

A Study on Improving the Mechanical Performance by Controlling the Halo Ring in the Q&P 980 Steel Resistance Spot Welds

Dileep Chandran Ramachandran ^{a,*}, Bruna Figueredo ^a, Oleksii Sherepenko ^a, Woosung Jin ^b,
Yeong-Do Park ^b, Elliot Biro ^a

^a Centre for Advanced Materials Joining, Department of Mechanical & Mechatronics Engineering,
University of Waterloo, Waterloo, Ontario N2L 3G1, Canada

^b Department of Advanced Materials Engineering, Dong-Eui University, Busan 47340, Republic of Korea

* Corresponding author: Dileep Chandran Ramachandran

Centre for Advanced Materials Joining (CAMJ),
Department of Mechanical & Mechatronics Engineering,
University of Waterloo,
Waterloo, Ontario,
Canada N2L 3G1

E-mail: dcramachandran@uwaterloo.ca

Abstract

This study focuses on improving the mechanical performance of third-generation Q&P steel resistance spot welds using a double-pulse welding cycle. Single and double-pulse welding schedules were implemented to assess the mechanical performance of the welds. Single-pulse welds exhibited poor cross-tension strength (CTS) values, failed around the fusion zone, and were accompanied by poor energy absorption capability. However, the double-pulse schedule showed improved CTS values by 33%, with an associated 110% increase in absorbed energy. The failure path observed from interrupted cross-tension tests showed that, in welds made using both pulsing schedules, failure proceeded along the fusion boundary and CGHAZ. In the single-pulse welds failed in brittle fashion, whereas the welds made with a double-pulse schedule exhibited a mixed (ductile and brittle) fracture morphology. The high-density microhardness mapping confirmed the presence of a localized softened zone (halo ring) adjacent to the fusion boundary in single-pulse welds. Strong elemental partitioning of Mn, Si, and C in the vicinity of the fusion boundary during long welding time was the primary cause for the halo formation. However, the halo ring was eliminated by performing a double-pulse weld schedule with 30 ms cooling time in between pulses; resulting in improved mechanical properties.

1. Introduction

Advanced high-strength steels (AHSS) are potential candidates for making new generation auto bodies to build economic and environmentally-friendly structures without compromising vehicle safety [1–3]. Among various grades, third-generation AHSSs, such as quenching and partitioning (Q&P) steels, are promising grades due to their exceptional combination of strength and ductility compared to the conventional dual-phase (DP) steels [4–6]. To produce body-in-white structures, resistance spot welding (RSW) is predominantly utilized in the automotive industry [7,8]. Hence, it is essential to investigate the resistance spot weldability of Q&P steels.

Knowledge of the weldability of Q&P steels is limited in the available literature [9–12]. Pouranvari *et al.* [13] reported that, the mechanical performance of the resistance spot welds was predominantly related to the failure location in the spot welds. The load-bearing capacity of the spot welds improves when the failure location shifts from sheet-sheet interface (interfacial failure) to through-thickness failure in either the base metal through heat-affected zones (HAZ), known as button pullout [13]. Recently, Liu *et al.* [11] examined the effect of double-pulse welding schedules on the mechanical properties of Q&P980 steel spot welds. Weld failures were concentrated on the fusion boundary and they associated the deterioration of mechanical properties of the investigated spot welds to liquation cracking. However, recent studies [14–16] showed that failure around the fusion zone was related to the localized softening of the fusion boundary called as ‘halo ring’. These studies hypothesized that the formation of this halo ring at the fusion boundary was due to partitioning of solute atoms, such as carbon, from the solid fusion boundary to the molten weld nugget during extended welding times.

Aside from local softening at the fusion boundary, softening has also been noted within the subcritical heat affected zone (SCHAZ) of the spot welds due to the tempering of martensite has been extensively reported in previous generations of AHSS such as DP steels [13,17–19]. Even though the SCHAZ softening is also prominent in most of the third-generation AHSS [10,11], recent investigations show the presence of a halo ring [14–16,20–22], which affects the mechanical performances of the spot welds. Raath *et al.* [20]

investigated the effect of welding schedules on the residual stress distribution of boron steel spot welds. They reported an increased hardness drop around the fusion zone with increasing welding time. Meanwhile, Li *et al.* [21] studied the magnetically assisted spot weldability of dual-phase steels. A clear drop in hardness around the fusion zone was detected. They suspected the microstructure of this region consisted of a mixture of bainite and austenite or ferrite and martensite. Whereas, Sherepenko *et al.* [14,15] investigated the effect of transient softening at the spot weld fusion boundary on mechanical properties by analyzing the weldability of 22MnB5 martensitic steel. The transient softening of the fusion boundary deteriorates the mechanical performance of the spot welds and the failure follows exactly through the fusion boundary. Mohamadizadeh *et al.* [22] examined the failure behavior of spot-welded ultra-high-strength 22MnB5 steel. Other than the reported softening in the SCHAZ, a transient softening at the fusion boundary was observed. The failure under tensile, and shear loading initiated and propagated through the softened fusion boundary, which deteriorated the mechanical properties of spot welds due to the formation of shear bands. Furthermore, EftekhariMilani *et al.* [23] pointed out that implication of short secondary pulse current on the third-generation 1 GPa AHSS anneals the primary nugget at the weld edge and it homogenizes the alloying elements and as a result, it significantly improves the cross-tension strength (CTS). Recently, Figueredo *et al.* [24] also observed the failure around the fusion boundary on the Q&P980 spot welds under cross-tension tests, which deteriorates the strength compared to the same strength DP steels.

Based on the literature, it can be summarized that localized softening around the fusion zone could be detrimental to the mechanical performance of spot welds. To date, only a handful of studies have been reported that looked at the effects of double-pulse welding on the mechanical performance of third generation AHSS. Hence, this study focuses on improving the mechanical performance of Q&P980 steel spot welds via welding schedule optimization (chemical homogenization) and possibly eliminating the influence of localized softening on the mechanical performance.

2. Materials and methods

The present study examines an industrially produced zinc-coated 1.2 mm thick Q&P980 steel with a composition of 0.23 wt.% C, 2.0 wt.% Mn, 1 wt.% Si, and other alloying additions. The microstructure of the base material consists of a combination of about 38% ferrite (F), 47% tempered martensite (TM), 9% fresh martensite (M), and 6% retained austenite (RA), as shown in Figs. 1(a) and (b). The microstructural constituents were identified according to their morphology as reported in previous literature [1,18,25,26]. The RA was characterized using the EBSD technique; the phase map is shown in Fig. 1(c). The carbon equivalent of the materials was calculated according to the method described by Yurioka *et al.* [27], where it was found that the carbon equivalent for Q&P980, is 0.65. The steels' material properties, including ultimate tensile strength (UTS), yield strength (YS), elongation, and hardness, are presented in Table 1.

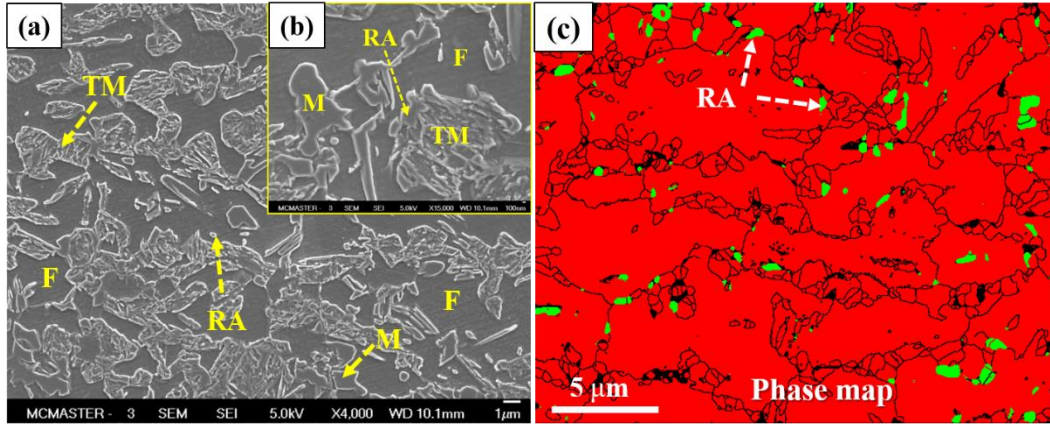


Fig. 1: Base metal micrographs of the Q&P 980 steel (a) electron micrograph (b) EBSD phase map to quantify retained austenite (RA).

Table 1: Mechanical properties of the Q&P 980 steel

Material	YS (MPa)	UTS (MPa)	Elongation (%)	Hardness (HV)
Q&P 980	640	1013	20	323

RSW was carried out using a robotic spot welder with a C-type transformer spot welding gun and a Rexroth Bosch weld controller. A 6 mm dome-shaped Cu-Cr electrode with an electrode force of 4 kN was used for the welding. AWS D8.9 recommended a single-pulse schedule [28] and a novel double-pulse welding schedule was experimented as well [29]. Both welding schedules are shown schematically in Fig. 2. For the single-pulse schedule, the welding current was selected as 8.0 kA, which is 0.2 kA less than the expulsion current (I_{max}), to maximize the nugget diameter (see Fig. 2(a)). For double-pulse schedules, a pause in the welding current (7.8 kA) between 15 to 60 ms was allowed before initiation of the second current pulse, then a constant current pulse of 8.2 kA was applied. Since the primary current was used to maximize the nugget diameter in this study, 7.8 kA was the maximum current that can be used without the expulsion, combined with a constant 8.2 kA secondary current, as shown in Fig. 2(b). These conditions provide a chemical homogenization effect in double-pulse welds.

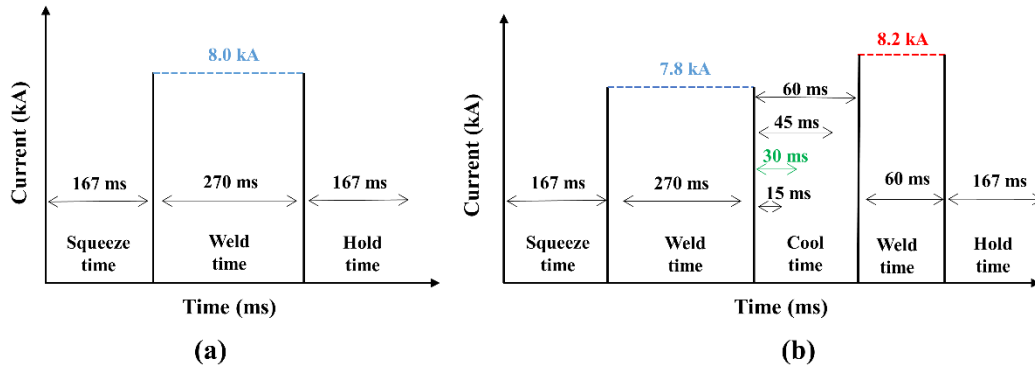


Fig. 2: Schematic diagram showing the (a) single, and (b) double-pulse welding schedule parameters.

Metallography was characterized using optical microscopy (Olympus BX51M) and scanning electron microscopy (SEM) after etching the samples with a 3% nital solution. The samples were prepared with conventional metallographic methods and etched with 3% nital to reveal the microstructure. A JEOL JSM 7000F field emission SEM was used for imaging and electron backscattered diffraction (EBSD) techniques. The EDAX orientation image microscopy (OIM) software was used for the EBSD data acquisition and post-processing. For EBSD analysis a step size of 0.2 μm was used for quantitative characterization of retained austenite. Electron probe micro analysis (FE-EPMA, model: JEOL JXA-8530F,

Japan) was performed to characterize the elemental distribution across the fusion boundary. Hardness measurements were carried out using an automated Clemex microhardness machine interfaced with a Vickers indenter. Hardness line scans were made using a 0.2 kg force and a spacing of 200 μm between the consecutive indents, whereas a 0.05 kgf force was used for hardness mapping with a spacing of 50 μm between the consecutive indentations. The hardness values are reported with a 95% confidence interval.

The CTS was measured according to the AWS D8.9 standard [30]. Mechanical tests were performed at a constant crosshead speed of 10 mm min⁻¹ using a universal tensile machine. The peak load and failure energy were extracted from the load-displacement curves. Three repetitions were done for each condition of the sample for consistent results. To identify the crack initiation and propagation mechanism during the CTS test, interrupted CTS tests were carried out. The tests were stopped after reaching 90% of the weld's peak load, after which, the weld was cross-sectioned and analyzed by optical microscopy.

3. Results

The weld made using a single-pulse condition was cross-sectioned. It had a weld nugget with a nugget diameter (ND) of 6.02 ± 0.11 mm, as shown in Fig. 3(a), which is the maximum achievable nugget diameter without an expulsion. On the other hand, Fig. 3(b) shows the macrostructure of the double-pulse schedule (7.8 kA-30 ms-8.2 kA) with an ND of 5.64 ± 0.14 mm. The double-pulse welds exhibited two distinct nuggets; a primary weld nugget produced by 7.8 kA of welding current followed by a secondary weld nugget produced by 8.2 kA of secondary welding current, as shown in Fig. 3(c). The secondary nugget grows by remelting the primary nugget with the secondary current pulses, as the cooling time between the pulses is comparatively shorter (30 ms). Furthermore, an equiaxed grain structure was seen in the interface between the primary and secondary nugget (see Fig. 3(d)) which confirms the secondary current reheats the primary nugget FZ nearer to the fusion line, making a chemically homogeneous final nugget.

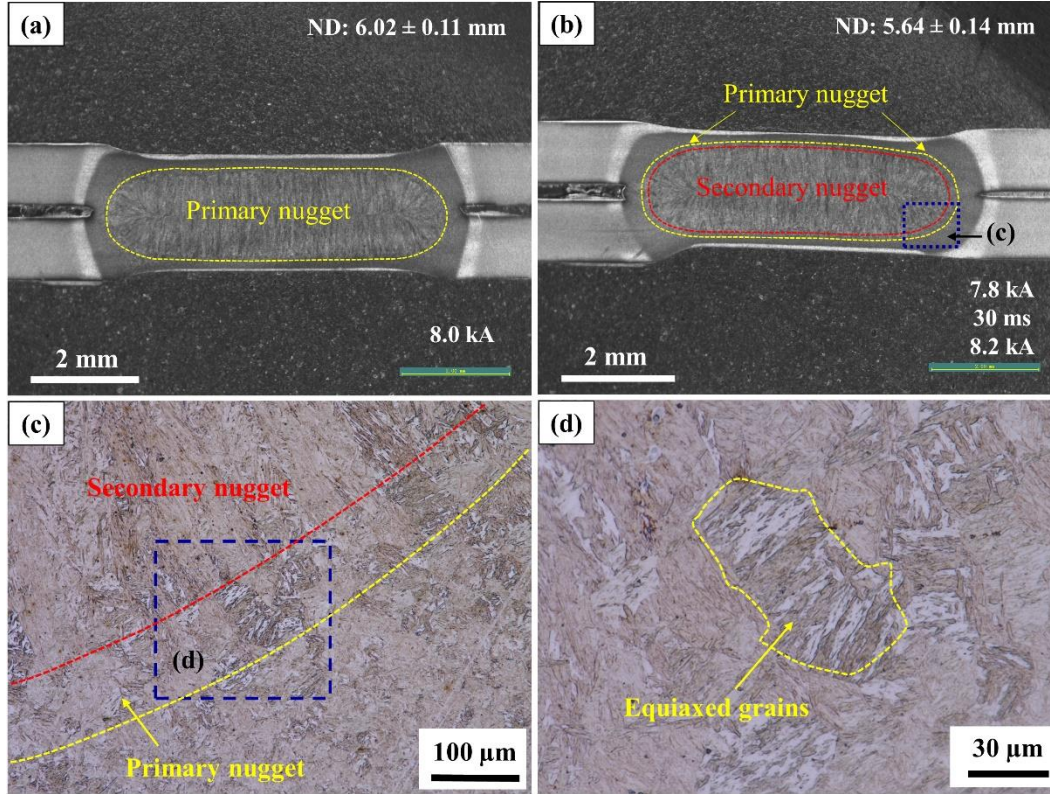


Fig. 3: Cross-sectional micrographs of (a) single-pulse, and (b) double-pulse with 7.8 kA primary current. (c) Higher magnification image showing the primary secondary nugget interface, and (e) equiaxed grain present in the interface

The optical micrograph of the single-pulse weld in Fig. 4(a) shows different areas of heat-affected zones (HAZ). The microstructures were characterized by analyzing their morphology and identified using similar micrographs found in the literature [26,31]. The fusion zone (FZ) consisted of a fully martensitic structure, oriented towards the weld centerline as shown in Fig. 4(b). It has been reported that the cooling rate after spot welding is on the order of 2000 °C/s [13]. Hence the microstructures of the FZ and HAZ, resulting from both welding schedules, were predominately martensitic. Similarly, the coarse-grained HAZ (CGHAZ) and fine-grained HAZ (FGHAZ) are predominantly martensite (see Fig. 4(c) and (d)).

Intercritical HAZ (ICHAZ) (where the peak temperature during welding ranged between the A_{c1} and A_{c3} temperatures of the material) consisted of a combination of complex microstructures such as TM,

F, and RA, as shown in Fig. 4(e). The intercritical temperature, rich chemistry, and high cooling rate stabilize the RA even at room temperature. Fig. 4(f) shows that in subcritical HAZ (SCHAZ) there is a combination of TM and F is the predominant microstructures. This tempered martensite is a mixture of TM from the base material, and TM formed during the welding cycle, as it has been widely reported that temperatures between 300 to 650°C are highly susceptible to the tempering of martensite [18,26]. The double-pulse condition showed a similar microstructure as the weld made using the single-pulse condition within the FZ and HAZ.

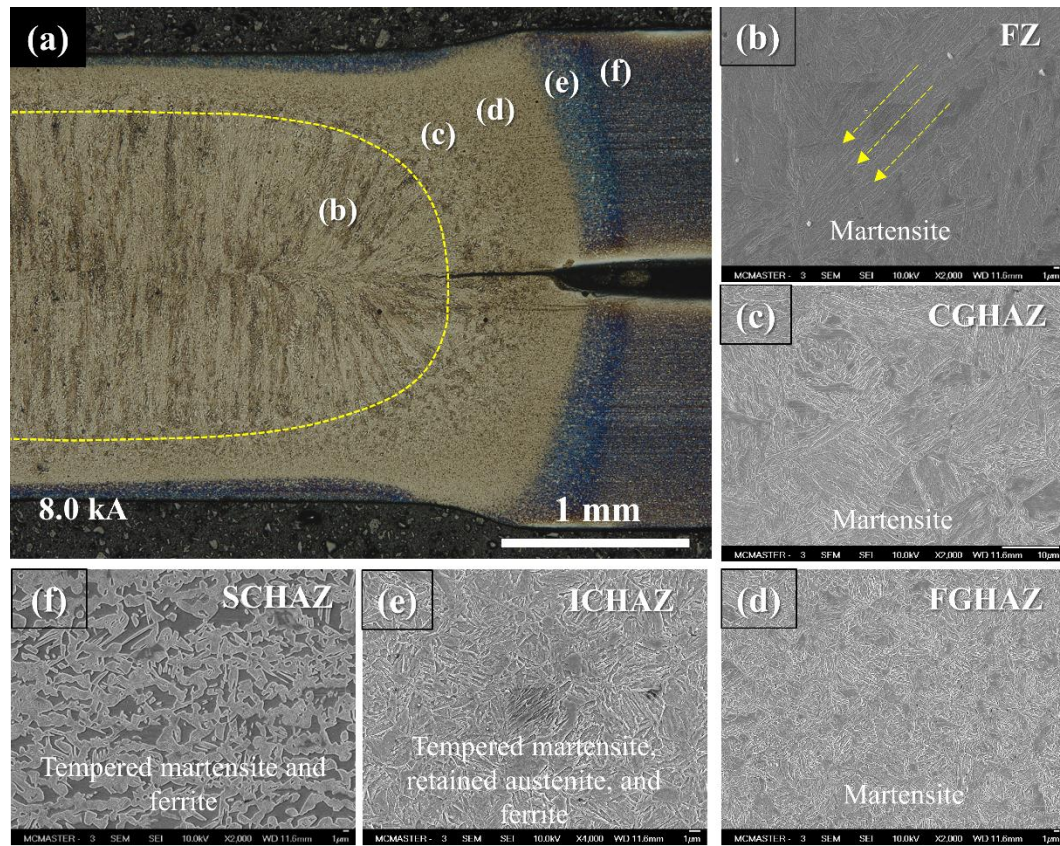


Fig. 4: Microstructure analysis of the Q&P 980 spot weld. (a) Macrograph showing the overall view of the FZ, HAZ, and, base material, (b) FZ, (c) CGHAZ, (d) FGHAZ, (e) ICHAZ, and (f) SCHAZ

Fig. 5(a) shows the load-displacement curve during the cross-tension test of both single and double-pulse welds. Pulsing significantly improved the mechanical performance of the welds. Moreover, as the

cooling time between the pulses increases, the CTS values increase to a maximum and subsequently decrease after 30ms cooling time. According to the AWS D8.1, the area under the load-displacement curve up to failure can be considered as the absorbed energy on the spot welds during loading until fracture [32]. The higher absorbed energy represents the improved weld performance. Hence, the absorbed energy is also in-line with the higher CTS values, as shown in Fig. 5(b). It has been reported that CTS values of the spot welds are predominantly related to the nugget diameter of the welds, as the nugget is the primary load-bearing area in the cross-tension test [33]. Considering this, it should be noted that even though the nugget diameter (ND) decreased from 6.02 ± 0.11 mm to 5.64 ± 0.14 mm (see Fig. 3), as the interpulse cooling time increased and the ND of the double-pulse weld made with 30 ms cooling time is smaller than the single-pulse condition, the CTS and absorbed energy values for the double-pulse condition using the 30 ms cooling time are approximately 33% and 110%, respectively, higher than that of the single-pulse condition. In other words, the mechanical performance of the welds was improved to 33% by implementing double-pulse welding schedules, specifically in the case of 30 ms cooling time even though the nugget diameter for the double-pulsed condition was smaller. It has been noted that the authors tried to match the single and double-pulse welds nugget diameters. However, early expulsion occurring in double-pulse welds due to the high current near to the maximum possible current ($I_{\max}$) made this impossible.

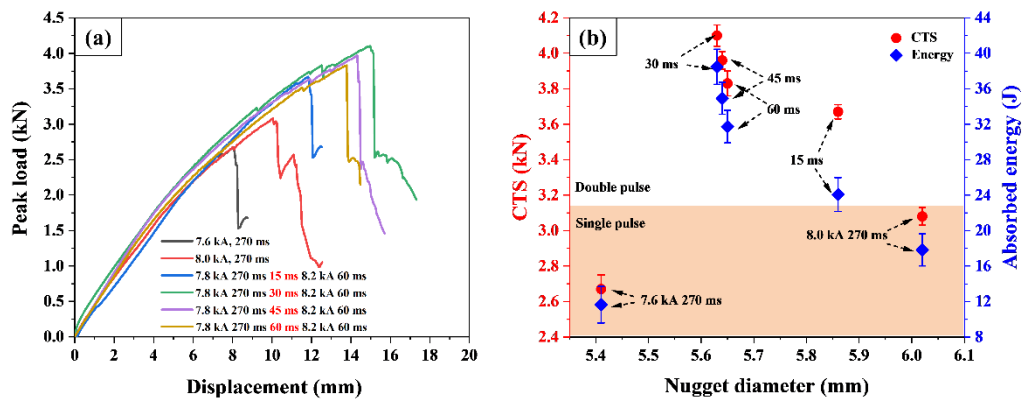


Fig. 5: The cross-tension test results of the spot welds. (a) load-displacement curves, and (b) relationship between nugget diameter, CTS, and absorbed energy

An interrupted cross-tension test was also conducted to understand the failure behavior of the weld joints. The single-pulse and double-pulse conditions with a 30 ms interpulse cooling time were chosen for the comparison as these conditions exhibit the lowest and highest CTS, respectively. Although in past studies RSW made with high-strength steels have failed in the SCHAZ [13,17,34], this was not seen in the present work. Examination of the welds in the current study that were loaded to 90% of the failure load showed that failure initiated at the fusion boundary for both the single and double-pulse welding conditions, as seen from Figs. 6(a-f). Recent studies show that in high-strength steels with rich chemistry, stress concentration always occurs at weld notch due to the geometrical factors, especially in steels that do not exhibit SCHAZ softening [16]. Hence it is likely that failure can also initiate at the sheet-sheet interface. The cross-sectional micrographs after the interrupted cross-tension test are shown in Figs. 6(e) and (f) confirm that the failure occurred along the fusion boundary or through the CGHAZ despite the welding conditions. In other words, failure occurred along the fusion boundary in single-pulse weld, whereas the failure likely occurred through the CGHAZ in double-pulse weld. How pulsing and this failure initiation location resulted in the difference in peak load and absorbed energy between the single and double-pulse welds will be discussed in the following sections.

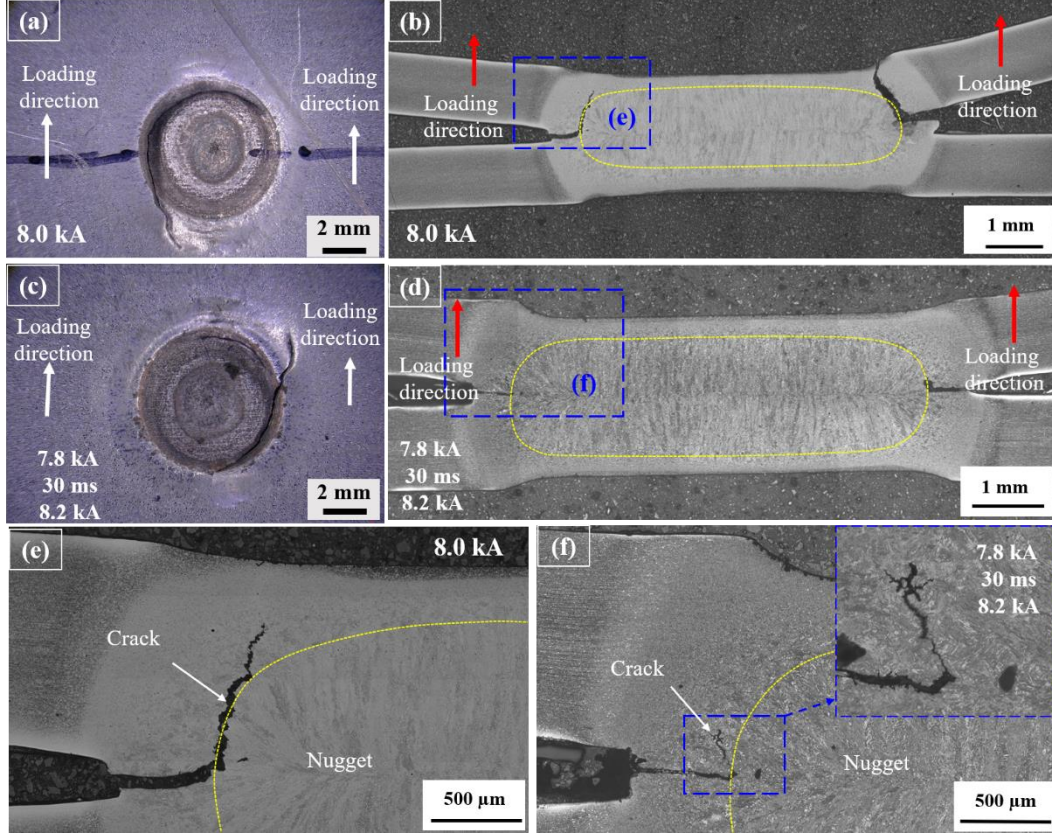


Fig. 6: Macroscopic images and weld cross-section after the failure and interrupted cross-tension test respectively. (a) and (b) single-pulse weld, (c) and (d) double-pulse weld. Optical micrograph showing the (e and f) failure path in single and double-pulse welds after the interrupted test.

4. Discussion

The high-density microhardness map and corresponding microstructure in Fig. 7 show that the soft zone exists at the nugget periphery in the single-pulsed weld (Figs. 7(a) and (c)). The hardness of this soft zone was still relatively higher than the base material hardness. Interestingly, the double-pulse welding schedule using a 30 ms cooling time, greatly reduces the severity of the softened zone along the fusion boundary, having a hardness of 519 HV compared to 436 HV in the softened zone of the fusion boundary of the single-pulsed weld, as seen from Figs. 7(b) and (d). These features are similar to those observed by

Sherepenko *et al.* [14,15] who reported that, in ultra-high-strength steels, localized softening may occur at the fusion zone boundary. It was claimed that this occurs by diffusion of carbon from the solid heat-affected zone to the molten weld pool causing the transient softened zone around the nugget [14,16], which has been called the ‘halo ring’.

The average base metal and CGHAZ hardness values for weld made in both the single and double-pulsed conditions are 314 ± 3 HV, and 540 ± 5 HV, respectively, as seen in Fig. 7(e). However, the average fusion zone hardness is slightly different in both welds, where the average fusion zone hardness of the single-pulse and double-pulse welds are 550 ± 5 HV, and 536 ± 5 HV, respectively. The width of the halo ring is approximately 50 μm and the hardness of the halo ring region is around 428 ± 37 HV, which is $\sim 20\%$ lower than the fusion zone hardness in the single-pulsed condition. Furthermore, as mentioned above, only a minor hardness decrease was detected at the fusion boundary of the weld made using the double-pulsed condition. The high-density hardness map along the fusion boundary is also shown in Fig. 7(e). It is worth noting that, the difference in fusion zone hardness between the two conditions could be attributed to the cooling rate associated with the two different fusion zones.

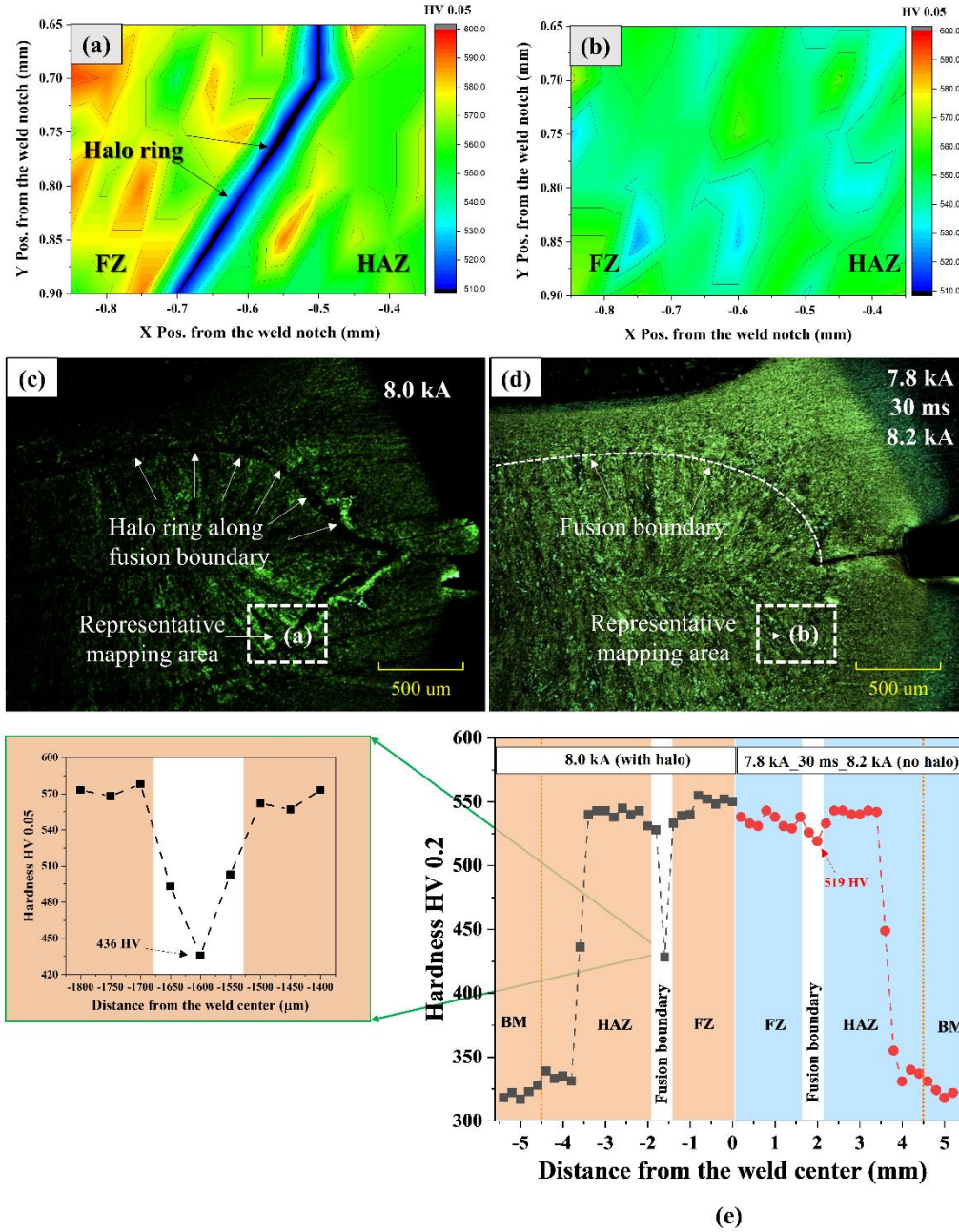


Fig. 7: High-density microhardness map and corresponding macrostructure of (a) and (c) single-pulse weld, (b) and (d) double-pulse weld, and (e) quantitative hardness map from center of the weld to base metal for both single and double-pulse condition.

Recall, cross-sections of the interrupted cross-tension tests (see Fig. 6) showed both welds failed near the fusion boundary. However, a clear difference in fracture morphology can be seen when the fracture

surfaces were viewed at higher magnification in Figs. 8(b) and (c). Figs. 8(a) and (d) show the electron micrographs of the fracture surface of single and double-pulse conditions, respectively. The crack initiation area has been identified by the interrupted CTS and is shown in Figs. 6(e) and (f). It is worth noting that failure surface resulting from the area where the crack starts at the sheet-sheet interface and propagates towards the fusion boundary through CGHAZ shows a small area of ductile dimple fracture in both the conditions (see Figs. 6(a) and (b)). In the single-pulse welds, once the crack reaches the halo ring, it propagates through the fusion boundary in a brittle manner as shown in Figs. 8(e) and (f), due to the halo ring's mechanical contrast with the adjacent microstructure. Even though the crack mainly propagates through the halo ring, some portion of crack (~21% of total crack length) also went through the CGHAZ, in the single-pulse weld (see Fig. 9(a)). The described small changes in fracture path towards the FZ or CGHAZ during normal loading are also reported in the literature [11,14,22]. The areas where the crack propagated through the CGHAZ failed in an intergranular fashion, as shown in Fig. 9(c). The failure mostly occurred (~ 54% of total crack length) through the halo ring, where it propagated in a brittle (cleavage) fashion (see Fig. 8(e) and Fig. 9(d)). Once the sample exhibits significant bending at the end of the test failure proceeds by shearing through the remainder of the area (~25% of total crack length). The double-pulse weld, failed through a similar path, however, the fracture mode differed. When the crack propagated through the CGHAZ, failure progressed in a mixed failure mode (brittle and ductile), as shown in Figs. 8(b), (d), (g), and (h). Despite the failure location, all the failure surfaces show a mixed mode of fracture (mixed with cleavage and dimple fractures). The failure path is chaotic and passes through fusion boundary, CGHAZ, and FGHAZ with a failure percentage of ~7%, ~59%, and ~34% of total crack length, respectively (see Figs. 9(b) and (e)). Failure through the FGHAZ is complete shearing as shown in Fig. 8(b). Moreover, the load-bearing capacity (Fig. 5) of the welds shows in line with the current observations.

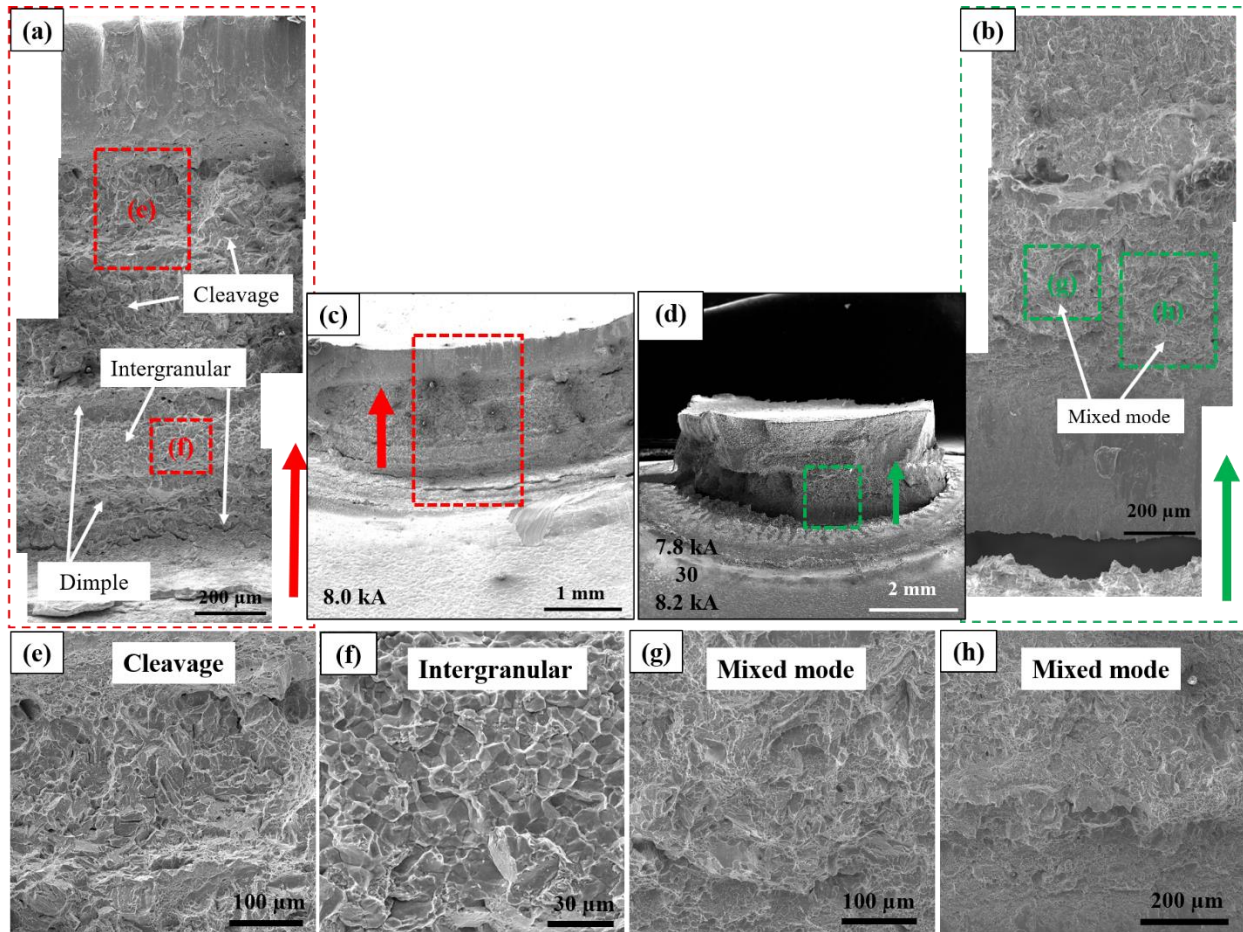


Fig. 8: Fractographic analysis of (c) and (d) single-pulse welds and double-pulse weld, respectively. (a) and (b) Panoramic images of the highlighted area in Fig. 8(C), and Fig. 8(d), respectively. High magnification images of the highlighted area (e) and (f) in Fig. 8(a) show cleavage and intergranular fracture, and (g) and (h) in Fig. 8(b) shows a mixed mode of fracture.

The absorbed energy (see Fig. 5(b)) of the joints agrees well with the above observations on fracture morphology. In other words, the fracture surface is brittle in the welds made with a single-pulse, whereas mixed mode fractures with significant ductile components in the welds were made in the double-pulsed condition, as shown in Fig. 9. Moreover, secondary cracks through the grain boundary also can be seen on the single-pulse weld, as seen from Fig. 9(c). It should be noted though, Mohamadizadeh et al. [22] observed shear banding along the halo ring in 22MnB5 spot welds during normal loading, which was contrary to the brittle fracture observed through the fusion boundary in the present study. It can be hypothesized that the

observed brittle intergranular fracture in the present study occurred through the near side of the CGHAZ or hard interface adjacent to the halo ring. EftekhariMilani *et al.* [23], reported that the Mn, Si, and P segregation at dendritic boundaries of third-generation 1GPa AHSS during single-pulse welds, embrittled the grain boundaries, thereby deteriorating the CTS values. This phenomenon could be the reason responsible for the weakening of CGHAZ in the current study. Due to the microstructural complexity and small width of the halo ring, further investigation of the fracture surface is necessary to find out the exact crack initiation and propagation regions and will be presented in future works.

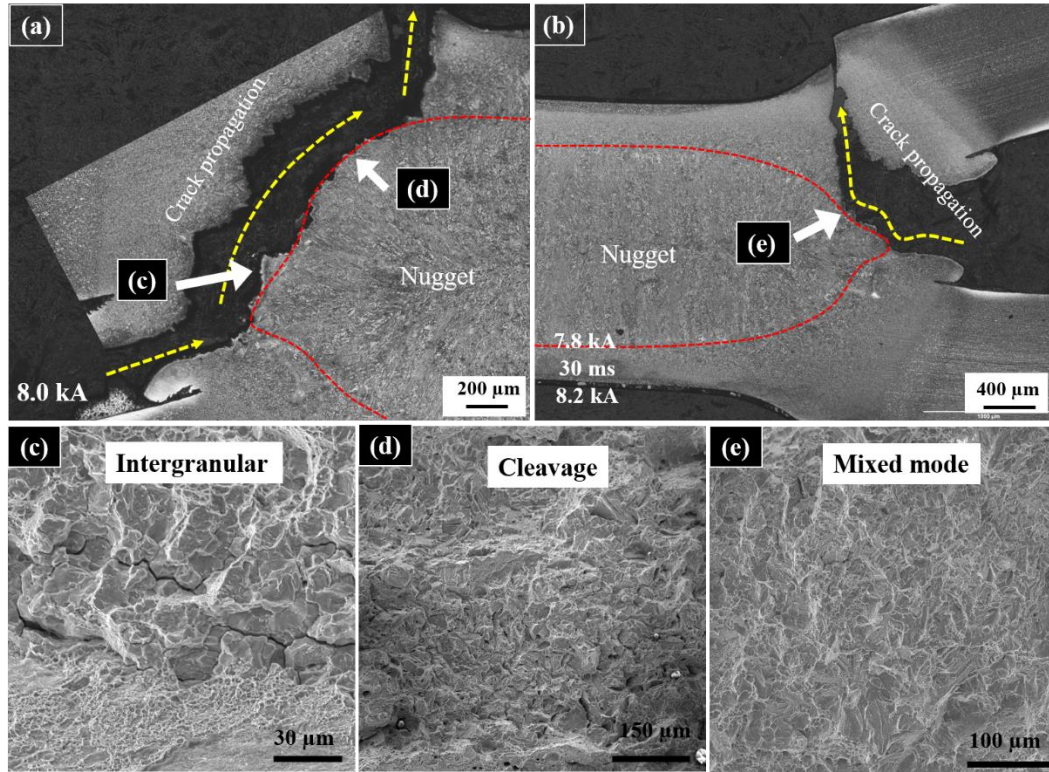


Fig. 9: Cross-sectional images showing the samples after CTS (a) single-pulse weld, (b) double-pulse weld, (c) and (d) brittle intergranular and cleavage fracture observed in Fig. 9(a), and (e) mixed-mode fracture observed in Fig. 9(b).

According to the previous research, Q&P steel spot welds fail along the SCHAZ due to martensite tempering or along the partially melted zone (PMZ) due to liquation cracking [9,11]. Interestingly, in the

current work, it has been observed that failure likely occurred along the fusion boundary and CGHAZ. The presence of a halo ring along the fusion boundary is the primary cause for the decreased CTS of single-pulse welds. However, a 33 % increase in peak load and a 110% increase in absorbed energy were observed in double-pulse (30 ms cooling time) welds, as shown in Fig. 5(b). It has been hypothesized that prolonged heating of the weld nugget by providing a long welding time could be the main reason to form the halo ring around the nugget. It can be either controlled by remelting the primary nugget boundary by secondary nugget for a very short time or the chemical homogenization (reheating) of primary nugget by secondary nugget. In this research, the authors approached the latter idea.

Several hypotheses have been asserted to explain the formation of the halo ring recently. Li *et al.* [21] suggested that the carbon depletion from the fusion boundary during welding could be responsible for this localized softening. Whereas, Sherepenko *et al.* [14,15] proposed that formation of δ -ferrite due to the long isothermal soaking of the fusion zone in the peritectic region caused this softening. Nevertheless, it can be hypothesized that fundamentally the halo formed due to diffusion of carbon and other alloying elements from the solid fusion boundary into the molten nugget. To validate this hypothesis, the elemental analysis across the fusion boundary was investigated using the electron probe micro analysis (EPMA). The EPMA measurements showed strong segregation of Mn, Si, and C at the fusion boundary (Fig. 10). The locations of this EPMA line analysis on single and double-pulse welds were shown in Fig. 10 (a) and (b). In the single-pulse weld, the segregation of Mn, and Si were seen in the fusion boundary, which is approximately 10 μm in width towards the center of the nugget (see Fig. 10(c)). Moreover, the C depleted region was seen inside the halo ring, and a strong C gradient was seen near the fusion boundary (see Fig. 10(e)). The average C content in the FZ is 0.25 wt.%, whereas in the halo it is 0.21 wt.% (~16% reduction). This C gradient forms due to a relatively long time that the fusion boundary remains stationary when welding for long weld times, such as 270 ms (see Fig. 2(a)). During the welding cycle, nugget growth ends before the termination of the welding current due to the cooling from the electrode. The growth of the fusion boundary stops when heat generation slows to the point where it equals the rate of heat extraction at the

electrodes. At this point elements in the solid HAZ will diffuse to the liquid nugget, where their solubility is higher, as proposed by Sherepenko *et al.* [15]. The strong partitioning of Mn, Si, and C in the vicinity of fusion boundary creates a hardness gradient which can potentially act as local weak point during deformation.

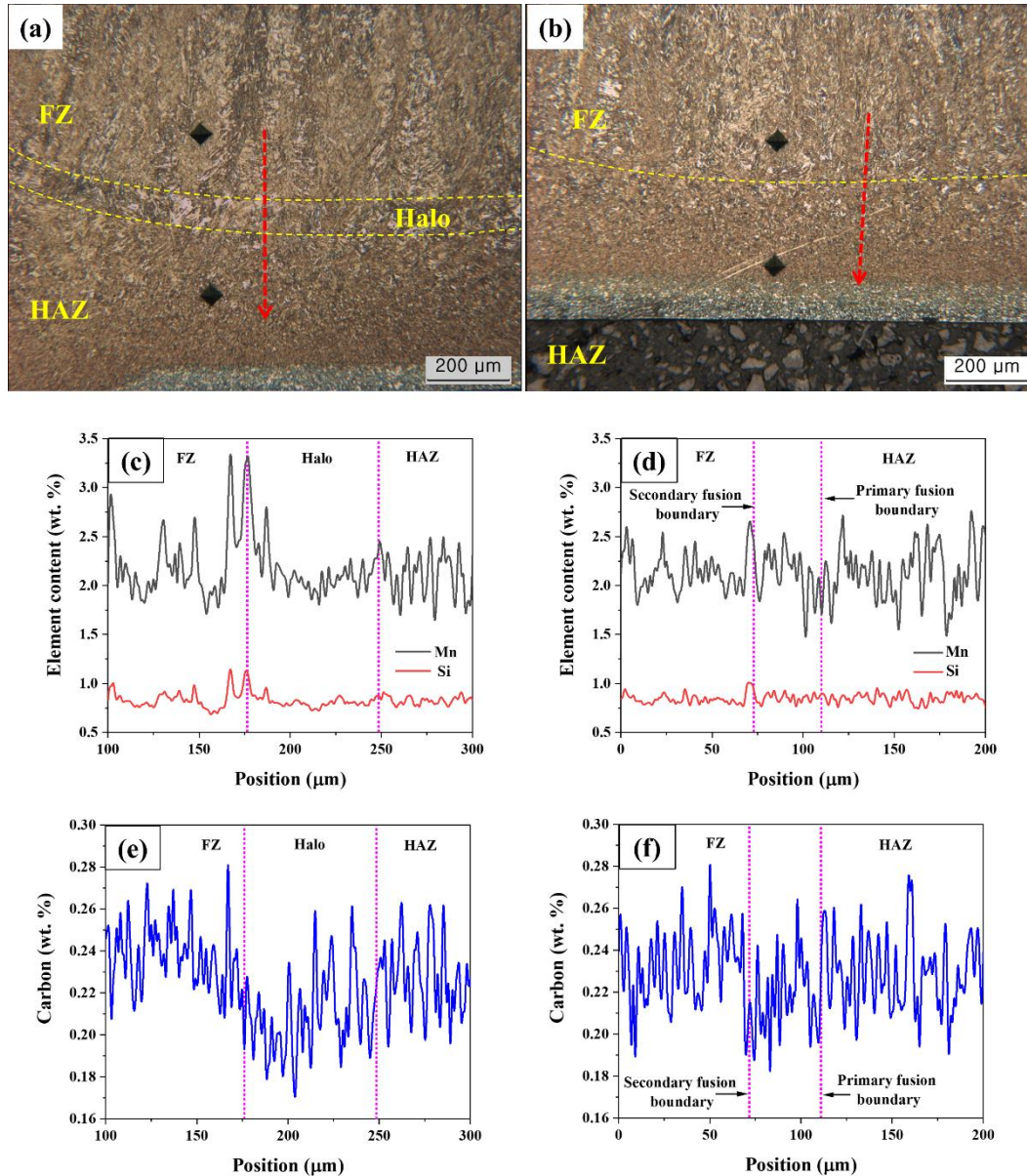


Fig. 10: Optical micrographs and corresponding electron probe micro analysis of Mn, Si and C across the fusion boundary: (a, c, and e) single-pulse weld with halo ring, and (b, d, and f) double-pulse weld without halo ring

Elemental partitioning has been considerably reduced in the fusion boundary of double-pulse welds (see Fig. 10 (b, d, and f)). However, there was still a small area ($<5\text{ }\mu\text{m}$) which shows the segregation of Mn, and Si near the secondary fusion boundary, as seen from Fig. 10(d). The region between the primary and secondary fusion boundaries were shown in Fig. 10(d) and (f). Furthermore, in the double-pulse weld, the average C content was 0.23 and 0.22 wt.% in the secondary nugget, and between the primary and secondary nuggets, respectively. This lower C content in the FZ also justifies the lower hardness value of the double-pulse weld compared to the single-pulse weld (see Fig. 7(e)). Hence, C partitioning was not significant in the fusion boundary of double-pulse welds, as compared to single-pulse welds, as the second pulse homogenizes the alloying elements in the primary nugget boundaries, redistributing the alloying elements at the primary and secondary nugget interfaces. The homogenization process improves the mechanical performance of the double pulse welds. Chen *et al.* [35,36] also reported that the implementation of stepped current pulses (more than one pulse) in welding schedules can improve the mechanical performances of the spot welds. It should also be noted that the double-pulse welding schedule implemented in this study needs further improvement to completely eliminate this elemental partitioning.

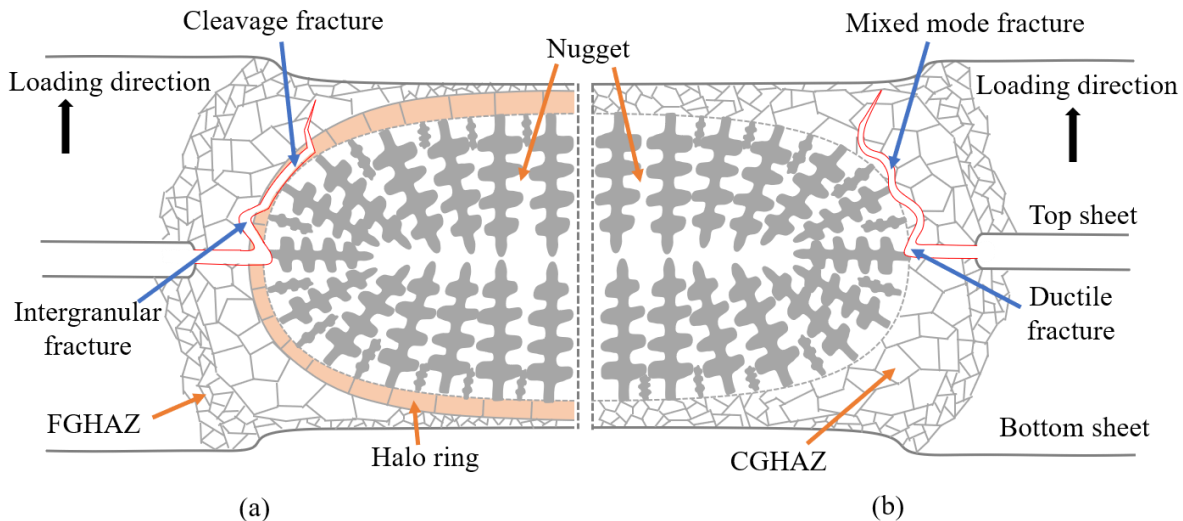


Fig. 11: Schematic representation of the fracture path in (a) single-pulse weld showing fracture along the halo ring, and (b) double-pulse weld showing fracture along the CGHAZ

The failure path of single and double-pulse spot welds is schematically shown in Fig. 11. The failure path resulted in a full pullout failure mode in both conditions. However, the fracture path is significantly different in single and double-pulse conditions as shown in Fig. 11(a) and (b), respectively. In single-pulse condition, the failure predominantly progresses through the fusion boundary (halo), whereas in double-pulse it remained in the CGHAZ. It is worth noting that, in both conditions, the fracture initiated at the sheet-sheet interface propagated towards the fusion boundary before following an independent path based on the microstructure resulting from the welding parameters used.

5. Conclusions

A systematic study was conducted on improving the mechanical performance by implementing a double-pulse welding schedule to minimize softened zone around the fusion boundary of Q&P 980 steel spot welds, and the following conclusions were drawn from the study. Detailed microstructural characterization and mechanical property correlation were also done in this study.

- 1) Double-pulse welding schedules were successfully implemented in Q&P 980 steels. Applying a higher second current with a short cooling time (30 ms) between the current pulses improves CTS and absorbed energy by 33 % and 110%, respectively, compared to the mechanical properties of a single-pulse weld.
- 2) Spot welds made with a single current pulse had poor CTS in the order of 3.1 ± 0.05 kN, whereas in double-pulse welds CTS improved to 4.1 ± 0.06 kN.
- 3) The halo ring was seen in the single-pulse welds, and it was eliminated through the pulsing technique (high secondary current for a short period of time).
- 4) The halo ring was chemically characterized by EPMA analysis. Carbon partitioning (carbon depleted halo ring), accompanied by Mn and Si segregation, was observed at the fusion boundary, which was attributed as the responsible mechanism for halo formation

- 5) The strength of the Q&P980 spot welds is predominantly related to the fracture type and its location. Both the spot welds show plug failure mode, however, the spot weld made with a single-pulse condition shows brittle fracture along the fusion boundary (halo), whereas the spot welds made with double-pulse condition failed along the CGHAZ in a mixed mode manner.

Acknowledgment

The authors would like to acknowledge the National Science and Engineering Research Council (NSERC) of Canada, for providing financial support to carry out this work. Electron microscopy was performed at the Canadian Centre for Electron Microscopy (also supported by NSERC and other government agencies).

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