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Elevated-temperature mechanical stability and transformation behavior of retained austenite in a quenching and partitioning steel

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ABSTRACT

The mechanical stability and transformation behavior of both film and blocky retained austenite (RA) in a quenching and partitioning steel are investigated at 293 K, 423 K and 573 K with X-ray diffraction measurements and transmission electron microscopy. Blocky RA both completely and incompletely transforms to twinned martensite during deformation at 293 K and 423 K, respectively, and completely transforms to lath martensite during deformation at 573 K. At 293 K and 423 K, only the film RA with widths larger than ~70 nm transforms to twinned martensite. However, film RA incompletely transforms to lath martensite at 573 K. Hence, RA transformation is non-monotonic with temperature. Significant carbide formation at 573 K, and therefore less carbon to stabilize RA, overcomes the increase in austenite stability due to the decrease in the temperature-dependent chemical driving force for the martensite transformation.

1. Introduction

The TRIP-effect (TRIP=transformation induced plasticity) is the diffusionless shear transformation of metastable retained austenite (FCC) into martensite (BCT), denoted by $\gamma \rightarrow \alpha'$, with plastic straining [1]. The anisotropic carbon distribution from the shear transformation leads to a tetragonal structure, a very small grain size, and high dislocation density, all characteristics of martensite. Exceptional mechanical properties (ductility and strength) result since necking and fracture are delayed. Distinct from the first generation, low alloyed TRIP steels, the TRIP-assisted steels have higher Mn content (~2 wt% or greater), e.g. high Mn TRIP steels [2], medium Mn TRIP steels [3], quenching and partitioning steels (also known as QP steels) [4], possess higher strength levels and contain a higher retained austenite (RA) volume fraction (V_{RA}).

Since the RA in TRIP steels is carbon enriched, the martensite start temperature (M_s), at which the RA transforms to martensite during cooling, is lower than room temperature (RT) [5]. When the temperature goes below M_s (e.g. <233 K, for a QP steel [5]), the RA is not thermally stable and transforms to martensite without plastic straining. At temperatures above M_s, the RA is not mechanically stable, and may transform to martensite with plastic straining: we refer to this as deformation induced martensite transformation (DIMT). The mechanical stability of RA, usually measured in tension as the retained austenite volume fraction, V_{RA}, with strain, is of considerable importance for both alloy design and material end-use. The amount of transformed RA generally increases with increasing plastic strains and the transformation kinetics of RA is dependent on the strain path, e.g. it can vary from tension to compression [6]. The transformation of RA with plastic straining is also affected by the morphology of the RA grains [7,8] and their surrounding matrix [9]. Xiong et al. [10] reported that there are two types of RA in the QP980 steel: high carbon content (~1.14 wt%) blocky RA surrounded by ferrite or ferrite/martensite, and lower carbon content (~0.64 wt%) film RA between martensite laths. Note that atom probe tomography measurements in Poling et al. [11] gave 1.28 (± 0.006) wt% carbon in blocky RA in the QP980 steel from a different heat of QP980 material. Xiong et al. [10] found that the film RA transformed at larger strains and hence was more mechanically stable than the blocky RA, which transformed at smaller strains during RT, quasi-static deformation. This is counterintuitive from the standpoint that the carbon concentration in the film RA was less than that in the blocky RA. As discussed by Xiong et al. [10], it is the strain level of the lath martensite surrounding the film RA that is much lower than that of the relatively softer blocky ferrite next to the blocky RA. Similar behavior was found by Park et al. [12] in a TRIP steel produced with intercritical annealing and isothermal bainitic

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transformation. Zhao et al. [13] reported that RA transforms in a discontinuous way when the plastic incompatibility between the ferrite and martensite which surround RA is significant at room temperature.

It is worth noting that the product of DIMT is dependent upon the carbon concentration of the RA. For instance, RA with a carbon concentration smaller than 0.4 wt% transforms to twinned martensite and lath martensite, respectively [14,15]. The maximum temperature at which DIMT occurs is M_d ($> M_s$). Generally, M_d is above RT for TRIP steels. As the diffusion rate of a carbon atom increases with increasing temperature and the carbon concentration in RA becomes super saturated, ε-carbide (M_3C, where M indicates Fe or Mn) can be precipitated from RA during tempering at 573 K in CMnSiAl TRIP steels, even without plastic deformation, according to Park et al. [16].

The local temperature of steel sheet can rise as high as ~550 K due to plastic work in a quasi-adiabatic process [17], e.g. an automotive stamping process. In general, the higher the strength of the steel sheet, the higher will be the local temperature during forming [17]. In addition, it has been reported that under hood temperatures can reach 423 K under normal operation and paint bake temperatures can achieve 463 K [18]. Despite these observations, there is minimal information in the literature on austenite mechanical stability, as V_{RA} vs. strain, and the RA transformation mechanism at temperatures above RT for TRIP-assisted steels, in particular, QP steels.

This paper presents an experimental investigation of the elevated temperature mechanical stability (V_{RA} vs. tensile strain) and transformation behavior of a three-phase TRIP-assisted steel. For this purpose, 980-grade QP steel was quasi-statically deformed at three temperatures, viz. 293 K (i.e. RT), 423 K and 573 K. This temperature range covers much of that for quasi-adiabatic plastic deformation in automotive stamping processes and paint baking. After tensile specimens deformed at 423 K and 573 K were cooled to RT, the V_{RA} was measured by X-ray diffraction and the resulting microstructures were investigated with transmission electron microscopy to determine how RA transforms at elevated temperatures. The effects of temperature and RA morphology on the RA mechanical stability and transformation behavior are discussed in detail.

2. Experimental details

2.1. Material

The material investigated in this study, hereinafter referred to as QP980, is a 1.2 mm thick, 980-grade TRIP-assisted steel which was subject to a two-step quenching and partitioning (QP) heat treatment [5]. The QP heat treatment involves heating to a temperature above AC_3 (the austenitization temperature), followed by an immediate quench to a temperature between the start and finish temperatures of martensite transformation. After quenching, the material is heated again to a partitioning temperature (PT) which is below the martensite start temperature and maintained at the PT for a while to allow greater carbon enrichment of the remaining austenite while minimizing carbide precipitation. The initial microstructural constituents of the QP980 steel is 35 vol% ferrite, 53 vol% martensite, ~12 vol% retained austenite (RA), which has been reported in a previous study [20]. The chemical composition of the QP980 steel is listed in Table 1.

Table 1
Chemical composition (wt%) of the QP980 steel.

C	Si	Mn	Al	P	N	S
0.195	1.38	1.90	0.036	0.009	0.0045	0.0005

Nine-QP980 tensile specimens following ASTM E8/E8M-11 [19] were strained to fracture at 293 K, 423 K and 573 K (three specimens at each temperature) in an Instron 5568 universal testing machine equipped with an environmental chamber having a transparent glass window. The crosshead speed was set to 10 mm/min at each temperature, corresponding to a nominal quasi-static strain rate $5 \times 10^{-3}\ s^{-1}$. Digital images of the specimen gauge section, captured at 20 frames/s during testing, were post-processed for strain fields via digital image correlation (DIC) following each tensile test. Procedures related to DIC preparation were detailed in a previous study [20] where the glass window of the environmental chamber and temperature were found to have negligible effects on the accuracy of DIC strain measurements. After the specimens tested at the two elevated temperatures were air cooled to RT, a 10 mm (along the tensile axis direction) × 6 mm (along the width direction) sample was cut by wire electrical discharge machining within the uniformly deformed section of each (i.e. outside of the neck, ~15 mm away from the fracture). The neck region was avoided since the strain distribution within the neck was nonuniform. One-half of each sample (5 mm × 6 mm) was used for X-ray diffraction measurement of V_{RA}, and the other half was used to investigate microstructural changes associated with the RA transformation with TEM. Digital image correlation post processing confirmed that no necking occurred on these extracted samples, and the average true major strains (ε_1) along the tensile axis of three samples deformed at 293 K (Sample A), 423 K (Sample B) and 573 K (Sample C) were 0.129, 0.107 and 0.157, respectively.

2.2. X-ray diffraction measurements

Measurement of V_{RA} as a function of deformation temperature [21] is critical not only for alloy development but also for the construction of finite element models of sheet forming and component performance that rely upon accurate material data [22]. In the present study, X-ray diffraction (XRD) measurements in reflection mode were conducted at 45 keV, 40 mA with a Cu target on a Bruker D8 DISCOVER AXS X-ray diffractometer, with a 1.54052 Å wavelength and 2 mm × 2 mm spot size. Note that the Bruker diffractometer has been used in other studies of TRIP steels to measure V_{RA} [23]. Each XRD sample was ground and polished with a cloth and diamond paste (6 μm, 3 μm, 1 μm), and then electropolished. Samples were scanned at a step size of 0.01° in the 2θ range from 40° to 105° with a dwell time of 10 s. V_{RA} was determined from the integrated intensity of austenite peaks $(111)_\gamma$, $(200)_\gamma$, $(220)_\gamma$, $(311)_\gamma$, and ferrite peaks $(110)_\alpha$, $(200)_\alpha$, $(211)_\alpha$, $(311)_\alpha$, following YB/T 5338-2006 [24].

2.3. TEM observations

A transmission electron microscopy (TEM) study of the microstructures of the samples extracted from the uniformly deformed section of the tensile specimens deformed at the three temperatures was performed with a HITACHI 800 electron microscope. Of special interest here was the effect of deformation at the two elevated temperatures on the stability of both blocky and film RA [5,10]. Thin foils were prepared from the samples by grinding to a nominal thickness of ~50 μm. Discs of 3 mm diameter were punched from the foil and ion milled to a condition suitable for TEM analysis.