A very-low-carbon microalloyed chemistry is used for the development of family of high-strength dual-phase (DP) steels with enhanced global and local formability characteristics. A clean steel practice combined with suitable choice of chemistry has resulted in excellent mechanical properties and forming characteristics of high-strength DP steels. Enhanced balance of formability is attributed to the optimization of microstructural constituents through refinement, uniformity and limiting the imbalance of microconstituent mechanical response. In this article, production strategy and advanced material characterization of DP steels with tensile strength of 600 MPa and above relevant to automotive body structure applications are discussed.

Advanced high-strength steels (AHSS) continue to enjoy significant material share of the total automotive market segment, driven by their enhanced safety and fuel economy offerings. AHSS steels reduce vehicle body-in-white mass by 25–39% compared to conventional automotive steels, thereby promoting downsizing and optimization of other system components that reduce carbon footprint due to improved fuel efficiency. Reduction in associated carbon footprint is increasingly dictating the global steel production technology road map.

AHSS steels are used in every new vehicle design and there is a rapid advancement of ultrahigh-strength Gen 3/third-generation/3G ultra-high-strength steels (UHSS) (e.g., medium-Mn, quenched and partitioned steels) offering superior tensile strength (up to 1,500 MPa) and elongation combine compared to first- or second-generation steels to further effect reduced carbon footprint and increased passenger safety. These steels are touted as probable replacement for press hardening steels. While most of the lower-strength AHSS steels are produced by conventional galvanizing lines with leaner mill-specific chemistry and processing, production of UHSS Gen 3 steels is daunting for discrete processing lines as these steels require disciplined steelmaking, casting and processing routines, and advanced dedicated annealing and coating lines to cope with heavy alloying and multi-step complex thermal cycles. These steels, while promising superior product performance, are extensively investigated to circumvent challenging product application issues such as weldability due to high alloying, hydrogen embrittlement due to higher fractions of retained austenite, and liquid metal embrittlement (LME) during spot welding. A select few companies have engaged in the production of Gen 3 steels in strengths up to 1,200 MPa almost exclusively through the integrated manufacturing process, and the automotive steel research has opened up possibilities for many variants of UHSS steels to be developed in the near future. In this respect, thin-slab casters have also taken up an ambitious program of developing Gen 3 steels as thin-slab casters offer favorable operational dynamics on the primary steelmaking side.

Mid-strength AHSS steel’s performance, on the contrary, is time tested and has secured a niche position in overall materials usage in passenger vehicles. It is stated that of the
total AHSS steel usage in a passenger car, dual-phase (DP) steels with tensile strength up to 1,000 MPa retain a significant portion of steels for body structures and may go up to 30% mass weight of a passenger vehicle, as indicated in Fig. 1.1

Some typical examples of dual-phase steel applications3 are indicated in Table 1.

Many mini-mills with thin-slab casters have successfully added some AHSS grades in their portfolio and stretched their production lines to make high-strength DP steels up to 800 MPa,5 though guaranteeing a superior finish surface quality remains a challenging task for thin-slab casters. Current research work has indicated choice of suitable casting powder can successfully produce acceptable surface quality results.

Big River Steel (BRS), being the latest advanced Flex mini-thin-slab casting mill in North America, has engaged in the production of automotive-grade AHSS steels and DP steels in particular. BRS has the advantage of its unique combination of electric arc furnace (EAF) steemaking with RH degassing technology, which enables it to produce advanced grades of steels with low gas contents (H, N) and enhanced internal cleanliness. A detailed description of its steemaking technology has been published.6,7

From a series of communications with automotive original equipment manufacturers (OEMs), it was realized that AHSS preferences were shifting toward a low carbon equivalence and high stretch flange-ability attributes for edge retention during stretching applications. Big River Steel, therefore, engaged in a product development program to develop families of high-strength DP steels catering to the emerging need of the automotive industries.

In the present article, alloy design and processing approaches for very-low-carbon base chemistries are discussed together with successful commercial development of high-strength DP steels with tensile strengths of 800 MPa and beyond, including DP grades with high hole expansion ratio (HER) for stretch flangeability.

**Alloy Design Approach** — Dual-phase steels consist of hard martensite constituents in a softer ferrite matrix promoting high work hardening, high tensile strength (TS)/yield strength (YS) associated with excellent total elongation. These steels also offer significant increase in strength due to strain aging.8 The strength of the composite microstructure is dictated primarily by the martensite volume fraction and less likely influenced by martensite composition and strength.9,10 Following a treatise of plastic instability in a composite matrix by Mileiko,10 the flow stress of DP steel at any strain (ε) and strain rate (˙ε) can be given as:

$$\sigma_{flow}(DP) = V_M \sigma_{f}(\epsilon, \dot{\epsilon}) + V_F \sigma_(\epsilon, \dot{\epsilon})$$

(Eq. 1)

**Table 1**

<p>| Selected Application Examples of Dual-Phase (DP) Steels in Automotive Structures |</p>
<table>
<thead>
<tr>
<th>DP grades</th>
<th>Applications</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP 350-600</td>
<td>Floor panel, hood outer, body side outer, cowl, fender, floor reinforcements</td>
</tr>
<tr>
<td>DP 500-800</td>
<td>Body side inner, quarter panel inner, rear rails, rear shock reinforcements</td>
</tr>
<tr>
<td>DP 600-1000</td>
<td>Safety cage components (B-pillar, floor panel tunnel, engine cradle, front subframe package tray, rail components, seating)</td>
</tr>
<tr>
<td>DP 700-1000</td>
<td>Roof rails</td>
</tr>
<tr>
<td>DP 800-1200</td>
<td>B-pillar upper</td>
</tr>
</tbody>
</table>
where $V_M$ and $V_F$ are volume fraction of martensite and ferrite, respectively.

Hence, key to achieving higher strengths in dual phase will be to increase martensite volume fraction and flow stress of ferrite. Ferrite exhibits low strain hardening because of dynamic recovery by cross-slip, and the effective way to increase flow stress of ferrite is to decrease ferrite grain size. Finer ferrite grain size decreases the dislocation mean free path, which is proportional to the inverse of the square root of dislocation density, thereby increasing the flow stress. A fine ferrite also has the capability to increase dynamic energy absorption, a critical requirement for current vehicular safety design. Depending on carbon content, processing is designed to control martensite volume fraction through control on intercritical temperature to develop steels with various tensile strengths.9,12

Ferrite grain refinement is usually inherited from hot-rolled microstructure resulting from the prior austenite grain size. Hence, microalloying with Nb was in consideration when necessary to refine the austenite grain size. Nb precipitates as carbonitrides during hot rolling, producing fine austenite grain sizes.13,14 The fine austenite grains in turn induce fine ferrite formation during cooling, resulting in carbon enrichment in retained austenite.

Choice of Carbon Content: In the current developmental study, a very-low-carbon approach unusual of conventional mills with thick-slab casters was adopted to produce DP steels in accordance with OEM preference and market trends. Low carbon guarantees low carbon equivalence, ease of welding, and higher and more balanced formability.4 Low carbon results in low carbon content in transformed lath martensite, which facilitates generation of increased mobile dislocations at the ferrite-martensite interface due to transformation volume strain. Choice of low carbon in DP steels is intimidating as it has been stated there is a limit to tensile strengths that can be achieved with low carbon contents without compromising ductility.15,16

For most CSP mills, low-carbon steel has 0.06 wt.% C or less and the carbon range from 0.07 to 0.17 wt.% is conscientiously avoided for fear of surface cracks development and caster breakouts. This fear was circumvented in the current study through the choice of steel composition–specific mold powder. The fear factor emanates from the fact that peritectic reactions occur between carbon contents of 0.09 to 0.51 wt.%. The reaction is associated with a large volumetric contraction due to $\delta$-$\gamma$ transformation with maximum volume contraction occurring at peritectic C of 0.17%. The transformation is also reported to be massive, unlike that controlled by diffusion transformation.17

The massive transformation and associated thermal shrinkage (up to 4%) between 0.09 and 0.17% C is critical as it results in longitudinal surface cracks, and for a Compact Strip Production (CSP) mill there is no opportunity to repair the surface as the cast slab is continuously fed to the hot strip mill through the tunnel furnace. Even when hypoperitectic carbon (<0.09 wt.%) is chosen, its equivalent carbon content is carefully evaluated as alloying additions have been found to shift the peritectic carbon $C_A$ (Fig. 2) toward lower values. For multi-component Fe-C system, earlier works have indicated an empirical peritectic carbon equivalence based on elemental coefficients as follows:

$$C_p = [\%C] + 0.047Mn + 0.1Ni + 0.7N - 0.14Si - 0.04Cr - 0.1Mo - 0.24Ti$$

(Eq. 2)
Recently, through numerous thermodynamic calculations based on published Fe-C-X (X-alloy element) phase diagrams, elemental influence on the peritectic carbon equivalence has been critically evaluated. Mn and S have been found to significantly influence shifting of peritectic carbon. An increase in Mn content from 0 to 1.4% shifts the C_A from 0.090% to 0.075% and C_p from 0.170% to 0.143%. An increase in S from 0 to 0.015% will decrease C_A from 0.09 to 0.08%.

In summary, in designing the low-C chemistry for DP steel, consideration was given in calculating solidification behavior inside mold, critically assessing carbon equivalence as per Eq. 2 and ferrite potential (FP) of the designed composition and a suitable mold powder appropriate of calculated ferrite potential (FP) of the melt was worked out where FP is estimated by Eq. 3 as:

\[
FP = 2.5(0.5 - [\%C_p])
\]  
(Eq. 3)

Table 2 lists the typical characteristics of the appropriate mold powder used after detailed research for casting low-C DP steels to ensure a smooth surface of the slabs. Adaptation to a suitable basicity and crystallization temperature aiming at optimum horizontal heat transfer were key to achieving the best mold powder for ease of casting and associated good surface of slabs.

| Chemical Composition of Mold Powder Used for the Current DP Steels, wt.% |
|-----------------------------|-----------------------------|-----------------------------|-----------------------------|-----------------------------|-----------------------------|
| CaO | SiO₂ | MgO | Na₂O | F | CaO/SiO₂ | Melting point, °C | % free carbon |
| 31.50 | 32.95 | 3.00 | 11.00 | 6.50 | 0.95 | -1,095 | 2.50-2.70 |

Alloy Additions – Influence of alloying elemental additions to DP structure-property correlations have been widely studied over time and a few elemental influences pertinent to the current alloy design and operations will be discussed.

A baseline chemistry of 0.06C-1.6Mn (wt.%) was chosen as a starting point for developing a family of DP steels. Mn is an austenite stabilizer and facilitates enhanced martensite volume fraction for a given intercritical annealing temperature. It also strengthens ferrite. The Fe-C diagram for the eutectoid portion of Fe-1.6Mn calculated using Thermo-Calc software is shown in Fig. 3a. It can be estimated that carbon content of austenite at intercritical annealing temperature of 780–830°C may vary between 0.43 and 0.22 wt.% C with a corresponding austenite volume fraction of 38% and 76%, respectively (computed using Lever Rule). In the current study, a final microstructural martensite fraction of 20–70% is aimed to develop DP steels of various tensile strengths.

Although the hardenability of low-C austenite when annealed at higher intercritical temperature will be less than the austenite formed at lower intercritical temperature, the low-C martensite will result in more mobile dislocations at the ferrite-martensite interface, as discussed earlier. To compensate for reduced hardenability, additions of Mn, Mo and Cr were done. Additions of boron have been intentionally avoided. Nb, though primarily added for austenite refinement, also acts as a hardenability agent as it inhibits carbide precipitation.

The rationale for current alloy additions is better understood from an understanding of current continuous galvanizing line (CGL) configuration and intended processing approach. As indicated in Fig. 3b, the time-temperature traverse through the galvanizing line is superimposed schematically on the continuous cooling transformation (CCT) diagram obtained using JMatPro software. The steel is fast-cooled from the soak temperature to Zn pot and then cooled after hot dipping. Mn, Mo and Cr additions are added to lower the martensitic start temperature below the hot dip temperature (~463) and is estimated using following equation:

\[
Ms(°C) = 539 - 423C - 30.4Mn - 17.7Ni - 12.1Cr - 7.5Mo
\]  
(Eq. 4*) *elements in wt.%

Secondly, the amount of Cr, Mo is adjusted to shift the bainite nose toward right so that bainite transformation is avoided before hot dipping. In assessing both bainite start temperature and critical cooling rate (CR) required to achieve this goal, the following deliberations by Furukawa et al. were used:

\[
Bs(°C) = 830 - 270C - 90Mn - 37Ni - 70Cr - 83Mo
\]  
(Eq. 5)

\[
\log CR(°C / s) = 5.36 - 2.36Mn - 1.06Si - 2.71Cr - 4.72P
\]  
(Eq. 6)

It can be estimated that for an aim Cr+Mo addition of <0.75 wt.%, a critical cooling rate of ≥35°C/second will be necessary from the soak temperature to avoid bainite start at 463°C. The CGL at Big River Steel is equipped with a rapid cooling section with a cooling rate up to 100°C/second and hence avoidance
Fe-C diagram of a Fe-0.06C-1.6Mn steel calculated using Thermo-Calc software (a) and continuous cooling transformation (CCT) diagram (using JMatPro) and schematic continuous galvanizing line annealing path for DP steel processing (b).

Details of mill trial were presented earlier but the takeaway salient features of the successful development are presented here.

The steel was cast into 65-mm-thick slabs and was hot-rolled using a finishing temperature range of 860–890°C and a coiling temperature of 560–600°C. After pickling and cold reduction, full hard samples were first prepared for laboratory annealing simulation before running through the CGL. Strips from cold-rolled coils were subjected to annealing simulation in a Gleeble Strip Annealing Simulator for a soak temperature of 780, 790 and 810°C and then tensile properties were evaluated. Following laboratory simulations and results, the coils were run through the actual galvanizing line using the best annealing cycle indicated through simulations.

Microstructures of the DP 590 steel for 1.0-mm processed coils are shown in Fig. 4 examined under...
optical as well as scanning electron microscopes. As indicated in both micrographs, martensite constituents are observed as fine islands with an average area fraction of approximately 24% as measured using image area analysis.

**Development of Cold-Rolled Galvanized DP780 Grade** — With the success of producing DP590 using the chemistry approach as outlined above, the next objective was to develop DP780 steel through the current line. A natural extension of the alloying approach was to increase austenite stabilizer Mn for an increase in the martensite fraction and a slight increase in Mo and Cr for more hardenability. Realizing increased ferrite strength might be a necessity, two heats, one without Nb and one with Nb microalloying, were tried. Table 5 indicates the alloy composition of two trial heats differing mainly in Nb.

Two trial heats were made with a slab thickness of 55 mm and width of 1,420 mm. During casting, dynamic soft reduction was employed at the last solidification point of the caster for segregation control. Full-width slab samples were collected at caster exit to study slab internal quality. Fig. 5 shows full-width microstructures of 1.0-mm-thick DP590 cold-rolled galvanized steel as examined under optical microscope showing fine ferrite grains and straw-colored martensite constituents revealed using step etching technique (a) and (b) scanning electron microscope (2% Nital etch).

**Table 5**

<table>
<thead>
<tr>
<th>Heat</th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Si</th>
<th>Ni</th>
<th>(Cr + Mo)</th>
<th>N</th>
<th>Nb</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>0.0600</td>
<td>&lt;2.0000</td>
<td>0.0070</td>
<td>0.0010</td>
<td>&lt;0.6000</td>
<td>Res</td>
<td>&lt;0.7500</td>
<td>0.0064</td>
<td>Res</td>
</tr>
<tr>
<td>B</td>
<td>0.0600</td>
<td>&lt;2.0000</td>
<td>0.0090</td>
<td>0.0007</td>
<td>&lt;0.6000</td>
<td>Res</td>
<td>&lt;0.7500</td>
<td>0.0068</td>
<td>0.0220</td>
</tr>
</tbody>
</table>

**Figure 5**

Macrograph of full-width transverse section of DP780 steel slab using Hatch Electrolytic Etcher; Slab thickness: 55 mm. Slab width: 1,429 mm.
macroetched transverse slab section revealed by electrolytic etching using Hatch Electrolytic Etcher. Fewer centerline segregation spots could be detected and mostly were less than 1 mm in diameter.

The slabs were hot-rolled using a finishing temperature of 880–890°C and coiling temperature of 550–565°C. The hot-rolled coils were further cold-rolled to 0.80-mm, 1.35-mm and 1.40-mm coils. Full hard samples were initially annealed in a Surtec Multifunctional Annealing Simulator for best inter-critical temperature assessment. Based on annealing simulations, a typical annealing cycle used for actual mill trial for DP780 steel production is shown in Fig. 6.

Mechanical properties for cold-rolled galvanized steel (thickness 0.80 and 1.35 mm) from two DP780 heats (with and without Nb addition) are listed in Table 6. Tensile strengths obtained in steel without Nb addition fell short of the stipulated DP 780 specification but the steel with Nb successfully met the tensile strength and other tensile properties. Both steels were processed with similar processing parameters.

OEMs also refer to high-yield DP780 for increased stretchability or stretch flangeability properties. In order to achieve higher yield strength in the steel, the effect on in-line skinpass (SPM) and tension leveling (TL) elongation was further studied by processing coils from heat B at different setpoint elongations keeping the continuous galvanizing line heat cycle the same. Mechanical properties listed in Table 6 represent average results obtained with total elongation (SPM+TL) greater than 0.5%. A yield strength increase of 67 MPa could be imparted when comparing with material processed at low total (SPM+TL) elongation (<0.5%).

The microstructure of 1.35-mm-thick Nb-added DP780 steel is shown in Fig. 7. Both optical and scanning electron micrographs revealed fine ferrite grains as well as fine martensite constituents. It is interesting to note that pronounced banded microstructural feature could not be observed. Martensite fraction estimated using image analysis was found to be approximately 34%.

### Formability Evaluations

#### Hole Expansion Ratio Testing:

There are various material formability attributes that are currently evaluated to determine AHSS/UHSS steel’s overall press or stamping performance such as bend test, formability limit diagram evaluation, stretch flangeability evaluation, etc. Of all the tests, stretch flangeability estimated through a hole expansion test is one of the critical evaluations that predicts edge splitting behavior during stamping. Forming limit diagram (FLD) cannot predict this edge splitting or failure mechanism as it

<table>
<thead>
<tr>
<th>Table 6</th>
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<table>
<thead>
<tr>
<th>Mechanical Properties of DP780 Cold-Rolled Galvanized Product With and Without Nb Addition</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
</tr>
<tr>
<td>---</td>
</tr>
<tr>
<td>Low yield strength DP780</td>
</tr>
<tr>
<td>Typical automotive spec 780T/420Y-DP</td>
</tr>
<tr>
<td>Heat A (No Nb)</td>
</tr>
<tr>
<td>Heat B (with Nb)</td>
</tr>
<tr>
<td>High yield strength DP780</td>
</tr>
<tr>
<td>Typical automotive spec 780T/500Y-DP</td>
</tr>
<tr>
<td>Heat B (with Nb)</td>
</tr>
</tbody>
</table>
only outlines a limiting strain value between successful forming and necking or excessive thinning when sheets are deformed through various strain paths under plain stress conditions. To study material local formability, HER testing is performed.

In the current study, hole expansion tests were done using an Interlaken SP400 test equipment at AMT-Fadi LLC. Square sheets of usually 100 mm x 100 mm had a 10-mm hole punched at the center and were clamped between a holder and die with a clamping force of 100 kN. A clearance 12±1% of nominal sheet thickness was adopted conforming to ISO 16630:2009(E) specification. A conical punch with 60° angle was pierced through the hole at a speed of 0.25 mm/second and the crack appearance during piercing was monitored using a digital imaging system. Schematics of the HER testing are shown in Fig. 8 for reference. The piercing was done at least 4–5 hours after punching the hole. Samples from different locations across the width were taken for this study to examine property homogeneity across the width and average data is indicated in Table 7 for various steel grades and thicknesses tested.

Table 7 also indicates hole expansion data evaluated for the hot bands of the steels. Fig. 9 shows the HER tested blanks of DP590 and DP780 galvanized sheets at edge, center and edge location of the coils and a typical close-up view of the HER tested sample. The holes retained their circular configuration after piercing and the cracks appeared in either transverse to or along rolling direction.

The HER data in Table 7 was plotted as a function of ultimate tensile strength and the trend is shown in Fig. 10. HER value increases as ultimate tensile strength decreases in DP steels. Lower HER value for high-ultimate-tensile-strength (UTS) DP grades are attributed to higher martensite volume fraction and has been reported by many others.28,29

**Forming Limit Diagram Analysis — FLDs were evaluated for the DP590 as well as DP780 galvanized steels at seven**
different strain paths (from uniaxial-tension to plane-strain to balanced-biaxial) as per ISO 12004:2008 standard using Marciniak (in-plane) approach. Material strain/deformation history was produced by performing digital image correlation (DIC) to map the evolution of surface strains and the section method was used to extract the FLC points per tested sample.\textsuperscript{27} The extracted FLCs of DP590 steel for thickness of 1.0 and DP780 steel for thickness of 0.8 mm are shown in Fig. 11. FLC points show low scatter and good repeatability, indicating material mechanical property homogeneity.

**Discussion**

Mechanical properties obtained in both DP grades of steels successfully met the OEM property specifications and hence validated alloy design and processing

**Table 7**

<table>
<thead>
<tr>
<th>Thickness, mm</th>
<th>Grade</th>
<th>Yield strength, MPa</th>
<th>Ultimate tensile strength, MPa</th>
<th>Elongation, %</th>
<th>Average HER, %</th>
<th>Comments</th>
</tr>
</thead>
<tbody>
<tr>
<td>2.61</td>
<td>DP590-HR</td>
<td>386</td>
<td>659</td>
<td>21</td>
<td>101±15</td>
<td>Tested after 4 hours of punching</td>
</tr>
<tr>
<td>1.00</td>
<td>DP590-CR</td>
<td>355</td>
<td>644</td>
<td>26</td>
<td>54.6±4.3</td>
<td>Tested after 4 hours of punching</td>
</tr>
<tr>
<td>1.2</td>
<td>DP590-CR</td>
<td>421–434</td>
<td>658–664</td>
<td>–</td>
<td>50.2±7.0</td>
<td>Tested after 4 hours of punching</td>
</tr>
<tr>
<td>1.73</td>
<td>DP590-CR</td>
<td>400</td>
<td>620</td>
<td>27</td>
<td>52.8±4.2</td>
<td>Tested after 4 hours of punching</td>
</tr>
<tr>
<td>2.38</td>
<td>HR skelp for DP780</td>
<td>655</td>
<td>781</td>
<td>12</td>
<td>78.8±5.5</td>
<td>Tested after 4 hours of punching</td>
</tr>
<tr>
<td>0.82</td>
<td>DP780-CR</td>
<td>543–568</td>
<td>843–847</td>
<td>16–19</td>
<td>36.8±4</td>
<td>Tested after 4 hours of punching</td>
</tr>
<tr>
<td>1.35</td>
<td>DP780-CR</td>
<td>492</td>
<td>820</td>
<td>20</td>
<td>25.8±2.3</td>
<td>Tested after 4 hours of punching</td>
</tr>
<tr>
<td>1.40</td>
<td>DP 780 CR</td>
<td>465</td>
<td>800</td>
<td>18</td>
<td>34.9±2.6</td>
<td>Tested after 30 minutes of punching</td>
</tr>
</tbody>
</table>

**Figure 9**

Pierced blanks of DP 590 and DP780 steel sheets after hole expansion tests. Pictures also show the close-up view of the hole after fracture appeared.

**Figure 10**

Hole expansion ratio of DP steels as a function of tensile strength.
considerations. The postulations of alloy design with a very low carbon content establishes the fact that DP590 and DP780 grade of steels can be successfully produced through mini-CSP mill with conventional CGL lines where soak temperatures are limited. A comparison of chemistry formulation with that practiced at conventional integrated thick-slab caster mills will indicate the current alloy design to be leaner in carbon contents but effective. The choice of small addition of Nb helped to get the required yield and tensile strength in DP780 grade steel.

Chemistry-specific use of mold powder guaranteed good surface quality of steels produced during the trials with a casting speed of 4.3–4.6 mpm. Surface quality obtained in DP780 grade 1.35-mm-thick steel is presented in Fig. 12 with a comparison to that typically obtained in extra-deep drawing steel (EDDS) sheet with equivalent coating weight. The surface presented excellent surface finish with no coatability or Zn wettability problem and compares well with EDDS sheet surface quality.

Microstructures presented in Figs. 4 and 7 for DP590 and DP780 steels, respectively, revealed a finer ferrite and martensite constituents. In both of the steels, martensite particles did not pose a banded orientation in spite of higher Mn contents. Probably, low carbon content and molybdenum contributed to the martensite morphology and distribution. This was also revealed in the retention of circular shape of the punched holes in both DP590 and DP780 steels after piercing in hole expansion tests (Fig. 9).

HER values obtained and shown in Table 7 are significant in the hot-rolled samples as well as in the cold-rolled galvanized DP samples. Excellent HER measured using dynamic image capture indicated high degree of cleanliness, finer microstructural constituents and material property homogeneity. A typical cleanliness of the steels is presented in Fig. 13 showing mostly globular oxides and oxy-sulfide inclusions indicating effective internal cleanliness control.
As elaborated earlier, ferrite strength was increased through an in-line SPM and TL of higher than 0.50%, which introduces dislocations strengthening. This was done to study influence of dislocation density of ferrite on the HER values, as opposed to solid solution strengthening.

Increased ferrite strength thus obtained minimizes the ferrite to martensite strength difference and hence resulted in higher HER values, as reflected in Fig. 14. Within the experimental data obtained from the current trial, it was observed that HER increased with an increase in YS/TS ratio and is reflected in Fig. 14.

FLDs of DP590 and DP780 steels presented in Fig. 11 indicated excellent material property homogeneity as there are less scatter in the strain data obtained for various strain paths.

Conclusions

A very-low-carbon-alloy concept for the production of DP steels of various strengths up to 780 MPa through a thin-slab caster was undertaken. Low-carbon-alloy concept is challenging for a mini-CSP mill considering highest levels of strength that can be achieved without heavy alloying notwithstanding operational hindrances. However, current study demonstrated that DP steels with tensile strengths up to 800 MPa can be successfully achieved in fully annealed galvanized conditions with leaner alloying than practiced possibly in many other mills. The study revealed potential for making DP980 fully annealed galvanized steel with a very-low-carbon approach.

Results of mechanical properties indicated excellent formability and stretch flangeability can be achieved that can compete with any such steels made elsewhere. HER tests and FLD evaluations indicated outstanding formability that was due to the internal cleanliness and property homogeneity in the steels. Finer grain structure with finer martensite constituents together with a near absence of banded structure probably resulted in outstanding formability properties. Surface quality attributes demonstrated chemistry-specific mold powder usage has an important role in guaranteeing quality of steel even when produced through EAF-based steelmaking in a CSP mill.

Acknowledgment

The authors are thankful to all those contributed to the successful development of the DP steels in the shortest possible time and resources. Big River Steel is thankful to research support provided by CBMM for the conduct of this successful development. The authors are grateful to the management of Big River Steel for their support.

References

Did You Know?

Metalloinvest Launches Ore Beneficiation Pilot Plant

Metalloinvest announced the launch of a pilot plant at Mikhailovsky GOK intended to model iron ore beneficiation processes and develop new high-quality products.

In a press release, the company said it developed the unique pilot plant with specialists from the TOMS Institute. “A beneficiation plant in miniature has been created. It allows us to simulate any existing or prospective beneficiation scheme in a short period of time and produce a prototype product,” said Rinat Ismagilov, director of the Mining Production Department of Metalloinvest. “We are able to trial the technology under conditions that are as close to industrial ones as possible, with minimal financial expenditure.”

According to the company, the pilot plant makes it possible to redirect product flows and include or exclude individual operations from the production process. This significantly reduces the time and cost of evaluating, designing and testing new ideas. The plant, which can be disassembled into individual modules for ease of transport, has, according to Metalloinvest, world-class equipment, automation systems and visualization processes.

The pilot plant is currently set up to develop new production processes for the concentrate post-treatment facility under construction at Mikhailovsky GOK and the flotation facility at Lebedinsky GOK, as well as oxidized quartzite enrichment technology.