

# Strain Partitioning & Mechanical Stability of Retained Austenite

Joo Hyun Ryu<sup>a</sup> Dong-Ik Kim<sup>b</sup> Hyoung Seop Kim<sup>c</sup>  
H. K. D. H. Bhadeshia<sup>a,d</sup> Dong-Woo Suh<sup>a</sup>

<sup>a</sup>*Graduate Institute of Ferrous Technology, POSTECH, Pohang 790-784, Korea*

<sup>b</sup>*Korea Institute of Science and Technology, Seoul, 136-791, Korea*

<sup>c</sup>*Department of Materials Science and Engineering, POSTECH, Pohang 790-784, Korea*

<sup>d</sup>*Materials Science and Metallurgy, University of Cambridge, CB2 3QZ, U.K.*

---

## Abstract

The mechanical stability of austenite in steels which rely on transformation-induced plasticity is usually attributed to its chemical composition, size and shape. We demonstrate here that another factor, the partitioning of strain between phases with different mechanical properties, can dramatically influence the stability.

*Key words:* TRIP steel, retained austenite, stability, strain partitioning

---

Many modern steels rely on a microstructure containing some retained austenite which transforms into martensite during the course of deformation [1,2]. This stress or strain-induced transformation leads to additional plasticity which enables strong steels to be formed into complex shapes. Critical in this behaviour is the mechanical stability of the austenite; if transformation is exhausted at small plastic strains then the protection against a necking instability is lost, and if the austenite is too stable then it does not contribute to strain hardening at stress concentrations.

The importance of the stability of retained austenite in the design of transformation induced plasticity (TRIP) steels is well understood and the chemical composition [3–8], size [9,10] and shape [11] of the austenite are the main factors that have been researched in this context.

In most cases, TRIP steels are only partly austenitic because this phase is retained without using expensive solutes by ensuring that the carbon that is

---

*Email address:* dongwoo1@postech.ac.kr (Dong-Woo Suh).

partitioned into the parent phase during the course of the bainite transformation does not precipitate as cementite [1,2]. The austenite therefore occurs in mixed microstructures where the majority phase is the mechanically softer allotriomorphic ferrite. During deformation, it is the softer phase which yields first and after appropriate work hardening, sufficient stress is transferred onto the harder regions to cause them to yield [12]. The distribution of plastic strains is therefore nonuniform [13]. In the present work we demonstrate that the heterogeneous distribution of plastic strain is an important factor in determining the mechanical stability of the retained austenite.

The experimental plan was to characterise *in situ*, the partitioning of plastic strain in two microstructures with similar retained austenite characteristics, but which in macroscopic tensile tests have exhibited quite different martensitic transformation behaviour as a function of strain, Fig. 1 [14]. These two samples are made of aluminium-containing TRIP steels [14], with chemical compositions (wt%) and designations as follows:

|      | C    | Mn  | Si   | Al  |
|------|------|-----|------|-----|
| L-Al | 0.12 | 4.6 | 0.55 | 1.1 |
| H-Al | 0.12 | 5.8 | 0.47 | 3.1 |

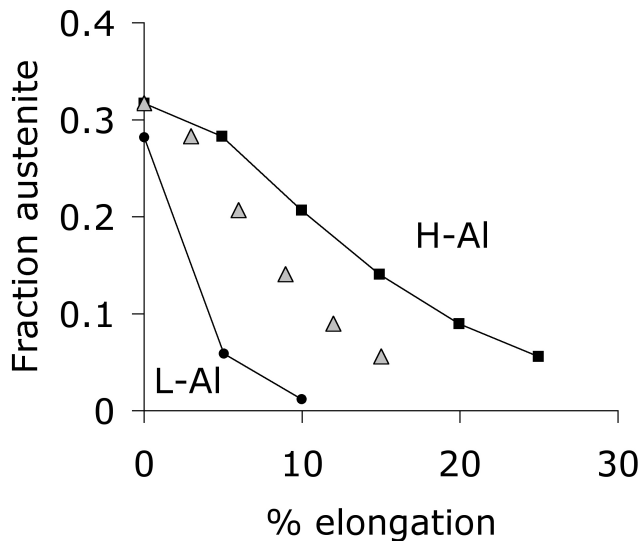


Fig. 1. Showing a remarkable difference in the mechanical stability of the austenite even though the chemical composition and size of the retained austenite have been verified to be similar [14]. The triangular points are discussed later in the text.

The details of the alloy preparation have been presented elsewhere [14]. Both alloys, after cold-rolling and heating to 720°C (L-Al) and 780°C (H-Al) for two minutes contained a volume fraction 0.3 of retained austenite on cooling to ambient temperature at 10°Cs<sup>-1</sup>. A field emission scanning electron microscope equipped with electron back scatter diffraction and an *in situ* tensile-test attachment was used for the experiments. Tensile test specimens with a gauge length of 15 mm and width 4 mm were mechanically polished with colloidal silica in the final stage. Four square dots of amorphous carbon were deposited

using a focused ion beam facility in a rectangle  $2 \times 2 \times 0.5 \mu\text{m}$  to mark the region observed in the scanning electron microscope during tensile testing.

The step size for EBSD measurements was 70 nm and tension was applied using a crosshead speed of  $5 \mu\text{m s}^{-1}$ , with images captured every 100  $\mu\text{m}$  of displacement. The strain distribution on the marked area in the scanning microscope images was analysed using *ARAMIS* software which uses digital image correlation.

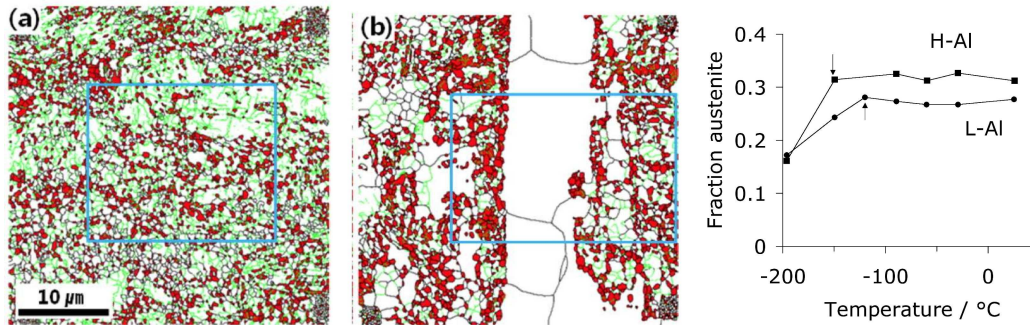


Fig. 2. EBSD phase mappings in which austenite is red and the rectangles indicate the areas designated for strain analysis. Deformation is applied along horizontal direction. Crystallographic misorientations in the range  $2\text{--}15^\circ$  are marked by green boundaries and larger deviations by black interfaces; (a) L-Al, (b) H-Al. (c) Thermal stability of the retained austenite. The arrows mark the onset of martensitic transformation.

Figs 2a,b show the phase maps for the two alloys, the details of which have been described elsewhere [14]; the important point to note is that the L-Al microstructure contains a uniform dispersion of ferrite and austenite, whereas that of H-Al has a bimodal size-distribution of ferrite grains, some of which are very coarse. Fig. 2c shows data from X-ray measurements done as a function of sub-zero temperatures; the two samples have similar martensite-start ( $M_S$ ) temperatures in the vicinity of  $-120^\circ\text{C}$  with a maximum difference of about  $30^\circ\text{C}$  due to the slightly higher manganese concentration in H-Al. The similar  $M_S$  temperatures are expected since the austenite has comparable composition and size [14].

It is evident from Fig. 3 which shows the microscopic strain distributions as a function of the macroscopic tensile elongation, that the distribution of strain is much more heterogeneous in H-Al, with the highest strains located in the coarse ferrite and that in the austenite is much lower than the mean strain. In other words, the apparent large mechanical stability of the austenite as illustrated in Fig. 1 is a reflection of the fact that the austenite experiences less deformation than the average elongation recorded on the tensile specimen.

These results could at least partly explain why steel-specific [15], or even generic theory [16], for the mechanical stability of austenite is not able to

explain the experimental data on TRIP-assisted steels to an accuracy of better than  $\pm 0.05$  volume fraction of austenite; this is a large discrepancy given that the fraction of retained austenite is of the order of 0.15. Such theories test experimental data based on the assumption that the plastic strain experienced by the austenite is identical to the overall elongation of a tensile-test specimen.

Judging from the data in Fig. 3b, the strain experienced by the austenite is a factor of 0.6 less than the overall elongation. Assuming that this is representative of all the austenite present in the H-Al steel, the triangular points in Fig. 1 represent the fraction of austenite as a function of the strain in austenite. It is seen that the discrepancy in the mechanical stabilities of austenite in H-Al and L-Al is substantially reduced when a comparison is made using the strain in austenite. We have made the reasonable assumption in this comparison that the distribution of strain in L-Al is homogeneous.

To summarise, it appears that the partitioning of strain between the different phases present in TRIP-assisted steels can lead to a misleading impression of the mechanical stability of retained austenite when comparisons are made between steels with substantially different microstructures, or when attempts are made to explain the stability in terms of chemical composition alone.

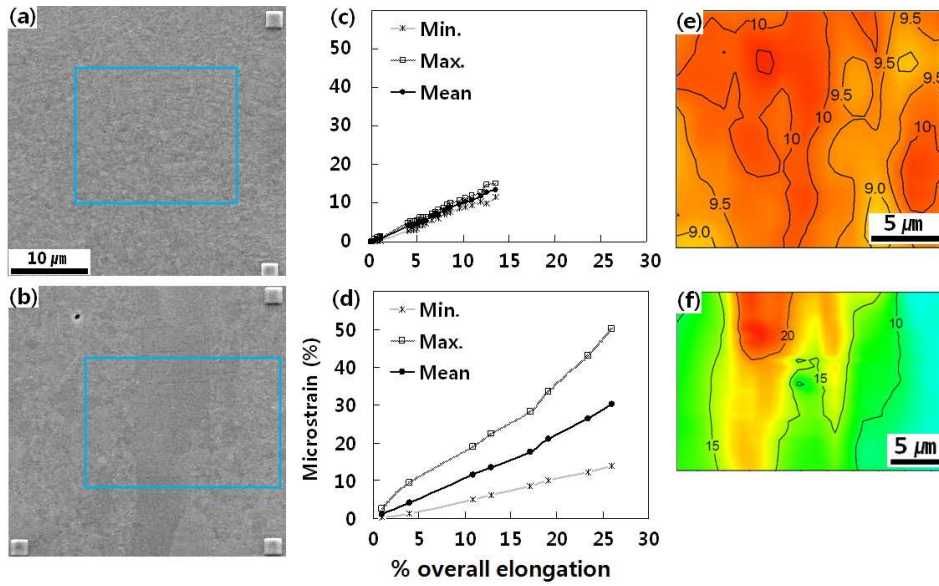


Fig. 3. (a,b) The maximum, minimum and mean microstrains recorded as a function of the elongation of the tensile sample, for samples L-Al and H-Al respectively. (c) The strain map at a tensile elongation of 9.6% in sample L-Al and (d) the corresponding map at a tensile elongation of 12.8% in sample H-Al.

## References

- [1] P. J. Jacques: *Current Opinion in Solid State and Materials Science* 8 (2004) 259–265.
- [2] B. DeCooman: *Current Opinion in Solid State and Materials Science* 8 (2004) 285–303.
- [3] R. Le-Houillier, G. Begin, A. Dubé: *Metallurgical Transactions* 2 (1971) 2645–2653.
- [4] C. K. Syn, B. Fultz, J. W. Morris, Jr.: *Metallurgical Transactions A* 9 (1978) 1635–1640.
- [5] D. Duchateau, M. Guttman: *Acta Metallurgica* 29 (1981) 1291–1297.
- [6] M. Takahashi, H. K. D. H. Bhadeshia: *Materials Transactions JIM* 32 (1991) 689–696.
- [7] Y. Tomita, T. Okawa: *Materials Science & Engineering A* 172 (1993) 145–151.
- [8] D. W. Suh, S. J. Park, C. S. Oh, S. J. Kim: *Scripta Materialia* 57 (2007) 1097–1100.
- [9] W. C. Leslie, R. L. Miller: *ASM Transactions Quarterly* 57 (1964) 972–979.
- [10] D. Q. Bai, A. D. Chiro, S. Yue: *Materials Science Forum* 284–286 (1998) 253–260.
- [11] H. K. D. H. Bhadeshia, D. V. Edmonds: *Metal Science* 17 (1983) 411–419.
- [12] Y. Tomota, K. Kuroki, T. Mori, I. Tamura: *Materials Science and Engineering* 24 (1976) 85–94.
- [13] Q. Furnémont, G. Lacroix, S. Godet, K. T. Conlon, P. J. Jacques: *Canadian Metallurgical Quarterly* 43 (2004) 35–42.
- [14] D. W. Suh, S. J. Park, T. H. Lee, C. S. Oh, S. J. Kim: *Metallurgical & Materials Transactions A* 41 (2010) 397–408.
- [15] K. Sugimoto, N. Usui, M. Kobayashi, S. Hashimoto: *ISIJ International* 32 (1992) 1311–1318.
- [16] M. Sherif, C. Garcia-Mateo, T. Sourmail, H. K. D. H. Bhadeshia: *Materials Science and Technology* 20 (2004) 319–322.