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Strain partitioning in ultra-fine grained medium-manganese transformation induced plasticity steel

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ABSTRACT

A 7.1-Mn 0.1-C transformation-induced plasticity steel was intercritically annealed at 600 °C and 650 °C for 168 h. Ultra-fine-grained microstructures with annealing temperature dependent retained austenite fractions and tensile properties were produced. *in situ* neutron diffraction was used to investigate the change in tensile properties *via* measurement of phase fractions, elastic phase strains, and diffraction peak broadening during deformation. Austenite transformation to martensite controlled initial yielding in the 650 °C annealed steel and stress induced transformation was observed. In contrast, yielding after annealing at 600 °C was controlled by plastic deformation of ferrite, with austenite transformation initiating only after yield point elongation. The sequence of deformation between constituents was readily apparent in the lattice strain and peak width data. During deformation, compressive lattice strains were always developed in austenite, ferrite plastic deformation initiated around 700 MPa in both steels, and tensile stress was preferentially transferred to deformation-induced martensite. The development of compressive strains in austenite was related to constraint of the volume expansion during austenite transformation to martensite.

1. Introduction

Intercritical annealing of medium-manganese (Mn) (*i.e.* 5–10 wt pct. Mn), low carbon, transformation-induced plasticity (TRIP) steels is an effective methodology to make steels of interest to meet third generation advanced high strength steel (AHSS) properties goals [1,2]. During annealing in the two-phase ferrite and austenite region, enrichment of Mn to austenite from ferrite stabilizes austenite to room temperature on subsequent cooling [3,4]. The annealing temperature controls the relative Mn-enrichment of austenite and thereby determines the subsequent austenite stability during deformation. Previous work employing this processing methodology [5–14] has highlighted the potential to produce high tensile strength and ductility combinations through systematic variations in heat treatment methods, intercritical annealing temperature, and retained austenite content. The resulting properties clearly correlate with variations in austenite stability [3,13,14]; however, the fundamental interactions between phases during deformation and austenite transformation to martensite have received limited attention.

in situ neutron diffraction provides a method to directly monitor lattice plane spacings in multiple phases as a function of applied stress and/or strain during deformation, and thus is attractive as a method for investigating deformation. Small changes in interplanar spacing serve as microstructurally scaled internal 'strain gauges' to monitor deformation in crystalline materials. In multiphase materials, differences between material properties (*e.g.* elastic modulus, yield stress, or work hardening rate) of the constituent phases result in a divergence between the lattice strains of individual phases with deformation. Three distinct regimes of deformation in a two phase composite may be defined [15]: Stage 1 deformation where both phases deform reversibly, resulting in bulk linear and elastic loading; Stage 2 deformation is marked by the initiation of plastic flow in the lower strength constituent, the elastic lattice strain of each phase

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continues to increase, however, changes in the stress on each phase due to yielding of the 'soft' phase results in a decrease in the slope of the stress–lattice strain relation for the 'hard' phase and increase in the slope of the 'soft' phase; Stage 3 deformation occurs when the 'hard' phase also deforms plastically resulting in work hardening rate dependent stresses in each phase and corresponding changes in the lattice strains. Observations of stress loci for changes in the slope of the lattice strain–applied stress behavior yield valuable information about the sequence in which various constituents yield plastically and the influence of the flow strength of individual constituents on the mechanical properties of the multiphase microstructure [15,16].

TRIP steels present a special situation when considering lattice strain partitioning effects due to the dynamic transformation of austenite to martensite with deformation. As austenite is replaced by martensite, the physical force on austenite will decrease, however this decrease may not be proportionate to the change in austenite volume fraction due to the load redistribution to the hard martensite, and the resulting changes in the stress applied to austenite. Additionally, the large volume change (approximately 3 pct.) and lattice shear associated with the diffusion-less austenite to α' martensite transformation will also affect the elastic lattice strains [17,18]. Previous *in situ* diffraction studies of lattice strains in TRIP steels have highlighted results that are highly sensitive to the specific processing methodology, austenite morphology, and matrix microstructure studied [19,20]. Diverse experimental results are reported and vary from lattice strains being preferentially transferred to austenite as a 'hard' phase after the onset of plastic deformation [19,21–23] to a slight austenite relaxation after the onset of transformation [15,24].

In the present study, *in situ* neutron diffraction was performed on a 7.1 Mn-0.1C sheet steel to provide direct measurement of mechanical interactions between constituent phases by recording the phase-specific elastic lattice strains during uniaxial tensile deformation on samples heat treated to produce significantly different austenite stability conditions. The observed differences in macroscopic yielding and work hardening are related to changes in the austenite transformation mechanism in each alloy. The sequence of deformation between phases explains the work hardening behavior observed in ultra fine-grained Mn-TRIP steels.

2. Experimental methods and materials

The experimental steel with the composition of 0.099C–7.09Mn–0.13Si–0.031Al–0.008N (wt pct.), was the subject of a recent study considering systematic variations of tensile properties with changes in austenite fraction and stability [14]. The steel was cold rolled, intercritically annealed in the ferrite-austenite region at 600 °C and 650 °C for 168 h, and water quenched. These two temperatures were selected from the previous work to highlight material with displaying pronounced differences in austenite stability, initial yielding, work hardening behaviors [14] and transformation mechanism [3]. The long annealing cycle was used to provide sufficient time for diffusion to produce nearly equilibrium C and Mn concentrations in austenite and ferrite that stabilize austenite to room temperature on final cooling [3,14]. Tensile properties were measured using ASTM E-8 sub-sized samples with a 25.4 mm gauge length tested at a constant engineering strain rate of $5.74 \times 10^{-4}\ s^{-1}$ continuously deformed to failure at room temperature [25].

in situ neutron diffraction, performed on the SMARTS diffractometer [26] at the Lujan Center at Los Alamos National Laboratory. Measurements of phase fractions, elastic lattice strains, and diffraction peak width were made during tensile deformation. SMARTS has two detector banks oriented at $\pm 90^{\circ}$ to the incident beam; one detector collects data for crystal orientations in the direction normal to the specimen thickness (*i.e.* in the transverse direction) and the second in the plane of maximum tension (*i.e.* in the axial direction), shown schematically in Fig. 1 [26]. Tensile specimens for diffraction measurements were incrementally deformed: the actuator displacement paused, holding the sample at constant displacement, and diffraction patterns were recorded. Applied stresses for the *in situ* diffraction data presented here are engineering stress on the sample at the end of the hold for neutron diffraction measurement.

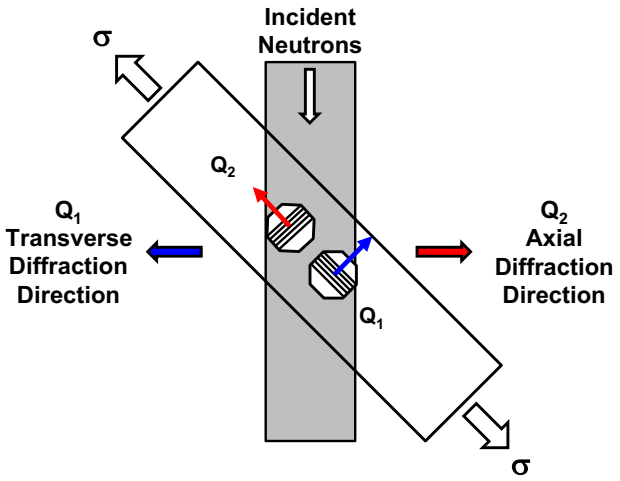

Fig. 1. Schematic of the SMARTS diffractometer highlighting the orientation of planes diffracting in the axial and transverse orientations [26].

Four phases were identified during the analysis of the diffraction data: thermodynamically stable body centered cubic (BCC) ferrite (α), metastable face centered cubic (FCC) austenite (γ), hexagonal close packed (HCP) epsilon (ε) martensite, and body centered tetragonal (BCT) alpha prime (α') martensite. Representative indexed diffraction patterns from the axial diffraction direction are shown in Fig. 2a and b for the 600 °C and 650 °C annealed steels, respectively. Data from the as-annealed samples and after successive increments of tensile strain are shown in Fig. 2. Note, the diffraction peaks corresponding to the ε martensite $\varepsilon_{\{101\}}$ planes were mislabeled previously [14] and are correctly indexed here. As the sample was deformed, metastable austenite transformed to ε-martensite and α'-martensite and the phase fractions of these constituents varied with strain. Whole pattern Rietveld analysis, performed with the GSAS software package, was employed to determine the austenite fraction at each strain increment using data from both diffraction directions [27,28]. The amount of ε martensite was calculated using only data collected from the axial diffraction direction, as ε martensite was not quantifiable in the transverse diffraction data. The amount of α' martensite was estimated by subtracting the sum of the amounts of austenite and ε martensite (measured with neutron diffraction) and the predicted intercritical ferrite amount (estimated with ThermoCalc software [29]) from the whole.

Single peak fitting using the Rawplot subroutine of GSAS was performed to measure the interplanar spacing as a function of applied stress and served to highlight representative orientation dependent lattice strains. A Gaussian function was fit to select peaks for each phase to determine the interplanar spacing and peak width at each increment of deformation. Elastic lattice strains for each set of diffraction planes (ε_{hkl}) were calculated using Eq. (1)

$$\varepsilon_{hkl} = (d^{\sigma}_{hkl} - d^{0}_{hkl})/d^{0}_{hkl} \quad (1)$$

d^{σ}_{hkl} is the $\{hkl\}$ interplanar spacing averaged over a set of grains with $\{hkl\}$ plane normals parallel to the diffraction vector measured