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Global and local deformation behavior and mechanical properties of individual phases in a quenched and partitioned steel

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ABSTRACT

Third generation advanced high strength steels produced via quenching and partitioning (Q&P) treatment are receiving increased attention. A 0.25C–3Mn–1.5Si–0.023 Al steel was subjected to Q&P processing (with varying partitioning temperature and time) resulting in the formation of complex multi-phase microstructures. The effect of Q&P parameters on the microstructure and morphology of microconstituents was analyzed. Mechanical properties of the material and of its individual microconstituents were studied via tensile testing and nanoindentation on individual microconstituents, which were identified *a priori* by electron back-scattered diffraction analysis. Special attention is paid to the effect of the morphology of retained austenite on its transformation stability. *In situ* tensile tests and digital image correlation analysis were performed to study deformation behavior of the Q&P processed steel at the micro-scale with respect to the local microstructure. The effect of local microstructure and properties of individual phases on the degree of strain partitioning is discussed.

1. Introduction

Over the past decade, with increasing demands for weight reduction in the automotive industry a significant research effort has been directed towards the development of several types of advanced high-strength steels (AHSS) [1]. The main objective of the development of AHSS has been to obtain a good combination of strength and ductility. AHSS steels are usually multiphase and, thus the combination of different phases (such as ferrite, martensite, retained austenite (RA), etc.) leads to unique mechanical properties [1,2]. The high strength martensitic or bainitic constituents contribute to a simultaneous increase of strength and toughness, whereas RA provides an improvement of strength and ductility through the strain-induced transformation of the metastable austenite to hard martensite, i.e., the transformation induced plasticity (TRIP) effect [2,3]. This microstructural heterogeneity also results in significant stress and strain partitioning among the phases during plastic deformation.

A novel thermal processing route, "quenching and partitioning" (Q&P), was proposed by Speer et al. [4] in 2003. It was used to develop the third generation of AHSS, based on a new understanding of carbon diffusion from martensite into austenite for the production of martensitic microstructures containing enhanced levels of retained austenite [2–4]. The Q&P process consists of a two-step heat treatment. First, after heating in order to obtain a fully austenitic or intercritical microstructure, the steel is quenched to a suitable pre-determined temperature (*QT*) below the martensite start (M_s) but above the martensite finish (M_f) temperatures to form a pre-determined amount of martensite. Then, the steel is either held at this quenching temperature or brought to a higher, partitioning temperature (*PT*), where the untransformed austenite enriches with carbon through carbon depletion of the supersaturated martensite [5,6]. In this way, a complex microstructure which contains metastable RA and martensite (together with, in some cases, ferrite and bainite) is obtained after final quenching to room temperature [7,8].

In AHSS multiphase steels, including Q&P steels, the deformation behavior and the mechanical properties in bulk strongly depend on mechanical properties, morphology and spatial distribution of individual phases [7–10]. Strain partitioning between phases during plastic deformation is greatly affected by the

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difference in strength of the individual microconstituents [11,12]. However, investigations of the relation between the local microstructure and local plastic deformation in Q&P steels have just begun. Very recently in [13], de Diego-Calderón et al. have reported the effect of the microstructure on the local mechanical behavior of an intercritically annealed Q&P steel. A strong partitioning of plastic strain between the phases (microconstituents) was observed during deformation. It was found that the plastic strain accommodated by a given phase or constituent is inversely proportional to its nanohardness [13]. It should be noted that transformation of the metastable austenite may also lead to increased strain hardening in these AHSS if the TRIP effect is operating [2,14,15], so the volume fraction and stability of RA strongly affect mechanical properties of Q&P steels [14,16–18]. On the other hand, the transformation stability of RA is determined by many factors, among others the carbon concentration in the RA grains [19,20], the RA grain size [21], the constraining effect of the surrounding phases [22] and the morphology of the RA [14]. It can be concluded that fundamental understanding of the local plastic deformation behavior of Q&P steels could lead to further improvement of their mechanical properties via intelligent microstructural design. Therefore, the objective of the present work is to relate the micromechanics of deformation of a Q&P steel with its phase composition, Q&P parameters and mechanical properties of its individual phases.

2. Experimental

2.1. Material and processing

The chemical composition of the studied steel is presented in Table 1. The material was cast in a laboratory vacuum induction furnace. After casting, the steel slabs were hot rolled to a final thickness of 2.5 mm, accelerated cooled by water jets to 600 °C and transferred to a furnace for coiling simulations at 560 °C. The hot rolled plates were pickled and cold rolled to a thickness of 1 mm imposing a total reduction in thickness of 60%. The obtained strips were cut perpendicular to the rolling direction and subsequently subjected to Q&P heat treatment cycles in the thermomechanical simulator Gleeble™ 3500. The specimens were heated to 850 °C for full austenitization and quenched to 244 °C at quenching rate of 20 °C/s to obtain microstructures consisting of martensite and austenite. Then the samples were reheated with a heating rate of 10 °C/s, kept isothermally at 300 °C or 400 °C for 100 s or 500 s for partitioning and quenched with a rate of 20 °C/s. Data for the Q&P parameters are presented in Table 2, where the indication of the samples is also shown.

2.2. Microstructural characterization

Specimens for microstructural characterization were ground and polished to a mirror-like surface using standard metallographic techniques. For scanning electron microscopy characterization, specimens were etched with 2 vol% HNO_3 in ethanol (nital 2%) solution. Examination of the microstructure was performed using a scanning electron microscope (SEM) EVO MA15 operating at an accelerating voltage of 20 kV. Specimens for EBSD analysis were prepared using standard metallographic techniques with final polishing with OP-U for 20 min. Orientation imaging microscopy (OIM) studies were performed using a FEI Quanta™ 450 FEG-SEM equipped with a Hikari detector controlled by the EDAX-TSL OIM-Data Collection (version 6.2®) software. The data were acquired at an accelerating voltage of 20 kV, a working distance of 16 mm, a tilt angle of 70° and a step size of 40 nm. The orientation data were post-processed with TSL-OIM Analysis 6.2© software.

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

C	Mn	Si	Al	Cr
0.25	3	1.5	0.023	0.015

Table 2
Q&P processing parameters applied to the studied steel grade.

Specimen	Quenching temperature (QT) (°C)	Partitioning temperature (PT) (°C)	Partitioning time (Pt) (s)
QP4-100	244	400	100
QP3-500	244	300	500
QP4-500	244	400	500

The volume fractions of RA and its average carbon content at room temperature were measured by X-ray diffraction (XRD) experiments performed on a Siemens Kristalloflex D5000 diffractometer equipped with a Mo-Kα source operating at 40 kV and 40 mA. A 2θ-range of 25–45° was scanned using a step size of 0.01°, dwell-time of 2 s and a rotation speed of 15 rpm. The data were post-processed by subtracting the background radiation and $K\alpha_2$ influence. The volume fractions of RA were determined by the Cullity formula [23], and the austenite lattice parameter a_γ was determined from the extrapolation function of the lattice parameter vs. $\cos^2(\theta)/\sin(\theta)$ of the (200), (220) and (311) austenite peaks. The carbon concentration X_C was obtained according the procedure proposed in [24,25] where the link between the lattice parameter of the retained austenite is presented as

$$a_\gamma = 0.3556 + 0.00453X_C + 0.000095X_{Mn} + 0.00056X_{Al}, \quad (1)$$

where a_γ is the austenite lattice parameter in nm and X_C, X_{Mn} and X_{Al} are the concentrations of carbon, manganese and aluminum in austenite in wt%. Volume fractions of tempered martensite (TM) and untempered martensite (UM, which is often referred as fresh martensite) were estimated from the OIM images using the free ImageJ software [26].

2.3. Nanoindentation and atomic force microscopy studies

Nanoindentation tests were performed on a Hysitron TI950 Triboindenter using a Berkovich tip on square areas having a size of $\sim 12 \times 12\ \mu m^2$, which were *a priori* analyzed by EBSD (as described in Section 2.2). Five areas were subjected to testing for each material's condition. In order to target specific phases within well-defined grains, these square areas were scanned, prior to nanoindentation, using the scanning probe microscopy (SPM) mode of the instrument. Nanoindentation tests were carried out in displacement control mode, at a constant strain rate ($\dot{\varepsilon} = \dot{h}/h$) of 0.07 s^{-1}, where h is the penetration depth and $\dot{h}$ the penetration rate of indenter. At least 20 indents were performed on each phase, at an imposed maximum depth of 100 nm. At least 8 indentation tests were found to be located strictly within each individual microconstituent with good quality load–displacement depth curves. The nanohardness was determined from the analysis of the load–displacement curves using the Oliver and Pharr method [27]. Atomic force microscopy (AFM) topography scans were performed in contact mode on selected residual imprints using a Park XE150 instrument. Surfaces for nanoindentation had a low roughness (≤ 5 nm), suitable for performing nanoindentation measurements at the prescribed maximum depth of 100 nm.