be used concurrently for reliability of the results.

3.1. Transmission electron microscopy

Here we restrict the description only to the method of SFE determination using the extension widths of dissociated $\frac{a}{2}\langle 110\rangle$ dislocations since extended isolated dislocation nodes were not readily observed under present deformation conditions. In equilibrium configuration of dissociated dislocations, the repulsive elastic force between the partial dislocations balances the attractive force resulting from increase in the area of the SF. The following equation based on an approximation of anisotropic dislocation theory is used to determine the SFE from the experimental separations for few dissociated dislocation orientations [26].

$$\gamma_{eff} = \frac{Gb_s^2}{8\pi x_{eq}(1-\nu)}(2-\nu-2\nu\cos 2\beta) \tag{1}$$

where γ_{eff} is the effective SFE, $G = 72$ GPa is the shear modulus of the steel, b_s is the Burgers vector of the Shockley partials, x_{eq} is the equilibrium separation distance of the partial dislocation pairs, β is the angle between the perfect dislocation line and its Burgers vector and $\nu = 0.24$ is the Poisson's ratio of the steel considered for an identical composition of steel in the literature [27].

3.2. X–ray diffraction

The newer method of X-ray estimation of SFE [25] is based on the consideration of strain field anisotropy induced by the contrast factor of dislocations [28] and the effects of dislocation density and arrangement parameters on the SFs that supersede the original approach of Schramm and Reed [24]. Taking these effects into consideration, the expression for SFE is given as [25]:

$$\gamma_{eff} = \frac{6.6\times A_i^{-0.37}Ga}{\sqrt{3}\pi P_{sf}}\left\{\left(\frac{b}{2\pi}\right)^2\rho\overline{C}_{111}\pi ln\left(\frac{R_e}{L}\right)\right\}\left(\mathrm{mJ/m^2}\right) \tag{2}$$

where, P_{sf}, a and ρ are stacking fault probability, lattice parameter and dislocation density of deformed austenite, respectively. b is the magnitude of Burgers vector of $\frac{a}{2}\langle 110\rangle$ perfect dislocations, L is the Fourier variable, R_e is the effective outer cut–off radius of the dislocations, signifying the range over which the distribution of dislocations can be treated as random, A_i is an anisotropy factor