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A study on the effect of chemical composition on the microstructural characteristics and mechanical performance of DP1000 resistance spot welds

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ABSTRACT

This paper reports on the factors governing the mechanical properties of resistance spot welded hot dip galvanized DP1000 under tensile-shear and cross-tension loading. In particular, the effects of chemical composition on the microstructural evolution and mechanical properties of DP1000 resistance spot welds are studied thoroughly, by comparison of a higher and lower carbon alloying approach. It is shown that DP1000 steel with higher carbon content attains a martensitic microstructure in the weld nugget with smaller prior austenite grains and finer block sizes. The intervariant boundary fraction analysis also reveals that DP1000 steel containing lower carbon content shows stronger variant selection as the fraction of variants belonging to the same Bain group is higher for this steel. Intervariant plane distribution also reveals that the most of intervariant boundaries for both steels terminated at or near (011) slip planes. Mechanical testing of the welds reveals that the steel with higher carbon content shows a better mechanical performance in tensile-shear test, whereas the DP steel with a lower carbon content exhibits higher maximum load of cross-tension test. The key factors controlling the mechanical response of resistance spot welds during two different mechanical tests are explored via nanoindentation, slit-milling method combined with digital image correlation and micro-cantilever bending. It is demonstrated that the strength and/or hardness of the weld nugget is the key parameter governing the tensile-shear strength of the spot welds, while the fracture toughness of the weld is the dominant parameter that determines the cross-tension strength.

1. Introduction

Dual phase (DP) steels are one of the most commonly used group of advance high strength steels (AHSS) in automotive industries. DP steel microstructure consists of a hard phase of martensite dispersed in the softer matrix of ferrite offering an excellent combination of strength, ductility and formability. Their excellent mechanical properties make them an ideal candidate to be used in car body structure providing a good crashworthiness while reducing the vehicle weight. Resistance spot welding (RSW) is the predominant joining method in automotive industry and its rapid thermal cycle combined with higher alloying element of DP steels compared to classic mild steels can lead to the formation of brittle microstructure in the weld nugget.

Both the mechanical properties and chemical composition of the base material are important parameters affecting the mechanical performance of the weld. It was shown that due to ultra-fast cooling of RSW process fully martensitic structure can be easily form in the weld zone. Oikawa et al. investigated the effect of base material strength on the cross-tension strength (CTS) of resistance spot welds [1]. It was found that the CTS increases with base material strength up to 590 MPa and decreases noticeably from 780 MPa upward. Radakovic and Tumuluru also showed that the CTS for the 980 MPa DP steel was slightly lower compared to that of the 780 MPa steel [2]. They speculated that the decrease in CTS is due to lower ductility of the 980 MPa steel. It was also reported that the base material strength affects the stress condition at the weld edge as the mild steels with lower strength are easy to bend. It shows lower shear stress at the edge of the weld nugget and thus a lower tendency to interfacial failure mode compared to AHSS [3]. AHSSs are also more susceptible to void and shrinkages in the fusion zone (FZ) because of their higher content of alloying elements.

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Pouranvari and Marashi studied the tensile shear mechanical properties of three different grades of DP resistance spot welds [6]. They found that interfacial mode susceptibility increases in the order of DP600, DP980 and DP780. Lower tendency of DP980 to interfacial failure mode than DP780 was attributed to higher HAZ softening of DP980.

The desired volume fraction of ferrite and martensite in DP steel can be obtained using a combination of chemical composition and heat treating parameters. As a result, the same grade of DP steel can have significant difference in chemical composition among different steel makers. While the strength and formability of DP steels have drawn many attentions in automotive industries, correct material selection for different part of the car body based on the spot weldability of DP steels must also be taken into account. Development of low carbon DP1000 steel aims at applying it to structural parts that protect the cabin when the vehicle crashes together with fulfilling the requirements of low carbon equivalents for spot welding for heavy gauge up to 2 mm in platform [7].

Although extensive effort was made to clarify the mechanical behavior and failure mechanism of DP steel resistance spot welds, it still lacks the detailed observation on the effect of base material chemical composition on the microstructural evolution of the weld nugget and its effect on the mechanical response of the weld. Besides, because of sample size constraints, it is not feasible to measure the local mechanical properties of the weld such as tensile and fracture toughness properties.

This work provides an insight into the microstructural characteristics of two DP1000 steels with different chemical compositions. Effects of carbon content as the main difference in the chemical composition of two steels on the crystallographic features of martensite in the weld nugget of two steels are studied in details via orientation imaging microscopy (OIM). The mechanical properties of resistance spot welds are evaluated via tensile-shear and cross-tension tests. For the first time, micromechanical testing combined with nanoindentation and slit milling methods are utilized to evaluate the factors controlling the failure behavior of the resistance spot welds.

2. Experimental

Two DP1000 hot dip galvanized steels labeled as LC (low carbon) and HC (high carbon) with the same thickness of 1.5 mm and different chemical compositions were examined. The chemical composition of the two DP1000 steels is given in Table 1. The carbon content of the LC steel is 0.061 wt% in contrast to the higher carbon content of 0.157 wt% for HC steel. Carbon equivalent (CE) numbers for LC and HC are calculated as 0.29 and 0.33 using the equation proposed by Ito et al. [12].

Resistance spot welds were produced using a 1000 Hz MFDC pedestal welding machine with constant current regulation and constant load of 4.5 kN. Welding electrodes (F1 16-20-5.5) and the weld scheme were taken from the VDEh SEP1220-2 welding standard [13]. For tensile-shear test a range of welding current from 4.8 to 8.4 kA was used to make weld nuggets of different sizes. In the case of cross-tension test maximum and minimum welding currents were selected for each material to produce the minimum weld nugget size proposed by standard ANSI/AWS/SAE [14] and maximum weld size before splash, respectively. For HC weld an extra medium welding current was also used as shown in Table 2.

The cross-tension properties were evaluated through the average value of four specimens with the same weld nugget size. Fig. 1 shows the schematic of the samples used for tensile-shear and cross-tension tests.

Cross sections of the welds were prepared with conventional metallographic methods and the microstructure was studied via optical microscopy (OM) and scanning electron microscopy (SEM). For OIM analyses, the samples were mechanically polished and then electropolished using a solution of 90% CH₃COOH + 10% HClO₄ at 20 V voltage and 21°C for a period of 25 s. The OIM characterization was carried out by electron back scatter diffraction pattern using a Philips ESEM-XL30 scanning electron microscope equipped with a field emission gun operating at 20 kV. Vickers microhardness measurements were performed at 200 g load for a loading time of 15 s. In order to extract the tensile properties of the weld nugget using the algorithm presented in Ref. [15], nanoindentation test was performed with a Berkovich indenter at the constant maximum load of 50 mN. A minimum number of 20 indentations were conducted for each sample. Micro-slit milling combined with digital image correlation (DIC) method presented in detail in Ref. [16] was used to measure the residual stress normal to the plane of the pre-crack at the weld edge.

### Table 1
Chemical composition of studied steels.

|          | C (wt.%) | Mn + Cr + Mo (wt.%) | Si + Al (wt.%) |
|----------|----------|---------------------|----------------|
| DP1000   |          |                     |                |
| LC       | 0.061    | 2.865               | 0.414          |
| HC       | 0.157    | 2.785               | 0.142          |

### Table 2
Welding parameters.

|          | Squeeze time (ms) | Weld current (kA) | Weld time (ms) | Hold time (ms) |
|----------|-------------------|-------------------|----------------|---------------|
| Tensile-shear welding | 550               | 4.8-8             | 380            | 300           |
| Cross-tension welding  | 550               | 6.4-7, 8          | 380            | 300           |

*a Only for HC welds.*
To evaluate the local fracture toughness, notched micro-sized cantilevers were milled in front of the pre-crack at the weld edge as described in Ref. [11]. Because of large plastic deformation during bending of micro-cantilevers, linear elastic fracture mechanics could not be used. Thus, cyclic loading was applied to measure the J-integral value at micro-scale.

The thermal history of the resistance spot weld was simulated using SORPAS® software, a dedicated commercial finite element code for resistance spot welding. The welding current, welding time, electrode force, sheet geometry and coating type were served as input data for the simulation. ISO5821 F1-16-20-5.5 CuCr1Zr electrodes with the geometry similar to the experimental configuration was defined. Besides, using SORPAS standard data base, 1000 Hz DC arbitrary projection/spot welding machine was selected. The material data for two steels including temperature-dependent and tensile properties were created using the combination of SORPAS material data base and experimental results. The tensile properties of the two steels are given in Table 3.

3. Results and discussion

3.1. Microstructural evolution

The image quality (IQ) maps of the LC and HC steels are shown in Fig. 2a and b, respectively. Both ferrite and martensite were indexed as bcc phase as the carbon content of both steels was not high enough to make large difference in the tetragonality of the structure in order to distinguish between bcc and bct phases directly. However, in the IQ map, martensite phase appears darker due to its higher dislocation density and lattice distortion, which enables to differentiate two phases. The LC steel yielded a fine and banded structure of martensite uniformly distributed in the ferrite matrix (Fig. 2a). The average diameter of the martensite islands was measured as 1.37 ± 0.57 μm. As shown in Fig. 2b, HC steel revealed a coarser blocky morphology of martensite along the grain boundaries and triple point of ferrite phase. The average diameter of blocky martensites for HC steel was measured as 3.03 ± 1.15 μm. Fig. 2c and d show the inverse pole figure (IPF) texture of ferrite and martensite phases of the two steels with respect to the rolling direction.

![Image Quality Maps](image1.png)

![Inverse Pole Figures](image2.png)

Table 3

|          | Yield Strength (MPa) | Ultimate Tensile Strength (MPa) | Elongation (%) |
|----------|----------------------|---------------------------------|----------------|
| LC steel | 683                  | 969                             | 13.8           |
| HC steel | 779                  | 999                             | 9              |

Fig. 2. Image quality map of the base metal LC steel (a) and HC steel (b), and inverse pole figure texture of ferrite and martensite phases of the LC steel (c) and HC steel (d) with respect to the rolling direction.

Fig. 3. Resistance spot weld cross section of LC (a) and HC (b) welds. IPF map of the fusion zone for LC (c) and HC (d) welds. (e) and (f) the corresponding reconstructed maps of PAGs. Black lines represent the grain boundaries with a misorientation angle ≥15°.
As illustrated, both steels show a strong texture of $<101>_{//}$ the rolling direction for both martensite and ferrite phases.

The cross section of the weld nuggets of LC and HC welds produced at 8 kA are shown in Fig. 3a and b. The dotted rectangle schematically illustrates the location of OIM scans. Fig. 3c and d show IPF maps of the weld nugget for LC and HC resistance spot welds, respectively. Both show the typical directional solidification structure, in the direction of the highest cooling rate. Black lines represent the grain boundaries with the misorientation angle of $>15^\circ$. The higher carbon content in the HC steel leads to the formation of a martensitic microstructure with much finer blocks and prior austenite grains (PAGs). Fig. 3 (e) and (f) show the reconstructed PAGs maps for LC and HC welds, respectively. PAG columns in the LC steel are wider up to $\sim 100$ $\mu$m and elongated along the radial direction of the weld nugget. In the case of HC steel, the PAGs become narrower ($<50$ $\mu$m) and the dendrites of the same morphology contain several PAGs inside.

Fig. 4. (a, c) IPF maps of the selected area in the weld nugget of LC and HC steels, respectively; (b, d) misorientation profile through the arrows in the LC and HC IPF maps. PAG, block and lath boundaries are marked by bold black, light black and white lines, respectively.

Fig. 5. PAG size distribution (a), block thickness distribution (b), misorientation angle (c) and intervariant length fraction (d) of LC and HC weld nuggets.
Fig. 4 shows magnified IPF maps for two welds together with the point to point and point to origin misorientation along the vectors. Blocks of martensite are separated by the grain boundaries with misorientation angle around 60° (black lines). Each block is composed of laths of martensite that are misoriented by low angle grain boundaries smaller than 15° (white lines). The misorientation profile in the HC weld reveals multiple peaks at 60° with a few microns width as opposed to the LC weld that shows few wider peaks at 60°. Several peaks are shown inside the blocks of LC weld corresponding to the lath substructure of martensite with low misorientation angle.

Statistical analysis of the PAG size is shown in Fig. 5 (a). The area of each grain is calculated by summing up the number of pixels in a grain (N), multiplied by the product of the square of the step size (600 nm) and a factor depending on the type of scan grid. The used scan grids were hexagonal and the factor for this kind of grid is one half times the square root of 3. The area fraction is obtained then by the area of each grain divided by the total area of all grains. The average PAG size decreases from 137 μm for LC steel to 67 μm for HC steel weld. The smaller PAGs sizes of the HC steel weld can be attributed to the segregation of carbon atoms at the boundaries leading to dragging effect on the grain boundary movement. Activation energy for grain growth was found to increase by addition of alloying elements like carbon [17]. Fig. 5 (b) shows the distribution of the measured block thickness in the nugget of two steels. Apparently, addition of higher carbon content to the chemical composition of the steel results in a reduction in the thickness of martensite blocks. The average block thickness in the HC steel is 2.4 μm as opposed to the thicker blocks of LC steel with an average thickness of 3.8 μm. In low carbon steels, the formation of martensitic lath in a large block is associated with large plastic accommodation in the parent austenite matrix. By addition of carbon, the austenite matrix becomes harder due to solid solution hardening. Thus the strain of the martensitic transformation cannot be relieved easily and self-accommodation by combination of martensite laths is intensified leading to the formation of finer blocks and packets [18]. Similar results were reported on the effect of carbon content on the martensite hierarchical structure by Zhang et al. [19] and Ryuo et al. [20]. Besides, Liang et al. [21] showed that the decrease in the PAG size leads to a reduction in packet and block size of martensite, which is in consistence with the current results of HC and LC martensitic welds.

Misorientation angle distribution shown in Fig. 5 (c) reveals a bimodal distribution for both weld with two peaks at low (∼5–10°) and high (∼50–60°) misorientation angles. Misorientation distribution of HC weld shows a weaker peak at low angle grain boundaries and stronger at higher misorientation angles. The increase in the fraction of high angle grain boundaries might be attributed to the finer blocks of the HC weld. The length fraction of intervariant boundaries between V1 and other variants were also measured assuming the K-S orientation relationship between prior austenite and martensite (Refer Fig. 5 (d)). More details on the grouping of martensite variants and their orientation have been presented in Ref. [22]. HC weld shows a reduction in the length fraction of intervariant boundaries shared between V1 and V4, V8 variants. These variants belong to the same Bain group and their misorientation angle is ∼10.5°. However, the fraction of V1/V3–V5 intervariant boundary with high misorientation angle of 60° is the largest for the HC weld. By contrast, the most frequently observed intervariant boundary for LC weld is V1/V4. The results obtained clearly show that the solidification structure of the resistance spot weld transforms to a microstructure with finer blocks of martensite separated by high angle grain boundaries and less tendency to variant selection by addition of higher carbon content.

The intervariant character distribution was carried out using a stereological five-parameter procedure deeply discussed in Ref. [23]. The characterization is based on the observation of many boundaries or segments (more than 50000 traces for cubic materials) in the 2D EBSD plane section. Each segment with a given misorientation is characterized by its own great circle. The single correct habit plane is appeared in the great circle of every segment/boundary with the same misorientation where the great circles intersect each other. Grain boundary plane orientation distribution was computed for all 24 misorientation angle/axes pairs proposed by K-S orientation relationship. The grain boundary plane distribution about [111] axis with different misorientation angles for two welds are shown in Fig. 6. In the case of 10.5° misorientation angle, LC weld shows a maxima at the position of the plane normal of twist boundaries at which the boundary plane normal is parallel to the misorientation axis (the circle mark). For HC weld, the distribution shows also multiple peaks on the zone axis of tilt boundaries as the zone of normals is perpendicular to the misorientation axis of <111>. For misorientation angles of 49.5° and 60°, the maximum distribution is mainly centered on the zone axis of tilt boundaries in the absence of any intensity at the zone axis of twist boundaries. Misorientation angle of 60° for two welds shows the highest peak at (110)//(111) symmetric tilt boundary as both side of the boundary have the same surface. Besides, the population of the symmetric tilt boundaries is higher for HC weld.

The intervariant plane distribution around [011] misorientation axis is shown in Fig. 7. Qualitatively, similar distribution of grain boundary planes is obtained for both welds for a given misorientation angle. The maximum intensity increases for two welds with increase in misorientation angle from 10.5° to 60°. Multiple peaks appear in the misorientation angle of 10.5° mostly centered on the (110) twist boundaries. A significant change in the distribution is observed for the misorientation angle of 49.5° as the maxima is only centered on the (110) twist boundary. Almost similar distribution is achieved for the misorientation angle of 60° for two welds, although HC weld shows much higher population at the position of (110) twist boundary.
The grain boundary plane distribution for misorientation angle/axis with higher index is shown in Fig. 8. The distributions of two welds mostly show a peak at or around (110) plane position. The maximum intensities of the distribution are lower for higher indices compared to the misorientation axis of [111] and [011]. Furthermore, no tilt or twist character is observed for the high index misorientation. The analysis reveals that the most of the intervariant planes in the microstructure of LC and HC welds are terminated at or near to (110) planes. The obtained results are in agreement with the reported grain boundary character distribution of lath martensite [23] and bainite [24]. It is attributed to the crystallographic constraint of shear transformation of martensite that leads to the formation of (110) planes that are not necessarily favorable from boundary energy point of view. It arises from the fact that the martensitic microstructure of resistance spot welds is evolved from a transformation process that is different from typical grain growth phenomenon, which mainly promotes the boundaries with less energy (i.e. (112) tilt boundaries for polygonal ferrite) [25]. As the majority of the intervariant boundaries end up on the (011) plane, the linear intercept between the boundaries proposed by K-S orientation relationship can be used as a measure to estimate the distance between (011) plane boundaries. The mean linear intercept between the intervariant boundaries with misorientation axis of [011] would represent the distance between (011) twist type planes as most of these intervariant boundaries end on twist type boundaries of (011) plane. Accordingly, the mean distance between the intervariant boundaries with misorientation axis of [111] can be used to estimate the distance between (011) symmetric tilt plane types. Fig. 9 shows the mean linear intercept of the (011) tilt and twist plane type for two microstructures as a function of misorientation threshold. In general, HC steel weld shows a smaller mean linear distance between (011) twist and tilt plane types compared to the weld of LC steel.

### 3.2. Mechanical properties

Tensile-shear and cross-tension tests were conducted to evaluate the mechanical performance of the resistance spot welds for two steels. Loading condition and stress distribution during mechanical loading can be summarized as shown in the schematic sketches of Fig. 10.
tensile-shear testing, the weld nugget is subjected to shear stresses, while the HAZ and BM experience shear in the thickness direction and tensile in loading direction (Refer Fig. 10 (a)). In cross-tension test, the main loading type in the weld nugget is tensile and mode I crack tip opening occurs at the weld edge. In the HAZ and BM main loading state is shear as well as bending moments during testing (Refer Fig. 10 (b)).

3.2.1. Tensile-shear results

The growth curve of weld nugget size versus welding current for two steels is presented in Fig. 11 (a). As shown in Fig. 3a, the weld nugget size is defined as the diameter of the weld at the sheet/sheet interface. Expectedly, the weld nugget size becomes larger with increase in welding current. Similar increase trend and also comparable weld nugget size are observed for both steels at different welding currents. Change in maximum peak load of tensile-shear test with weld nugget size is shown in Fig. 11 (b). A gradual increase in peak load with increase in weld nugget size is observed for two steels. The HC steel shows a higher strength than the LC steel during tensile-shear mechanical test.

Fig. 8. Plane normal distribution for the boundaries with higher index.
The LC steel shows interfacial failures (IF) for all the weld nugget sizes. The HC steel has a higher tensile-shear strength (TSS) for all the currents and its failure mode changes from interfacial (IF) to pullout failure (PF) mode as the weld size reaches to ~7 mm.

Fig. 12 shows the Vickers hardness distribution over the weld zones for two steels. The average hardness of the weld nugget for the LC is 361 HV, which is lower than the average hardness value of 415 HV for the HC steel. Upper critical heat affected zone (UPHAZ) of the HC steel also yield higher hardness compared to the LC steel. Both samples show softening at sub-critical heat affected zone (SC-HAZ) as there is decrease in hardness with respect to the hardness of base metal. However, the degree of softening is higher for the HC steel.

Based on the simplified stress distribution model during tensile-shear testing of resistance spot welds, the sheet interface plane in the weld nugget is subjected to a shear stress (see Fig. 10 (a)). Thus, the failure mode during tensile-shear is the result of the competition between shear plastic deformation at the weld nugget and necking outside the weld in the HAZ or base metal [4]. If the shear stress reaches its critical value before the necking in HAZ or base metal, the weld will fail in IF mode. A simplified analytical model can be developed to estimate the maximum load for IF mode during tensile-shear testing by assuming a cylindrical weld nugget with diameter of \(D\) [4]:

\[
F_{IF} = \frac{\pi}{4} D^2 \tau_{FZ}
\]

where \(\tau_{FZ}\) is the shear strength of the weld nugget. In order to evaluate the shear strength of the weld nugget, nanoindentation test were carried out and the obtained data were processed by the algorithm described in Ref. [15]. The average yield strength of the weld nugget for HC and LC steels are measured as 1435 and 1136 MPa, respectively. According to Mises-Hencky theory, the shear yield strength can be estimated as 0.577 \(\sigma_y\). Therefore, the shear yield strength of the HC and LC welds would be calculated as 827 and 655 MPa, respectively. It is already documented that the hardness and consequently the strength of the resistance spot weld does not change remarkably with change in welding current [27].

In order to assess the controlling factor of the peak load in the IF mode of spot welds during TSS test, the analytical fit was made using Eq. (1) as shown in Fig. 11 (b) for two welds. A very good agreement is found between the analytical fit obtained from nanoindentation test and the peak load of tensile-shear test. This confirms that the tensile shear and/or hardness of the weld nugget is the dominant controlling factor for the IF fracture during tensile-shear test of spot welds.

The yield strength of martensitic microstructure without considering precipitation strengthening is directly proportional to \(\Delta \sigma_{GB}\) (grain boundary strengthening) and \(\Delta \sigma_{dis}\) (dislocation strengthening). As indicated by OIM analysis, an increase in carbon content results in a pronounced decrease in the block size of martensite, increase in the fraction of high angle grain boundaries and less tendency for variant selection. Our previous work [22] has shown that the decrease in variant...
selection tendency results in smaller Bain packets, which are separated by high angle grain boundaries. Thus, the grain boundary strengthening factor is more effective in the case of the HC weld with smaller effective grain size. Disordered structure of high angle grain boundaries along the blocks, Bain packets and PAGs are strong barrier against dislocation movement leading to difficult dislocation slip along the grain boundaries. By contrast, low angle grain boundaries composed of aligned edge or screw dislocations have weak strengthening contribution against the dislocation mobility. Besides, it was already shown that the lath martensite with higher carbon content yields a substructure with higher density of dislocations [28]. High carbon martensitic microstructure with lower Ms, exhibits higher density of accommodated dislocations in PAGs, which leads to the increase in the energy barrier for the lath boundary migration. Thus, the LC weld with lower carbon content and coarser structure of martensite shows lower strength compared to HC weld containing higher carbon concentration and finer microstructure. As a result, TSS of LC steel is inferior to the HC resistance spot weld.

3.2.2. Cross-tension results

The cross-tension strength (CTS) for the minimum and maximum weld nugget sizes are shown in Fig. 13. As opposed to the tensile-shear test, LC steel exhibits better mechanical performance during cross-tension testing compared to HC steel. At smaller weld nugget size, HC fails in partial IF (PIF) mode, whereas the LC steel fails in PF mode both at small and large weld sizes. The difference in the mechanical behavior of the two steels welds during two different mechanical tests arises the key question about the controlling factor that determines the failure mode and peak load of spot welds during cross-tension test.

The main loading mode during cross-tension test at the weld edge is mode I under which the tensile stress normal to the plane of the crack is applied (see Fig. 10 (b)). Our previous investigation showed that the state and magnitude of the residual stress in front of the pre-crack may affect the crack opening and propagation during cross-tension test [9]. Slit milling method was used to evaluate the residual stress magnitude in front of the pre-crack for the two welds. Micrometer-sized slit was made parallel to the pre-crack at the weld edge and the residual stress normal to the plane of the slit and/or pre-crack was measured. Fig. 14 shows the surface displacement field measured by DIC after stress release for LC and HC welds with the nugget diameter of 7 mm. As shown, the decorating particles are displaced toward the slit after milling, which shows the presence of compressive residual stress normal to the plane of the pre-crack at the weld edge for both welds. The magnitude of residual stress perpendicular to the plane of slit was measured by empirically fitting the experimentally detected displacements with the displacements calculated from analytical solution for an infinite length slit in an isotropic linear elastic material [16].
$U_{dir}(d) = \frac{2.243}{E'} \sigma_{dir} \int_0^{\theta_0} \cos \theta \left( 1 + \frac{\sin^2 \theta}{2(1-\nu)} \right)^\frac{1}{2} \left( 1.12 + 0.18 \text{sech}(\tan \theta) \right) \text{d}u$

where $a_f$ is the depth of the slit, $E' = E/(1 - \nu^2)$, $E$ is the Young’s modulus, $\nu$ is the Poisson’s ratio, $\theta = \arctan(d/a)$ with $d$ the distance to the slit and $a$ changing between 0 and $a_f$. The elastic modulus of the welds was measured as 231 ± 15 GPa for HC steel and 228 ± 25 GPa for LC steel using nanoindentation. The fitted $\sigma$ value for the slit made in the nugget edge of LC and HC welds are shown in Fig. 14 (c) and (d), respectively. The larger displacement field obtained for the HC weld slit is because of larger depth of milling (3 μm) compared to the milling depth of LC weld slit (2.5 μm). Nevertheless, as illustrated, the magnitude of the fitted residual stresses for the two welds are very close, ~410 MPa. The thermal history of the welding process of two steels including peak temperature and cooling time (800-500 °C) was simulated using Sorpas software, and the result is shown in Fig. 15. A quite similar peak temperature and cooling time for different weld zones are obtained for the steels. Thus, the obtained results from residual stress measurement are not surprising as both welds are subjected to very similar thermal history leading to negligible difference in the residual stress.

Fracture toughness of the weld is another important parameter that can heavily influence the crack opening and propagation during mode I loading of cross-tension test. Notched micro-cantilever bending can be a versatile method to simulate the response of the microstructure zones of a weld to crack opening mode and subsequently to evaluate the fracture toughness of the weld quantitatively. Fig. 16 (a) and (b) show the location of the cantilever and loading direction schematically and a fabricated notched micro-cantilever, respectively. As the bending of the cantilevers is associated with large plastic deformation, cyclic loading was applied to measure the conditional fracture toughness value using $J$-integral method. Several loading and unloading steps with the rate of 20 nm/s were applied to monitor the crack propagation during bending. Fig. 17 (a) and (b) illustrate the load-displacement curves for the LC
Fig. 16. Fabricated micro-cantilever (a), the location of the micro-cantilever and loading direction showing schematically (b).

and HC resistance spot welds, respectively. Both cantilevers show strain hardening before reaching the maximum load followed by gradual decrease in load with further displacement. It is assumed that no crack propagation occurs during strain hardening, before reaching the maximum load. As seen, the crack propagation starts at almost same displacement for two cantilevers. However, the crack propagation for the LC cantilever is accompanied by higher load compared to the HC displacement curve excluding the triangle part defined by the stable crack propagation stages was linearly fitted. The intersection of the two lines holds an estimate for the critical J that indicates a transition from one stage to another. Once the J_{\text{cr}} is extracted from J curve versus crack extension, the conditional fracture toughness can be achieved by \( K_{\text{c}} = \sqrt{\frac{\pi J_{\text{cr}}}{\delta}} \). The \( K_{\text{c}} \) value for the LC and HC weld is measured as 43.7 and 35.9 MPa m^{1/2}, respectively.

The LC resistance spot weld shows higher fracture toughness compared to the HC weld. It can be attributed to the higher carbon content of HC steel that results in higher brittleness of the martensitic structure in the fusion zone of the weld. In fact, dislocation mobility is obstructed by high density of high angle grain boundaries in the fine structure of the HC weld. This leads to higher stress concentration at the grain boundaries and thus higher strength but low fracture toughness because of higher susceptibility for crack initiation. According to intervariant character distribution the (011) inter-planar distance for the HC weld is smaller than the LC weld. (011) is the slip plane of bcc structure, and since the lamellar structure of martensite is highly misorientated along these planes, the slip is likely to take place along <111> direction. Therefore, the slip distance is limited by the spacing between the boundaries that are terminated at (011) planes. 2 slip systems out of 12 equivalent (110)<111> slip system are activated at this situation. However, based on general plasticity, 5 independent active slip systems are required for a successful slip. Therefore, for the HC weld with a structure with smaller (011) inter-planar distance, the stress relaxation at the crack tip due to slipping is restricted and the crack propagation by fracture becomes more likely. Deteriorated fracture toughness properties were similarly observed for the steel with dense lamellar layers of (011) plane boundaries [24]. Although not shown here, grain boundary segregation of the weld structure can play an important role affecting the fracture toughness and crack propagation during cross-tension test. It was shown that during equilibrium solidification of the steel with carbon content of 0.07 wt% (similar to LC) liquid completely solidifies to \( \delta \) ferrite, whereas for the steel with higher carbon content of 0.14 wt% (similar to HC) a peritectic reaction occurs first during which the austenite forms from liquid/\( \delta \) ferrite. This leads to a higher segregation of alloying elements such as Mn and P at the solidifying grain boundaries of the weld and thus, deteriorated mechanical performance for the steel with higher carbon content [29]. The results obtained suggest that the failure mechanism and mechanical properties of the steel welds during cross-tension test are mainly governed by the fracture toughness of the weld.

A correlation was made between the beam stiffness for each unloading segment and the crack size using a finite element model. The measured crack size for every unloading segment for the two cantilevers are shown in Fig. 17 (c) and (d). Initial slow crack growth followed by stable crack propagation is shown for both cantilevers. However, the final crack size for the HC cantilever is larger compared to the cantilever of LC weld. The J-integral of ith unloading segment is calculated using:

\[
J_{Q_i} = J_i^\text{el} + J_i^\text{pl} = \frac{(K_{Q_i})^2(1 - \nu^2)}{E} + \left[ \frac{\eta(A_i - A_{i-1})}{w(t - a_{i-1})} \right] \left[ 1 - \frac{a_i - a_{i-1}}{(t - a_{i-1})} \right]
\]  

(3)

The initial elastic part of j-integral is calculated using \( K_{Q_i} \) that is obtained by setting \( F_Q = F_{Q,95} \) in the \( i^{\text{th}} \) unloading part. \( F_{Q,95} \) is the load obtained by making a construction line with 95% of the slope of the reloading part of every unloading segment. In the plastic part, \( \eta \) is a constant and equals to 2, \( A_i \) represents the area beneath the load displacement curve excluding the triangle part defined by the \( i^{\text{th}} \) unloading line, \( w \) is the width and \( t \) the thickness of the micro-cantilever and \( a_i \) represents the crack size for each unloading segment. Fig. 17 (e) and (f) show the plots of J value versus crack size for LC and HC micro-cantilevers, respectively. The data for two initial slow crack growth and stable crack propagation stages was linearly fitted. The intersection of the two lines holds an estimate for the critical J that indicates a transition from one stage to another. Once the J_{\text{cr}} is extracted from J curve versus crack extension, the conditional fracture toughness can be achieved by \( K_{\text{c}} = \sqrt{\frac{\pi J_{\text{cr}}}{\delta}} \). The \( K_{\text{c}} \) value for the LC and HC weld is measured as 43.7 and 35.9 MPa m^{1/2}, respectively.
4. Conclusion

The effects of the chemical composition on the microstructural evolution and mechanical properties of two resistance spot welded DP1000 steels are investigated. A new approach was utilized to measure the yield strength and fracture toughness at the weld edge of the resistance spot welds. The results obtained can be summarized as:

(1) The average PAG size and block thickness of martensite were measured as 137 μm and 3.8 μm, respectively, in the weld nugget of the LC weld. Higher carbon content of the HC steel led to a reduction in PAG size to 67 μm and block thickness to 2.4 μm in the weld nugget.

(2) OIM studies showed that the HC weld yields a martensitic structure with higher fraction of high angle grain boundaries and less tendency to variant selection.

(3) Intervariant character distribution analysis revealed that for both steel welds most of the intervariant boundaries terminate at {011} slip plane of bcc structure. It was shown that the {011} inter-planar distance for the HC weld is smaller than that of the LC weld.

(4) The HC steel welds show better mechanical performance during tensile-shear test, whereas the LC steel welds outperform the HC welds during cross-tension mechanical test.

(5) It is found that the controlling factor of interfacial fracture mode in tensile-shear test is the shear yield strength of the weld. The shear yield strength of the weld nugget was obtained using

Fig. 17. Load-displacement curves for LC (a) and HC (b) micro-cantilevers. Corresponding crack extension size for each unloading step (c, d) and J-integral plot versus crack size (e, f).
nanoindentation as 827 and 655 MPa for the HC and LC welds, respectively.

(6) Higher shear yield strength of the HC weld was attributed to the finer martensitic microstructure, higher fraction of high angle grain boundaries and less tendency to variant selection with smaller Bain packets.

(7) It was proved that the fracture toughness of the weld is the dominating parameter determining the strength and failure mode during cross-tension test. Using notched micro-cantilever bending enabled direct measurement of the fracture toughness at the weld edge as 43.7 and 35.9 MPa m$^{1/2}$ for the LC and HC welds, respectively.

(8) The higher fracture toughness of the LC weld was ascribed to the larger (011) inter-planar distance of the microstructure leading to stress relaxation at the crack tip due to easier activation of slip systems.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

CRediT authorship contribution statement

Ali Chabok: Conceptualization, Data curation, Investigation, Methodology, Formal analysis, Writing - original draft. Ellen van der Aa: Data curation, Investigation, Writing - original draft. Yutao Pei: Funding acquisition, Conceptualization, Methodology, Formal analysis, Writing - review & editing, Supervision.

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