



## Induced martensitic transformation during tensile test in nanostructured bainitic steels

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### ABSTRACT

Retained austenite in nanostructured bainite is able to undergo mechanically induced martensitic transformation. However, the link between transformation and deformation mechanisms involved makes difficult the understanding of the process. In this work, a model has been developed to assess the effect of the external stress itself on the martensite phase transformation. In addition, after a detailed initial microstructural characterization, the martensite fraction evolution during tensile deformation has been obtained by means of X-ray diffraction analyses after interrupted tensile tests in several nanostructured bainitic steels. Experimental results have been compared to the outputs of the model, as a reference. They suggest that stress partitioning between phases upon tensile deformation is promoted by isothermal transformation at lower temperatures.

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### 1. Introduction

TRIP-assisted multiphase steels are advanced microstructures which present an improved balance of strength and ductility thanks to the so-called TRIP-effect, i.e. transformation induced plasticity (TRIP) of the retained austenite into martensite. These microstructures normally consist of a soft matrix of allotriomorphic ferrite and a dispersed microconstitute, e.g., bainite, consisting of bainitic ferrite plus retained austenite [1]. Apart from the TRIP effect, other mechanisms are involved in these steels during tensile deformation, including the well-known Hall-Petch strengthening mechanism, and also the composite-type of strengthening plays a big role. The coupling between plasticity and the mechanically induced transformation is thus complex, often without a clear distinction between cause and effect [2]. There is a strong relation between strain hardening and TRIP effect, but it is difficult to know the extent at which the austenite evolution occurs as a consequence of the stress-strain behaviour, and how this transformation affects the work hardening mode in return.

Two different martensitic transformation mechanisms have been reported in TRIP steels: stress-assisted and strain-assisted TRIP effect. In the first situation, stress-assisted, martensitic

nucleation takes place on the same heterogeneous sites responsible for the transformation on cooling. In this case, kinetics of transformation governs kinetics of macroscopic deformation. However, in the other case, strain-assisted TRIP effect, martensitic transformation takes place thanks to the new nucleation sites being created, shear-band intersections; that is, prior plastic strain is necessary to trigger TRIP under this condition. Regardless of the TRIP mode, stress- or strain-assisted, the presence of an external mechanical stress has an effect on the total driving force for martensitic transformation,  $|\Delta G^{y-\alpha}|$ , which must include a mechanical term,  $\Delta G_{\text{mech}}$ , dependant on that stress [3,6]. There is some controversy on whether martensitic transformation occurs either by a stress-assisted or by a strain assisted mode depending on the microstructure [3].

Recently, many works on TRIP-assisted steels have been devoted to develop constitutive models, normally based on finite element analyses multiscale simulations, for the formulation of the flow strength of the evolving multi-phase composite. For this purpose, the knowledge of the transformation kinetics law describing the evolution of the volume fraction of austenite is compulsory, which is addressed by implementing different models [4–9]. The complexity arises from the fact that the stability of the retained austenite depends on many factors, as its chemical composition, morphology and size. A low stability for retained austenite can be ascribed to its low carbon content as well as to a relatively large grain size [10]. Moreover, stability is highly affected

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by the relative mechanical properties of the austenite and the surrounding phases, and the corresponding stress partitioning between them [5,11–13].

In this work, a set of microstructures belonging to a new generation of steels, nanostructured bainite, have been studied. The presence of retained austenite and the occurrence of TRIP effect in these microstructures is well-known [14,15]. However, the major phase of these nanostructured steels is not allotriomorphic ferrite, as in conventional TRIP-assisted steels, which is absent, but bainitic ferrite. Bainitic ferrite presents, instead, a high strength due to its nanometric size, a high dislocation content, and the tetragonality associated to its C supersaturation [16–18]. Thus, the deformation mechanisms of these microstructures are thought to differ from those of the conventional TRIP-assisted steels.

It is not the purpose of this work to design a constitutive model for nanostructured bainite. Instead, a model will be implemented to isolate and assess the effect that mechanical stress alone has on the martensitic transformation start temperature,  $M_s$ , and thus, on the martensitic transformation evolution.

## 2. Materials and experimental procedure

A total of three steels have been used for this work, and their chemical compositions are given in Table 1. The heat treatment consisted in a first stage of austenitization, a subsequently cooling down to a temperature between the bainitic start temperature ( $B_s$ ) and the martensitic start temperature ( $M_s$ ), an isothermal holding at that temperature for the bainitic transformation, and finally an accelerated cooling down to room temperature. As shown in Table 2, for each steel two isothermal temperatures were selected to assess differences in the nanostructured bainite features and its performance.

Tensile tests were performed at room temperature in specimens of 8 mm diameter and 20 mm of gauge length at a deformation rate of  $0.004 \text{ s}^{-1}$ . The load-displacement data during tests was obtained from an extensometer fitted to electronic equipment. Apart from tensile tests performed until fracture, some tensile tests were intentionally interrupted at the uniform deformation region in order to track the austenite fraction evolution. For that purpose, selected cross-sections of the gauge length, perpendicular to tensile direction, were extracted.

The microstructure was observed by secondary electron scanning electron microscopy (SE-SEM). Metallographic samples were cut, ground and polished following the standard procedures, including a final step of polishing with colloidal silica suspension. A 2% Nital etching solution was used to reveal the phases. Scanning electron microscopy (SEM) observation was carried out on a JEOL JSM-6500F field emission gun scanning electron microscope (SEM-FEG) operating at 10 kV.

The determination of the fraction of retained austenite ( $V_{F_\gamma}$ ) of undeformed and deformed samples was achieved by means of X-ray diffraction (XRD) analyses. Specimens were prepared by a standard grinding and polishing procedure, finishing with a step of polishing with  $1 \mu\text{m}$  diamond paste. Several cycles of etching and polishing were applied in order to remove the deformed layer.

**Table 1**  
Chemical composition of the studied steels.

	Chemical composition [wt%]								
	C	Si	Mn	Ni	Mo	Cr	V	Cu	Al
Steel 1	1.0	1.50	0.74	0.12	0.03	0.97	0.00	0.17	0.025
Steel 2	0.6	1.67	1.32	0.20	0.15	1.73	0.12	0.18	0.03
Steel 3	0.6	2.5	1.32	0.20	0.15	1.73	0.12	0.18	0.03

**Table 2**  
Heat treatment settings of the studied samples.

	Samples	Bainitic transformation	
		Temperature [ $^{\circ}\text{C}$ ]	Time [h]
Steel 1	1C1.5Si_250	250	16
	1C1.5Si_220	220	16
Steel 2	0.6C1.5Si_250	250	16
	0.6C1.5Si_220	220	22
Steel 3	0.6C2.5Si_250	250	16
	0.6C2.5Si_220	220	24

XRD measurements were performed with a Bruker AXS D8 diffractometer equipped with a Co X-ray tube, Goebel mirror optics and a LynxEye Linear Position Sensitive Detector for ultra-fast XRD measurements. A current of 30 mA and a voltage of 40 kV were employed as tube settings. Operational conditions were a  $2\theta$  range of  $35\text{--}135^{\circ}$  and a step size of  $0.01^{\circ}$ . The volume fraction of the retained austenite was calculated from the integrated intensities of (111), (002), (022) and (113) austenite peaks, and those of (011), (002) and (112) planes of ferrite, with the equation for the ratio of these experimental values to the normalization factors for peaks intensity (R) given in the ASTM E975-08 [19]. More details on the XRD experimental procedure can be found in ref. [16].

All the necessary thermodynamics calculations were performed using MtData in combination with the SGSOL-SGTE Solution database [20]. Crystallographic simulations have been coded using the free and open-source MTEX toolbox [21] running in MATLAB [22].

## 3. Results and discussion

### 3.1. Initial microstructure

For all heat treatments the microstructure consists of a mixture of two phases, bainitic ferrite and retained austenite. The addition of Si avoids the massive carbide precipitation [23]. Further details on the characterization of these microstructures can be found elsewhere [18,24,25]. Fig. 1(a) and (b) shows scanning electron micrographs of the microstructures corresponding to samples 1C1.5Si\_250 and 1C1.5Si\_220, respectively. The etched phase corresponds to bainitic ferrite plates and the higher relief to retained austenite, this latter present in two different morphologies, as films ( $\gamma_f$ ) and as blocks ( $\gamma_b$ ), Fig. 1. Due to the nature of the bainitic transformation and geometrical restrictions of the mentioned austenite features, films of austenite are more enriched in C in solid solution than blocks of austenite [10]. In these microstructures, C does not only lay at defect-free solid solution but also at defects such as twin and phase boundaries, clusters or dislocations [18,26].

The initial fractions of austenite,  $V_{F_0}$ , are summarized in Table 3. Considering each steel, for Steel 2 and Steel 3, the initial volume fraction of austenite is similar in both samples regardless of the treatment temperature. However, for Steel 1,  $V_{F_0}$  is considerably higher in the sample treated at lower temperature, 1C1.5Si\_220, than in 1C1.5Si\_250. This is due to an isothermal treatment time insufficient for the completion of the bainitic reaction in the case of 1C1.5Si\_220.

### 3.2. Mechanical properties

Table 3 gathers the results from tensile tests. YS increases as the treatment temperature does. In nanostructured bainitic steels, YS has been proven to depend mainly on the volume fraction of the

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