

On the origin of dynamic strain aging in twinning-induced plasticity steels

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Abstract

Room temperature dynamic strain aging (DSA) is often observed from the beginning of plastic deformation in high Mn Fe–18% Mn–0.6% C and Fe–22% Mn–0.6% C twinning-induced plasticity steels. Although the phenomenon is in many cases very pronounced, there have up to now been no attempts to explain the phenomenon of room temperature DSA in TWIP steel containing solute C. It is proposed that DSA occurs by a single diffusive jump of the C atom of the point defect complex in the stacking fault region. DSA is only observed when the C atom reorientation time is smaller than the residence time of the stacking fault at the location of the point defect complex. The latter interaction can explain the DSA-suppressing effect of Al, which increases the stacking fault energy.

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1. Introduction

Dynamic strain aging (DSA) is observed in high Mn twinning-induced plasticity (TWIP) steels containing solute C. The contribution of DSA to the flow stress is clearly visible in many published stress–strain curves of TWIP steels, which often show pronounced type A serrations. In the case of Fe–18% Mn–0.6% C and Fe–22% Mn–0.6% C TWIP steels the deformation is entirely due to the formation and motion of localized Portevin–Le Châtelier (PLC) deformation bands. The properties of PLC bands have been studied in detail [1–3]. It is noteworthy that most models for the mechanical properties of TWIP steels have hitherto not taken into account the pronounced localization associated with DSA [4,5]. Although the phenomenon is in many cases very pronounced, there have up to now been no attempts to explain the phenomenon of room temperature DSA in TWIP steels. The phenomenon is also very sensitive to the addition of small amounts of Al.

Alloying additions of 1.5 mass% Al suppress DSA and effectively increase the critical strain for the occurrence of DSA [6].

The origin of DSA occurrence in high Mn TWIP steels has not yet been clarified. DSA is observed in well-annealed TWIP steels with a low equilibrium concentration of vacancies. In addition, the DSA in Fe–18% Mn–0.6% C TWIP steels starts directly with plastic yielding of the material. The presence of dislocation-generated vacancies and their possible effect on DSA can therefore be discarded. In addition, DSA is observed at room temperature, i.e. at a temperature at which the diffusivities of solutes, including that of solute interstitial C, is extremely low and diffusion of C to temporarily arrested dislocations is not possible. Even if local heating due to the enhanced strain rate within in a PLC band is taken into consideration, the C diffusion distance is insufficient at room temperature to allow for a DSA mechanism based on large diffusion distances.

Several models for DSA have been proposed. In the original Cottrell model [7] it was assumed that solute atoms are mobile and travel along with moving dislocations above a certain temperature. At low dislocation velocities

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their presence leads to an increase in the friction force. At high deformation rates the dislocations lose their solute atmosphere and this leads to a lower resolved shear stress for dislocation glide. As a consequence, DSA leads to a negative strain rate sensitivity. The Cottrell model is physically not realistic. It does not take into account that dislocation motion is “jerky”, i.e. dislocations frequently stop, having to overcome barriers, and travel rapidly between obstacles.

In the DSA model of van den Beukel [8] solute clustering takes place at temporarily arrested dislocations. The dislocations are forced to wait in these locations for the stress to increase so that they can be unpinned. The pinning force exerted by the solute clusters is a function of their size, which depends on the solute diffusivity and the waiting time of the dislocation. The van den Beukel model is also not realistic in the case of TWIP steels as it does not take into account that the bulk diffusion is too limited to lead to clustering. In particular, substitutional alloying element diffusion requires a large non-equilibrium vacancy excess.

This latter limitation has been addressed by Mulford and Kocks [9]. They suggested that pipe diffusion of solutes along the dislocation core could result in a high diffusivity without the need for extra vacancies. Calculations have shown, however, that pipe diffusion is also too slow to result in a significant clustering at arrested dislocations.

An alternative model has been proposed by Picu et al. [10,11]. They have suggested that a pre-existing inhomogeneous solute distribution could cause DSA and influence the strain rate sensitivity. In this alternative DSA mechanism small solute clusters, which exist at room temperature, interact with forest dislocations. Cluster formation is by lattice diffusion, so the resulting cluster size depends on the dislocation arrest time. The average size of clusters on new forest dislocations is therefore smaller for larger strain rates.

According to Rose and Glover [12] the presence of interstitial C is essential in order to observe DSA in austenitic fcc alloys. They attribute DSA to the reorientation of C–vacancy complexes in the stress field of dislocations. They point to the low activation energy for DSA (83 kJ mol^{-1} , 0.86 eV) to argue that reorientation is due to the movement of the vacancy rather than the C. The activation energy for vacancy diffusion is approximately one-third that of self-diffusion (280 kJ mol^{-1} , 2.90 eV).

In the present contribution internal friction (IF) and Mössbauer spectroscopy (MS) were used to characterize the types and properties of point defect complexes involving carbon atoms in TWIP steels. On the basis of the experimental observations it is argued that whereas isolated C atoms cannot interact with dislocations, point defect complexes involving carbon atoms can interact with partial dislocations and, more importantly, with stacking faults in TWIP steels. The importance of the interaction between the point defect complex and the stacking fault can explain the DSA-suppressing effect of Al, an alloying element which increases the stacking fault energy (SFE) of TWIP

steels. It is proposed that DSA occurs by a single diffusive hop of the C atom in the point defect complex in the stacking fault region. DSA is only observed when the C atom rearrangement time is shorter than the residence time of the stacking fault at the location of the point defect complex. The model is similar to that proposed recently by Curtin et al. [13] to explain the occurrence of room temperature DSA in Al alloys.

2. Experimental

Austenitic TWIP steels with the compositions Fe–18% Mn–0.6% C and Fe–18% Mn–1.5% Al–0.6% C (in mass%) were used in the present study (here termed FeMnC and FeMnAlC). The content of other alloying elements was very low, less than 100 ppm. Cold rolled sheets, 1.4 mm thick, were provided by POSCO. The materials preparation is discussed in detail in Chin et al. [14]. Both materials show a fully austenitic microstructure at room temperature and no occurrence of strain induced martensitic transformation during tensile deformation was observed. ASTM E8 standard samples of the annealed materials were machined along the rolling direction. Room temperature tensile tests were carried out at strain rates of 10^{-4} – 10^{-2} s^{-1} using a Zwick universal tensile testing machine.

IF measurements were carried out in free flexural vibration mode at the resonant frequency. The specimens for IF measurements, 80 mm in length and 25 mm in width, were tested in vacuo. The specimens were suspended between two thin thermocouples and were excited to oscillate at their resonant frequency by means of impulse excitation. The specimen vibration amplitude was detected by a laser vibrometer. The specimens were heated from $-100 \text{ }^\circ\text{C}$ to $+550 \text{ }^\circ\text{C}$ by infrared radiation at a heating rate of $3 \text{ }^\circ\text{C s}^{-1}$. IF measurements were carried out on as-annealed FeMnC and FeMnAlC TWIP steels.

Transmission MS was carried out at room temperature and at 18 K. As-annealed and 10% strained samples of the FeMnAlC TWIP steel were used. The MS spectra were recorded in constant acceleration mode using a ^{57}Co :Rh source. The MS spectra consisted of various components which were fitted assuming crystalline components using the NORMOS program [15].

3. Results

3.1. Strain localization

Fig. 1 shows the stress–strain curve for the FeMnAlC and FeMnC TWIP steels. Whereas there are multiple type A serrations on the curve for the FeMnC TWIP steel, the curve for the FeMnAlC TWIP steel is smooth over a large strain range. In the case of the FeMnC TWIP steel the serrations initiate at the start of plastic deformation. In contrast, the critical strain for DSA of the FeMnAlC TWIP steel is approximately 14.1 and 17.0% for strain rates of 10^{-4} and 10^{-2} s^{-1} , respectively.

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