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# In situ synchrotron study on the interplay between martensite formation, texture evolution and load partitioning in low-alloyed TRIP steels

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ABSTRACT

We have studied the micromechanical behaviour of two low-alloyed multiphase TRIP steels with different aluminium contents by performing in situ high-energy X-ray diffraction experiments at a synchrotron source under increasing tensile stress levels. A detailed analysis of the two-dimensional diffraction data has allowed us to unravel the interplay between the martensite formation, the texture evolution and the load partitioning, and to correlate the observed behaviour to the macroscopic response of the material. The high aluminium content TRIP steel grade presents a higher volume fraction of retained austenite at room temperature that transforms more gradually into martensite under deformation, providing a larger uniform elongation. The comparison between the observed transformation behaviour and the texture evolution indicates that the ⟨1 0 0⟩ component along the loading direction corresponds to a low critical stress for the transformation. The evolution of the elastic strains revealed the occurrence of a significant load partitioning before reaching the macroscopic yield strength, which becomes more pronounced in the plastic regime due to the progressive yielding of the different grains in the polycrystalline material. This opens the door to tailor the austenite stability by altering the distribution in grain size, local carbon content, and grain orientation in order to produce the optimal load partitioning and work hardening for improved combinations of strength and formability in low-alloyed TRIP steels.



## 1. Introduction

Current trends in modern vehicle concepts steer the ongoing material research for automotive applications towards new solutions for future lightweight vehicle designs. The criteria for a proper (multi-)material selection include a significant vehicle weight reduction, thus decreasing the energy consumption and environmental impact, without either increasing the production costs or compromising the passengers' safety and comfort [1–3]. This has triggered the development of a new generation of high-strength steels with improved formability, so that thinner metals sheets can be used to produce lighter auto bodies [4]. Low-alloyed TRansformation-Induced Plasticity (TRIP) steels are considered promising high-strength automotive materials, where the key to attain high formability levels seems to reside in the presence of a relatively small fraction (<20 vol.%) of metastable austenite retained at room temperature within the complex ferrite–bainite–(martensite) microstructure of low-alloyed TRIP steels [5–7]. The metastable austenite phase will respond to mechanical and/or thermal stimuli by transforming progressively into harder martensite [8].

In order to design and control the mechanical response of TRIP steels for selected automotive components, a thorough understanding of the processes occurring at the micro-scale induced by the applied stress is clearly required. In view of the strong correlation between the complex deformation/transformation behaviour of TRIP steels at a micro-scale and the macroscopic material response [5,9,10], a great effort has been made in recent years to develop TRIP microstructures with the adequate combination of phases [11–13], and to collect information about their behaviour under deformation using a broad range of experimental techniques including conventional X-ray diffraction [8,14,15], Mössbauer spectroscopy [16], atomic force microscopy [17], transmission-electron microscopy [18,19] and electron back-scattered diffraction [20,21]. However, most of the available information has been derived from ex situ and/or (near-) surface studies, and do not accurately capture the complex deformation/transformation

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process that takes place in TRIP steels. As a consequence, relevant questions for the design and production of new TRIP steel grades with the desired mechanical properties do still not have a unified explanation, such as what the exact contribution of the TRIP effect is to the high work hardening rate and the large ultimate elongation observed in these materials, or what the local microstructural characteristics of the austenite grains and their surrounding evolving matrix are at the onset of the transformation and how they affect the stability of the austenite grains. In parallel to these experimental efforts, multi-scale models have been developed in the last decade to link the microstructural evolution of TRIP steels to their macroscopic response under different loading conditions [22–26]. However, these multi-scale models still cannot reliably predict key material properties, such as the delay of necking or the ultimate elongation, based on an accurate description of the deformation/transformation behaviour at the micro-scale.

The availability of intense neutron and high-energy synchrotron X-ray beams at several large scale facilities worldwide has opened the door to probe the bulk mechanical behaviour of polycrystalline materials during in situ diffraction studies under deformation [27,28]. In recent years, the use of neutron and/or high-energy X-ray diffraction has led to new insights into the deformation behaviour of a wide range of technologically relevant materials, ranging from stainless steel [29], copper [30] and magnesium alloys [31] to shape memory alloys [32] and nickel-base superalloys [33]. In situ neutron and high-energy X-ray diffraction experiments have in the last years also been reported on low-alloyed TRIP steels [34–38]. However, due to the composite-like TRIP microstructure and the complex deformation/transformation behaviour in these materials, the reported neutron and synchrotron experiments were designed to focus on selected microstructural aspects of the material during deformation, especially on the stress partitioning between the constituent phases.

The aim of the present paper is to assess in situ the changes induced by the applied tensile stress on the main microstructural parameters governing the mechanical behaviour of both the metastable austenite phase and the surrounding ferritic matrix in low-alloyed TRIP steels: phase fractions, texture effects and occurrence of strains. The characteristics of each phase and the interaction between phases will be correlated to the macroscopic response of the material at each deformation step up to rupture. In order to achieve this goal, we have studied two TRIP samples differing in their aluminium content by performing in situ high-energy X-ray diffraction at a synchrotron source during deformation. The purpose of using two aluminium-containing samples is twofold: (1) to obtain two different starting microstructures and (2) to link the observed micro- and macro-scale behaviour to an industrial process parameter, i.e. the overall aluminium content. In recent years, we have succeeded in monitoring the temperature-dependent martensitic transformation of metastable austenite in TRIP steels on both an average [39] and single-grain level [40–42]. In the present work, we characterize in detail the effect of deformation on these complex multiphase TRIP steels.

## 2. Experimental

### 2.1. Sample preparation

The chemical composition of the two studied TRIP steel grades is shown in Table 1. Cylindrical dog-bone-shaped tensile specimens with a gauge length of 10 mm and a diameter of 1 mm were machined from 6-mm thick hot-rolled sheet materials. The cylindrical axis of the samples was selected to be parallel to the rolling direction (RD) of the sheet material. A mark parallel to the normal direction of the sheet material was made on the top part of the samples, in order to keep track of the normal (ND) and transverse (TD) directions of the starting sheet material when studying texture effects during the in situ X-ray diffraction experiments under deformation. In order to generate the starting (non-deformed) TRIP microstructures at room temperature with a significant amount of retained austenite, the samples were initially annealed in a salt bath at the intercritical temperature $T_i$ to obtain a ferrite–austenite microstructure. The intercritical holding time $t_i$ was selected to be 30 min for both materials. The samples were subsequently quenched to a lower temperature in order to trigger the bainitic transformation of part of the intercritical austenite. After a holding time ($t_{bh}$) of 60 s at the bainitic temperature of $T_{bh}$ = 673 K, the samples were quenched in water to room temperature. The relevant parameters for the heat treatment of the two specimens are collected in Table 2. The intercritical temperature $T_i$ for each chemical composition was chosen to obtain the maximum fraction of metastable austenite in the room-temperature microstructure. Fig. 1 shows an optical micrograph of the $Al_{1.8}$ TRIP material at room temperature. The average austenite grain size is about 3 μm in both samples. For a detailed characterization of the initial microstruc-

**Table 1**
Chemical composition of the two TRIP steels used in this study (wt.%), with balance Fe.

| Material | C | Mn | Si | Al | P |
|---|---|---|---|---|---|
| $Al_{0.4}$ | 0.188 | 1.502 | 0.254 | 0.443 | 0.015 |
| $Al_{1.8}$ | 0.218 | 1.539 | 0.267 | 1.750 | 0.018 |

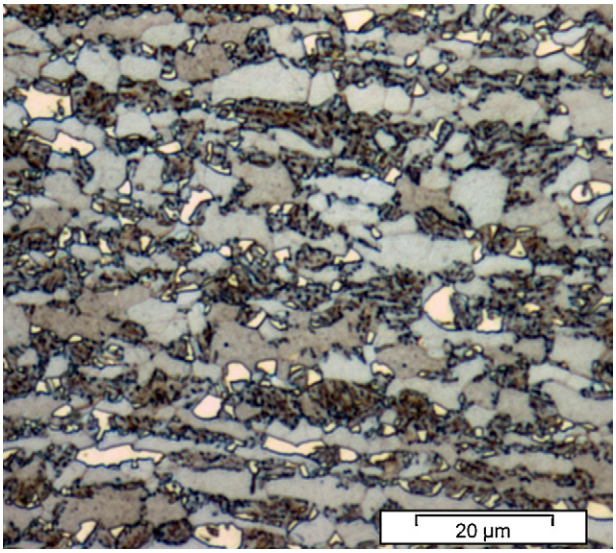


**Fig. 1.** Optical micrograph of the $Al_{1.8}$ TRIP material. The room-temperature multiphase microstructure contains a metastable austenite phase (white) within a matrix of ferrite (grey) and bainite (dark grey) phases.

**Table 2**
Relevant parameters describing the heat treatment of the two TRIP samples: the thermodynamic transformation temperatures ($A_1^-$, $A_1^+$ and $A_3$) calculated using the thermodynamic database MTDATA, together with the intercritical annealing temperature ($T_i$), time ($t_i$) and austenite fraction ($f_\gamma^i$). The samples were quenched to a bainitic holding temperature $T_{bh}$ and held for a period $t_{bh}$, before finally quenching to room temperature.

| Material | $A_1^-$ (K) | $A_1^+$ (K) | $A_3$ (K) | $T_i$ (K) | $t_i$ (min) | $f_\gamma^i$ (%) | $T_{bh}$ (K) | $t_{bh}$ (s) |
|---|---|---|---|---|---|---|---|---|
| $Al_{0.4}$ | 967 | 983 | 1127 | 1073 | 30 | 37 | 673 | 60 |
| $Al_{1.8}$ | 1016 | 1035 | – | 1173 | 30 | 52 | 673 | 60 |

*Note*: $Al_{1.8}$ steel cannot be made in the pure austenite phase.