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The effect of morphology on the stability of retained austenite in a quenched and partitioned steel

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Synchrotron X-ray diffraction and transmission electron microscopy experiments are employed to investigate the mechanical stability of retained austenite in a quenching and partitioning steel. It is found that martensitic transformation takes place in high carbon blocky retained austenite at the onset of deformation, while low carbon film-like retained austenite, despite having a much lower carbon content, is stable at strains up to 12%. Two possible mechanisms are employed to explain these interesting findings.

Keywords: Q&P steel; Austenite stability; Morphology effect; Synchrotron XRD; TEM

Transformation-induced plasticity (TRIP)-assisted multiphase steels have been the constant focus of much research interest from both academia and industry due to their excellent combination of strength and ductility [1–3], which are essential in automotive applications for reducing a vehicle's weight. Such good mechanical properties are due to the high work hardening rate induced by the austenite-to-martensite transformation during deformation. The quantity and stability of the retained austenite in TRIP steels play important roles in determining their mechanical properties. The stability of the retained austenite can be affected by its chemical composition [4–12], size [13–19] and morphology [9,20,21]. It is a challenging task to distinguish which factor is the dominant one when a range of factors are varying. The present work intends to distinguish the roles of the carbon content and morphology of retained austenite on its stability in a new TRIP steel grade.

The chemical composition of the present steel is Fe–0.22C–1.40Si–1.8Mn (in wt.%). A quenching and partitioning (Q&P) process [3,22–24] was employed to produce a multiphase microstructure containing a considerable amount of retained austenite. The steel was austenitized at 860 °C for 5 min, then cooled slowly at a rate of 5 °C s^{-1} to ∼725 °C, at which temperature two phases (ferrite and austenite) are present. It is worth noting that this slow cooling was deliberately designed to allow not only for the formation of proeutectoid ferrite but also for establishing a non-uniform carbon distribution in the retained austenite. After cooling to ∼725 °C, the steel was rapidly quenched to 280 °C, reheated and held at a partitioning temperature of 350 °C for 10 s, then quenched again to room temperature. The resulting microstructure contained a significant amount of retained austenite.

Tensile samples with a gauge length of 15 mm, width of 5 mm and thickness of 1.2 mm were machined from a 1.2 mm thick steel sheet. The tensile axis was selected along the rolling direction. Interrupted tensile tests, stopped at different engineering strains (0, 2, 4, 6, 12 and 24% (fracture strain)), were performed at a quasi-static strain rate at room temperature. After the tensile tests, the samples were subjected to synchrotron X-ray diffraction (XRD) experiments to measure the volume fraction of retained austenite. The synchrotron XRD experiments were carried out at the beamline no. B13 at Shanghai Synchrotron Radiation Facility. The energy of the monochromatic X-ray beam was 10 keV, corresponding to a wavelength of 0.1239 nm. The reflection mode was applied. An image plate detector was used to record the two-dimensional patterns of diffraction

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rings. The distance from the sample to the detector was 200 mm. The pixel size was 100 μm × 100 μm and the detector size was 3450 × 3450 pixels. The beam size was 0.5 mm × 1 mm. Transmission electron microscopy (TEM) experiments were carried out on samples strained at 0, 2, 12 and 24% on a JOEL 2100 microscope operated at an accelerating voltage of 200 kV. TEM samples were prepared using a twin-jet technique in a 10% perchloric acid +90% acetic acid solution.

The volume fractions of retained austenite as a function of engineering strain were calculated by integrating diffracted intensities of corresponding Bragg reflections and the results are plotted in Figure 1a. The engineering stress–strain curve of the sample strained to fracture is shown in Figure 1b. The non-deformed microstructure contained a considerable amount of retained austenite (12%) which, however, showed a significant decrease at the early stage of deformation. At 6% of engineering strain, one-third of the retained austenite had transformed into martensite.

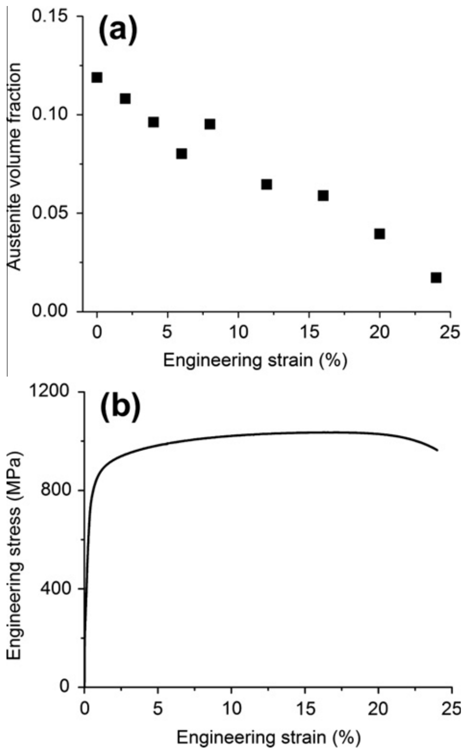

Figure 1. (a) Evolution of the retained austenite volume fraction with engineering strain; and (b) engineering stress–strain curve.

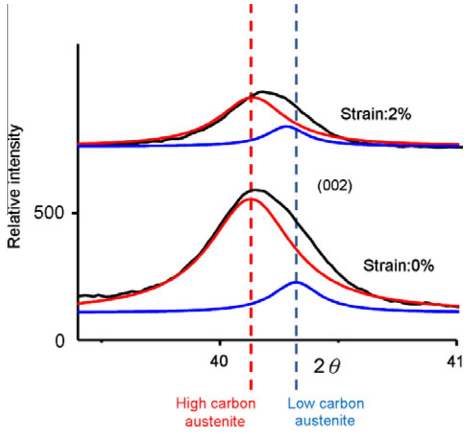

Figure 2. (002) austenite peaks at 0 and 2% strain. Black curves are the original experimental data showing asymmetry characteristics. Two sub-peaks (red and blue curves) can be derived from the black curves. The red curve represents high carbon austenite and the blue curve corresponds to low carbon austenite. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

As demonstrated in Figure 2, the austenite (002) peaks at 0 and 2% strain both showed asymmetry (black curves) and thus could be deconvoluted. A Gaussian multi-peaks fitting method was employed. Due to the specific Q&P treatment process [3,22–24], it is reasonable to assume that deconvolution can be performed by choosing two sub-peaks. The peak (black curve) is divided into two sub-peaks (blue and red curves), indicating that the retained austenite can be divided into two different subgroups: one with a smaller lattice parameter (low carbon content, blue curve) and the other with a larger lattice parameter (high carbon content, red curve). The respective lattice parameters can be calculated by the respective 2θ values shown in Figure 2. In TRIP steels, the effect of carbon on the lattice parameter is more profound than other alloying elements (e.g. Si, Mn) such that the lattice parameter a of retained austenite can be approximately expressed as $a = 0.3573 + 0.00075$ at.% C (nm) [25,26]. Using this relationship, the carbon contents of the high and low carbon retained austenite phases were estimated as 1.14 and 0.64 wt.%, respectively. Assuming that all alloying elements except carbon were homogeneously distributed during slow cooling, quenching and partitioning, it is possible to estimate the martensitic transformation start (Ms) temperature for both high and low carbon austenite by the following empirical equation [27]:

$$\mathrm{Ms}\ (^{\circ}\mathrm{C}) = 539 - 423\mathrm{C} - 30.4\mathrm{Mn} - 7.5\mathrm{Si} + 30\mathrm{Al}\ (\text{in wt.\%}) \tag{1}$$

From Eq. (1), the Ms temperatures for high and low carbon austenite are estimated as −8.4 and 203 °C, respectively, which indicates that the high carbon austenite should be much more stable than the low carbon austenite during deformation. However, it is surprising that the present synchrotron XRD experiments demonstrate that the low carbon austenite is more stable than the high carbon austenite. For instance, comparing the two austenite (002) peaks at 0 and 2% strains (Fig. 2), one can see that the volume fraction of the high carbon austenite (the area under the red curve) decreases more than that of the low carbon austenite (the area under the blue curve). This illustrates that the low carbon austenite, despite having lower chemical stability, appears to be more resistant against martensitic transformation. Therefore, one can conclude that the carbon content is not the main factor determining the stability of the retained austenite in the present TRIP steel, despite the difference in carbon content being large (1.14–0.64 wt.% = 0.5 wt.%). Other factors should be more important for the stability of retained austenite, which will be clarified by the following TEM investigations.