



## Retained austenite: transformation-induced plasticity

Citation of the final chapter:

Pereloma, Elena, Gazder, Azdiar and Timokhina, Ilana 2016, Retained austenite: transformation-induced plasticity. In Colás, Rafael and Totte, George E. (ed), *Encyclopedia of iron, steel, and their alloys*, Taylor and Francis, New York. N.Y., pp.3088-3103.

This is the **accepted manuscript** of a chapter published by Routledge in *Encyclopedia of iron, steel, and their alloys* in 2016, available at: <https://doi.org/10.1081/E-EISA>

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## **THE TRANSFORMATION-INDUCED PLASTICITY EFFECT AND THE STABILITY OF RETAINED AUSTENITE IN STEELS**

**Elena Pereloma<sup>1</sup>, Azdiar Gazder<sup>1</sup> and Ilana Timokhina<sup>2</sup>**

<sup>1</sup>School of Mechanical, Materials and Mechatronic Engineering, University of Wollongong, Wollongong, NSW 2522, Australia

<sup>2</sup>Centre for Material and Fibre Innovation, Deakin University, Geelong, VIC 3217, Australia

**KEYWORDS:** transformation-induced plasticity, retained austenite, chemical stability, mechanical stability, microstructure, TRIP steels, carbide-free bainite

### **ABSTRACT**

The deformation-induced phase transformation of metastable austenite to martensite is accompanied by macroscopic plastic strain and results in significant work hardening and the delayed onset of necking. Steels that exhibit such transformation-induced plasticity (TRIP) effect possess high strength-ductility ratios and improved toughness. Since the stability of the retained austenite phase is the rate controlling mechanism for the TRIP effect, the factors affecting the chemical and mechanical stability of retained austenite in CMnSi TRIP steels are discussed. It was suggested that chemical stability plays a more important role at low strains, whereas other factors become responsible for the RA behavior at higher strains. The importance of optimising the processing parameters to achieve the desirable level of austenite stability was highlighted. Finally, the influence of mechanical testing conditions and the interaction between the phases during tensile testing are also detailed.

### **INTRODUCTION**

Since the latter part of the 1980s, significant research efforts have been directed towards the development, understanding and improvement of a new class of advanced high strength steels

for automotive applications; the range of so-called Transformation-Induced Plasticity (TRIP) or lean TRIP-assisted steels. These steels have typical compositions of 0.12-0.55C, 0.2-2.5Mn, 0.4-1.8Si (wt.%)<sup>1</sup> along with possible alloying additions of Al, Nb, Mo and P (<sup>1</sup>Thereafter in the text concentration of alloying elements in % always indicates wt.%). The multi-phase microstructure of TRIP steels is responsible for the high strength to ductility ratios (yield strength >550MPa, tensile strength >1000 MPa and total elongation between 25-40%) and significantly improved energy absorption capabilities. The above mechanical properties are the result of the complex interaction between the polygonal ferrite, carbide-free bainite and martensite phases as well as the gradual transformation of metastable retained austenite (RA) to martensite during straining. The aim of this chapter is to present the latest views on controlling the TRIP effect and enumerate the factors governing the stability of RA. The first section briefly summarises the deformation-induced martensite transformation, whereas the second one introduces the TRIP phenomenon and provides a historical perspective on the pathway that led from highly alloyed, fully austenitic steels to lean, multi-phase TRIP-assisted steels.

The subsequent sections focus on the factors affecting the chemical (composition and C content) and mechanical (size, orientation, morphology, stress state and location) stability of the RA which need to be controlled in order to achieve the desired strength-ductility balance in TRIP steels. Since variations in the processing schedules tend to affect the microstructure in general and the state of the RA in particular, the last section details the influence of the thermo-mechanical processing parameters.

## BRIEF DESCRIPTION OF DEFORMATION-INDUCED MARTENSITE TRANSFORMATION

Martensitic transformations are first order, diffusionless, solid state transformations during which a shear-dominant lattice distortion accompanies the cooperative movement of atoms. This results in an increase in the strain energy which in turn, dictates the kinetics of phase transformation and determines the morphology of the product phase. The transformation proceeds by the migration of relatively sharp interfaces at the speed of sound [1]. Based on the defects dissociation model for fcc to bcc transformations, Bogers and Burgers [2] predicted the essential characteristics of the barrierless heterogeneous nucleation of martensitic transformations such that the control of the transformation kinetics was governed by interfacial mobility. More recently, a dynamic model applicable to both, athermal and deformation-induced martensite transformation has been proposed. The dynamic model accounts for the supersonic growth of the martensite crystal through a master wave process initiated by an excited (oscillatory) state from within the elastic field of a dislocation [3]. From a thermodynamic point of view, a decrease in the Gibbs free energy is essential for the phase transformation from austenite to martensite to take place (Fig. 1a). At  $T_0^{\sigma=0}$  the free energies and chemical composition of both phases are identical and the Gibbs energy barrier ( $\Delta G^*$ ) opposes any phase transformation. Thus, austenite will only transform to martensite on undercooling to temperatures below  $T_0$  to the martensite transformation start temperature ( $M_S$ ); at which the excess free energy ( $\Delta G_{M_S}^{\gamma\sigma=0}$ ) to overcome the nucleation barrier becomes available. At any temperature between  $T_0$  and  $M_S$ , this chemical driving force ( $\Delta G_T^{\gamma\sigma=0}$ ) is insufficient to trigger austenite to martensite transformation on cooling ( $\Delta G_T^{\gamma\sigma=0} < \Delta G_{M_S}^{\gamma\sigma=0}$ ). However, the application of an external stress can provide the additional mechanical driving force ( $U'$ ) for transformation [4-6] such that [7]:

$$U' = \sigma_N \gamma_0 + \tau \epsilon_0 \quad (1)$$

where,  $\sigma_N$  is the normal component and  $\tau$  is the shear component of the applied stress resolved along the direction of the shear displacement of the shape deformation while  $\gamma_0$  and  $\epsilon_0$  are the normal and shear strains associated with phase transformation, respectively. The parameters  $\sigma_N$  and  $\tau$  act on the crystallographic plane common to both austenite and martensite phases. However, depending on the nature of the external load, the mechanical energy ( $U'$ ) either contributes to or detracts from the thermodynamic driving force. Thus, for a favourable martensite transformation stress state, the  $M_s^\sigma$  should be higher than  $M_s$ .

The combination of favourable chemical and mechanical driving forces could achieve the total driving force required for the martensitic transformation within the  $M_s - M_d$  temperature interval; where  $M_d$  is the temperature above which no martensite transforms before material fracture is reached [8] (Fig. 1a):

$$\Delta G = \Delta G_{M_s^\sigma}^{\gamma \rightarrow \alpha'} + U' \quad (2)$$

In addition to the chemical and mechanical driving forces, a crystallographic orientation or Magee effect [9, 10] also affects transformation kinetics such that the austenite grains that are more favourably oriented with respect to the macroscopic applied stress undergo phase transformation first. It should be noted that other non-chemical contributions to the total free energy change such as the strain energy due to volume change, the change in the interfacial energy and the kinetic energy due to martensite crystal formation are not addressed here.

The mechanically-assisted formation of martensite differs at temperatures below and above the  $M_s^\sigma$ . Below the  $M_s^\sigma$  temperature and similar to the nucleation of martensite on cooling, the nucleation of martensite takes place at pre-existing defects in austenite; as the applied stress is still below the yield stress of austenite (Fig. 1b). It follows that the higher the temperature of the applied deformation above  $M_s$ , the higher is the stress required for the initiation of this stress-induced  $\gamma \rightarrow \alpha'$  transformation.

In the temperature regime between  $M_s$  and  $M_s^\sigma$  in which plastic flow is controlled by stress-assisted transformation, the volume fraction of transformed martensite ( $f_{\alpha'}$ ) is proportional to the plastic strain ( $\varepsilon$ ) [11]:

$$f_{\alpha'} = k\varepsilon \quad (3)$$

At the temperatures between  $M_s^\sigma$  and  $M_d$  the applied stress required to initiate the martensitic transformation exceeds the yield stress of the parent austenite which in turn, starts to deform plastically and leads to the formation of new defects such as dislocation intersections, microbands, in-grain shear bands and stacking faults. Martensite nucleates predominantly at these abundant new nucleation sites and this mode of phase transformation is termed as strain-induced transformation. Two alternative explanations have been put forward for the formation of strain-induced martensite: (i) the higher potency of the nucleation sites created by plastic strain [8,12] and, (ii) the stress concentration at defects which on a local scale, may reach the critical stress required for transformation even when the magnitude of the applied macroscopic stress is less [13]. The kinetics of strain-induced martensite formation at the intersection of shear bands is usually described as [12]:

$$f_{\alpha'} = 1 - e^{-\beta(1-e^\alpha)^n} \quad (4)$$

where, the dimensionless parameter ' $\alpha$ ' describes the rate of shear band formation with respect to strain, is dependent on the stacking-fault energy (SFE,) of the steel and that its value increases with higher temperature; the parameter ' $\beta$ ' is linearly related to the probability that a shear band intersection will generate a martensitic embryo, and is governed by the chemical driving force for the martensitic reaction  $\Delta G_{>M_s^\sigma}^{\gamma \rightarrow \alpha}$ ;  $n$  is a constant.

In the vicinity of  $M_s^\sigma$ , both stress and strain-induced modes of transformation are operational.

Above  $M_d$ , the driving force for phase transformation is lowered and austenite undergoes

deformation either by preferential slip and/or twinning when the SFE of the steel is high or low, respectively (Fig. 1b).

## TRANSFORMATION-INDUCED PLASTICITY EFFECT

The TRIP phenomenon is defined as the manifestation of significantly increased plasticity resulting from a phase change during deformation [10]. Since the first observations of the TRIP effect in metastable austenitic steels [14, 15], a large volume of modelling [5-6, 10, 11, 16-18] and experimental work [5, 15, 16, 18, 19] has been aimed at understanding the mechanism. During the austenite to martensite transformation, the macroscopic plastic strain, arises from the shape change as determined by the preferential selection of favourable crystallographic variants (Magee effect [9]) and from the plastic accommodation processes which occur around the forming martensite grains (the Greenwood-Johnson effect) [6, 12, 16, 19]. During deformation-induced transformation, two main factors control the plastic flow behaviour of the material: (i) the dynamic softening arising from plastic straining due to the dilation effect upon formation of martensitic grains [11, 17] and, (ii) the static hardening caused by the increasing volume fraction of the harder martensitic phase. Dynamic softening is dominant at low strains, whereas the static hardening effect is related to the fraction of martensite formed at the stress and strain-assisted regimes via Eqs. (3) and (4), respectively.

Research on the mechanical behaviour of iron-based alloys exhibiting the TRIP effect has shown that with a decrease in temperature, the increase in tensile strength is very nearly proportional to the amount of martensite formed and that a maximum in the total elongation is attained at temperatures just above the  $M_s^\sigma$  temperature [11]. Overall, the strength-ductility relationships in TRIP-assisted metastable austenitic steels have shown much improvement when tested at temperatures between  $M_s$  and  $M_d$  due to the concurrent changes in work hardening rate [6, 11, 17, 20-22]. Deformation at elevated temperatures further

stabilises austenite due to a reduction in the driving force for transformation and an increase in the stacking fault energy (SFE). As the TRIP phenomenon requires austenite to undergo a phase transition, the term 'austenite stability' is used to describe the resistance of the metastable austenite to form martensite. To this end, a large variety of factors such as the steel composition, the processing history, the stress state and the test environment have been shown to affect austenite stability [15, 19-23].

The concept of transformation plasticity has been applied to the development of formable austenitic stainless steels. However, these TRIP steel grades are expensive as they are highly alloyed and contain significant amounts of Ni and Cr. Consequently, low-cost TRIP steels with lean compositions that typically contain ~0.12-0.55C, 0.2-2.5Mn, 0.4-1.8Si (wt%) were first developed by JFE Corporation in the 1980s [24-26]. The TRIP behaviour in these lean steels differs somewhat from highly alloyed austenitic steels as the microstructure before straining consists of ferrite and bainite; with a 5-20% volume fraction of austenite retained as the minor phase. Therefore, these lean steels have also often been referred to as TRIP-assisted steels. The subsequent sections of this chapter deal with the main factors controlling the retained austenite stability in these modern TRIP-assisted steels.

## FACTORS AFFECTING THE RETAINED AUSTENITE STABILITY

Early research on the mechanical properties of CMnSi TRIP-assisted steels had focused on the relationship between the enhanced mechanical properties and the volume fraction of the retained (residual) austenite (RA) present in the microstructure. It was considered that the volume fraction of the RA is a critical feature providing the TRIP effects such that a high volume fraction of retained austenite results in an increase in the strain-hardening coefficient which in turn, enhances the total elongation of the steel [25-27]. However, further research led to the conclusion that while the volume fraction of RA is an important factor, the

RA also had to possess a specific (optimum) stability in order to sustain the TRIP effect over a wide range of strains [28-30].

Consequently, if the RA phase is inherently unstable, then it easily transforms to martensite during the early stages of deformation and does not provide the TRIP effect at higher strains.

On the other hand, the overstabilisation of the RA phase also serves to restrict its transformation to martensite without providing the desirable TRIP effect [30-34]. Thus, in addition to Eqs. 3 and 4, constitutive approaches have been developed that link the plastic stability of the RA with the kinetics of its transformation during straining. By modifying earlier proposals [22-23], Matsumura *et al.* [26] described the transformation kinetics as a function of strain as follows:

$$1/f_{\gamma} - 1/f_{\gamma_0} = (k_p / p) \varepsilon^p \quad (5)$$

where,  $k_p$  is a constant related to the stability of RA against deformation,  $f_{\gamma}$  is the volume fraction of retained austenite at true strain ( $\varepsilon$ ),  $f_{\gamma_0}$  is the initial volume fraction of retained austenite and  $p$  is the strain exponent related to the autocatalytic effect ( $p= 3, 2$  and  $1$  for austenitic stainless, duplex stainless and dual-phase steels, respectively).

Alternatively, Sugimoto *et al.* [31] stated the following relationship for a reduction in the volume fraction of the RA with strain:

$$\log f_{\gamma} = \log f_{\gamma_0} - k\varepsilon \quad (6)$$

where,  $k$  is a constant whose value above zero is small when the strain induced transformation of retained austenite is suppressed. Other developments of such empirical relationships included expressing the fraction of untransformed austenite as a function of the chemical composition, the amount of original austenite, the deformation temperature and the strain [35]:

$$\log f_{\gamma_0} - \log f_{\gamma} = k_1 \Delta G \varepsilon \quad (7)$$

where,  $k_1 = 0.002017 \text{ Jmol}^{-1}$  is independent of the steel composition.

Recently, McDermid *et al.* [36] showed that while the strain-based model (Eq. 4, [12]) could successfully describe the kinetics of the RA to martensite transformation, a simple normalised plastic flow stress model (Eq. 8) also provides excellent agreement with the experimental data. The latter fit indicates that for the temperatures above  $M_s^\sigma$ , the transformation of tends to be stress-induced only:

$$\frac{f_\gamma}{f_{\gamma 0}} = \delta \exp\left(-\lambda \left[ \frac{\sigma_{flow} - \sigma_{YS}}{\sigma_{YS}} \right]\right) \quad (8)$$

where,  $\delta$  and  $\lambda$  are fitting parameters such that  $\delta$  is the untransformed fraction of RA at yield stress and  $\lambda$  is the rate constant for the martensite transformation.

The overall composition of the TRIP steel tends to define the chemical stability of the RA; with the C content in the RA playing the more dominant role. However, the stacking fault energy, substructure and solid solution strengthening are also contributing factors towards the overall stability of the RA. The processing history and the presence of other phases in the microstructure of the steel also control the chemical stability of the RA; as such variables define the stress state and composition of the RA and affect the  $M_s$  temperature. The condition necessary for the TRIP effect is that the temperature of athermal martensite formation ( $M_s$ ) remains below room temperature; otherwise the thermal formation of austenite occurs. On the other hand, the mechanical stability of the RA is dependent on its size, morphology, orientation and location within the microstructure. This means that processing history once again plays a very important role in RA stability; with the deformation mode, temperature and environment helping to control the TRIP effect. Another important feature that affects the RA stability and dictates the mechanical properties in general, is the interaction between the various phases during deformation [28,30]. Fig.

2 provides a summary of the factors affecting the RA stability, which will be discussed in more detail below.

### **Overall Chemical Composition and Carbon Content**

The main alloying elements in lean TRIP-assisted steels are Mn, Si and C, with possible additions of Al, P, Mo, Nb, V and Ti. Some steels could also have deliberate additions of Cu, Cr and Ni, or these elements could be present as tramp elements. In any case, the lower the  $M_s$ , the larger is mechanical driving force  $U'$  needed to cause the deformation-induced transformation at room temperature or above it. Thus, the RA is more stable and will transform at higher strains. However, if the difference between the  $M_s$  and the deformation temperature is too large, then the RA may not be able to transform to martensite and contribute to the TRIP effect. It was also shown that in the steel with ~10 vol% retained austenite the maximum uniform elongation was achieved during testing in the range of 20-70 °C [37]. At the same time, De Cooman [38] suggested that for automotive applications, TRIP steels should be designed with room temperature falling within the  $M_s$  and  $M_{30}^d$  temperature range; where  $M_{30}^d$  should preferably be below 100°C and is the temperature at which 50% of the austenite is transformed to martensite at a strain of 30%. Consequently, the above temperatures and austenite stability strongly depend on steel composition. However, the alloying elements also affect the austenite SFE and it is well known that high SFE ( $>20\text{mJ/m}^2$ ) suppresses the martensite formation and that the difference between the  $M_s$  and  $M_d$  is also related to the SFE of the steel [39].

Si plays an important role as a ferrite stabiliser and as a suppressor of cementite formation during bainite transformation [27, 32, 38, 40]. Si helps raise the temperature of ferrite formation, increases the chemical potential of C in ferrite and assists in more C to be rejected

into the remaining austenite; thereby providing its stabilisation. Similar to Si, both Al and P also inhibit the formation of cementite [41], decrease the C activity coefficient and increase the solubility of C in ferrite and lead to a higher enrichment of C in RA [33, 38, 41, 42].

However, Al also has the potentially deleterious effect of making the retained austenite less stable by increasing the  $M_s$  temperature to above room temperature [41]. Al also provides lower levels of solid solution strengthening in comparison with Si [43]. On the other hand, the beneficial effects of P as a carbide inhibitor are found only in the presence of Si or Al [44].

Mn is an austenite stabiliser which reduces the  $M_s$ , the activity of C in austenite and the kinetics of ferrite growth [45, 46]. In a C-Si-Mn system, Si enhances Mn segregation to austenite grain boundaries and inhibits the carbides precipitation; thereby increasing the C content of the untransformed austenite and decreasing the sympathetic nucleation rate of the bainitic ferrite (BF) such that it results in an incomplete bainite transformation phenomenon [40, 46]. Mn is also a strong solid solution hardener. It is well known that the presence of strong solid solution strengthening elements also influences the transformation characteristics [47]. High Mn content (>2.5 wt%) over-stabilises the RA [38] possibly through excessive solid solution strengthening effects and by increasing the SFE.

Other alloying elements such as Nb, Ti, Mo and V are carbide-forming elements and are typically used for ferrite grain refinement and strengthening. They also increase hardenability and thus suppress the undesirable pearlite formation. Although, these elements reduce the  $M_s$  when present in solute form in the austenite due to solid solution strengthening, they also take out C from the solid solution when they form carbonitrides and carbides [47, 48-50]. Thus, if precipitation is avoided, the volume fraction of the RA increases in such alloyed TRIP steels. Otherwise the chemical stability of the RA is reduced due to the lower C content (for example from 1.8 to 1.6% with addition of Nb and below 1.1% with addition of Mo [30, 51]) and leads to faster initial kinetics of martensite formation upon straining and a reduction

in the total elongation compared to the non-alloyed TRIP steel (Fig. 3a). In addition, changes in the kinetics of bainite transformation due to alloying could also adversely affect the rate of enrichment of the RA with C (*for e.g.*-Nb and Mo slows down the kinetics but increase the volume fraction of the RA [50, 52]).

Similar to Si and Al, Cu slows the precipitation of cementite and increases the amount of the RA. Cu is an austenite stabiliser, solution and precipitation strengthener [53,54]. Although Ni is an austenite stabiliser and a solid solution hardener [54] and Cr increases hardenability and assists in BF formation [54,55], their additions are expensive. Furthermore, in order to stabilise the RA to room temperature ~10% of Cr is required.

The most detailed studies were conducted on the effect of C, which is the strongest austenite stabiliser amongst all the alloying elements in TRIP steels. The condition that is necessary for the achievement of significant C enrichment in the RA is to prevent the formation of competing phases which could consume C such as pearlite and carbide-containing bainites. Sakuma *et al.* [56] revealed the strong dependence between the amount of C in the steel composition and the volume fraction of the retained austenite such that the most stable RA corresponded to the nominal concentration of 0.2-0.3% C. However, it is not the overall C content in steel but the C content in the RA that plays one of the dominant roles in the TRIP effect. Itami *et al.* [57], showed that the volume fraction of the RA with lower C content (~0.9%) reduces at a faster rate at relatively low strains compared to the more gradual transformation of the RA with higher C content (~1.1%) to martensite. At the same time, it was noted that the first portion of the RA that transformed first had a smaller lattice parameter, which corresponds to a lower C content compared to the remainder RA at higher strains.

Just as a fast rate of transformation of the RA with low C content (<0.6wt%) during plastic straining does not contribute to an increase in elongation [18], a much higher C content

(>1.8wt.%) results in the incomplete transformation of the RA to martensite and also does not lead to an increase in elongation [33]. In a steel with an average C content of 1.8% [58], the operation of twinning in austenite has been noted at a strain of 0.17. It could be speculated that in such a steel some RA grains may have a C content even higher than the average 1.8% and due to this, an increase in the SFE and a reduction in the  $M_s$  temperature made twinning the preferential behaviour at these strains.

Furthermore after twinning, the RA grains are stabilised even further due to matrix constraints to martensite transformation. However, such grains could contribute to mechanical properties of the steel via the twinning induced plasticity effect. Although the recent observations on the effect of strain on the RA stability confirmed a faster kinetics of martensite formation when the average C content in the RA was low (~0.7%) and a more gradual transition when the C concentration was within the 1.1-1.8 % range (Fig. 3a) [33, 51, 59], they indicate that contrary to previous reports [57, 60, 61]- the C content does not continue to increase with a decreasing volume fraction of the RA during tensile testing (Figs. 3a and 3b). The main change in the average C concentration of the RA takes place early on in the deformation process at strains below ~0.5. Thereafter, C content either remains constant (within the experimental error) or reduces while the volume fraction of the RA continues to decline. Using in-situ neutron diffraction studies, Tomota *et al.* [62] have also shown fast transformation with strain of RA with 1.57% C, which was present in the form of interlayers between the BF laths, compared to equiaxed RA with ~1.05% C which was located in the ferrite matrix. This indicates that while the chemical stability and C content in the RA in particular, plays an important role in the RA stability at low strains, other factors may become more important at higher strains. These points will be addressed in later sections.

### **Inhomogeneity of C content between and within the RA Crystals**

It is highlighted that the vast majority of studies on the chemical stability of RA are based on the average composition of the RA and that they do not provide a full account of all the processes taking place during deformation. This is associated not only with the inhomogeneous distribution of C between the various RA grains but also within a single RA grain itself. With the application of more advanced experimental techniques such as Electron Energy Loss Spectroscopy (EELS), Convergent Beam Electron Diffraction (CBED) and Atom Probe Tomography (APT) the quantitative analysis of local C concentrations in the RA has become achievable [30, 50, 51, 63-65]. It is accepted that RA of different morphologies, defined by the various phases neighbouring it, exist within the TRIP steel microstructure (Fig. 4) (i) as blocky grains enclosed by polygonal ferrite (ii) as islands within the granular bainite or at the interfaces between the polygonal ferrite and bainite (iii) as films between the BF laths/plates (iv) as blocky crystals enclosed by bainite/ferrite sheaves and (v) as an austenite/martensite constituent in granular bainite [30, 66, 67]. The RA crystals located within the polygonal ferrite are C enriched as a result of only austenite to ferrite phase transformation, whereas those within bainite benefit from the additional intake of C rejected from the BF. However, after the isothermal bainite transformation, the RA in bainite is also not homogeneously enriched with C; e.g. higher C films and lower C blocky austenite [40, 68]. In addition, and due to the presence of defects in austenite, the formation of C depleted zones was observed [64] which may have resulted in a partial transformation of the RA to martensite on straining (Fig. 4d). Furthermore, if the size of the RA blocks exceeds  $\sim 1\mu\text{m}$  then the typically available time for C diffusion is insufficient for its uniform distribution; especially in the centre of such grains. A wide variation in the C content in the RA crystals within the same microstructure together along with an inhomogeneous distribution of C within the RA (Fig. 4h) was observed experimentally using EELS and APT [50, 51, 58, 63, 65]. Scott and Drillet [63]

found that the C concentration varied between individual austenite grains in the range of 0.9-1.5%. As a result of such inhomogeneous C distribution between and within the RA crystals, those with lower C content tend to transform at low strains. Here the transformation will be more rapid the lower the C concentration and the larger the volume fraction of such RA crystals is in the microstructure. Thus, by controlling the fraction of the RA crystals of a particular morphology and size, the overall TRIP response could be improved upon. However, it should be kept in mind that chemical stability is only one of the factors influencing TRIP behaviour and may not be the most dominant controlling one.

## MECHANICAL STABILITY

The RA size [29, 32, 69, 70, 71], morphology and distribution [30, 32, 34, 59, 62, 71], orientation [18, 60, 73], the nature of neighbouring phases [16, 29, 30, 59, 62], defect structure [47, 48, 66], rotation [74] and stress state [18, 73, 75-77] control its mechanical stability.

### **Effect of grain size**

A number of experimental and modeling studies indicate that the stability of austenite is inversely proportional to its grain size such that the smaller the austenite grain size, the higher is its mechanical stability during straining. Coarse RA grains are unstable, transform to martensite at low strains [69, 70] and provide only a limited contribution to the total ductility. On the other hand, very fine grained RA does not transform to martensite even at fracture; thus making itself useless for the TRIP effect [30, 32]. Based on Electron Back Scattering Diffraction studies, Fig. 5a shows that RA grains before tensile testing had an average grain size of up to 2  $\mu\text{m}$  whereas the vast majority of the remaining RA grains had sizes  $<0.7 \mu\text{m}$  at fracture. If the role of chemical stability is disregarded, the grain size of the RA itself

restricts the number of available nucleation sites for martensite which is proportional to the surface area [78], hinders the plastic accommodation accompanying the martensite formation [79] and martensite plate/lath propagation, as well as controls the interfacial energy of martensite and by this affecting the  $M_s$  [80]. If austenite was subjected to significant deformation resulting in austenite subdivision by stacking faults, microbands, shear bands or by formation of subgrains, then the RA resistance against deformation-induced martensite transformation is increased due to: (i) the presence of areas that are free of effective heterogeneous nucleation sites [81,82], (ii) an increased number of matrix constraints to transformation [82], and (iii) mechanical constraint by the semi-coherent nature of the faceted ferrite/RA interface which is incapable of the inhomogeneous shear-type lattice invariant deformation necessary for a martensitic transformation to take place [83].

### **Role of Neighbouring Phases**

As CMnSi TRIP steels have a multi-phase microstructure, the properties, volume fractions, morphology of all phases, as well as interactions between the phases during deformation have a pronounced influence on the TRIP phenomenon [28, 30, 51]. These phases affect both chemical and mechanical stability of the RA. Although a significant enrichment of the RA with C takes place during austenite to polygonal ferrite (PF) formation, which is a function of the PF volume, the maximum amount of the RA that corresponds to excellent strength and ductility combinations is achieved at 40-50% of the PF [25, 49, 66]. For example, the presence of ~30% PF in thermo-mechanically processed NbMoAl TRIP steel resulted in ~8.5% RA, whereas 50% of the PF helped preserve ~12% of the RA in the microstructure. Compared to the former steel, the latter exhibited better elongation (~35% vs 21%) [51]. As seen in Fig. 5b, the RA located between the PF grains in thermo-mechanically processed TRIP steels transforms first at low strains due to both lower chemical stability and coarser

grain size. However, some relatively coarse RA grains remain untransformed among the PF grains at fracture [30]. Similarly, the high stability of submicron sized blocky RA located in the PF was reported by Chen *et al.* [42] and Tomota [62] whereas Sugimoto *et al.* [34] found that the RA grains in ferrite matrix were the least stable. As discussed previously, all the RA crystals that remain untransformed after fracture are overstabilised either due to high C content and/or small grain size. The RA grains that transform next are located at the interface between the PF and carbide-free bainite while the most stable are the fine layers of RA between the BF laths [30] (Fig. 5b). The size, morphology and C content of the RA located in the bainitic microstructure is affected by the BF lath/plate dimensions such that finer bainite results in similarly fine RA. However, the stiffness/strength and work hardening behaviour of the phases also define the kinetics of the RA transformation to martensite during straining. According to Radcliffe and Schatz [84], if the surrounding phase is hard, it exerts a hydrostatic pressure ( $\sigma_p$ ) on the RA such that:

$$\sigma_p = \frac{2}{3}(Y_o + 2H_o \varepsilon^*) + \frac{2}{3}Y_o \ln\left(\frac{E\varepsilon^*}{(1-\nu)Y_o}\right) \quad (8)$$

where  $Y_o$  and  $H_o$  are yield stress and strain hardening rate of the surrounding austenite matrix,  $E$  is Young modulus and  $\nu$  is Poisson ratio,  $\varepsilon^*$  is strain. Radcliffe and Schatz [84] postulated that an increase in the hydrostatic pressure by 100MPa results in an effective lowering of  $M_s$  by 6°C. Sugimoto *et al.* [34] explained that the much higher stability of the RA films located between the BF laths as compared to the lower stability of the blocky or acicular RA located in the PF is due to the prevalence of a much higher  $\sigma_p$  in the former case. However, if the RA is in contact with hard phases such as martensite or strong BF, which do not accommodate external stress, then the direct propagation of stress to the RA will increase the total driving force for it to transform to martensite [85].

Tomota *et al.* [62] linked (i) the transformation of high-C RA films to martensite with the stress localisation at the RA/BF interface resulting from the slip activity in the BF and (ii) the more stable behaviour of the lower C content RA in ferrite with the more uniform stress relaxation around the blocky/spherical RA. In the former case, the BF laths were parallel; whereas when they are randomly oriented, they provide a physical constraint to RA transformation [86]. Based on the above, the kinetics of RA transformation in three CMnSi thermo-mechanically processed TRIP steels with and without Nb, Mo and Al additions is considered (Steels 3-5, Fig. 3a). In Steel 4 alloyed with Nb and Mo, most of the RA transformed to martensite at 0.05 strain. However, ~ 2% of RA remained in the microstructure after tensile testing and was located predominantly between the BF laths. In this steel, the PF is rigid due to the solution strengthening effect from Si, Mo and Nb. While this should delay the rate of work hardening of PF and load transfer to the RA, an early transformation of the RA takes place due to the significant number of martensite crystals neighbouring the RA. On the other hand, the PF and bainite in Steel 3 alloyed with Nb and Al, were relatively soft since Al does not provide a strong solution strengthening effect. Although, the rate of the RA transformation to martensite was slower than in Steel 4 [51]), all of it transformed to martensite at low strains due to its very low chemical stability (0.7 wt.% C) and relatively fast load transfer from ferrite. The transformation of the RA to martensite in Steel 5 containing Nb, Mo and Al occurred gradually with ~4.5% of RA remaining in the microstructure at fracture. More RA was untransformed in Steel 5 compared to Steel 4 due to: (i) the higher C content in Steel 5 and, (ii) the majority of the BF laths having random orientations in Steel 5; contrary to the parallel arrangement of the BF laths in Steel 4. These results highlight the complexity of the RA behaviour during straining and its dependence on the properties and morphology of the surrounding phases which in turn, are defined by steel composition and processing, as well as on the interaction between all the phases.

### **The Effect of RA Orientation and Stress State**

Research addressing the effect of RA orientation on its stability is very limited. Using micromechanical modelling, Reisner *et al.* [18] showed that orientation affects the tendency to strain-induced transformation such that the formation of martensite variants having the largest mechanical driving force is favoured. Kruijver *et al.* [60], Blonde *et al.* [61] and Oliver *et al.* [73] have shown that grains with {200} planes oriented under  $0^\circ$  and  $90^\circ$  to the tensile direction are more susceptible to martensite transformation. On the other hand, Tirumalasetty *et al.* [74] showed that RA crystals that are fully embedded in ferrite and possess the highest Schmid factor first undergo a rotation that causes a lowering of their Schmid factor and consequently, delays their strain-induced transformation.

The conditions under which deformation takes place will either aid in or detract from the RA stability as the  $M_s^\sigma$  and  $M_d$  temperatures will differ because of the strong dependence of the mechanical driving force on the triaxiality of the applied stress. Modelling has shown that simulated pure shear (torsion) yields the highest mechanical driving force whereas plane-strain compression offers the lowest mechanical driving force for the austenite to martensite transformation [18].

In this regard, experiments have confirmed that RA is less stable under tensile loading than under compressive loading as the normal component (hydrostatic) of the applied load is positive in tension and negative in compression. Thus, by applying hydrostatic pressure during a tensile test, the stability of the RA could be enhanced and result in a higher uniform elongation [75]. On the other hand, since martensite transformation is associated with a volume increase, compressive loading suppresses it and stabilises the RA [7]. The strain-induced martensite transformation is retarded under plane-strain conditions in comparison

with tensile testing [76] whereas biaxial stretching results in faster transformation of the RA than under uniaxial tensile conditions [77].

## EFFECT OF PRIOR PROCESSING CONDITIONS

### **Deformed and Non-deformed Austenite**

In intercritically annealed steels or in steels rolled within the recrystallization region, the RA will have a low density of defects such as dislocations or shear bands, which are abundant in the RA deformed in the non-recrystallisation or intercritical regions. The work hardening of austenite makes the plastic accommodation of shape strain more difficult and lowers the  $M_s$  temperature [87]. Thus the RA which was deformed could be considered to be more stable than the non-deformed one. Moreover, the amount of deformation will affect the overall microstructure refinement; *e.g.* with an increase of deformation, the availability of heterogeneous nucleation sites in the austenite increases. This results in finer ferrite grains, accelerated ferrite formation, the formation of fine bainite and consequently, the refinement of the RA. An indirect effect of austenite deformation on the RA stability is also associated with the influence of deformation on the temperature of bainite transformation and thus, on the nature of the bainitic phase. However, these are nonlinear relationships; as shown in [29], at both relatively low and very high strains (0.47 and 3.0) applied in the non-recrystallisation region, carbide-free morphologies are formed whereas at intermediate strains, the carbide-containing ones were present. The nature of bainite is of critical importance for the formation of a required volume fraction of the RA which in turn, is capable of delivering a TRIP effect on straining. Fig. 6 shows that with an increase of deformation in either recrystallised or non-recrystallised regions from the optimum ones, the carbide-free bainite changes to upper bainite and is accompanied by a reduction of the RA amount. There are additional effects of deformation for the steels containing strong carbide-forming elements which are able to form

strain-induced precipitates. Strain-induced precipitation could play a two-fold role: (i) by pinning the dislocations, fine precipitates preserve the substructure and aid in the strengthening of austenite and by this, delaying the transformation to martensite and, (ii) reducing the chemical stability of the RA by making less C available in the solid solution. In addition, at high temperatures, the fast diffusion of alloying elements to grain and sub-grain boundaries or deformation bands may lead to the localised strengthening of austenite; thus delaying the transformation on straining.

According to Godet *et al.* [88], deformation in the intercritical region leads to a significant improvement in the mechanical properties compared to intercritically annealed steels without deformation. However, no changes in the volume fraction of the RA were observed nor was a change in its behavior recorded during subsequent straining at room temperature. With greater deformation, the strain hardening, tensile strength and total elongation increases. This has been attributed to the difference in the interaction of deformation-induced martensite transformation with annealed and recovered ferrite substructures. On the other hand, Basuki and Aernoudt [89] found that deformation in the intercritical region reduces the volume fraction of bainite, increases the volume fractions of ferrite and retained austenite and refines the RA while its C content remains high.

It is clear that the effect of deformation is a complex one and that there is no linear relationship between the amount of deformation and the stability of the RA. A summary of our current understanding into the effects of deformation is given in Fig. 7.

### **Effect of Isothermal Bainite Transformation Temperature and Holding Time**

The isothermal bainite transformation (IBT) temperature determines the morphology of the bainite formed and thus, the RA fraction varies strongly with the IBT temperature (Fig. 8). The formation of carbide-free bainitic phases aids in increasing the volume fraction of the RA as well as its stability. On the other hand, the formation of carbide-containing bainite (upper

or lower bainite) reduces the volume fraction and the chemical stability of the remaining RA as C is taken up by the cementite present in bainite. In general, depending on composition and prior processing, the volume fraction of the RA and the resultant total elongations will peak at a moderate IBT temperature (~350-450°C) [27, 42, 47, 49, 66, 84]. For thermo-mechanically processed 0.2C-1.5Mn-1.5Si (wt%) TRIP steel with and without addition of 0.037Nb (Figs. 8 and 9a), the optimum temperature for carbide-free bainite formation was 400-450°C; as higher or lower temperatures increased the volume fractions of upper or lower bainite, respectively. However, the addition of Al shifted the optimum temperature for carbide-free bainite formation to 470°C [50].

Typically, the volume fraction of RA first increases with holding time and then decreases (Fig. 9b) and this change is also reflected in the concurrent deterioration of mechanical properties [48, 57, 90, 91]. During the very early stages of holding, the initial low value of RA volume fraction with relatively low C content are associated with the beginning of bainite transformation. With increasing time, the incomplete bainite transformation progresses further and results in more C being rejected into residual austenite. This makes the RA more stable on cooling and thus, its volume fraction and C content reaches maximum values. At longer holding times, the decomposition of the austenite which is saturated in C into ferrite and carbides takes place and leads to a reduction in the RA volume fraction and its C content and an increase in the amount of martensite (Fig. 10). In situ observations of bainite transformation and RA stability [92] have also confirmed that with increase in holding time from 20 s to 1000s both, the volume fraction of the RA and its resistance to martensite transformation (higher C content) increases. The transformation of the RA with low C content to martensite on cooling was reported for holding times up to 240s. The fact that no decomposition of the RA was detected was probably due to insufficient holding time. However, when already processed TRIP steels [93] were subjected to heating up to 500°C, the

decomposition of RA was observed to have started from 350°C; with the initial formation of ferrite accompanied by C enrichment of the remaining austenite followed by the formation of carbides in the temperature range of 370-450°C and the depletion of C in the RA. C-depleted RA tends to transform to martensite upon cooling. Using the assumption that the films of the RA between the BF laths are more enriched in C compared to the blocky-type of the RA, synchrotron data by SahaPodder *et al.* [94] showed that during holding at 400°C, the films decompose into ferrite and carbides faster than the blocky RA. The effects of IBT parameters on the RA chemical stability and subsequent TRIP behavior are summarised in the schematic diagram shown in Fig. 11.

## CONCLUDING REMARKS

This chapter intends to provide an overview of the TRIP effect with a particular attention on its importance for modern TRIP-assisted multiphase steels with lean compositions. It shows the complexities related to the issue of RA stability with the interplay between several factors such as alloying elements and C, size, orientation, stress state and location of the RA, and properties of the neighbouring phases playing significant roles. The importance of creating a microstructure with the right combination of volume fractions and properties of all phases present has been highlighted. From this overview it is clear that the development of new models and the application of advanced experimental techniques in the last three decades have led to tremendous advances in our understanding of the TRIP phenomenon and the issue of RA stability. However, our understanding to-date has been limited by the deficiency of the majority of applied experimental techniques, using two-dimensional cross-sections of the microstructure. More importantly, differences in the behaviour of the grains at the surface and in the specimen bulk must be considered; as the exact morphology of the RA crystals in the bulk and the other phases with which it is coming into contact as its nearest neighbours

remains unknown. In future, 3D EBSD or a combination of tomography with 3D XRD could provide further insights into the interplay between the various phases.

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## FIGURE CAPTIONS

Fig. 1 Change in Gibbs free energy (a) and schematic diagram showing transformation and deformation modes of austenite (b) as a function of temperature.

Fig. 2 Schematic diagram of the factors affecting the stability of retained austenite.

Fig. 3 Effect of the initial average C content in RA on the kinetics of transformation during straining (a) and variations of the RA and C content as a function of strain in intercritically annealed CMnSi TRIP steel (Steel 6 in a) and in thermo-mechanically processed TRIP steel alloyed with Nb, Mo and Al (b).

Fig. 4 TEM (a-f) and C atom maps (g,h) images showing different morphologies of retained austenite in TRIP-assisted steels: (a) blocky in the PF; thin films between BF laths (b,h); lath with martensite and RA (d); blocky/plate at PF/Bainite interface (e); (f) twinned austenite

(TA); (g)- blocky RA. PF is polygonal ferrite, BF is bainitic ferrite, RA is retained austenite, M is martensite, TM is twinned martensite. Figures (b) and (c) are reprinted from [51] with permission from Trans. Tech. Publishing.

Fig. 5 Effect of grain size on the RA stability (a), and fractions of the RA at different locations in the microstructure as a function of strain (b).

Fig. 6 Effect of deformation on the morphology of bainite and volume fraction of RA.

Fig. 7 Schematic diagram showing the effects of deformation on RA volume fraction and stability.

Fig. 8 Effect of isothermal holding temperature on morphology of bainite in CMnSiNb steel: (a) - 550°C, (b) - 450°C and (c)- 350°C. Zone axis in (b) is  $[112]_{\gamma}$ .

Fig. 9 Effect of isothermal holding temperature (a) and time (b) on the RA volume fraction and carbon content.

Fig. 10 Effect of isothermal holding time on the microstructure of CMnSiNb steel (a) 600s, (b,e) 1800s, (c) 2400s, (d,f) 3600 s. Zone axis in (e) is  $[2\bar{1}\bar{1}]_c$ .

Fig. 11 Effect of Isothermal Bainite Transformation parameters on the RA behaviour.