On Local Formability/Ductility of New Advanced High-Strength Steels: Temperature, Bake Hardening and Strain Rate Effects

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ABSTRACT

New advanced high-strength steels (AHSSs) based on different microstructure-based concepts have been developed in the past decades to meet various needs in the automotive industry. Nevertheless, cracking issues in many practical forming and crash scenarios still limit the applications of these new AHSSs, despite their relatively high tensile elongations measured under laboratory test conditions. Accordingly, and inspired by the concept of the Local-Global Formability Diagram, this work started from comparing true fracture strain (TFS), a typical representative parameter of the local formability/ductility, of different AHSS sheet-metal grades. Furthermore, possible TFS evolution by some critical application-based effects, such as temperature, bake hardening, and strain rate was investigated. Explanations and discussions of the material mechanisms based on the test results were also presented in this work. The ultimate objective of this study is to highlight the importance of local formability/ductility and provide the corresponding references to the future AHSS development strategy.

INTRODUCTION

Steel Strength-Ductility Chart

As a ubiquitous illustration of visualizing various automotive sheet steel grades, the Steel Strength-Ductility Chart, also known as the ‘Banana Chart,’ has been popularly quoted by both industry and academia to facilitate steel comparisons and selections. Slightly evolved from the original version, the abscissa of the current ‘Banana Chart’ is based on ultimate tensile strength (UTS) values, and the ordinate is based on total elongation (TE) values, both of which can be directly obtained from engineering stress-strain curves. For AHSSs in particular, the ‘Banana Chart’ not only highlights the variety of the AHSS family, but also aids to categorize different AHSS generations based on either the microstructures or the UTS × TE values. Furthermore, the empty space in the ‘Banana Chart’ directs the development opportunities of new AHSSs of further enhanced tensile properties. Accordingly, a representative microstructure-based concept highlighted the importance of retaining austenite, which not only possesses a sufficient combination of ductility and strength, but also transforms to extra-high-strength fresh martensite during plastic deformation to retard strain localization, namely the transformation-induced plasticity (TRIP) effect. Additionally, by controlling the austenite stabilities, multiple strength-ductility combinations can be possibly achieved for different steel development objectives. Inspired by this concept, multiple new AHSS grades have been developed and commercialized by the steel industry in the past decades. A representative grade is quenched-and-partitioned (Q&P) steel, which was developed based on a carbon-stabilized austenite retaining technique during steel processing, rather than the costly conventional means of heavily alloyed chemical compositions in austenitic steels. A precedent work compared the representative stress-strain curves of two Q&P steels (Code 1 and 2) with two Dual Phase (DP) steels (Code 3 and 4) and two metastable austenitic steels (Code 5 and 6) as shown in Figure 1(a), and accordingly their distributions in the ‘Banana Chart’ in Figure 1(b). Each of the two Q&P steels contains about 14% retained austenite in the microstructure, while the DP steels do not contain any austenite and each of the austenitic steels contains more than 90% austenite yet of different stability for comparison. Indeed, the more austenite in the microstructure, the higher UTS × TE values, and the more favorable positions in the ‘Banana Chart.’
Figure 1. (a) Representative engineering stress-strain curves of two Q&P steels (Code 1 and 2), compared with two DP (Code 3 and 4) and two metastable austenitic steels (Code 5 and 6), and (b) their distributions in the ‘Banana Chart’

Local-Global Formability Diagram

Nevertheless, the ‘Banana Chart’ simplifies many critical characteristics of steels, such as the local and global formability. In fact, some early empirical literature already distinguished DP980 with enhanced stretch flange-ability from DP980 with enhanced TE as two different sub-grades to meet the corresponding applications. Accordingly, Hance introduced the Local-Global Formability Diagram, also known as the ‘Hance Diagram,’ which was composed of true uniform strain (TUS) as the abscissa to represent the global formability, an indicator of material necking resistance, and true fracture strain (TFS) as the ordinate to represent the local formability, an indicator of material fracture resistance. Unlike the TUS that can be directly obtained from a true stress-strain curve, the TFS can be represented by major strain, thickness strain, or effective (plastic) strain at fracture, and requires a procedure of measurement and derivation. A precedent work studied and compared two commonly-used TFS measurement-derivation methods based on either digital image correlation (DIC) or ASTM E8 standard, and accordingly proposed a new hybrid method to avoid the drawbacks of these two conventional methods. All the TFS results in this work are represented by the von Mises effective strain at fracture and concluded via this hybrid method.

For AHSSs ranked orderly in the ‘Banana Chart,’ their distributions in the ‘Hance Diagram’ can be reshuffled, especially along the local formability axis. For example, the six selected AHSS grades from Figure 1 are replotted in Figure 2 based on their anisotropic local and global formability, which illustrates such a reordered result. Notably, Code 5 exhibits the highest global yet the lowest local formability among the six AHSSs, Code 3 exhibits the lowest global yet the highest local formability, while Code 1 possesses a balanced formability property. Such a comparison proves no correlations between either the local and global formability, or the local formability and the retained austenite contents.
Nevertheless, a positive correlation was experimentally identified between the TFS/local formability with many application-based performances, such as hole-expansion ratio (HER), fracture toughness, bendability, edge-crack sensitivity (flangeability), and damage tolerance\textsuperscript{1,14}. In other words, based on the TFS results from the fundamental tensile tests, it is possible to estimate and compare some critical mechanical performances of different AHSS grades in the automotive applications. Take the hole-expansion performance and the six selected AHSS grades from Figure 1 for example. As shown in Figure 3, the higher TFS results from the uniaxial tensile tests, generally the higher HER results based on the ISO 16630 standard tests. Notably, the anisotropic local formability in Figure 2 shows that all these steels are comparatively more vulnerable to fracture in the T direction, and in the actual hole expansion tests, the initial cracks are indeed typically observed in the T direction. Therefore, here the TFS results from the T-direction tensile test samples are plotted together with the HER results to indicate the correlation between them. A similar study in the literature\textsuperscript{16} also highlighted such a correlation based on more AHSS grades, despite the local formability being represented by the thickness strain there.

Figure 2. The six selected AHSSs in the ‘Hance Diagram’ and their anisotropic formability along the longitudinal (L), diagonal (D), and transverse (T) directions relative to the sheet metal rolling orientation

Figure 3. A positive correlation between the HER and the TFS-T results
Scope of Work
Essentially, either the ‘Banana Chart’ or the ‘Hance Diagram’ is based on highly simplified material tensile test data, which are defaulted to be characterized in a laboratory circumstance at a quasi-static uniaxial strain rate under room temperature. Nevertheless, in the actual automotive applications, either the typical stamping strain rate (order of $10^1$ s$^{-1}$) or the crash strain rate (order of $10^3$ s$^{-1}$) is much higher than the quasi-static strain rate (order of $10^{-3}$ s$^{-1}$), let alone the large amount of adiabatic heating from the AHSSs plastic deformation that can rapidly elevate the local temperature. A precedent study investigated the temperature and strain rate effects on the tensile properties of the six selected AHSSs in Figure 1, compared their evolving distributions in the ‘Banana Chart,’ and highlighted the sensitivity of the austenite in the new AHSSs to these effects. As a subsequent study, this work focuses on the temperature and strain rate effects on the TFS/local formability. In addition, as a branch direction of the temperature effect, the bake hardening effect on the TFS is also investigated in this work. Correspondingly, the representation of the TFS shall be generalized from the local formability to local ductility and becomes the indicator of the material fracture resistance in both the forming and crash scenarios.

As for the target materials, Code 1 (Q&P980-GI) and 3 (DP980-GI with low carbon equivalent, LCE) are selected in this study, also as a continued effort after the precedent publications. Their representative microstructures are shown in Figure 4. Comparatively, the Q&P980 possesses a multi-phase microstructure of ferrite, austenite, and martensite, while the DP980LCE consists of mainly ferrite and martensite. Based on either the UTS × TE values or the microstructures, the Q&P980 belongs to the third-generation AHSS category, while the DP980LCE is theoretically just a conventional first-generation AHSS grade. Nevertheless, the DP980LCE was developed based on a metallurgical concept of reducing the carbon content in its chemical composition to decrease the hardness difference between its ferrite and martensite grains, and thereby exhibit enhanced local formability as shown in Figure 2. In contrast, the retained austenite in the Q&P980 on the one hand indeed increases the tensile elongation via the TRIP effect, but on the other hand, transforms to fresh martensite during plastic deformation and thereby increases the hardness difference in the microstructure, which tends to create micro-voids that can easily coalesce to cracks. Notably, as an effective austenite stabilizer, carbon is relatively rich in the austenite in the Q&P steels, and hence can further harden the fresh martensite transformed from the austenite during the plastic deformation, which consequently accelerates the ‘martensite-matrix de-cohesion’ to become micro-voids.

![Figure 4. Representative microstructures of (a) Q&P980 and (b) DP980LCE](image)

TEMPERATURE EFFECTS

Experimental Procedure
In this study, both the temperature-dependent tensile testing procedure and sample geometry are consistent with the precedent publications. Only L-direction ASTM E8 standard geometry samples of at least three repeats were tested at a nominal strain rate of 0.001 s$^{-1}$ in this section. Regarding the testing setup, concisely, an environmental chamber was fitted on an electromechanical universal testing machine. This chamber can create a quasi-isothermal inner environment of up to 250°C using convection heating. During a test, in-situ temperature monitoring was based on not only thermocouple readings, but also sample surface temperature imaging via an infrared camera through a side window of the chamber. Two customized grips were designed for fast loading in extreme temperature testing and exert the tension load on the samples through shoulders. In addition, a stereo DIC system was applied to monitor the sample deformation through a glass window on the chamber door. The glass window and black-body radiation emitted at elevated temperatures are expected to distort the captured images to...
some extent in the DIC measurements, and the relative in-depth research on analyzing and offsetting the consequent errors has been published in a precedent work. The DIC system includes a pair of four-megapixel cameras, which are operated at an adjustable frame rate of a maximum of 672 Hz to record sufficient temporal resolution of strain-field evolution. The testing machine and the DIC system were triggered simultaneously at the beginning of each test, and the force signals measured in situ by the load cells were continuously transferred into the DIC software to synchronize with the corresponding images during the testing. In this way, the force-displacement curves can be plotted in the DIC software, and accordingly, other significant values representing material characteristics can be post-processed.

Results and Discussions
The experimental results of the temperature effects on the global and local formability of the two target materials are shown in Figure 5. Generally, and comparatively, the DP980LCE exhibits higher TFS, yet its tensile elongation (either the UE or the TE) is much lower than the corresponding properties of the Q&P980. In addition, none of the plotted properties exhibits a linear tendency with the elevating temperature. Especially at 150°C, both materials exhibit a valley of their tensile elongation yet discrepant TFS tendencies. In contrast, at 250°C, their tensile elongation properties are at peaks, while their TFS levels significantly drop down. Such wavy temperature dependency can be attributed to multiple effects, either favoring or opposing, which affect the material formability dynamically in different temperature ranges.

• Thermal softening on both materials: when the temperature rises, the increased external thermal energy boosts mobile dislocations to diffuse through grain boundaries, and thereby this effect is consistently favoring the tensile elongation, while its impact on the TFS is unclear.

• Dynamic strain aging (DSA) on both materials: at the applied strain rate in this study, this effect is explicitly active at 100-150°C in the two target materials. A typical DSA manifestation is Portevin-Le Châtelier (PLC) banding, which contributes to serrated flow stress on Curve 3 and 4 in Figure 6(a) and Curve 4 in Figure 6(b). Compared with the DP980LCE, the Q&P980 contains more carbon as the austenite stabilizer in the processing, so its Cottrell atmosphere (carbon atom clusters) is easier to form and ‘arrest’ mobile dislocations, and consequently shorten its tensile elongation more rapidly. Such a mechanism can also explain the TFS decline of the DP980LCE at 100-150°C, owing to the early strain localization and piled dislocations inside the PLC bands. Nevertheless, for the Q&P980, the TFS peak at 150°C implies the DSA-induced suppression is overcome by a unique favoring effect, which is, the suppressed martensitic transformation.

• Martensitic transformation in the Q&P980: according to the Olson-Cohen theory, when the temperature is elevated, the plasticity-induced transformation from austenite to martensite gradually decays, due to its increased inducing stress, which, at a critical temperature, eventually exceeds the UTS of the material and consequently terminates the transformation. According to the literature, for TRIP and Q&P steels, such a critical temperature is typically around 100°C. Nevertheless, discrepant from the Olson-Cohen theory, the martensitic transformation can be reactivated at around 200°C, which so far has only been observed on TRIP and Q&P steels, yet the corresponding explanations are still not unified. Based on a series of interrupted tensile tests and X-ray diffraction inspections, Figure 7 illustrates the volume fraction of retained austenite at multiple plastic strain levels at different temperatures. At 100-150°C, the transformation significantly slows down (rather than ‘be terminated’), and thus few fresh martensite grains of high hardness are nucleated, which creates minimum ‘martensite-matrix de-cohesion’ and consequently elevates the TFS level to the peak as shown in Figure 5(a). Vice versa, the reactivated transformation at 200-250°C resumes the fresh martensite nucleation during plastic deformation and to some extent contributes to the TFS decline. Precipitation strengthening effect on both materials above 250°C leads to the accentuated TFS decline in Figure 5. Inspired by the literature, carbides precipitating into the martensite in this temperature range once again increases the hardness difference among the multi-phases.
Figure 5. Temperature effects on the global and local formability/ductility of (a) Q&P980 and (b) DP980LCE

Figure 6. Temperature-affected representative engineering stress-strain curves of (a) Q&P980 and (b) DP980LCE
BAKE HARDENING EFFECTS

Experimental Procedure

As an application-inspired branch topic of the temperature effects, the impacts of the paint-baking cycles on the target steels are investigated in this section. Here the concept of ‘formability’ needs to be generalized to ‘ductility,’ since the paint-baking cycles are applied after the forming procedures in the actual vehicle manufacturing. In this work, the test procedure generally follows the DIN EN 10325 standard: 1) pre-straining the tensile samples (ASTM E8 standard geometry in this work) to 5% (instead of 2% in the standard) major strain level, 2) baking the pre-strained samples in a temperature chamber at 170 ± 2°C for 20 ± 0.5 minutes, and 3) tensile testing the bake-hardened samples to fracture. For each material, one L-direction sample with at least two replicates was tested at a nominal strain rate of 0.001 s⁻¹. Additionally, the DIC system was deployed to not only measure the strain evolution, but also control the pre-straining levels with spring-back compensation via communicating the in-situ strain signals with the test frame controller.

Results and Discussions

The comparative engineering stress-strain curves of the two target materials are shown in Figure 8. In particular, the TE results of both materials are decreased by the pre-straining plus bake-hardening (PS+BH) cycle. Due to the remained austenite with the TRIP effect after the PS+BH cycle, the Q&P980 exhibits a yield-point-elongation (YPE) stage and then a brief strain hardening before fracture, while the DP980LCE exhibits a direct yield-to-fracture procedure after the cycle due to local residual and ferrite-martensite shear stress. In contrast to the decreased TE, the TFS values are not affected by the PS+BH procedure as shown in Figure 9. Such a phenomenon was also reported briefly in the literature yet without detailed explanations. Essentially, static strain aging (SSA) is the primary material mechanism of the bake hardening effect. Corresponding to the three test steps, 1) the pre-straining generates many dislocations, 2) the baking boosts the carbon atoms to gather around the dislocations to form the Cottrell atmosphere due to energy equilibrium, and 3) the hindered dislocations need extra external stress to move in the continued tensile testing. Similar to the DSA introduced in the last section, the SSA can effectively accelerate the necking but hardly trigger the early fracture.
Figure 8. Comparison between the as-received and the pre-strained plus bake-hardened (PS+BH) engineering stress-strain curves of (a) Q&P980 and (b) DP980LCE

Figure 9. Discrepant bake-hardening effects on the TE (left) and TFS (right)

**STRAIN-RATE EFFECTS**

**Experimental Procedure**

Regarding the strain-rate dependent tests, the test system, procedure, and sample geometry were adjusted based on the ISO 26203-2 standard and have been introduced in the precedent work. Briefly, a servo-hydraulic high-speed test system was applied to perform a maximum 160 kN axial load at a maximum 20 m/s speed in either tension or compression direction. The tension test fixture based on this system consists of a static grip, which is connected to a piezo-electric load washer mounted on the base of the system, and a dynamic grip with a slack adaptor, which is connected to a hydraulic actuator. During each tensile test, the slack adaptor is first accelerated to a target speed and then impacts with the dynamic grip to stretch a mounted sample all the way to fracture. When the target speed exceeds a certain limit (1 m/s in this system), the impact usually causes system ringing and consequently oscillations in the force signals. To mitigate such an issue, the sample geometry in this study was optimized based on some suggestions in the literature. In addition, two strain gauges were attached to proper deformation zones of both sides of each sample to function as an additional load transducer as the ISO 26203-2 standard suggested. The slack adaptor also allows installing damping washers to improve the force signal quality, despite some sacrifices on the dynamic grip speed. As for the strain measurement, a 2D-DIC system based on a high-speed camera was deployed.
normally to each gripped sample. Such a camera can record continuous images at a frame rate of a maximum of 120,000 Hz with sufficient resolution. At each test strain rate, the frame rate of the camera was adjusted to synchronize with the signal acquisition rate of the load washer based on a 1:8 ratio. Hence, the strain results from the DIC could be synchronized with the stress results from the force signals to process stress-strain curves in the subsequent processing.

Results and Discussions

The experimental results of the strain-rate-affected formability/ductility of the two target steels are shown in Figure 10. In this section, because of the downsized sample geometry\(^{11}\), the TFS results at 0.001 s\(^{-1}\) of both materials are lower than the TFS results at 25°C in Figure 5, and the TE values from the engineering stress-strain curves in Figure 11 are converted based on the ISO 2566 standard to be comparable to the TE results in Figure 5. Generally, there are also multiple material mechanisms affecting the tendencies of the formability/ductility. In the ‘quasi-static’ strain-rate range\(^ {13}\) from 0.001 to 1 s\(^{-1}\), adiabatic heating from the material plastic deformation is gradually elevated and eventually alters the isothermal condition, especially in the period between the sample necking and fracture. According to infrared measurements, at the strain rate of 1 s\(^{-1}\), the peak temperature on a Q&P980 tensile test sample can rise to around 150°C, and on a DP980LCE sample to around 100°C. As discussed in the previous section, combined effects in this temperature range suppress the tensile elongation (either the UE or the TE) of both materials as shown in Figure 10 and Figure 11. Such a thermal effect based on the adiabatic heating can also explain the tensile elongation increase in the ‘dynamic-low’ strain-rate range\(^ {14}\) from 1 to 1000 s\(^{-1}\), in which the adiabatic heat has less time to dissipate to the atmosphere and thus the temperature on a sample is elevated more severely than that in the ‘quasi-static’ strain-rate range. Nevertheless, the strain-rate-affected TFS values of the two target materials do not exhibit similar tendencies. For the Q&P980, at no matter which strain rate, the retained austenite transforms to martensite at a similar transformation rate as shown in Figure 12 before the adiabatic heat can suppress it, and thus the carbon-rich martensite can always create the ‘martensite-matrix de-cohesion’ that eventually evolves to fracture. This mechanism thereby limits the TFS of the Q&P980 changing with the strain rate. Notably, the retained austenite contents in Figure 12 were inspected via neutron diffraction at Oak Ridge National Laboratory at different locations of the tested samples, which were mapped to the corresponding strain gradients in the DIC results. Similar behavior of strain-rate insensitive transformation was also reported in the literature\(^ {45}\). In contrast, the TFS of the DP980LCE consistently increases with the elevated strain rates. This is caused by the higher strain hardening rate in the ferrite than that in the martensite decreases the hardness gap between the low-carbon martensite and the matrix, and as a result, further retards the nucleation of micro-voids.

![Figure 10](image_url). Strain-rate effects on the formability/ductility of (a) Q&P980 and (b) DP980LCE
CONCLUSIONS

To summarize, this work started from reviewing the Steel Strength-Ductility Chart and the Local-Global Formability Diagram to acknowledge their remarkable roles in categorizing and developing the generations of AHSSs. Extended from the local formability, this work highlighted the importance of the TFS (representing the local formability/ductility) in correlating the tensile testing results with the hole-expansion ratios, which represent a series of highly demanded material performances in the automotive applications. Furthermore, multiple practical conditions, such as the temperature, bake hardening, and strain rate, were investigated to reveal their effects on the TFS as well as the tensile elongation. Two AHSS grades, Q&P980 and DP980LCE, were studied under these conditions because of not only their comparative global and local formability advantages over each other, but also the discrepant microstructure development concepts behind them. The Q&P980 contains retained austenite in its multi-phase microstructure and thereby can take advantage of the TRIP effect to exhibit remarkable global formability/ductility under an isothermal quasi-static laboratory test condition. Yet in either a certain temperature range (around 150°C) or a strain-rate range (around 1 s⁻¹), the TRIP effect is suppressed due to the temperature-sensitive martensitic transformation. Such unfavorable temperature or strain-rate ranges are within the practical sheet metal stamping conditions. Besides, the rich-carbon martensite transformed from the austenite in the plastic deformation increases the hardness difference among the multi-phases and consequently tends to trigger fracture and suppress the local formability/ductility of the Q&P980.
Such suppression can be released at 150°C due to the inactive transformation, yet not significantly affected by the bake hardening or the strain rates. In contrast, the DP980LCE contains mainly ferrite and martensite in the microstructure and thereby exhibits very limited global formability/ductility. However, the decreased hardness difference between the ferrite and the low-carbon martensite retards the nucleation of micro-voids and eventually contributes to the enhanced local formability/ductility of the DP980LCE. Such an enhancement is not significantly affected by the bake hardening or the temperature conditions below 200°C and even further leveled up with the elevated strain rates due to the different strain-hardening rates of the multi-phases. Nevertheless, a common suppression condition for the TFS of both steels is the temperature range of 200-250°C, and the mechanism is believed to be the carbide precipitation into the martensite, which again increase the hardness difference between the martensite and the matrix.

REFERENCES


