Hole expansion ratio in intercritically annealed QP 980/1180 steel grades as a function of testing condition

M Madrid1, C J Van Tyne1,2, S Sriram3, E J Pavlina4, J Hu1, and K D Clarke1

1 Colorado School of Mines, 1500 Illinois St., Golden, CO 80401
2 Virginia Tech, 445 Old Turner St., Blacksburg, VA 24061
3 ArcelorMittal, 3001 E. Columbus Dr., E. Chicago, IN 46312
4 AK Steel Research and Innovation Center, 6180 Research Way, Middletown, OH 45005

Email: mmadrid@mines.edu

Abstract. United States Department of Transportation fuel economy and safety standards have led to the development of economically viable steels with excellent combinations of strength and ductility, better known as 3rd generation advanced high strength steels (AHSS). Quench and partitioned (QP) steels are of interest due to their excellent mechanical properties, though forming may be a challenge since stamped parts with a sheared edge may experience cracking at low strains. Hole expansion testing (HET) was performed on intercritically annealed QP 980 and QP 1180 steel sheets with two punch geometries (conical and flat bottom), edge conditions (sheared and machined), and complementary microstructural analysis was performed to better understand the effects on hole expansion ratio (HER). X-ray diffraction and electron backscatter diffraction experiments were performed to better understand factors affecting retained austenite (RA) stability as a function of strain. In this study, the QP 980 and QP 1180 steel grades had similar HERs for a majority of the testing conditions. Conical and flat bottom punches resulted in similar HERs despite varying edge conditions for a majority of the testing conditions, and the machined hole samples resulted in a higher HER than sheared hole samples regardless of punch geometry. The RA in QP 980 transformed at a faster rate than QP 1180 as a function of tensile strain. The relative stability of RA in QP 1180 was attributed to both RA morphology and the surrounding microstructure.

1. Introduction

Corporate average fuel economy standards and passenger safety requirements set out by the United States Department of Transportation, have driven the development of 3rd generation advanced high strength steels (AHSS) [1-3]. The automotive industry needs materials with higher strengths while maintaining levels of formability. One potential processing pathway to achieve the desired strength and ductility combinations in sheet steels is the quench and partitioning (QP) process, which has the capability to produce excellent combinations of strength and ductility during tensile deformation [4]. The QP processing path was developed by Speer et al. [4-5] with the goal of producing microstructures with controlled martensite and retained austenite (RA) phase fractions from low-alloyed steels. Processing may also be tailored to include ferrite and non-martensitic constituents (NMC) such as bainite [6], allowing great flexibility in microstructural design. The process begins with either a full austenitization or intercritical annealing step followed by quenching to a temperature (QT) between the martensite start and martensite finish temperature, transforming a controlled fraction of the austenite to
martensite. After equilibrating at QT, an isothermal hold, or partitioning step, is conducted as either a one-step [4] or two-step process, where one-step maintains the quench temperature for partitioning while the two-step process uses a temperature higher than QT. The intent of the partitioning step is to diffuse carbon from the martensite to the RA, increasing the stability of the austenite at room temperature. Although these steels have shown excellent tensile properties, other challenges exist such as stamping parts with a sheared edge which may experience cracking at strains lower than those predicted by forming limit diagrams [7]. Stretch-flangeability is a measure of edge formability and cracking resistance and is typically measured by hole expansion testing (HET) [8-17].

Typical HET configurations utilize a conical [8,10,16-17] or flat bottom punch [8, 11, 15] where the conical punch causes the sheet to undergo stretching and bending and the sheet only exhibits stretching when testing with the flat bottom punch. The literature has shown the conical punch results in a higher HER than the flat bottom punch [8, 18-19]. The higher HER observed with the conical punch is attributed to the bending component which has been shown to delay failure in stretch-bending experiments [20]. Hole expansion samples have also been prepared with different edge conditions (i.e., sheared or machined) [8, 11-13, 15]. Generally, sheared holes result in lower HERs than machined holes because of a shear affected zone (SAZ) at the edge which may result in pre-established voids or cracks at the sheared edge [8]. Shearing holes creates a burr and removing this region has shown to increase HER, perhaps by reducing potential crack nucleation sites [21]. Machining holes leaves a surface without a SAZ and has been shown to increase HER [13].

There is limited published research on QP steels and HET [15-16], but there are studies on transformation induced plasticity (TRIP) steels that can help explain the formability in QP steels as they experience the TRIP effect [14-17, 22-24]. Sevuma stated in their review of high strength steel sheets that the TRIP effect enhances the difference in strength of the constituents leading to a greater chance of micro-crack formation during the hole shearing process, but the TRIP effect can prevent the growth of micro-cracks during HET and ultimately improves HER [22]. Studies on the effects of RA volume fraction and HER concluded that RA stability played a larger role in determining HER [15, 23], as relatively unstable RA can lead to lower HERs. Sugimoto et al. suggested that 2-4 pct. untransformed RA during testing is good for HER in TRIP bainitic steels, and that less stable RA can lead to void formation that lowers HER [14].

The stability of RA in steels that exhibit the TRIP effect is affected by chemical composition [25], grain size [26], morphology [27], and surrounding microstructure [28]. Xiong et al. compared film-like and blocky RA morphologies in a QP 980 steel (0.22C-1.8Mn, wt. pct.), and showed that film-like RA located in martensite/austenite islands had a higher stability than the blocky RA that was surrounded by proeutectoid ferrite despite having the lower carbon content [27]. This was attributed to two possible mechanisms: 1) the lath martensite surrounding the film-like RA has a higher yield strength than the proeutectoid ferrite surrounding the blocky RA and 2) a higher hydrostatic pressure created by the volume expansion during the formation of RA to martensite on the film-like RA could be suppressing the transformation. Chiang et al. compared RA stability with tensile strain for equiaxed and lamellar microstructures with the same chemical compositions [28]. The equiaxed microstructure consisted of RA surrounded by intercritical ferrite while in the lamellar microstructure the lamellar RA was surrounded by bainite and led to a slower rate of transformation. They concluded that the RA stability was controlled by the carbon content of the austenite (higher in lamellar structure due to lower RA volume fraction), surrounding microstructure (RA in lamellar surrounded by bainite), and austenite grain size (smaller in lamellar structure).

De Moor et al. performed HET with a flat bottom punch and reamed edge conditions on QP 980 steels that were intercritically annealed (IA) and QP 1180 and QP 980 steels that were fully austenitized [15]. HER performance of the QP 980 steels that were IA were compared to dual-phase (DP) and TRIP steels. The IA QP steels resulted in a higher HER than the DP steel and showed a slightly lower HER than the TRIP steel. The TRIP steel consisted of a microstructure containing a combination of blocky RA and lath RA in-between bainitic and ferritic laths, while the QP steel typically exhibited film-like RA between martensite laths. Both the QP and TRIP steel exhibited similar mechanical properties and
RA volume fractions. Typically, lath RA in the TRIP steel is surrounded by bainite and probably led to more stable RA. Their results were attributed to the possibility that hardness differences between constituents (bainite/ferrite and martensite/ferrite) played a role in HER between QP, DP, and TRIP steels, while the RA morphology controlled the void formation behavior in the QP and TRIP steels [15]. The fully austenitized QP 980 and QP 1180 steels were compared to quench and tempered and a bainitic steel and resulted in a higher HER.

Jin et al. focused on the mechanical properties and HER of IA QP 980/1180, a DP 980, and TRIP steels of different thicknesses with different edge conditions using a conical punch [16]. The QP 1180 in their study had a higher HER than the QP 980 and was attributed to its high YS/TS ratio. The QP 1180 also showed the least HER loss between punched and wire-cut edge samples because of a low strain-hardening exponent, n, according to the authors. The low n-value led to the smallest hardness increase at the sheared edge and therefore, the HER difference between a punched edge and wire cut edge was low. The QP steels were generally less sensitive to edge conditions in comparison to the DP 980 tested in their studies. The sample thickness effects on HER were not clarified in their study, though the work of Comstock et al. showed that thickness has an effect on HER [29].

The QP process is still a fairly new concept and there are few HET studies that vary testing conditions and delve into microstructural effects on the stretch-flangeability of QP steels. Here, we focus on varying HET conditions (punch geometry and edge conditions) and their effect on HER for IA QP steels. The effect of microstructure (specifically RA volume fraction and stability) on the HER of IA QP 980/1180 steel grades is also evaluated.

2. Procedure
IA QP 980 (0.17C-2.06Mn wt. pct.) and QP 1180 (0.17C-2.38Mn wt. pct.) steel sheets (nominally 1.4 mm thick) were examined. Standard metallographic preparation techniques were used to prepare samples for characterization of the microstructure in the planar and longitudinal orientation with a field-emission scanning electron microscope (FESEM). The samples were etched with 2 pct. Nital. Standard metallographic preparation techniques include grinding steps that could induce the TRIP effect, therefore the microstructures observed may include some newly transformed martensite.

Tensile tests were performed according to ASTM E8. Standard ASTM E8 sheet tensile samples were waterjet cut from the as-received sheet to prevent the RA from transforming, as may occur during mechanical cutting or machining. The mechanical properties were determined from an average of four tensile samples tested along the rolling direction. The strain-hardening exponent, n, was calculated from a true strain of 0.02 up to the strain at ultimate tensile strength (UTS). The normal anisotropy coefficient, R\text{nn}, was determined for samples pre-strained to 10\% engineering strain.

RA volume fractions for the QP 980 and QP 1180 were calculated according to the SAE method [30]. RA was measured using x-ray diffraction (XRD) using nickel filtered copper radiation operating at 45 kV and 40 mA. Four ferrite peaks (\{110\}, \{200\}, \{211\}, \{220\}) and four austenite peaks (\{111\}, \{200\}, \{220\}, \{311\}) were used in the calculation of RA to account for crystallographic texture that may have been introduced during cold rolling. Samples were ground through 600 grit SiC grinding discs followed by being submerged in a solution of 2 parts hydrofluoric acid, 20 parts de-ionized water, and 20 parts hydrogen peroxide for 10 minutes to remove a minimum of 0.254 mm to assure no grinding damage remained and surface RA values reflect bulk values as closely as possible. The samples were measured using a two-theta range of 40° to 105° with a dwell time of 200s with a step size of 0.05° for a total of 35 minutes per scan. Samples that were strained to 5\%, 10\%, and failure (uniform elongation region) were also included in the RA calculations to see if the transformation rates of the RA differed between the two steel grades.

RA morphology in the QP steels was characterized by electron backscatter diffraction (EBSD). EBSD scans were performed along the rolling direction of the QP steels. The test samples were 10 x 10 mm with a wedge that were milled in a cross-section polisher for 8 hours to remove any deformed layers from the grinding process and to minimize the transformation of RA to martensite. Samples were analyzed at a 70° tilt, with a 10 mm working distance, 0.05 μm step size and 20 keV
accelerating voltage across scan areas of 15 x 15 microns. The carbon content was measured on 12 samples via XRD that were taken throughout the sheet using the \{220\} peak position of austenite:

\[ \text{a}_0 = 3.555 + 0.044C_\gamma \]  \hspace{1cm} (1)

where \( a_0 \) is the austenite lattice parameter in Å and \( C_\gamma \) is the carbon content in austenite in wt. pct. [31].

HET was performed according to ISO 16630 and the typical setup is shown in references [32-33]. Varying edge conditions (sheared and machined) and punch geometries (conical and 25.4 mm flat bottom) were tested. The nomenclature for the testing conditions is conical punch/machined edge (CM), conical punch/sheared edge (CS), flat punch/machined edge (FM), and flat punch/sheared edge (FS). Test blanks (1.4 mm thick) were waterjet cut to 101.6 x 101.6 mm (4 x 4 in.) and 177.8 x 177.8 mm (7 x 7 in.) geometry with a sheared or machined 10 mm hole with a die clearance of 11 pct. of the sheet thickness. All samples were tested in the burr-up condition at a punch speed of 0.5 mm/s with a camera system monitoring crack formation. When a through thickness crack was observed, the test was stopped and measurements were made using ImageJ analysis software [34].

3. Results

3.1. Microstructures

Figure 1 displays secondary electron images for QP 980 (1a and 1c) and QP 1180 (1b). QP 980 shows a microstructure consisting of intercritical ferrite (dark areas), martensite-austenite (M-A) regions showing substructure, and coarse lath-like NMC (seen most clearly in Figure 1c). It is presumed that the M-A constituent includes both blocky and lath RA. The QP 1180 shows a smaller amount of intercritical ferrite, but a greater amount of M-A constituent. There is also evidence of NMC though it is less than the amount seen in the QP 980. The calculated RA volume fractions from XRD were measured to be 16.0% and 12.6% for QP 980 and QP 1180, respectively.

![Secondary electron SEM micrographs of (a) QP 980 and (b) QP 1180 etched with 2 pct. Nital. Pictured is the planar orientation with the rolling direction horizontal. (c) QP 980 in the longitudinal orientation showing lath-like NMC.](image)

3.2. Mechanical Properties

Table 1 summarizes the measured mechanical properties of the QP steels. The QP 1180 has a higher yield and tensile strength, but lower elongation and reduction of area (ROA) relative to the QP 980. The mechanical properties that have been shown in the literature to affect HER are the \( n \)-value, \( R_m \), yield strength to ultimate tensile strength ratio (YS:UTS), and ROA.
Table 1. Mechanical Properties for QP 980 and QP 1180

| Material | YS (MPa) | UTS (MPa) | TE (MPa) | n | Rm | YS:UTS | ROA  |
|----------|----------|-----------|----------|---|----|--------|------|
| QP 980   | 698 ± 15 | 1057 ± 5  | 20.7 ± 0.2| 0.20 | 0.99 | 0.66   | 41.9 ± 1.9 |
| QP 1180  | 990 ± 27 | 1188 ± 8  | 16.2 ± 0.4| 0.11 | 0.93 | 0.83   | 36.0 ± 1.1  |

3.3. RA Stability

Figure 2 shows the RA volume fraction for as-received, 5% strain, 10% strain, and in the uniform elongation gauge of failed samples. It suggests that RA in QP 1180 has a higher stability than in QP 980, at least initially, indicating that RA in QP 980 is transforming to martensite at lower strains than QP 1180.

![Image of RA Stability](image)

Figure 2. Plot of RA volume fraction measure via XRD in QP 980 and QP 1180 as a function of strain, showing the transformation rate with tensile strain of RA in each material. The error bars are taken from the standard deviation of the RA variation.

The RA carbon contents were calculated as 1.14 and 1.15 wt. pct. for QP 980 and QP 1180, respectively. Figure 3 shows EBSD scans of QP 980 and QP 1180 with a confidence index (CI) ≥ 0.5. The scans have also been filtered to show areas ≥ 0.1 CI. The RA volume fractions in Figure 3 are 2% and 4% for QP 980 and QP 1180, respectively. The volume fractions are lower than the XRD measurements because EBSD only takes into account the scanned sample area and may not be representative of the bulk volume fraction. Both QP steels show evidence of blocky RA, with lath RA being harder to index due to the resolution limitations. In Figure 1, both steels show M-A constituent where both lath RA and blocky RA can occur with QP 1180 showing a greater fraction of M-A.

![Image of EBSD Scans](image)

Figure 3. Overlayed image quality and phase maps of (a) QP 980 and (b) QP 1180 obtained through EBSD. The red represents BCC-Fe (ferrite or martensite) and the green represents FCC-Fe (austenite).

3.4. HET

Figure 4 summarizes the HER for the QP steels as a function of edge condition and punch geometry. The QP 980 and QP 1180 have similar HER for all testing conditions with the exception of the FM
condition. The QP 1180 has higher strength and lower ductility and has similar HER results to the lower strength, more ductile steel. The QP steels do not seem to follow the trend reported in the literature [8, 18-19] that the conical punch results in higher HER as compared to the flat bottom punch for the CS and FS conditions. When considering edge condition, however, the machined QP samples show a higher HER than the sheared samples. The ratios in Figure 4 were calculated between the QP steels for a given test condition.

Figure 4. HER plotted for QP 980 and QP 1180 as a function of punch geometry and edge conditions. Ratio = QP 980:QP1180.

4. Discussion

4.1. Effects of Edge Conditions and Punch Geometry on HER
Generally, sheared holes lead to a lower HER than machined holes because the latter removes the SAZ resulting in improved HER [8]. In the current study, the samples with machined holes showed a higher HER than the sheared holes for both steel grades in agreement with the literature [8, 11, 13]. The punch geometry has also been shown to affect the HER where a conical punch generally produces a higher HER than a flat bottom punch. This is because of the different stress states each punch creates where the conical punch adds a bending component that creates a material constraint that has been used to explain higher HER [20]. In this study however, the QP 980 CM and FM condition could be considered to have similar HER within experimental error. The CS and FS conditions for both the QP 980 and QP 1180 also show similar HER despite different punch geometries. Therefore, the results show a different trend for the effects of punch geometry on HER than has been published [8, 18-19] with the exception of the QP 1180 in the CM and FM conditions. The QP 980 steels in this study are not sensitive to punch geometry, while the QP 1180 is sensitive for CM/FM, but not for CS/FS conditions, which warrants further investigation.

The ratios of HET calculated in Figure 4 for QP 980 and QP 1180 are independent of test conditions with the exception of the FM samples. The FM condition does not induce a strain gradient in the material due to punch geometry and has no strain hardening at the sample edge due to hole preparation method. AHSS of this strength grade and processing method are fairly new and the literature for HET with a flat bottom punch with varying edge conditions with these steels is sparse. Further research is needed to better understand HET condition effects on QP steels, as the trends shown in lower strength materials may not apply.

4.2. Microstructure and HER
The QP 980 and QP 1180 have similar HERs for all conditions with the exception of the FM condition. This is interesting because the QP 1180 has higher strength and lower ductility. The mechanical properties that have been shown to contribute to HER in the literature [16, 35-36] were analyzed. The QP 980 has a higher strain-hardening exponent and should show a greater difference in HER between sheared and machined edges according to studies done by Jin et al., though this is only seen in the QP 980 FM and FS conditions. Despite the higher strain-hardening exponent, both steel grades have similar HER differences between sheared and machined edges. A trend that was in agreement with Jin et al.
was that the QP 1180 has a higher YS/UTS ratio than the QP 980 and could be a reason for the good HER in comparison to the lower strength QP 980. Pathak et al. showed that greater ROA resulted in higher HERs [8], which is not seen in the current study. Therefore, many of the prior correlations of HER with mechanical properties do not appear to apply for the QP steels in this study. The RA stability may be a significant factor for stretch-flangeability.

Figure 2 showed that the RA transformation rates between the two QP steels were different, with the austenite in the QP 980 transforming at a faster rate as a function of tensile strain. The slower transformation rate and possibility of stable non-transformed RA present in the QP 1180 steel could be a reason for comparable HER to the QP 980. The RA stability was analyzed through RA carbon content and characterization of RA morphology. The measured RA carbon content between the two steels were almost identical, suggesting carbon content is not a contributing factor for the different transformation rates seen in the two QP steels during tensile testing. The QP process inherently aims at stabilizing austenite, though chemical composition can also help with stabilization. The QP 1180 has a higher Mn content than the QP 980 and it is possible that the higher content could lead to greater austenite stabilization.

In order to better understand RA morphology, EBSD scans were performed and are shown in Figure 3. Detecting lath RA is difficult due to resolution limitations, though a qualitative analysis can still be made. The darker areas in Figure 3 can be considered newly transformed martensite from sample preparation that was not indexed through EBSD. Both steels show evidence of blocky RA for the areas scanned through EBSD and the microstructures in Figure 1 show M-A constituents where both blocky and lath RA occur. Looking at the microstructures in Figure 1, the QP 1180 has a greater amount of M-A constituent and it is then possible that the QP 1180 has a greater ratio of lath RA to blocky RA. Literature has shown that lath RA is more stable than blocky RA. It is also possible for the blocky and lath RA to have different carbon contents, hence affecting their stabilities. Though for this study, the carbon content of blocky and lath RA was not further analyzed. The surrounding microstructure may also play a role. The RA in the QP 1180 is more than likely surrounded by martensite since there is little intercritical ferrite, while in the QP 980 the RA has a greater chance of being surrounded by intercritical ferrite and martensite. The EBSD scans show that the RA is in close proximity to ferrite in the QP 980 and more so to martensite in the QP 1180.

5. Conclusions
The effects of punch geometry and edge conditions on HET were analyzed for IA QP steels of the same thickness. In the current study, the QP 980 and QP 1180 had similar HERs for a majority of the testing conditions despite the differences in strength and ductility. The QP 1180 had a higher YS/UTS ratio than the QP 980 and could be a reason for the comparable HER. The QP 980 steel examined here was not sensitive to punch geometry and both QP steels were sensitive to edge conditions in that the machined holes did result in a higher HER than sheared holes regardless of punch geometry. The ratios of HET for QP 980 and QP 1180 were similar for all punch and edge combinations with the exception of the FM condition. This warrants further investigation of the FM testing condition as the factors controlling HER may be different than the rest of the conditions.

IA QP steels lead to a microstructure with intercritical ferrite while a fully austenitized QP steel does not have this phase. It is believed that the ductile intercritical ferrite can accommodate strain and enhance HER. The TRIP effect increases the strength of the material and delays strain localization, therefore, a delay of the TRIP effect through a higher stability of RA can lead to improved HER. Experiments showed that the RA in QP 980 transforms at a faster rate than QP 1180 as a function of tensile strain. The relatively stable RA in QP 1180 was attributed to RA morphology and the surrounding microstructure since measured carbon contents were similar. The QP 1180 showed a higher fraction of M-A constituent and therefore the possibility of a higher fraction of lath RA. The RA in QP 1180 also has a higher probability of being surrounded by martensite as opposed to intercritical ferrite helping in the stability of RA.
Acknowledgments
We thank Baosteel for providing the test materials. We also thank both ArcelorMittal (East Chicago) and AK Steel (Middletown, OH) who offered their research facilities for hole expansion testing. We also thank Dr. Kavesary Raghavan and Dr. Mai Huang for assisting with testing. The Advanced Steel Processing and Products Research Center (ASPPRC) at Colorado School of Mines is acknowledged for their support and funding. We also acknowledge the Bill and Melinda Gates Scholarship Foundation for their funding.

References
[1] US Dept. of Transporation, “Corporate Average Fuel Economy”, http://www.nhtsa.gov/laws-regulations/corporate-average-fuel-economy.
[2] US Dept. of Transporation, “Laws and Regulations”, http://www.nhtsa.gov/laws-regulations.
[3] Bouaziz O, Zurob H and Huang M 2013 Steel Research Int. 84 pp 937-47
[4] Speer J, Streicher A, Matlock D, Rizzo F and Krauss G 2003 TMS Conf Proc. pp 505-22
[5] Speer J, Matlock D, De Cooman B and Schroth J 2003 Acta Mat. 51 pp 2611-22
[6] Wang L and Speer J 2013 Metallogr. Microstruct. Anal. 2 pp 268-81
[7] Levy B and Van Tyne C 2008 J. Mater. Eng. Perform. 17 pp 842-48
[8] Pathak N, Butcher C and Worswick M 2016 J. Mater. Eng. Perform. 25 pp 4919-32
[9] Hasegawa K, Kawamura K, Urabe T and Hosoya Y 2004 ISIJ Int. 44 pp 603-09
[10] Terrazas O, Findley K and Van Tyne C 2017 ISIJ Int. 57 pp 937-44
[11] Taylor M, Choi K, Sun X, Matlock D, Packard C, Xu L and Barlat F 2014 Mater. Sci. Eng., A 597 pp 431-39
[12] Levy B, Gibbs M and Van Tyne C 2013 Metall. Mater. Trans. A 44 pp 3635-48
[13] Karelzova A, Krempaszky C, Werner E, Tsiopouridis P, Hebesberger T and Pichler A 2009 Steel Research Int. 80 pp 71-77
[14] Sugimoto K, Nakano K, Song S and Kashima T 2002 ISIJ Int. 42 pp 450-55
[15] De Moor E, Matlock D, Speer J, Fojer C and Penning J 2012 SAE Int. Technical Paper 2012-01-0530
[16] Jin X, Wang L and Speer J 2013 Proc. Int. Symposium on Automobile Steel pp 60-67
[17] Konieczny A and Henderson T 2007 SAE Int. Technical Paper 2007-01-0340
[18] Larour P, Freudenthaler J, Grunsteidl A and Wang K 2014 Proc. Conf. IDDRG pp 188-193
[19] Iizuka E, Uraba M, Yamasaki Y and Hiramoto J 2017 J. Phys.: Conf. Series 896 012008
[20] Neuhauser F, Terrazas O, Manopulo N, Hora P and Van Tyne C 2016 IOP Conf. Ser.: Mater. Sci. Eng. 159 012011
[21] Adamczyk R, Dickinson D and Krupitzer K 1983 SAE Int. Technical Paper 830237
[22] Senna T 2001 ISIJ Int. 41 pp 520-32
[23] Matsumura O, Sakuma Y, Ishii Y and Zhao J 1992 ISIJ International 32 pp 1110-16
[24] Pradhan R, Kelley S, Frailey R and Layland B 2004 SAE Int. Technical Paper 2004-01-0505
[25] Jacques P, Ladriere J and Delannay F 2001 Metall. Mater. Trans. A 32A pp 2759-68
[26] Yang H and Bhadesia H 2009 Scripta Materialia 60 pp 493-95
[27] Xiong X, Chen B, Huang M, Wang J and Wang L 2013 Scripta Materialia 68 pp 321-24
[28] Chiang J, Lawrence B, Boyd J and Pilkey A 2011 Mater. Sci. Eng. A 528 pp 4516-21
[29] Comstock R, Scherrer D and Adamczyk R 2006 J. Mater. Eng. Perform. 15 pp 675-83
[30] Jatczak C 1980 SAE Int. Technical Paper 800426
[31] Dyson D and Holmes B 1970 J. Iron and Steel Institute 208 pp 469-74
[32] Madrid M, Van Tyne C and Clarke K 2017 MS&T Conf. Proc. pp 1302-04
[33] Shih H and Shi M 2011 J. Manuf. Sci. Eng. 133 pp 520-32
[34] Schindelin J, Arganda-Carreras I, and Frise E 2012 Nature Methods 9 pp 676-82
[35] Chen X, Jiang H, Cui Z, Lian C and Lu C 2014 Procedia Eng. 81 pp 718-23
[36] Sadagopan S and Urban D Formability Characterization of a new Generation of High Strength Steels DOE Report No. 0012