Austenite stability and heterogeneous deformation in fine-grained transformation-induced plasticity assisted steel

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Abstract

We report the mechanistic explanation of the variation in Lüders strain in fine-grained transformation-induced plasticity assisted steel. The austenite stability is demonstrated to have a profound influence on the Lüders strain. Furthermore, it is shown unambiguously using a thermodynamic analysis that the transformation of austenite is strain-induced. The work results in a generic method of distinguishing the cause of martensitic transformation during tensile tests, given that both stresses and strains are necessarily present beyond the yield point.

Key words: TRIP steel, retained austenite, stability, heterogeneous deformation, digital image correlation

When austenite decomposes by a displacive mechanism, the change in crystal structure is accompanied by a shape deformation which is an invariant-plane strain with a large shear component and a volume change normal to the plane. This shape deformation is exploited in the design of TRIP steels, where the acronym stands for transformation-induced plasticity. Such steels contain retained austenite as a minor phase since their invention has been available in a commercially viable form some fifteen years ago [1–3]. The austenite transforms during deformation, thus leading to work hardening and hence delaying plastic instabilities. The major microstructural constituent is allotriomorphic

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ferrite, with the residue consisting of a mixture of bainitic ferrite and carbon-enriched retained austenite. But an important further characteristic is that the microstructure contains a significant density of mobile dislocations, introduced by the plastic accommodation of the shape deformation due to the bainite transformation. This means that the steel yields nearly continuously and uniformly until necking occurs. An alternative mode observed in mild steels involves localised deformation which once initiated, spreads at a constant stress throughout the specimen, i.e. by the propagation of Lüders bands [4, 5]. These are undesirable features in applications where surface quality is vital, since the bands become visible on the finished component. There has been a report of a significant Lüders strain in such multiphase steels [6] but only in circumstances where the heat treatment is manipulated to eliminate or greatly reduce the retained austenite content [7].

Recent work has attempted to create a different kind of TRIP-assisted steel, in which there are only two phases, a mixture of extremely fine grains of allotriomorphic ferrite and austenite generated by intercritical annealing [8–12]. The austenite is stabilised to transformation during cooling by the fine grain size and the partitioning of manganese in particular, into the austenite. The average manganese concentration in these steels is about 6 wt%. Given that there is no displacive transformation to generate a mobile dislocation structure, and because the intercritical annealing also leaves the structure with very low dislocation density, the material exhibits and upper and lower yield point together with a Lüders strain. However, the Lüders strain occurs in the presence of austenite which in turn is induced to transform into martensite. This must influence the nature of the deformation at the Lüders band, and was the purpose of the present work to investigate this interaction between the retained austenite and the heterogeneous deformation during tensile testing with the strains monitored *in situ* using digital image correlation [9, 13].

The chemical composition of the steel investigated is Fe-0.055C-5.6Mn-0.49Si-2.2Al wt%. An ingot $300 \times 150 \times 100$ mm was prepared by vacuum-induction melting, hot-rolled to 3 mm thickness and further cold-rolled to 1 mm thick sheet. Sample were intercritically annealed using an infrared heating furnace at 700 °C and 740 °C, respectively, the temperatures being chosen to vary the mechanical stability of the austenite. The heating and cooling rates were $\pm 10 °C s^{-1}$.

Tensile tests were conducted on a universal testing machine at a crosshead speed of 2 mm min^{-1}, using sub-sized coupons [14]. The local strain distribution during tensile test was measured quantitatively with digital image correlation. A black pattern was sprayed to white background on tensile specimen and its displacement of the pattern was recorded at 2 frames per second, which was analysed using *ARAMIS* software. The fraction of retained austenite was evaluated using X-ray diffraction with Cu K$_\alpha$ radiation. The integrated intensities

of $(200)\alpha$, $(211)\alpha$, and $(220)\gamma$, $(311)\gamma$ reflections were used to quantitatively determine the austenite content. The Vickers hardness was measured so that work–hardening by localised deformation could be characterised.

Fig. 1 shows the annealed microstructure consisting of fine grains of austenite and ferrite. The grain size of austenite is submicrometer for both annealing temperatures, which accords with earlier results [8]. The austenite fractions measured using X-ray diffraction are 0.15±0.02 and 0.21±0.03 when annealed at at 700 °C and 740 °C, respectively.
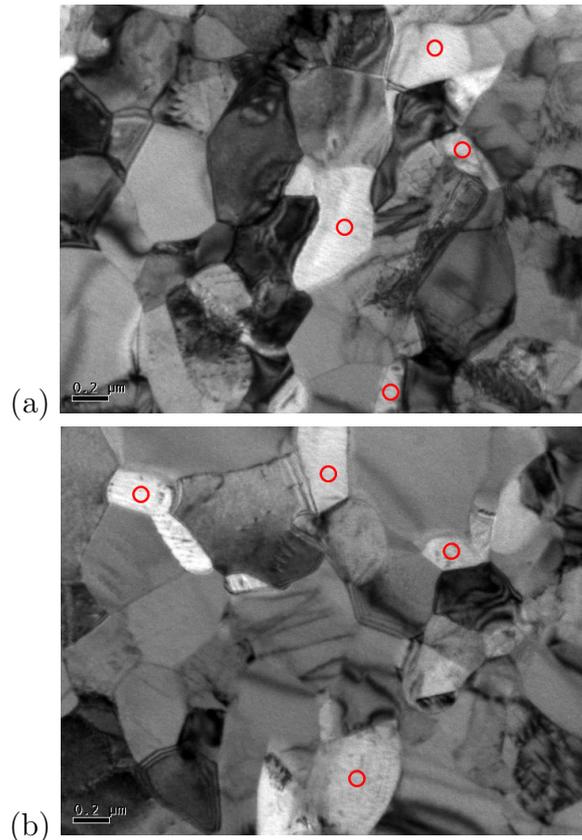
Fig. 1. Transmission electron micrographs of sample annealed at (a) 700 °C and (b) 740 °C. Austenite grains are identified with circles.

Figs 2 (a) and (b) show the stress-strain curves overlapped with the change of retained austenite as a function of tensile strain. Both samples show clear upper and lower yield stresses, with the onset of the Lüders strain once dislocations become mobile. Austenite content measurements are also illustrated, both behind and ahead of the Lüders front; of course, there is only the former measurement is possible once the sample begins to deform homogeneously following the exhaustion of the Lüders strain. It is fascinating that the austenite essentially remains intact until the passage of the band, indicating that much of the induced transformation occurs in the localised deformation during the passage of the Lüders band. This represents exceptional evidence that the martensitic transformation during deformation is strain-induced because the

stress alone has little influence on the austenite in the regions ahead of the Lüders band. This conclusion is in contrast to extensive evidence that in ordinary TRIP-assisted steels, the decomposition of the retained austenite during tensile testing where the deformation is homogeneous, can be explained on the basis of stress alone [15–17]. The reason is clear from Fig. 2, where it is apparent that the level of stress at yielding is insufficient to stimulate martensite, since the amount of austenite at zero strain does not change until the passage of the Lüders band.

In order to establish this hypothesis further, thermodynamic calculations were conducted in order examine the potential for stress alone to stimulate martensitic transformation. Stress-induced transformation is possible if the combination of mechanical (ΔG_{mech}) and chemical driving force ($\Delta G_{\text{chem}} = G^{\alpha'} - G^{\gamma}$) is less than critical value needed to induce martensite ($\Delta G_{M_S}^{\gamma\alpha}$):

$$\Delta G_{M_S}^{\gamma\alpha} \geq \Delta G_{\text{chem}} + \Delta G_{\text{mech}} \tag{1}$$

The mechanical driving force is given approximately by $-0.86\sigma \, \text{J} \, \text{mol}^{-1} \, \text{MPa}$ [18], where σ is the applied tensile-stress, and it is assumed that the martensite that will form is perfectly oriented for the maximum interaction with the stress. In general, for a polycrystalline material, the actual magnitude of ΔG_{mech} is likely to be smaller since the indices of the martensite habit plane are determined by crystallography, and the orientation of the austenite grains is not controlled. The critical value of the driving force necessary to stimulate martensite depends on two terms, G_1 due to the chemical composition of the steel [19] and on G_2 due to the austenite grain size [20–22], and can readily be calculated. The chemical driving force, which is purely a function of the chemical composition was calculated using standard software and thermodynamic database [23], assuming that the austenite has its equilibrium composition. The results are presented in Table 1. It is evident from these data, that $\Delta G_{\text{chem+mech}}^{\gamma\alpha}$ and $\Delta G_{M_S}^{\gamma\alpha}$ are almost identical for the 700°C, consistent with the fact that the stress does not stimulate martensitic transformation in the region not affected by the Lüders band. In the case of the 740°C treatment, some stress-induced martensite is possible since at face value $\Delta G_{\text{chem+mech}}^{\gamma\alpha} < \Delta G_{M_S}^{\gamma\alpha}$, the values differing by about 10%. Consistent with this, a small amount of stress–induced transformation is observed in the region where plastic strain is not yet implemented (the blue points in Fig. 2b); the influence of the stress is nevertheless small, probably because as pointed out previously, $|\Delta G_{\text{mech}}|$ is overestimated.

Martensitic transformation is therefore induced by plastic strain in the local region swept by localised deformation; this in turn leads to strengthening and hence the propagation of the band into previously unaffected material. The austenite in the sample intercritically annealed at 740 °C is much more sensitive to plastic strain, indicating that it is less thermodynamically stable than

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Table 1
Free energy terms describing the composition and tensile stress effects on martensitic transformation. Note that ΔG_{mech} assuming that σ is the upper yield strength.

Heat treatment	ΔG_{chem}	ΔG_{mech}	$\Delta G_{\mathrm{chem+mech}}^{\gamma\alpha}$	$\Delta G_{M_S}^{\gamma\alpha} = G_1 + G_2$
	$\mathrm{J\,mol^{-1}}$	$\mathrm{J\,mol^{-1}}$	$\mathrm{J\,mol^{-1}}$	$\mathrm{J\,mol^{-1}}$
700°C	-2782	-742	-3524	$-2162 - 1361 = -3523$
740°C	-3155	-674	-3829	$-2086 - 1352 = -3438$

the austenite formed by heat-treatment at 700 °C. This is related to the difference in the carbon and manganese concentrations of the austenite as a function of intercritical annealing temperature [8, 11]. The carbon and manganese contents in austenite are estimated to be 0.37, 10.7 wt.% after annealing at 700 °C and 0.26, 7.6 wt% at 740 °C, using the mass balance of carbon and the energy dispersive spectroscopy analysis on manganese, respectively.
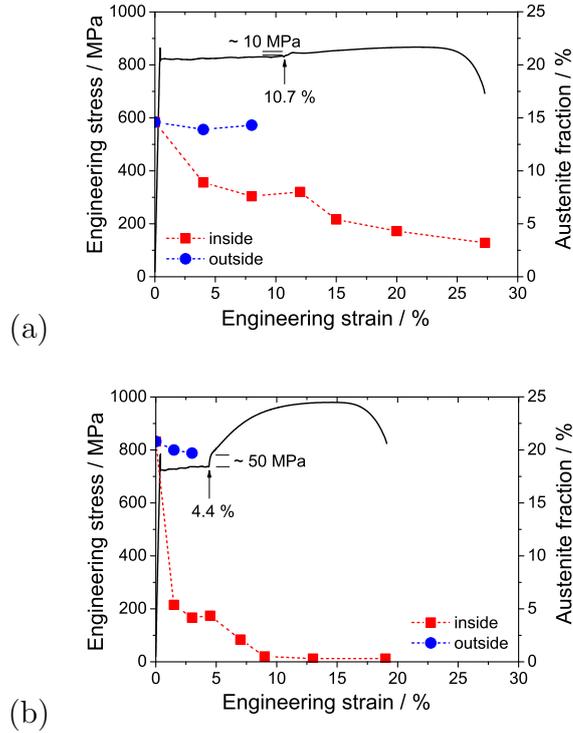
Fig. 2. Stress-strain curve and change of austenite content (points) during deformation after annealing at (a) 700 °C and (b) 740 °C. Sudden increase of stresses after the Lüders bands have traversed the entire gauge length are also indicated.

The plastic strain along the tensile axis as digitally determined is shown in Fig. 3. The strain distributions show vividly that the Lüders band propagates, with homogeneous deformation following once the Lüders strain is exhausted. The transition occurs at 10.9% and 4.2% for the samples annealed at 700 °C

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and 740 °C respectively, which agrees well with the Lüders strain observed in the stress-strain curves in Fig. 2. It is evident that the magnitude of the Lüders strain correlates with the mechanically stability of the austenite. What then is the physical reasoning behind this correlation?

Once the Lüders band developed in the initial stage of deformation, the work–hardening in the Lüders front counterbalances the stress concentration due to the reduction in cross-section behind the front, otherwise plastic instability would lead to failure. The grain size of the sample studied is extremely fine, and fine grains are associated with a reduced ability to work hardnen [24]. However, the TRIP effect is a work-hardening mechanism. The stress increment in the Lüders band due to the reduction in cross section is given by

$$\Delta\sigma = \frac{F}{A_0} \cdot \varepsilon_{\mathrm{L}} \tag{2}$$

where, F is the applied load, A_0 is cross-section outside of Lüders band and ε_{L} is amount of Lüder's strain. Approximating F/A_0 with the lower yield strength, the stress increment will be 88 MPa and 32 MPa for sample annealed at 700°C and 740°C. Fig. 4 shows the hardness profiles along the tensile axis with increase of overall tensile strain. A remarkable increase of hardness with the passage of Lüders band is verified. The hardness increment after propagation of Lüder's band is $\Delta H_{\mathrm{v}} = 27$ and 28. It corresponds roughly to strengthening of 90 and 93 MPa [25]. For the fine-grained steels studied here, this could be attributed largely to the TRIP effect. For the 700 °C sample, the extent of work–hardening clearly compensates for section reduction, indicating that the propagation of the band is controlled by the balance between these factors. On the other hand, rather larger strengthening is observed in the Lüders band in sample annealed at 740 °C, implying that more martensite forms than is necessary to compensate for section reduction. This in turn leads to an abrupt increase of stress by 50 MPa (Fig. 2b) once the Lüder's strain is exhausted, and general deformation begins. The increase of strength by 50 MPa matches well with the difference of estimated work–hardening from hardness measurement, 93 MPa, and the required strengthening for the propagation of the Lüders band, 32 MPa.

In summary, it is evident that when appropriate mixtures of austenite and ferrite are subjected to tensile test, there is a strong dependence of the Lüders strain on the mechanical stability of the austenite. A greater Lüders strain is obtained when the austenite transforms gradually, because the degree of work-hardening due to the TRIP effect is then smaller. It follows that it should be possible to completely eliminate Lüders banding by adjusting the stability of the austenite so that it undergoes stress-induced transformation (rather than one caused by plastic strain alone) and results in the hardening of any region which yields locally.

Acknowledgments

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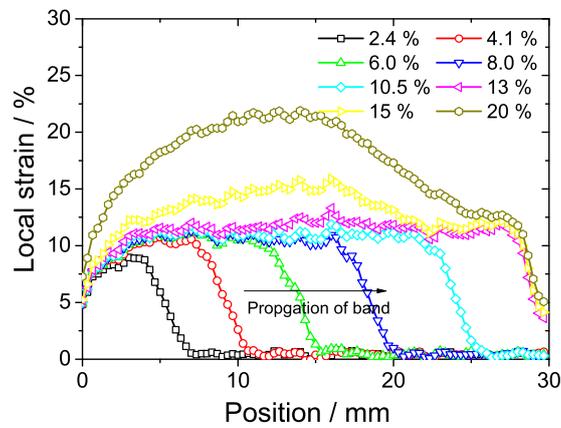
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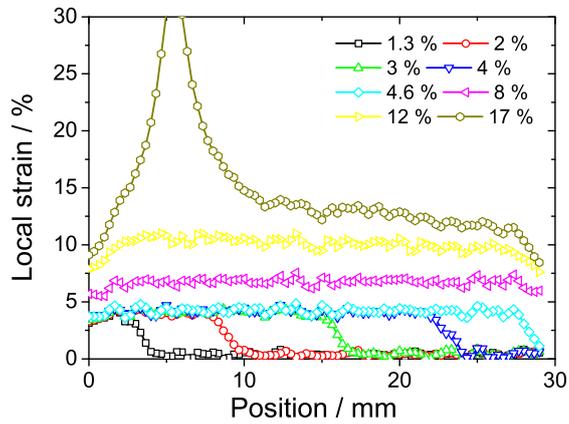
(b)

Fig. 3. Local strain distribution along tensile axis during deformation of sample annealed at (a) 700 °C and (b) 740 °C
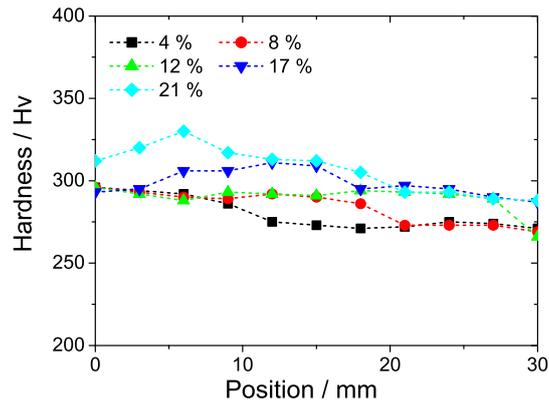
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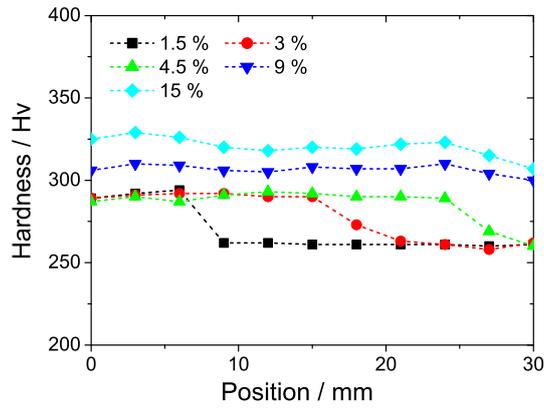
(b)

Fig. 4. Hardness change along tensile axis during deformation of sample annealed at (a) 700 °C and (b) 740 °C