

Retained Austenite Stability and Impact Performance of Advanced High Strength Steel at Reduced Temperatures

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ABSTRACT

Retained austenite stability to both mechanically induced transformation and athermal transformation is of great importance to the fabrication and in-vehicle performance of automotive advanced high strength steels. Selected cold-rolled advanced high strength steels containing retained austenite with minimum tensile strengths of 980 MPa and 1180 MPa were pre-strained to pre-determined levels under uniaxial tension in the rolling direction and subsequently cooled to temperatures as low as 77 K. Room temperature uniaxial tensile results of pre-strained and cooled steels indicate that retained austenite is stable to athermal transformation to martensite at all tested temperatures and pre-strain levels. To evaluate the combined effects of temperature and pre-strain on impact behavior, stacked Charpy impact testing was conducted on the same 980 MPa minimum tensile strength steel following similar pre-straining in uniaxial tension. A reduction in absorbed energy was observed with decreasing temperature and increasing pre-strain, indicating that thermal effects on plasticity, not athermal transformation to martensite, predominantly account for the observed reductions in impact energy at reduced temperatures.

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INTRODUCTION

Advanced and ultra-high strength steels are used increasingly in the construction of automotive body structures for both enhanced passenger safety and as part of larger lightweight vehicle strategies [1–3]. Typical and projected future applications of advanced steels include body structure components that require combinations of high strength and high ductility to both reduce intrusion and absorb energy during a crash event, e.g., B-pillar reinforcements, A-pillar reinforcements, tunnels, and roof rails, etc. Most importantly, components must retain this capability following forming. To overcome formability and energy absorption related challenges within the structure, advanced steels with tensile strengths exceeding 1000 MPa have been developed with the inclusion of metastable retained austenite [4–7]. The metastable retained austenite within the microstructure of newly-developed grades results in enhanced tensile elongation relative to previous advanced high strength steel grades of similar tensile strength as a consequence of the strain-induced transformation of retained austenite to martensite. Therefore, the relative stability of the metastable retained austenite phase to martensite transformation during vehicle operation is of great importance to the implementation and performance of retained austenite bearing advanced high strength steels.

The stability of retained austenite in a formed steel component can be considered with respect to either further plastic deformation, i.e., stability to further strain-induced transformation, or to cooling below an effective martensite start temperature, i.e., athermal martensite formation, for remaining austenite in the microstructure [8–13]. Previous investigations explain in varying detail the relative chemical and structural contributions to austenite stability in as-produced steels, but additional investigation is required to properly characterize the combined effects of deformation and temperature excursions during operation [14]. For a steel component to maintain consistency of performance within the body structure, properties of interest should be consistent to temperatures as low as -40 °C, and, therefore, athermal martensite generated by cooling to this temperature must be understood.

In this study, three selected retained austenite bearing sheet steels with minimum tensile strengths of 980 MPa and 1180 MPa were evaluated to determine retained austenite stability as evidenced by uniaxial tension and Charpy impact testing following combined pre-straining and cooling to temperatures to 233 K (-40 °C) and 77 K. Various retained austenite morphologies were chosen to determine if structural contributions to as-produced retained austenite stability are observable under tested conditions. Pertinent details of the production of quench and partition and TRIP-aided bainitic ferrite steels are

available in literature [4–7]. Cooling to 77 K prior to mechanical testing was included as an extreme assessment of the stability of retained austenite to athermal martensite formation, and to serve as a baseline for comparison to the conditions cooled to -40 °C. Results from combined pre-straining and cooling were then compared to those obtained from room temperature (298 K) testing, and a general assessment is made of the retained austenite stability characteristics for the selected alloys.

EXPERIMENTAL PROCEDURES

Materials and Sample Preparation

The three retained austenite bearing advanced high strength steels selected for evaluation from various steel suppliers are listed in Table 1 with general alloy chemistry ranges as specified by General Motors' worldwide specification, GMW17627. Minimum mechanical properties for the selected steels are 980 MPa ultimate tensile strength - 20% total elongation and 1180 MPa ultimate tensile strength - 13% elongation as tested according to ASTM E8 for the 980 RA and 1180 RA steels, respectively. The three selected steels were commercially-produced and received in the uncoated condition at a nominal thickness of 1.2 mm for uniaxial tensile investigations. Additional uncoated 1.8 mm thick steel was obtained for the 980 RA steel for use in Charpy impact testing.

Table 1. Chemical composition ranges (in wt. %) of selected steels retained austenite containing steels.

	C max.	Mn max.	Si max.	Al max.	N max.	Cr+Ni+Mo max.
980 RA	0.30	3.5	2.5	2.0	0.008	1.0
1180 RA (Q&P and TBF)	0.35	3.5	2.2	2.0	0.008	1.0

Samples were sectioned from the as-received steels for metallographic preparation using standard practices to a final step of 1 µm diamond suspension. Microstructural characterization was performed on etched samples (2% nital) using scanning electron microscopy (secondary electron imaging mode) at a magnification of 10,000x.

Tensile specimens were fabricated according to the geometry outlined in ASTM E8 (50 mm gauge length) for the three selected alloys at 1.2 mm by water jet cutting specimens with the tensile axis oriented parallel to the rolling direction. All tested conditions as subsequently described, i.e., each alloy at each pre-strain and cooling combination, were produced and tested in duplicate. Samples were retained without pre-strain or pre-strained to the levels of 2%, 5%, and 10% total (including elastic) engineering strain in uniaxial tension. Following pre-straining, samples from each pre-strain level were then either retained at room temperature or cooled to 233 K and held for 15 mins or cooled to 77 K by immersion in liquid nitrogen (LN2) for 5 mins. Following a 36 hr hold time at room temperature, a total of 48 samples (2 per each condition) were then tested to failure in uniaxial tension using a screw-driven tensile frame according to the standard method outlined ASTM E8. Strain rates were controlled to $\dot{\epsilon} = 0.004 / s$ for the first 2% of engineering strain followed by an increase to $\dot{\epsilon} = 0.01 / s$ for 2% engineering strain (from previously applied

pre-strain condition) to failure. Stresses for reloaded samples following pre-strain were subsequently normalized for proper representation based upon original cross sectional dimensions, and strains were adjusted to correspond to the proper pre-straining level. Here, and modulus of elasticity of 205 GPa was assumed for all samples.

Charpy impact specimens for 980 RA steel were fabricated similarly to the tensile specimens described above. Following initial blanking of as-received 1.8 mm sheet to 25 mm x 150 mm coupons, both pre-straining and cooling were applied as previously described. Again coupons were pre-strained along the material rolling direction. For Charpy impact specimens, however, pre-strain levels of 0% and 5% (excluding 2% and 10%) were employed and cooling was done only to 77 K (excluding 233 K) for comparison to samples pre-strained and retained at room temperature. The stacked Charpy V-notch impact test followed the standard test procedure per ASTM-E23, except that three identical sheet coupons are mechanically riveted together to make a sub-sized compact with nominal thickness of 5.4 mm. The notch of Charpy specimens was oriented transverse to the rolling direction of the tested steel. All stacked Charpy V-notch tests were conducted on a SATEC impact tester with a maximum energy of 400 J at the following temperatures: -196 °C, -125 °C, -75 °C, -40 °C, -25 °C, 0 °C, +25 °C and +75 °C (77 K – 348 K). The standard test procedure, as described in ASTM E23, was followed. Three replicates were tested at each condition. Additional details of the sample preparation method and Charpy impact specimen geometry are available in a previous study by Wang *et al.* [15].

Table 2 displays, in tabular form for clarity, the pre-strain and cooling temperature combinations of samples prepared for uniaxial tension testing for each selected steel and Charpy impact testing for the 980 RA steel only.

Table 2. Pre-strain and cooling combinations for steels tested in uniaxial tension (○) and both uniaxial tension and Charpy impact (⊗).

	0%	2%	5%	10%
RT (298 K)	⊗	○	⊗	○
-40 °C (233 K)	○	○	○	○
LN2 (77 K)	⊗	○	⊗	○

RESULTS AND DISCUSSIONS

Figure 1 shows the microstructures of the as-received 980 RA and two 1180 RA steels. All contain appreciable fractions of bainite and/or martensite of lath-type morphology with primary ferrite visible and designated accordingly. Retained austenite is visible both in inter-lath and blocky morphologies, and this blocky character is more visible in the 980 RA steel in comparison both the 1180 RA steels. Additionally, the relative microstructural feature size of the 1180 RA (Q&P) steel is smaller than that of the 1180 RA (TBF) steel.

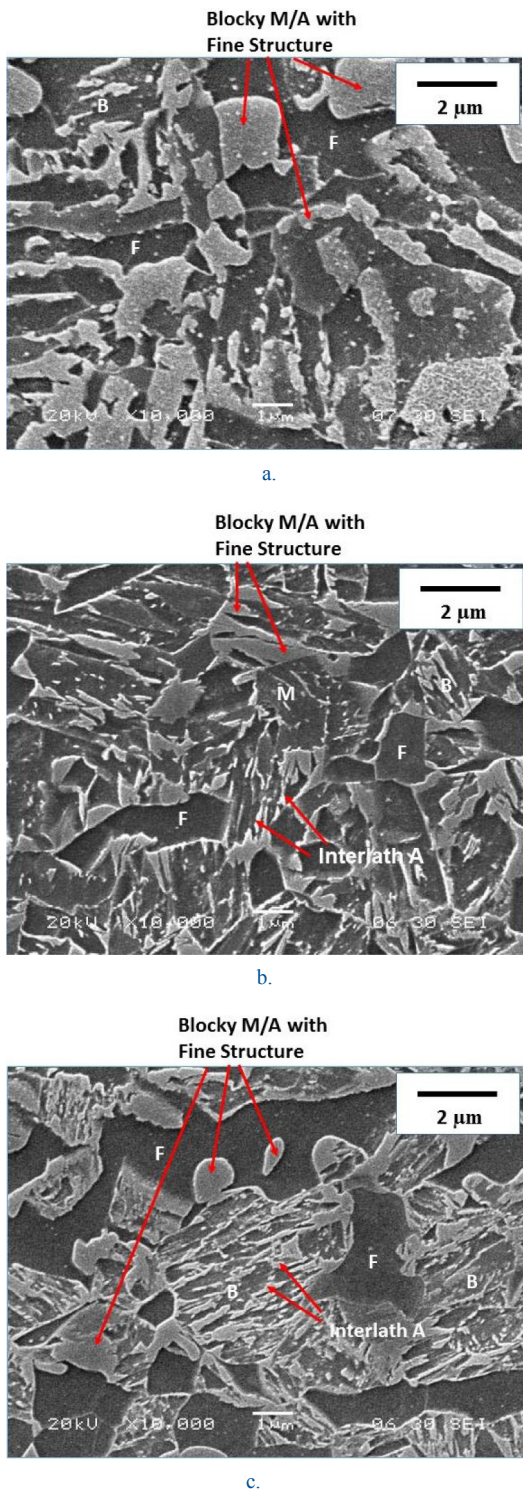


Figure 1. Scanning electron micrographs for (a) 980 RA, (b) 1180 RA (Q&P), and (c) 1180 RA (TBF). All micrographs were taken at 10,000× magnification. Note: A = Retained Austenite; B = Bainite; F = Ferrite; M = Martensite.

Results for steel specimens tested to failure in uniaxial tension following the previously outlined preparation methods (see Figure 1) are shown in Figure 2, Figure 3, and Figure 4 for 980 RA, 1180 RA (Q&P) and 1180 RA (TBF), respectively. An absence of significant effects on flow stress due to cooling is demonstrated for all alloys upon reloading following cooling to 233 K and 77 K for the 0%, 2%, 5%, and 10% pre-strained sample. By comparing selected and

representative pre-strained samples that have been cooled to those retained at room temperature as shown in Figure 5, it is evident that the low temperature thermal treatment has resulted in little to no athermal martensite formation at any pre-strain level for any of the tested alloys. Results for the 980 RA and 1180 RA (Q&P) shown in Figures 5(a) and 5(b), respectively, indicate no significant change in tensile flow behavior due to cooling. Figure 5(c), however, shows variation in total elongation within the presented sample set, but no clear trend with respect to an effect of thermal treatment is evident. Noticeable yield point differences exist upon comparison of the 0% pre-strained sample (initial quasistatic tensile testing to failure) and the yielding behavior of the reloaded samples, and this difference is attributed to room temperature strain aging following pre-straining as noted in Figure 2(a). Artifacts in the tensile flow curves are also present due to increases in the quasistatic strain rate at engineering strains of 2% upon loading – this behavior is also noted in Figure 2(a). Additionally, subtle increases in the combination of total elongation and flow stresses are observed upon reloading for samples pre-strained to 5%, and 10% for the 980 RA steel for all tested cooling temperatures. The same increases are apparent for both 1180 RA steels, especially the 1180 RA (TBF), for samples pre-strained to 2% and 5%. Similar observations have been previously reported for the room temperature aging response of 980 MPa minimum tensile strength quench and partition steels following pre-strain of 5% [16]. Samples of 1180 RA pre-strained to 10% and retested display a noticeable decrease in total elongation relative to uniaxial tests to failure without pre-straining. This observation is attributed, again, to room temperature aging; however, the aging effect of increased elongation following pre-strain is thought to be an effect of carbon segregation to and further stabilization of remaining austenite in pre-strained samples. The 10% pre-strained samples of 1180 RA steels likely retain little austenite following initial straining, and, therefore, would accommodate carbon diffusion at room temperature in the traditional manner resulting in reduced elongation upon reloading. For clarity and to further illustrate the observed changes in elongation and flow stress presumably due to strain aging, both the average ultimate tensile strength and average total elongation of each tested condition are shown in Figure 6 as a function of pre-strain.

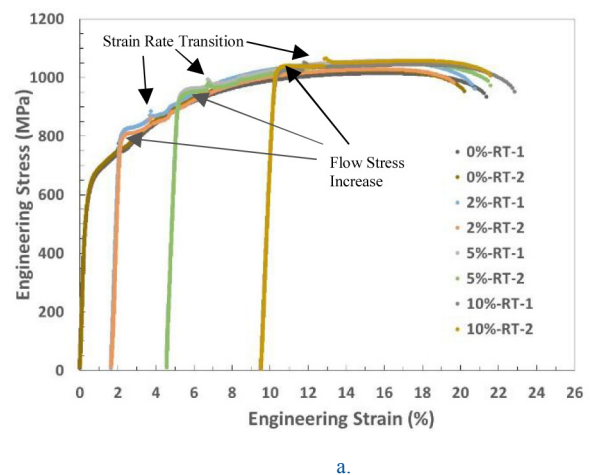


Figure 2. Engineering stress-strain curves for 980 RA steel pre-strained at 0, 2, 5, and 10%, followed by (a) 36 hr. room temperature aging, (b) a 15 min soak at 233 K (plus 36 hr room temperature aging), and (c) a 5 min soak at 77 K (plus 36 hr room temperature aging) prior to retesting. Duplicate samples were tested per condition.