Improvement of Microstructures and Mechanical Properties of Resistance Spot Welded DP600 Steel by Double Pulse Technology

Ning Zhong1,2, Xinshen Liao2, Min Wang3, Yixiong Wu3 and Yonghua Rong2,*

1Institute of Marine Materials Science and Engineering, Shanghai Maritime University, Shanghai 201306, P. R. China
2School of Materials Science and Engineering, Shanghai Jiao Tong University, Shanghai 200240, P. R. China
3Shanghai Key Laboratory of Materials Laser Processing and Modification, Shanghai Jiao Tong University, Shanghai 200240, P. R. China

The effects of the second pulse in resistance spot welding (RSW) on the mechanical properties of dual phase (DP600) steel were investigated. The experimental results show that the addition of the second pulse can give rise to the formation of acicular ferrite, retained austenite and chromium carbides, which can markedly improve the cross-tensile strength (CTS) comparing to that treated by the single pulse technology. Meanwhile, the formation mechanism of acicular ferrite, bearing the cubic to cubic orientation relationship with respect to the surrounding martensite lath, was revealed by the analysis of the thermal history of DP600 steel under different RSW processes. The effect of microstructure on the fracture behavior was also discussed.


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1. Introduction

In recent years, there is a tendency to increase the application of advanced high strength steels (AHSS) to reduce the automotive weight, improve fuel efficiency, and increase the crashworthiness of vehicles in automobile industry. Dual phase (DP) steel, as a member of the AHSS family, is very important for the automotive applications due to a lot of the unique mechanical properties, such as continuous yielding behavior, a lower ratio of yield/tensile strength, superior combination of strength and ductility, good formability, and a higher initial work-hardening rate compared to the other high-strength low-alloy (HSLA) steels with similar chemical composition.1–3)

As automotive material, DP steel must be weldable, coatable and repairable, which means that the welds fabricated in these automotive materials must fulfill the combination of high strength and good toughness. Generally, the change in microstructures during welding process dramatically affects the steels’ mechanical properties. Thus, improving welding performance is essential for the applications of the DP steel in automobile industry. Resistance spot welding (RSW) is a predominant welding technique used for joining steels in automotive applications due to its low cost and high robustness. Typically, a car body contains thousands of spot welds, whose mechanical strength and performance significantly affect the safety and reliability of the automobile structure.4) Therefore, the investigation on how to improve the mechanical properties of the spot weld is necessary for the safe integration into the automotive architecture. Recent publications reported that the relationship among the welding process parameters, microstructures and welding properties of the spot welded DP steels.5–7) For example, Marya et al.8) studied the effects of RSW process parameters (including weld current, weld time and weld force) on the mechanical properties, defects and microstructures in a DP steel.

Similarly, Ma et al.9) investigated the influence of both the microstructure and weld parameters on the interfacial fracture of spot welded DP600 steel. However, the above studies about the microstructures and mechanical properties of spot welded DP steels focused exclusively on the field of single-pulse RSW process technology. Moreover, the characterization of microstructures by transmission electron microscope (TEM), has not been sufficiently preformed, especially in the double-pulse RSW processes.

The present work aims to improve the mechanical properties of resistance spot welded DP600 steel by adding the second pulse based on a single-pulse technology. The microstructure of the spot welds is investigated by TEM in detailed, and the relationship between the microstructure and mechanical properties is analyzed.

2. Experimental Procedure

2.1 Materials

The material used in the present work is a cold-rolled DP600 steel sheet with a thickness of 1.4 mm. The chemical composition of the DP steel (in mass%) is 0.12C–1.8Mn–0.44Si–0.26Cr–0.016Ni–0.021P–0.006S. The yield strength, ultimate tensile strength and total elongation were determined as 432 MPa, 665 MPa and 29%, respectively.

2.2 Welding equipment and parameters

The spot welds were fabricated using a TDN-63 spot weld machine. Electrode with 6.0 mm face diameter flat tip was used and the cooling water flow rate was 200 L/min.

In our previous study,10) a series of single-pulse RSW specimens was produced according to the American Welding Society Standards. According to the welding standards, the welding parameters were optimized for improving the ratio of cross-tensile strength to tensile-shear strength (CTS/TSS). The current, time and weld electrode force for the single-
pulse RSW process were determined as 8.5 kA, 20 cycles (One welding cycle is equivalent to 0.02 s) and 3.92 kN, respectively.

In order to improve the mechanical properties of spot welds, three kinds of different double-pulse RSW processes were designed based on the single-pulse technology. The welding parameters for the second pulse were optimized by using the SORPAS software. The SORPAS is dedicated professional software for simulation and optimization of resistance welding processes.\textsuperscript{10) For the double-pulse RSW process, the welding parameters (the current, time and electrode force) of the first pulse were the same as those used in the single-pulse process. During the double-pulse RSW processes, after suspending as much time as $t_3$, the second pulse was carried out with the welding time of $t_4$ and the current of $I_2$. The sketch of double-pulse RSW process is shown in Fig. 1 and the corresponding welding parameters are listed in Table 1. The specimens treated by four different RSW processes were denoted as A, B, C and D, respectively.

### Table 1 Resistance spot welding parameters.

<table>
<thead>
<tr>
<th>Welding Process</th>
<th>$I_1$ (kA)</th>
<th>$t_0$ (cyc)</th>
<th>$t_1$ (cyc)</th>
<th>$t_2$ (cyc)</th>
<th>$I_2$ (kA)</th>
<th>$t_4$ (cyc)</th>
<th>$t_5$ (cyc)</th>
<th>$t_6$ (cyc)</th>
</tr>
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<tr>
<td>A</td>
<td>8.5</td>
<td>20</td>
<td>10</td>
<td>20</td>
<td>0</td>
<td>0</td>
<td>70</td>
<td>20</td>
</tr>
<tr>
<td>B</td>
<td>8.5</td>
<td>20</td>
<td>10</td>
<td>20</td>
<td>13</td>
<td>4.8</td>
<td>15</td>
<td>70</td>
</tr>
<tr>
<td>C</td>
<td>8.5</td>
<td>20</td>
<td>10</td>
<td>20</td>
<td>23</td>
<td>5.1</td>
<td>65</td>
<td>70</td>
</tr>
<tr>
<td>D</td>
<td>8.5</td>
<td>20</td>
<td>10</td>
<td>20</td>
<td>46</td>
<td>3.85</td>
<td>65</td>
<td>70</td>
</tr>
</tbody>
</table>

2.3 Mechanical and microhardness test

Both cross-tensile test and tensile-shear test were conducted on the Zwick T1FR020TNA50 electronic universal material testing machine. The specimens were prepared according to the Standard of GB/T 2651-2008.\textsuperscript{11) The geometrical shape and dimensions are shown in Fig. 2, and the spot welds were fabricated on the center of two identical overlapping DP steel plates.

Vickers hardness for the polished cross-section was tested using a HX-1000 Microhardness Tester (Shanghai, China) with a normal load of 300 g and a dwell time of 40 s.

2.4 Optical and electron microscopy investigation

The cross-section metallography of the welding area was prepared by using the standard metallographic technique, and then was observed by both a Nikon EPIPHOT-300 optical microscope and a FEI SIRION-200 scanning electron microscopy (SEM) operated at a 5 kV acceleration voltage with a field emission gun.

The TEM specimens were mechanically polished to a thickness of 50 \( \mu \)m and electro-polished in a twin-jet polisher using 5% perchloric acid solution at 20°C, and then finally thinned by using a Gatan ion-beam milling machine so as to achieve large thin areas for the observation. TEM investigation was carried out using a JEM-2010 microscope operated at 200 kV.

3. Results

3.1 Tensile-shear strength and cross-tensile strength

Both TSS and CTS were obtained to evaluate the mechanical properties of spot welds under different RSW processes, as shown in Fig. 3. From specimen A to specimen D, all the TSS curves exhibit a slight change around 21 kN, whereas the CTS curves show remarkable variation. The CTS of specimen C reaches the maximum value of 11.94 kN while the value of CTS of specimen A is 8.13 kN. As a result, the ratio of CTS/TSS of specimen A treated by single-pulse RSW is the lowest (0.39), while specimen C demonstrates the highest value of the ratio of CTS/TSS (0.59), indicating that the addition of the second pulse can improve the mechanical properties of spot welds.

3.2 Characteristics of fracture surface morphologies

Figure 4 and Figure 5 show typical characteristics on the fracture surface of the specimens treated by single-pulse and double-pulse RSW process, respectively. Full button pullout
fracture can be observed in both Fig. 4(a) and Fig. 5(a). The crack initiates and propagates only in the HAZ close to the weld nugget. However, as higher magnification of Fig. 4(a) and Fig. 5(a), Fig. 4(b) and Fig. 5(b) show obvious different characteristics, respectively. The most significant difference is that some second microcracks (as indicated by the arrow in Fig. 4(b)) exist in specimen A, while much fewer microcracks exist in specimen C. Another important difference is that the specimen A appears cleavage brittle fracture feature with smooth fracture surface, while the specimen C shows ductile fracture feature with the dimples and tearing ridge.

3.3 Distribution of the microhardness

The location of microhardness indentations is shown in Fig. 6(a) and the corresponding microhardness distribution is shown in Fig. 6(b). Obviously, the values of microhardness
dramatically increase from the base metal (BM) to the HAZ, and then slightly decrease in the fusion zone (FZ). The microhardness in HAZ of specimen C is 350 HV, which is lower than that of specimen A (400 HV). It is worthy to point out that all the specimens treated by double-pulse technology exhibit lower microhardness comparing to that of specimen A treated by single-pulse RSW.

3.4 Characterization of the microstructures

Figure 7 is the optical micrograph of specimen C, and it shows that the typical microstructure of spot weld including crystallized morphology in FZ. The dot-line 1 stands for the boundary between the BM and HAZ, and dot-line 2 stands for the boundary between the HAZ and FZ. In the HAZ near BM there is a fine-grained region, while in the HAZ near the fusion boundary there is a coarse-grained region.

The microstructure in HAZ near the fusion boundary was investigated by scanning electron microscope to analyze the effects of the second pulse on the mechanical properties of spot welds. Similar to specimen A (Fig. 8(a)), specimen C consists of fine and long martensite laths, as shown in Fig. 8(b). However, it is worthy to point out that there are needle-like and leaf-like morphologies within the martensite laths in specimen C, as indicated by the arrows in Fig. 8(b). TEM observation will be further carried out to characterize and analyze the microstructure.

A typical TEM result shows that there is only one [001] zone axis of bcc structure existing in the selected area electron diffraction pattern (SAED), rather than two zone axis of diffraction patterns produced by the needle-like microstructure and the surrounding martensite lath (Fig. 9). This suggests that the diffraction spots from the needle-like microstructure and the surrounding martensite lath overlap with each other. When the operating $g = \{110\}$ is selected, the needle-like microstructure shows bright contrast in dark field image (Fig. 9(b)). What is more, when $g = 110$ is selected, both the needle-like microstructure and the surrounding martensite lath show almost the same contrast, as indicated in Fig. 9(c). This suggests that the \{110\} planes of needle-like microstructure and the surrounding martensite lath are almost parallel to each other. Therefore, the experimental result indicates that the bcc needle-like microstructure and the surrounding martensite lath bear the cubic to cubic orientation relationship. Since the needle-like microstructure is bcc structure, it can be determined as ferrite. What is more, the needle-like ferrite appearing within martensite lath has similar morphology as that of the ordinary acicular ferrite. However, the size of the needle-like ferrite is much smaller than that of ordinary acicular ferrite. TEM observation shows that the acicular ferrite does not exist in specimen A or specimen B, but only exists in specimen C and specimen D. In general, the nucleation of the acicular ferrite needs the assistance of inclusions.\(^{12-14}\) However, no evidence of inclusion was found at the vicinity of acicular ferrite in our experiments, which may be attributed to the very rapid cooling rate in RSW.

The leaf-like microstructure in the martensite matrix is identified as primitive orthorhombic Cr$_3$C$_2$ chromium carbides based on our previous work,\(^{15}\) and there are three variants of Cr$_3$C$_2$ carbides marked by arrows in Fig. 10(a). In addition to the Cr$_3$C$_2$ carbides, another face-centered cubic CrC carbides with two variants are found (marked by arrows in Fig. 10(b)). The orientation relationship between the CrC carbide and the martensite matrix is determined by SEAD as \((111)_{\text{CrC}} \parallel (110)_{\text{M}}, [01\bar{1}]_{\text{CrC}} \parallel [00\bar{1}]_{\text{M}}\).

Besides the acicular ferrite and chromium carbides, film-like retained austenite between the martensite laths is an important characteristic in the microstructure, as shown in Fig. 11. The orientation relationship between the martensite and retained austenite is identified as the well-known

![Fig. 7](image-url)  
Fig. 7 Crystallized morphology and microstructure of spot weld of specimen C.

![Fig. 8](image-url)  
Fig. 8 Typical microstructure in HAZ near fusion boundary for specimen A (a) and specimen C (b).
Kurdjumov-Sachs (K-S) relationship: \[ \frac{1}{2} / \{111\} // [101], \]
\[ \{110\} // \{111\} \] and Nishiyama-Wasserman (N-W) relationship:
\[ \frac{1}{2} / \{111\} // [101], \]
\[ \{110\} // \{111\} \] by SAED (as inserted in Fig. 11(b)). The retained austenite was also observed by TEM in the specimen A treated by the single-pulse technology, but the amount of retained austenite of specimen A is far less than those in the specimens treated by double-pulse technology.

Fig. 9 TEM images of needle-like microstructure. (a) bright field image, (b) dark field image, \( g = \{110\} \) (SAED pattern inserted), (c) dark field image, \( g = 110 \).

Fig. 10 TEM bright-field images of leaf-like microstructure: (a) Cr\(_2\)C\(_2\) and (b) CrC.
4. Discussion

4.1 Effect of the second pulse on the amount of retained austenite

In order to analyze the effect of RSW processes on the microstructure of the spot welds, the welding metallurgy of the DP 600 steel must be investigated.

Figure 12 shows the thermal cycle curves in nugget center calculated by SORPAS software according to the welding parameters listed in Table 1. During the single-pulse RSW process, the DP steel was heated to the peak temperature of 2138°C at an average rate of about 5000°C/s. This temperature is much higher than the liquidus temperature (calculated as 1515°C by J-MatPro software), which will result in the rapid growth of columnar grain. After reaching the peak temperature, the steel was quenched to about 200°C during which the martensitic transformation can occur (the starting temperature of martensitic transformation calculated by J-MatPro software: \( M_s \approx 400°C \)).

After heating by the first pulse, specimen B was cooled to 740°C and then was reheated to a higher temperature of 860°C. During subsequent cooling to room temperature, most of the austenite can transform to martensite and only a little retained austenite may reserve. That’s why the microstructure of specimen B is almost the same as that of specimen A.

Specimen C was quenched to about 380°C after the heating of the first pulse, where partial austenite may transform into martensite, and then it was reheated to 670°C by adding the second pulse. During the heating of the second pulse, the “partitioning process” may happen and the retained austenite is enriched with carbon (the carbon comes from the martensite phase). As a result, certain amount retained austenite with sufficient carbon can be stable during the cooling process to room temperature, which results in higher amount of retained austenite in specimen C than that in specimen A. The thermal history of specimen D is similar to that of specimen C, and thus its microstructure is similar to that of specimen C except for the amount of acicular ferrite and retained austenite are less than those in specimen C. This is because that both the quenching temperature (220°C) and tempering (partitioning) temperature (620°C) of specimen D are much lower than those of specimen C.

4.2 Mechanism of the formation of acicular ferrite

Considering the thermal history of the double-pulse RSW process, there are two potential approaches for the formation of acicular ferrite in the FZ. The first one is that the acicular ferrite can nucleate and grow from austenite when the FZ is cooled from 2138°C to 200°C during the first pulse. However, no acicular ferrite was found in specimen A treated by the single-pulse technology. Therefore, we can assume that this thermal history is not favorable for the formation of the acicular ferrite since the very rapid cooling rate of 3000°C/s can make the temperature-time cooling curve escape from the nose of the time-temperature-transformation diagram of the DP steel. The second approach for the acicular
ferrite formation can be attributed to the addition of the second pulse. Taking specimen C for example, after the heating of the first pulse, the specimen C was quenched to about 380°C where partial austenite can transform into martensite. After then, the second pulse can reheat the fusion zone to the temperature of 670°C. Because there is still plenty of retained austenite in the specimen C, the acicular ferrite can directly nucleate and grow within the retained austenite. During the sequent cooling process from the tempering temperature of 670°C, the austenite surrounding the acicular ferrite may transform into martensite. As a result, the acicular ferrite can be observed to form within the martensite lath and the cubic to cubic orientation relationship between acicular ferrite and surrounding martensite can also be maintained, as shown in Fig. 9. Due to lower quenching temperature (220°C), the specimen D contains less amount of the retained austenite comparing to that of specimen C, so it is reasonably to believe that the amount of acicular ferrite in specimen D is also less than that of specimen C.

The origin of the orientation relationship between the acicular ferrite and the martensite can be understood as follows. The martensite laths show nearly K-S or N-W orientation relationship with respect to the surrounding austenite, while the acicular ferrite keeps cubic to cubic orientation relationship with respect to the surrounding martensite lath. This suggests that both the acicular ferrite and the surrounding martensite lath are transformed from the same austenite grain during the cooling process.

4.3 Effects of the second pulse on the mechanical properties of spot weld

The mechanical properties of the spot weld depend on both the geometry and the microstructure of the weld spot. In this paper, the geometry can be considered as constant which only depend on the welding parameters. Because the welding parameters of the single-pulse technology are similar to that of the first-pulse in double pulse technology, and what is more, the current of the second-pulse is too low to remelt the steel. Therefore, the mechanical properties of different specimens are mainly dependent on the microstructure of spot weld.

Among all the specimens, specimen C shows the best weldability. The CTS of specimen C is 11.94 kN, which is much higher than 8.13 kN of specimen A treated by the single-pulse technology. Meanwhile, the CTS of specimen B is 9.11 kN, increasing 12% compared with specimen A. The reason is that the specimen B exhibits almost the same microstructure as that of specimen A except for containing much less microcracks. It is worthy to point out that the CTS of specimen C is 47% higher than that of specimen A, and it can be roughly concluded that the increase in CTS for specimen C is mainly attributed to its microstructure which consists of the acicular ferrite, retained austenite and chromium carbides. In the field of welding, the acicular ferrite has been known to be the most desirable welding microstructure with good combination of high strength and high toughness because of its small grain size as well as having high density of dislocations.18–21) In the present work, the average length of the acicular ferrite is only 200 nm, such a fine acicular ferrite can result in better combination of high strength and toughness than those reported in the literatures.13,14,22) Morris et al.23–25) experimentally and theoretically verified that the steel is cleaved along {100} planes and the crack can pass across the laths with the same habit plane within a packet. A crack will branch at high angle packet boundaries where the orientation of the {100} planes changes. Since the lath boundary is low-angle boundary which does not impose significant crystallographic discontinuities. Meanwhile, the lath boundary tends to lay parallel to {110} planes, and the laths share common {100} planes that cut through the boundaries, as shown in Fig. 13. Therefore, the retained austenite between dislocated martensite laths can refine the effective grain size of martensite for transgranular cleavage (Fig. 13). Based on the above analysis, both acicular ferrite and CrC carbide with similar orientation relationship should be favorable for the improvement of the spot weld’s ductility. Because CrC carbide precipitates have the coherent interface with {110} planes of martensite, they have the same effect for transgranular cleavage as the interlath retained austenite. The acicular ferrite has cubic to cubic orientation relationship with the martensite lath and its long axis is along the (110) direction (Fig. 9). As a result, when the crack propagates along one of the {100} cleavage plane, it will be deflected and branched by some of acicular ferrite laths accompanying the formation of high density of tearing ridges, as shown in Fig. 5(b). Obviously, the retained austenite and CrC also have the same effect for the formation of tearing ridges. The main reasons for the good ductility of specimen C can be concluded as follows. (1) the amount of the microcracks in specimen C is much less than those in specimen A due to the decrease of internal stress by the addition of the second pulse; (2) the retained austenite and CrC can result in the canyon-like morphology of fracture surface in specimen C, while there are obviously flat cleavage planes in the prior austenite gain in specimen A; (3) there are more dimples in specimen C than that in specimen A. The formation of dimples is probably attributed to more Cr7C3 carbides precipitating dispersively in the martensite matrix in specimen C.

The experimental results indicate that the addition of second pulse can markedly improve the CTS of DP steel, but the TSS is almost the same as that treated by single-pulse technology. This can be attributed to the difference of stress states (normal stress for the second-pulse technology and shear stress for the single-pulse technology) in the above two tensile tests. The study on the effect of stress state on the mechanical properties of spot weld is underway.

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**Fig. 13** Schematic illustration of the effects of microstructures on the fracture behavior of spot weld.
5. Conclusions

The effects of the second pulse in RSW process on the mechanical properties of spot weld DP steel were investigated by characterization of microstructure, and the effects of microstructure on the fracture behaviors were analyzed. The main conclusions are summarized as follows.

(1) Mechanical properties of the spot welds can be improved by adding the second pulse based on the single-pulse RSW process. The CTS of specimen C treated by double pulse technology is 11.94 kN, which is 47% higher than that of specimen A treated by the single pulse technology.

(2) The needle-like microstructure distributed in the martensite lath was identified by TEM as acicular ferrite. By optimizing the welding parameters of the second pulse, more acicular ferrite and better mechanical properties can be obtained.

(3) The excellent ductility of specimen C was attributed to the microstructure which consists of lath martensite, acicular ferrite and retained austenite.

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