PRIOR MICROSTRUCTURE MODIFICATION AND SYNERGIC EFFECT OF DEFORMATION IN TRIP STEEL

Jozef Zrník, Ondrej Muransky, Petr Šittner

COMTES FHT Inc., 33441 Dobřany, Průmyslová 995, Czech Republic, jznik@comtesfht.cz
Nuclear Physics Institute v.v.i., 25068 Řež, CAS, Czech Republic, ondre.muransky@outlook.com
Institute of Physics, AS v.v., Na Slovance 2, 18221 Prague, Czech Republic

Abstract
The paper presents results of in-situ neutron diffraction experiments aimed on monitoring the phase evolution and load distribution in TRIP steel when subjected to tensile loading. Tensile deformation behaviour of TRIP steel with different initial microstructures showed that the applied tensile load is redistributed at the yield point and the harder retained austenite (Fey) bears larger load than ferrite (Feα) matrix. After load partitioning is finished, macroscopic yielding comes through simultaneous activity of the martensite transformation (in the austenite) and plastic deformation process in ferrite. The steel with higher volume fraction of retained austenite and less stronger ferrite proved to be a better TRIP steel with respect to strength, ductility and the side effect of the strain induced austenite-martensite transformation. The neutron diffraction proved to be efficient method to evaluate deformation behaviour of steel of different initial structures. The results have shown that in-situ neutron diffraction is sensitive method at monitoring the phase evolution and load distribution in time of deformation.

Key words: TRIP steel, microstructure, transformation behavior, mechanical properties, neutron diffraction

INTRODUCTION
Low alloyed TRIP (Transformation Induced Plasticity) steels remain one of the most challenging goals of the high strength advanced structural steel research. These steels are a class of multiphase steels (ferrite, bainite, retained austenite) offering an attractive combination of high strength and ductility [1, 2]. Their microstructure consists of a mixture of polygonal ferrite, bainite and metastable retained austenite [3]. Most publications on TRIP-assisted steels highlight the role of the retained austenite, its transformation to martensite during plastic deformation, which results in strength enhancement strength and improved formability of steel [3, 4]. Since the contribution of transformation strain due to formation of a newly formed martensite phase is much smaller than the achieved total sample elongation, the strain-induced transformation per se provides only a minor contribution to the uniform elongation [4]. However, on the other side it leads to the redistribution of stresses and at the same time appearance of composite effect [3, 5], which is behind the resulted higher uniform elongation of TRIP steels [4,6].

The paper presents the results on structure modification in TRIP steel resulting from different thermomechanical treatment applied and neutron diffraction results when monitoring the phase evolution and load distribution in TRIP steel, with slightly different microstructures, subjected to tensile loading. The results received have shown different load redistribution at the yielding point with respect to ferrite and austenite. The neutron diffraction results have shown that the applied tensile load is redistributed at the yielding point in such a way that the retained austenite bears a significantly larger load than the α-matrix during the TRIP-assisted steel deformation. Steel sample with higher volume fraction of retained austenite and less strong ferrite matrix proved to be a better TRIP steel with respect to strength, ductility and the side effect of the strain induced austenite - martensite transformation.
The low-alloyed Si-Mn-C steel with a chemical composition of 0.18 C, 1.4 Mn, 1.9 Si, 0.02 S, 0.02 P, 0.003 Nb, 0.02 Al (in mass%) was used in the investigation. The specimens in the form of bars 25 mm in diameter were subjected to two different thermomechanical (TM) controlled processing according to Fig. 1. The TM processing conditions were as follows:

**TM_A:**
1) heating to 850°C / 30 min. → 2) compression deformation $\varepsilon_1 = 55\%$ → 3) followed by isothermal holding at 750°C / 600 s (the first transformation $\text{Fe}_\gamma \rightarrow \text{Fe}_\alpha$) → 4) water cooling to bainite temperature and tempering at 420°C / 600 s → air cooling.

**TM_B:**
1) heating to 1000°C / 30 min. → 2) the first compression deformation $\varepsilon_1 = 50\%$ → 3) air cooling to $820°C$ followed by second compression deformation $\varepsilon_2 = 64\%$ → 4) isothermal hold at 750°C / 300 s → 5) water cooling to bainite temperature and tempering at 420°C / 300 sec. → 6) air cooling.

Two different structures with different morphology of bainite and retained austenite characterization was conducted by scanning electron microscopy. The mechanical stability of retained austenite was assessed in time of deformation process. *In-situ* deformation, using combined stress and strain control, were conducted at room temperature using the diffractometer at ISIS spallation neutron source, as shown in Fig. 2. Maceroscopic strain was monitored by means of a clip gage extensometer. The diffractometer operated in Time of Flight (TOF) diffraction mode and diffraction condition are referred to [7, 8].

The instrument is equipped with a 50 kN. Instron testing machine installed on the diffractometer, with its loading axis 45° turned to the incidental beam. Two detector’s banks measured time-resolved diffraction patterns, regarding grains oriented in axial and transverse radial geometry with respect to the applied tensile stress, at fixed horizontal scattering angles of ± 90°. Each (hkl) reflection in the diffraction pattern (see Fig. 3) is generated by a distinct family of polycrystal grains similarly oriented with respect to load axis.

**Fig. 1.** TM processing schedules for experimental TRIP steel.

**Fig. 2.** Experimental set-up of the ENGIN-X instrument on the beam line at ISIS[9].

**Fig. 3.** The full diffraction pattern of TRIP assisted steel analyzed by GSAS.
Strain/stress diffraction measurements in crystalline materials are based on the precise assessment of lattice plane spacing deviation ($d_{hhkl}^0$) of particularly oriented (hhkl) crystallographic planes or lattice phase parameter ($a_{ph}^0$) deviation due to residual or applied stresses [9]. The lattice plane strain ($\varepsilon_{hhkl}$) and the average phase strain ($\varepsilon_{ph}$) can be determined from the measured changes of the lattice plane spacing and lattice parameters, respectively, according to Eq.1 and Eq. 2, where $t_{hhkl}$ is time-of-flight and $d_{hhkl}^0$ are the particular lattice spacing and lattice parameter in the stress free sample, respectively [10]. Individual reflections in diffraction patterns from both detector banks (axial, radial) were analysed by single peak fits to give peak position and integrated peak intensity, Fig. 3. These values were used for lattice strains and austenite fraction evaluation during tensile test.

2. RESULTS AND DISCUSSION

2.1 Microstructure characterization.

Microstructural analysis (SEM) of both TRIP steel samples revealed multiphase structure consisting of polygonal ferrite, bainite (α-matrix) and retained austenite. The structure characteristics of individual samples appear to be different as concerns morphology of bainite phase and size and ferrite volume fraction, as seen in Fig. 4 (A, B). Apparently, the main difference between the structures is the size and ferrite volume fraction on one side and bainite morphology on the other. The retained austenite fraction, measured using X-ray and neutron diffraction was detected to be higher in Fig. 4A (0.08) and lower in B (0.04) samples. Due to a higher solutioning temperature in case of TM$_A$ sample (1000 ºC), the ferrite and bainite grains are of coarser morphology, Fig. 4B. As a fact, large amount of needle-like bainite is present in final structure and some scattered retained austenite particles are also found next to and/or inside the bainite islands (see arrows in Fig.4 B). As a fact, large amount of needle-like bainite is present structure and also some scattered retained austenite particles are also found next to and/or inside the bainite islands (see arrows in Fig. 4B). In the case that the TM$_A$ processing started at lower temperature of solutioning (800 ºC) the obtained structure is refined and consists of equiaxed ferrite and granular bainite (Fig. 4a). The grain size of refined retained austenite in sample A is of 1 – 2 μm and precipitates either on ferrite boundaries or in vicinity of granular bainite grains.

![Microstructures of TRIP steels resulting from different TM$_A$ and TM$_B$ processing procedures with modified bainite morphology.](image)

2.2. Macroscopic responses in tensile behaviour

The macroscopic stress-strain curves measured at tensile tests on samples A and B are shown in Figs. 5a,b. The Fig. 5b is an enlarged inceptive part of deformation records near the yield point. The short plateaus on the step - wise curves in Fig. 5b ($\varepsilon \leq 2 \%$) correspond to the temporary creep-like dwells in the stress controlled section of the tests, when diffraction data were collected. On the other hand, stress relaxations of
about 100 MPa (Fig. 5a) concurs during these dwells, in strain controlled part of the tensile tests records, when diffraction data were collected. The tensile properties results are summarized in Table 1. The sample B exhibits higher yield stress than sample A, but essentially lower elongation, whereas tensile strength of both samples are comparable. The higher yield stress of sample B can be attributed to different structure morphology and higher bainite volume fraction in microstructure.

Fig. 5. The strain-stress records of TRIP steels structures. The inset in a) shows record of stress relaxations during temporary dwells for neutron diffraction data collection. b) Linings marking the Young’s modulus, 0.2% yield stress and 0.2% plastic strain.

2.3 Load partitioning between retained austenite and ferrite-bainite constituents.

The evolution of the volume-averaged phase strains \( \varepsilon_{\text{ph}} \) in the ferrite-bainite (\( \alpha \)) matrix and retained austenite during tensile tests of both investigated TRIP steel samples (A, B) as functions of the applied stress, are plotted in Fig. 6 a,b. The stress free lattice parameters \( a_{\text{ph},0} \) were taken as those measured in the \( \alpha \)-matrix and retained austenite prior starting the uniaxial loading test. When studying tensile deformation behaviour of single phase materials by in situ diffraction method, the evolution of the volume-average phase strain \( \varepsilon_{\text{ph}} \) remains approximately linear with the applied stress even behind the onset of plastic flow. In contrast, the onset of plastic deformation of a multiphase material with different elasto-plastic properties of individual phases is accompanied by significant redistribution of stresses between the phases. The interphase stresses are evidenced by significant deviations of the lattice strain–stress dependencies from linearity [11]. The stress free lattice parameters \( a_{\text{ph},0} \) are those of measured in the \( \alpha \) matrix and retained austenite prior to tensile testing.

Fig. 6. Evolution of the volume averaged phase strain \( \varepsilon_{\text{ph}} \) in the \( \alpha \)-matrix and austenite with applied stress in time of deformation of TRIP steels samples A (a) and B (b) measured by neutron diffraction.
As seen in Fig. 6, the volume-averaged phase strains $\varepsilon_{ph}$ determined in the $\alpha$-matrix and retained austenite at applied stresses below the elastic limit (labeled as P1) are similar and they are proportional to the macroscopic applied stress. This suggests that the elastic properties of the $\alpha$-matrix and retained austenite are comparable. Above the elastic limit, however, the phase strains of $\alpha$-matrix and retained austenite deviate from linearity in opposite directions. This is due to the redistribution of stress from the $\alpha$-matrix (which starts to yield plastically at lower stress) towards the retained austenite. This implies that the stress needed to transform the retained austenite into martensite is higher than that needed to trigger plastic deformation in the $\alpha$-matrix. The tensile stress in the austenite increases with increasing applied stress up to the point where it starts to deform plastically and/or transform to the martensite phase. Macroscopic yielding of the sample may only start after the load has been redistributed and stress in the retained austenite has reached the transformation limit. In what follows, inelastic deformation processes proceed in both $\alpha$-matrix and retained austenite in a hardening manner which leads to further increase of the stresses (phase strains) in both microstructure components. The retained austenite thus provides the potential for high ductility of the TRIP steel but at the same time acts as a “reinforcement phase” during the plastic deformation of its complex multiphase microstructure. Beyond the macroscopic yielding point, when austenite volume fraction quickly decreases as a result of the strain-induced transformation, the data analysis becomes complicated by the fact that the reflections of the newly formed martensitic phase overlap with the ferrite-bainite reflections. In the last stages of the test, when significant martensite volume fraction is present, the stress is probably redistributed again from the austenite towards the new much harder martensite phase, which we refer to as “austenite load shedding” (point P2, Fig. 6a,b). The interpretation is not easy since the hardening behaviour is not known, and the retained austenite in the microstructure may exist in various morphological forms having different carbon contents. Similar qualitative conclusions on austenite being a harder phase in TRIP steel structure were drawn by [9] and [12].

CONCLUSIONS

Results of in situ neutron diffraction experiments on two TRIP-assisted steel with different microstructures after TM treatment have shown that the applied tensile load is redistributed at the yielding point in such a way that the harder retained austenite bears significantly larger load than the softer $\alpha$-matrix. Only after this load partitioning is finished, macroscopic yielding of the TRIP steel takes place through simultaneous activity of the martensitic transformation (in Fey phase) and plastic deformation processes (in Fe$\alpha$ matrix). The transforming retained austenite thus provides the potential for higher ductility of this TRIP steel but at the same time acts as a “reinforcement phase” during the further plastic deformation. The alloy with higher volume fraction of retained austenite and less strong $\alpha$-matrix (sample A) proved to be a better TRIP steel with respect to strength, ductility and side effect of strain induced austenite – martensite transformation.

Acknowledgement

The results presented in this paper arose under the project West-Bohemian Centre of Materials and Metallurgy CZ,1,05/2,1,00/03,0077 co-funded by European Regional Development Fund.

REFERENCES