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Viewpoint Paper

Internal stresses and the mechanism of work hardening in twinning-induced plasticity steels

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Abstract—An analysis of the loading–unloading behaviour of tensile tests of a 22Mn–0.6C (wt.%) twinning-induced plasticity steel shows the development of a very high level of internal stresses concomitant with its peculiar, nearly constant strain-hardening rate. A composite-type strengthening contribution from the increasing volume fraction of twin lamellae offers an explanation for an enhanced internal stress development in this type of materials. Proof of the quantitative significance of the composite-type strengthening requires, however, further study.

Keywords: Austenitic steels; Twinning; Plastic deformation; Internal stresses; Bauschinger effect

1. Introduction

Deformation twins contributing to the plastic strain of twinning-induced plasticity (TWIP) steels are of nanometric thickness ($\sim$100 nm). In polycrystals, after initial plastic deformation of a few per cent by crystallographic slip, deformation twins start to form in favourably oriented grains, where they occur simultaneously with dislocation-mediated slip in several systems in order to achieve the required macroscopic strain [1]. These twins cross through the austenite until they are stopped by high-angle grain or twin boundaries (one or two twin systems are activated per grain, rarely more than two). Due to their small thickness and because they embed a high density of sessile dislocations [2,3], the plastic strength of these twins is most probably very high. As a result of their high aspect ratio and continuity across the austenitic domains these twins are necessarily obliged to undergo almost the same deformation of the surrounding matrix in the subsequent elastoplastic deformation (at least for the in-plane strain components of the lamellae and matrix). A "composite strengthening" contribution to the flow stress of TWIP steels, derived from the thin twins generated by prior deformation, should not therefore be excluded. Likewise, it is also necessary to consider the contributions to strengthening from both the reduction of the dislocation mean free path in the matrix because of the presence of the twins (the so-called "dynamic Hall–Petch effect"), and the strengthening of the matrix by a dislocation density that increases as strain proceeds [4–8]. Given a TWIP composition and an initial grain size D_0, we can thus formally distinguish in the flow stress evolution the terms:

$$\bar{\sigma}(\bar{\varepsilon}) = \bar{\sigma}_0 + \Delta\bar{\sigma}[\rho_M[\bar{\varepsilon}, D_0, D_M(\bar{\varepsilon})]] + \Delta\bar{\sigma}\lfloor f_V^T(\bar{\varepsilon}), t_T(\bar{\varepsilon}), \rho_T(\bar{\varepsilon})\rfloor, \tag{1}$$

where $\bar{\sigma}$ is the macroscopic (equivalent) tensile flow stress, $\bar{\varepsilon}$ is the macroscopic (equivalent) tensile plastic strain, $\bar{\sigma}_0$ is the lattice friction stress plus other grain-size-independent and deformation-independent contributions to the flow stress of the austenite (solid-solution strengthening), ρ_M is the dislocation density in the untwinned matrix, D_M is the mean intercept distance between high-angle boundaries in the matrix (high-angle grains or twin boundaries), f_V^T is the deformation twin volume fraction, t_T is the average twin thickness and ρ_T is the dislocation density in the twins. Eq. (1) may be rewritten as a rule-of-mixtures addition:

$$\begin{aligned}\bar{\sigma}(\bar{\varepsilon}) &= \lfloor 1 - f_V^T(\bar{\varepsilon})\rfloor\bar{\sigma}_M(\bar{\varepsilon}) + f_V^T(\bar{\varepsilon})(\bar{\sigma}_T)_{eff} \\ &= \lfloor 1 - f_V^T(\bar{\varepsilon})\rfloor\bar{\sigma}_M(\bar{\varepsilon}) + \beta(\bar{\varepsilon})f_V^T(\bar{\varepsilon})\bar{\sigma}_T.\end{aligned} \tag{2}$$

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The strength of the matrix is assumed to arise from the "dynamic Hall–Petch" effect. The strength of the thin twin lamellae is supposed to be higher than the matrix strength, $\bar{\sigma}_T > \bar{\sigma}_M$, probably much higher, as long as the twin thickness remains, on average, smaller than the matrix domain size, $t < D_M$, i.e. as long as $f_V^T < 0.5$. The factor $0 < \beta(\bar{\varepsilon}) \leq 1$ is a factor accounting for the elasto-plastic misfit between the matrix and twins, plausibly an increasing function of the macroscopic strain, because, once formed, the twin lamellae progressively align with the tensile direction as strain increases (an alignment that results in an increasing effectiveness of the lamellar reinforcement). A plausible lower bound estimation of the twin strength is given by the critical stress for dislocation propagation confined to the intralamellar spacing left for the slip plane of the dislocation, of the order of Gb/t on top of $\bar{\sigma}_0$, with G the shear modulus (81 MPa [9]) and b the modulus of the Burgers vector (0.258 nm for a perfect dislocation [9]). Another estimation of the twin strength can be obtained from the Hall–Petch relationship applied to the twin thickness (the Hall–Petch slope is $0.36\,\mathrm{MPa\,m^{1/2}}$ for the tensile yield stress [10]). For plastic deformations above the twinning threshold, the compatible deformation of matrix (relatively soft) and twins (relatively hard) necessarily leads to the building-up of an heterogeneous field of internal stresses—on average, forward stresses in the twins (helping the applied stress) and backward stresses in the matrix (opposed to the applied stress). In the absence of other sources of internal stress:

$$\begin{aligned}
&\bar{\sigma}_M < \bar{\sigma} < (\bar{\sigma}_T)_{eff} \\
&\bar{\sigma}_M = \bar{\sigma} + (\sigma_{\mathrm{int}})_M \\
&(\bar{\sigma}_T)_{eff} = \bar{\sigma} + (\sigma_{\mathrm{int}})_T \\
&f_V^T(\sigma_{\mathrm{int}})_T = -\left(1 - f_V^T\right)(\sigma_{\mathrm{int}})_M
\end{aligned} \tag{3}$$

The internal stress level in the structure of a plastically deforming TWIP steel can be assessed from diffraction experiments or from the yield stress upon reversal of the sense of the strain in a test at a given forward pre-strain (Bauschinger-type test). In Bauschinger-type mechanical tests, assuming equilibrium, if the tensile flow stress is $\sigma_{fw} = \bar{\sigma} > 0$ after some given pre-strain, $\bar{\varepsilon}$, when the test displacement is reversed and the sample, after some elastic unloading, yields again at a stress level $(\sigma_{bw})_Y$, the level of average backward stress in the "soft" regions of the deforming structure is given by:

$$\sigma_b = \frac{\sigma_{fw} + (\sigma_{bw})_Y}{2} = \frac{\bar{\sigma} + (\sigma_{bw})_Y}{2}. \tag{4}$$

An important composite-type contribution of the previously created deformation twins to the current strength of deforming TWIP steel should lead to anomalously high levels of back-stress with respect to that measured in materials exclusively deforming by slip, i.e. $-(\sigma_{\mathrm{int}})_M$ should represent an important fraction of the measured back stress, σ_b, in TWIP steels. Of course, there are other micro- or mesoscopic contributions to the measured back-stress other than reinforcement by thin twin lamellae, including microscopic back-stresses from dislocation groups, and mesoscopic stresses originating from the plastic anisotropy of the different grain orientations; however, Bauschinger tests only give access to measurements of σ_b, not to $(\sigma_{\mathrm{int}})_M$ as used in Eq. (3). Bouaziz et al. [7] measured a back-stress to flow stress ratio of the order of 50% for a 22%Mn–0.6%C TWIP steel, and Gutierrez-Urrutia et al. [8] found a ratio of 20% for the same steel composition (both results were obtained by performing Bauschinger tests in shear). In this paper we present experimental evidence of still higher levels of back-stress present during the tensile deformation of the same TWIP steel. We show that conflict between these results is only apparent.

Obtaining the value of $(\sigma_{bw})_Y$ from a mechanical test using the loading–unloading Bauschinger scheme presents several difficulties or uncertainties [11], the main one probably being that, because of the gradual re-entry in plasticity of the sample upon strain reversal, defining $(\sigma_{bw})_Y$ depends on the selected offset deviation from the elastic unloading line. One common option is to use a conventional criterion of yielding for some fixed offset deformation, i.e. 0.2%; another one is to continue the reverse deformation well beyond such limit until any possible transient stage is surpassed and then to compute the back-stress from the difference between the monotonic curve and the reverse strain curve at equal strain-hardening rate ("permanent softening" criterion). Bouaziz et al. [7] employed the first criterion (0.2% offset) and found $\sigma_b/\bar{\sigma} = 0.5$, as mentioned. Gutierrez-Urrutia et al. [8] favoured the second approach ("permanent softening" criterion) and measured $\sigma_b/\bar{\sigma} = 0.2$; if they had used the 0.2% offset, their result would have been quite similar to that of Bouaziz et al. [7]. Here we have studied in detail the influence of the offset strain on the measurement of $(\sigma_{bw})_Y$; the results allow us to determine, by back-extrapolation to zero offset strain, the back-stress as monotonic forward plastic deformation proceeds.

2. Material, experimental results and discussion

Room-temperature tensile tests of a cold-rolled and annealed fully austenitic TWIP steel of composition 22.3% Mn, 0.59% C, 0.22% Si (wt.%) and several grain sizes (ranging from 1.5 to 19 μm) were run on an Instron 4505 testing machine according to ASTM E 8M-04 standard under displacement control at $10^{-3}\,\mathrm{s}^{-1}$. A 25 mm gauge length extensometer was used to record the extension of the central part of the 32 mm gauge length specimens. The sense of the displacement of monotonic tests was reversed after reaching different plastic strain levels in the range $0.09 \leq \bar{\varepsilon} \leq 0.27$ until full unloading using the same displacement rate (Fig. 1). Details of the material, thermal treatments and tests can be found elsewhere [12]. X-ray diffraction measurements and scanning electron microscopy electron back-scatter diffraction as well as transmission electron microscopy examination showed that this steel exclusively deforms by dislocation glide plus deformation twinning. Its deformation twins are lamellae of nearly constant nanometric thickness of very high aspect ratio (Fig. 2). The unloading stress–strain curves were analyzed to assess the internal stresses developed by the forward strain according to Eq. (4) for different fixed values of the offset strain