Microstructural analysis and the influence of shot peening on stress corrosion cracking resistance of duplex stainless steel welded joints

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Received 11 July 2012; accepted 19 November 2013

This paper aims to study the weldment of duplex stainless steel (DSS) AISI 2205. Tungsten inert gas (TIG) welding is performed with controlled welding parameters. The effect of cooling rate on microstructural changes is analyzed by varying the heat input during welding. The results show that high heat input and excessive nickel content in the filler wire, leads to excessive formation of austenite phases in the fusion zone. Lower heat input nucleates lesser amount of austenite phases and also very low heat input causes lack of fusion in the weld joints. Microstructural variations due to welding are assessed by conducting the micro hardness test, impact toughness test and tensile test. To study the effect of shot peening, the samples are prepared from the DSS welded plate. The chloride induced stress corrosion cracking (SCC) test is conducted and the result shows a noticeable improvement in the corrosion resistance of the weld zone due to shot peening. The findings of this study can be used in marine applications.

Keywords: Duplex stainless steel, Microstructure, Austenite, Ferrite, Secondary austenite, Shot peening

Duplex stainless steel (DSS) is a well-known material for its excellent strength and corrosion resistance nature. However, joining DSS plates by the fusion welding causes significant reduction in the mechanical properties, because of microstructural changes during weld solidification. It is very essential to maintain the characteristics of the weld zone to use DSS in servicing highly critical environments, such as ocean mining machinery, oil and gas pipe lines, desalination plants and chemical tankers of ships etc. DSS has ferrite (α) and austenite (γ) in an approximately equal proportion, which possess body centered cubic (BCC) and face centered cubic structure (FCC) respectively. During the controlled alloying process of the DSS, under equilibrium conditions, ferrite promoting elements (Cr, Mo, Mn, W, Nb, Si, Ti and V) will concentrate by diffusion into the ferrite. At the same time, austenite promoting elements (Ni, C, N, Co and Cu) will concentrate by diffusion into austenite phases. This gives the even formation of dual phase microstructure. But the welding of DSS forces the microstructure to remain in an excessive ferritic nature, because of the higher amounts of ferrite promoting elements in its chemical composition, and also due to faster cooling rate. Austenite usually nucleates in the temperature range 1200-900°C. During cooling, the weld zone remains in this range of temperature for a very short period of time, i.e., from 4 s to 15 s. Thus, the arc energy and filler metal composition play a major role in microstructural stability after welding. Low arc energy beam (higher intensity beam) welding processes, like laser beam welding (LBW), and electron beam welding (EBW) cause inadequate formation of austenite phases due to a faster cooling rate. Thus, higher arc energy (lower intensity) welding processes like gas tungsten arc welding (GTAW), gas metal arc welding (GMAW), and the shielded metal arc welding (SMAW) are preferred, to weld the DSS. Shielding gases such as argon and helium play a major role in the welding process to control the microstructure. Argon provides a large amount of ferrite phases in the DSS weld metal and also smoother arc during welding. Helium provides good penetration in faint surfaces due to higher arc energy. Various research works have been carried out in welding DSS, to control the microstructure. In general, to promote the nucleation of austenite phases in the weld zone, nickel enriched filler metal (ER 2209) is used in welding. It was reported, that there was a formation of coarser ferrite grains near the fusion line, thereby causing a reduction in the low

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temperature toughness. Thus, microstructural changes in the weldment, uneven segregation of alloying elements, coarser ferrite grains near the fusion line may not fully provide the efficient mechanical and metallurgical properties in the weld zone, when compared with base metal\(^{10-14}\). These contradictory observations make it essential to study the weld zone of the DSS by comparing it with the parent metal. In general, under compression there is no failure in the metal surface. Shot peening is a cold working process, which is generally used to reduce failures like stress corrosion cracking (SCC) in austenitic and ferritic stainless steel grades\(^{15-17}\). It hardens the surface, refines the surface grains and introduces high magnitude surface compressive stresses in the weld zone, by projecting high velocity cast steel shots. The surface hardening and the grain refinement on the weld metal surface induced by shot peening gives significant enhancement in the SCC resistance of DSS weldment.

In this work, an attempt has been made to study the mechanical and corrosion properties of DSS weld zone, by assessing the microstructure, hardness, impact toughness, and tensile behaviour and SCC behavior of the weld joint. Also, the effect of shot peening on the SCC resistance of the weldment have been discussed.

**Experimental Procedure**

**Material composition and welding process parameters**

The chemical composition of DSS 2205 and its filler metal ER 2209, used in this experiment is shown in Table 1. The percentage of alloying elements was found using the optical emission spectroscopy test. Tungsten inert gas (TIG) welding was carried out on four pairs of DSS plates of dimensions 150×140×8 mm, with a bevel angle of 60°. The recommended arc energy for welding DSS is 0.5 to 2.5 kJ/mm. During welding arc energies were controlled in between this range by varying the current, voltage and welding speed. The calculated arc energy values are given in Table 2. The time taken for each weld pass was calculated to evaluate the welding speed. The polarity used in the welding was direct current electrode negative (DCEN). ER 2209 filler wire with a diameter of 2.5 mm was used in welding. During welding trimix shielding gas (60% Ar + 38% He + 2% N\(_2\)) was used to shield the weldment and also to achieve the benefits of argon, helium and nitrogen gases. Nitrogen is added with the shielding gas to compensate the nitrogen loss during welding. Lower arc energy of 0.53 kJ/mm leads to insufficient side wall fusion during welding and also requires excessive reinforcement of filler metal deposition. The microstructure of the lower arc energy, i.e., 0.53 kJ/mm weld sample has not been presented here due to its insufficient side wall fusion.

**Metallography and mechanical testing**

After welding the ferrite/austenite ratio in DSS weld was measured using the magnetic method and point count method. In the magnetic method, Fisher Ferrite scope MP30E-5 was used to measure the ferrite content in six locations in each of the welded

<table>
<thead>
<tr>
<th>Welding parameters</th>
<th>Sample 1</th>
<th>Sample II</th>
<th>Sample III</th>
<th>Sample IV</th>
</tr>
</thead>
<tbody>
<tr>
<td>Current (I) (amps)</td>
<td>95</td>
<td>105</td>
<td>125</td>
<td>65</td>
</tr>
<tr>
<td>Voltage (V) (volts)</td>
<td>10</td>
<td>10.5</td>
<td>11.6</td>
<td>9.8</td>
</tr>
<tr>
<td>No of passes</td>
<td>2</td>
<td>3</td>
<td>3</td>
<td>3</td>
</tr>
<tr>
<td>Average welding speed (U) (mm/s)</td>
<td>0.504</td>
<td>0.587</td>
<td>0.523</td>
<td>1.19</td>
</tr>
<tr>
<td>Arc energy (Q) (kJ/mm)</td>
<td>1.88</td>
<td>1.88</td>
<td>2.77</td>
<td>0.53</td>
</tr>
<tr>
<td>Heat input (kJ/mm) (60% of arc energy)</td>
<td>1.13</td>
<td>1.13</td>
<td>1.66</td>
<td>0.32</td>
</tr>
</tbody>
</table>

**Interpass temperature**: 150°C to 200°C
samples. In the point count method, samples were etched by 10% NaOH. Five fields were examined in each zone with 100 test grid points and with magnification of 400×. The micro Vickers hardness test was performed in the transverse section of the weld plates for all the three samples, with a load of 20 g. The Charpy impact toughness test was carried out at room temperature and -40°C on the samples prepared from the welded plate. V-notches for the impact specimens were made in three zones, namely the base metal, the weld zone and the HAZ. The samples were brought to -40°C by using dry ice. Three samples were tested for each condition and the average of the three values was taken. Tensile behavior of the DSS weldment was analyzed by conducting tensile test on the samples prepared from the DSS weld plates as per ASTM A370. Two samples were prepared from each welded plate.

Shot peening and SCC test
To study the behavior of the weldment against stress corrosion cracking, a part of the weldment has been cut from the sample III and shot peening was performed. Grade 2, Class A Almen strips were prepared with standard dimensions as per J443 procedures, for using the standard shot peening test strip. The peening media used in this experiment were cast steel shots of different sizes. The hardness of the media used was 42 HRC and the diameter of the shot was 1 mm. The distance from the entry of the shots to the specimen was 750 mm. To study the surface morphology, the surface roughness of the peened sample was measured using contact and non-contact type surface roughness tester. The chloride induced SCC test was performed in the peened and unpeened U – bent samples of DSS weld joints as per the code ASTM G36 – 94. The samples were prestressed by bending to an angle of 180° using three point bending method. The test was carried out totally for 108 h in an Erlenmeyer flask, containing high purity magnesium chloride with a boiling temperature of 155°C ± 1.0°C, which was maintained by a suitable condenser.

Results and Discussion
Microstructural analysis
The microstructure of the base metal and welded samples captured using optical microscopy are shown in Fig. 1. The base metal of DSS shown in Fig. 1(a) reveals that the austenite (γ) phases are embedded in the ferrite (α) matrix. The austenite ferrite ratio measured in DSS base metal using point count method shows approximately 50:50. The weld zone microstructure of all the three samples shows that the increase in arc energy from low to high range promotes gradual increase in the amount of precipitation of austenite phases in the weldment. The average percentages of the ferrite – austenite ratio measured in the welded samples are shown in Figs 2 (a) and (b). The sample 3 which has been joined by using higher arc energy of 2.77 kJ/mm shows 77.8% of austenite phases. Nearly the same value of arc energy, i.e., 1.88 kJ/mm was used in the welding of samples 1 and 2, but with different number of weld passes. Sample 2 (3 pass welding) shows 68.7% of austenite phases and sample 1 (2 pass welding) shows 58.1% of austenite phases. This proves that the 3 pass welding nucleates higher amount of austenite phases, which is 10% more than that of 2 pass welding due to reheating of the deposited weldment. It has been observed macroscopically that the increasing arc energy leads to a wide HAZ in the welded samples.

The SEM-EDS measurements were taken of the austenite and ferrite phases of DSS base metal and weldment. It was clearly observed that in the base metal, ferrite promoting elements such as chromium, molybdenum and manganese were mainly segregated in the ferrite phase and similarly austenite promoting element nickel is mainly segregated in the austenite phases as shown in Figs 3 (a) and (b). But, in Figs 3 (c) and (d) the weldment shows that the segregation of alloying elements was not properly taking place due to insufficient time for diffusion in all the three samples. This is also one of the reasons for which the properties of the DSS weldment get sacrificed when compared with its base metal. The major alloying elements present in the ferrite and austenite phases of the DSS base metal and weld zones are shown in Table 3.

Microstructural evolution of DSS weld
The microstructure of the DSS weld zone usually evolved in three stages after welding. First the microstructure nucleates as allotriomorphs at the ferrite grain boundaries. Due to multipass welding, the weldment subjected to reheating, can result in widmanstätten side plates (needle like structured grains) that grow into the ferrite grains from the grain boundary allotriomorphs, and also as intragranular precipitates inside the ferrite grains.
Fig. 1 – Microstructure of base metal and fusion zones

(a) Base metal microstructure

(b) Fusion zone locations (Arc Energy = 1.89 kJ/mm)

(c) Fusion zone locations (Arc Energy = 1.87 kJ/mm)

(d) Fusion zone locations (Arc Energy = 2.77 kJ/mm)
The weld microstructures in Figs 1 (b), (c) and (d) show an elongated needle structure and also the intragranular austenite particles which are known as secondary austenite phases ($\gamma_2$). The transformation kinetics of secondary austenite is as follows:

Ferrite $[\alpha] +$ Austenite $[\gamma] + \text{Cr}_2\text{N} \rightarrow \text{Ferrite} [\alpha] + $ Austenite $[\gamma] + $ Secondary Austenite $[\gamma_2]

According to the kinetics of microstructure during welding, $\text{Cr}_2\text{N}$ nucleates at the boundaries of the ferrite austenite interface, which depletes the ferrite promoting elements Cr and Mo in the surrounding regions. Due to this depletion, the nucleation and growth of secondary austenite happens in the weldment by dissolving $\text{Cr}_2\text{N}$. The secondary austenite may reduce the pitting corrosion resistance, since pit nucleation seems to prefer the secondary austenite ferrite interface, because of the depletion of Cr and Mo. On the other hand, an excessive amount of austenite precipitation may leads to stress corrosion cracking. In addition, sometimes partial dissolving of $\text{Cr}_2\text{N}$ leads to the precipitation of metastable ferrite austenite boundaries.
Low temperature heat affected zone (LTHAZ)

After welding, the austenite phase percentage in the LTHAZ was found to increase by nearly 10%, when compared with the base metal. During welding, the temperature of this zone reaches nearly the range of 1000 and 1100°C, due to which the percentage of austenite phases increased. There was no observation of intermetallic sigma (σ) phases in LTHAZ. The microstructures of the LTHAZ in the welded samples are shown in Fig. 4. The measured ferrite-austenite ratio in this zone is shown in Fig. 5. There is no significant variation in the grain structure of LTHAZ due to varying arc energies between the three samples. The measured values of the austenite ferrite ratio in all the three samples are approximately around 60:40. But, in an extremely slow cooling rate, sigma (σ) can be precipitated in LTHAZ during the temperature range of 600-800°C. Therefore, the welding parameters should be carefully controlled to ensure, that the overall cooling conditions are fast enough to avoid deleterious precipitations in the LTHAZ. Because even very less amount of sigma precipitation may leads to detrimental effect in the mechanical and corrosion properties of DSS.

High temperature heat affected zone (HTHAZ)

During welding the zone nearer to the fusion line approaches the melting point and becomes almost

Table 3 – Segregation of alloying elements

<table>
<thead>
<tr>
<th>Phases</th>
<th>Fe</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>Mn</th>
</tr>
</thead>
<tbody>
<tr>
<td>Base metal ferrite</td>
<td>63.89</td>
<td>23.18</td>
<td>5.17</td>
<td>2.96</td>
<td>1.42</td>
</tr>
<tr>
<td>Base metal austenite</td>
<td>69.14</td>
<td>21.24</td>
<td>6.79</td>
<td>1.72</td>
<td>1.10</td>
</tr>
<tr>
<td>Weld sample – I austenite location</td>
<td>64.31</td>
<td>23.82</td>
<td>8.02</td>
<td>1.27</td>
<td>1.46</td>
</tr>
<tr>
<td>Weld sample – II austenite location</td>
<td>60.78</td>
<td>23.33</td>
<td>8.40</td>
<td>1.23</td>
<td>1.69</td>
</tr>
<tr>
<td>Weld sample – III austenite location</td>
<td>50.60</td>
<td>18.56</td>
<td>6.22</td>
<td>0.67</td>
<td>1.45</td>
</tr>
<tr>
<td>Weld sample – I ferrite location</td>
<td>56.38</td>
<td>22.97</td>
<td>6.71</td>
<td>1.49</td>
<td>1.99</td>
</tr>
<tr>
<td>Weld sample – II ferrite location</td>
<td>55.79</td>
<td>22.66</td>
<td>6.61</td>
<td>1.38</td>
<td>1.61</td>
</tr>
<tr>
<td>Weld sample – III ferrite location</td>
<td>51.24</td>
<td>19.40</td>
<td>5.91</td>
<td>0.84</td>
<td>1.03</td>
</tr>
</tbody>
</table>

Fig. 4 – Microstructure of the LTHAZ 10 mm from fusion zone

Fig. 5 – Ferrite-austenite percentage in LTHAZ
fully ferritic structure. There is an insufficient reformation of austenite phases in this zone due to the rapid cooling achieved in all the three welded plates. These regions are known as HTHAZ or overheating zone which are shown in Figs 6 (a), (b) and (c). Ferrite phase percentage in the HTHAZ reaches up to a range of 75-80% in all the three samples which are shown in Fig. 7. Higher amount of ferrite precipitation leads to a brittle nature at low temperature. Also this coarser ferrite grains increases the hardness and reduces the impact toughness particularly at the low temperature. This is due to the fact that the ferrite structure usually has less ductility and formability. In addition, this excessive ferrite zone will lead to the reduction in the corrosion resistance. Ferrite count method using ferrite scope was not used for measurement in HTHAZ due to the inability of locating the zone.

Microhardness test
The measured values of the hardness in the duplex weldment, HAZ and base metal are shown in Figs 8 (a), (b) and (c). It was found that the weldment of DSS has higher hardness than that of the base
material due to strain induced heating and cooling cycle and also due to the changes in microstructure. This strain induced hardening is caused by the compression of the weld region during solidification. It was observed that the hardness of the HTHAZ is slightly higher when compared with the base metal due to the formation of coarser ferrite grains near the fusion line. There was no variation in the hardness between the three samples even though there was a variation in the ferrite austenite ratio in the weldment. Muthupandi et al. reported that the ferrite austenite variation in the weld zone and HAZ does not have much effect on the hardness. This is because, during the solidification after welding, the austenite promoting elements and ferrite promoting elements do not have sufficient time to diffuse in FCC and BCC structures, respectively. However, there is a slight increase in the value of hardness mainly due to the strain induced hardening, weld induced residual stresses and secondary austenite precipitates in the weldment. Nowacki and Łukojć reported that the secondary austenite phases exhibited higher hardness compared to ferrite and primary austenite phases in the weldment. In the present study, the measured locations of secondary austenite phases in the weldment exhibit more hardness than the ferrite and primary austenite phases.

Charpy impact toughness
The impact test result shows that the base metal gives excellent impact toughness by absorbing nearly an average of 297 Joules at a room temperature of 24°C. It was also found that there was no significant reduction in the toughness of the base metal even at –40°C as shown in Fig. 9. The tested samples show that there was a metal flow in the base metal due to its ductile nature during the toughness test. This kind of metal flow behavior was not observed in the weld and HAZ samples, which reveals almost a brittle cleavage fracture. Significant reduction in the toughness was observed in the weld zone and HAZ when compared with the base metal. Nearly 40% of the impact energy gets reduced in the weld zone and HAZ. Uneven segregation of alloying elements, formation of the constitutional elements like harder secondary austenite phases and ferrite phases leads to ductile brittle transition at low temperature. Coarser ferrite grains near the fusion boundaries also a reason for this low impact energy.
Table 4 – Tensile properties of DSS weld

<table>
<thead>
<tr>
<th>Sample No.</th>
<th>Yield strength (0.2% proof stress) (MPa)</th>
<th>Ultimate tensile strength (MPa)</th>
<th>% of elongation</th>
<th>Max. load (kN)</th>
<th>Cross section area of the sample measured before test (mm²)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>530.50</td>
<td>661.81</td>
<td>12.00</td>
<td>56.31</td>
<td>85.09</td>
</tr>
<tr>
<td>2</td>
<td>544.29</td>
<td>729.54</td>
<td>27.40</td>
<td>60.22</td>
<td>82.55</td>
</tr>
<tr>
<td>3</td>
<td>545.16</td>
<td>683.82</td>
<td>15.20</td>
<td>59.05</td>
<td>86.36</td>
</tr>
<tr>
<td>4</td>
<td>564.84</td>
<td>715.33</td>
<td>23.60</td>
<td>62.26</td>
<td>87.04</td>
</tr>
<tr>
<td>5</td>
<td>547.71</td>
<td>657.48</td>
<td>11.40</td>
<td>57.61</td>
<td>87.63</td>
</tr>
<tr>
<td>6</td>
<td>572.34</td>
<td>725.85</td>
<td>25.80</td>
<td>64.11</td>
<td>88.32</td>
</tr>
</tbody>
</table>

Fig. 10 – Tensile test specimens after fracture

**Tensile behavior of DSS weld**

The yield strength and ultimate tensile strength (UTS) of base metal are 450 MPa and 655 MPa, respectively. DSS weldment exhibits higher yield strength and UTS in all the six tested samples when compared to the base metal. Also the test reveals that in most of the samples, fracture occurred in the region of base metal as shown in Fig. 10. Nickel enriched filler metal (ER 2209) and the trimix shielding gas is probably promoting the strength in the weldment of DSS. In addition, none of the tested samples got fractured in the HTHAZ region even though that region contains coarser ferrite grains. The higher hardness induced due to welding also be the reason for experiencing higher strength in the DSS weldment. During the tensile test, two samples 5 and 6 fractured in the weld region. Tensile properties arrived in the tested samples are shown in Table 4.

**Shot peening vs. surface roughness**

Shot peening was carried out in the welded sample III to study the influence of shot peening on the enhancement of corrosion resistance. The medium used for shot peening were cast steel shots of sizes S390 which is shown in Fig. 11. During peening, Almen strip curvature (arc height) was measured for every 2 min. The saturation curve obtained in the shot peening process is shown in Fig. 12. The Almen strip used in the peening process and the saturated almen
strip after 26 min peening are shown in Figs 13 (a) and (b). 100% coverage was achieved in the peening process after 15 min of peening. To achieve the beneficial effect of peening, fully the weld samples were peened during the exposure time of 1 h and 2 h even though the arc height of the almen strip was saturated after 26 min. After peening the roughness of the peened surfaces was measured. The measured values of the roughness parameter ($R_a$) using noncontact type roughness tester before and after peening in the weld zones are shown in Figs 14-16. It shows that there is no variation in the roughness of the peened surfaces even if the peening time was varied. The roughness values measured using probe type contact roughness tester is shown in Table 5. It was observed that the roughness of the shot peened surfaces is higher than that of the unpeened samples. Sanjurjo et al.\cite{19} reported that the enhancement in the surface characteristics of DSS is possible by shot peening with a minimum amount of roughness induced by peening process.

**Chloride induced SCC test**

The U-bent specimens used for SCC test are shown in Fig. 17. The test was carried out in the Erlenmeyer flask continuously for a period of 108 h at a temperature of 150°C as shown in Fig. 18. Frequent observations were made for every 36 h. During the CISCC test, after 36 h the grains were attacked by the corrosion media in the unpeened sample, but no significant attack was observed in the peened sample. After 76 h unpeened samples were attacked with very large pits with fissures were formed and in the peened samples weak attack was observed. After 108 h, in the peened samples the grains were slightly attacked, but in the unpeened samples the weld zone
Fig. 15 – Peened surface (2 h)

(a) Macro image (50x)       (b) Micro image (300×300 sq. microns)

Fig. 16 – Peened surface (1 h)

(a) Macro image (50x)       (b) Micro image (300×300 sq. microns)

Fig. 17 – SCC Test Specimens

Fig. 18 – Erlenmeyer flask
Table 5 – Roughness measurement after and before peening

<table>
<thead>
<tr>
<th>As received surface (Before peening)</th>
<th>After peening (1 h)</th>
<th>After peening (2 h)</th>
</tr>
</thead>
<tbody>
<tr>
<td>3.34</td>
<td>4.52</td>
<td>4.56</td>
</tr>
<tr>
<td>3.46</td>
<td>5.36</td>
<td>4.80</td>
</tr>
<tr>
<td>2.94</td>
<td>5.0</td>
<td>3.06</td>
</tr>
<tr>
<td>3.12</td>
<td>4.7</td>
<td>5.18</td>
</tr>
<tr>
<td>3.22</td>
<td>3.8</td>
<td>4.34</td>
</tr>
<tr>
<td>Average 3.21</td>
<td>Average 4.68</td>
<td>Average 4.38</td>
</tr>
</tbody>
</table>

Arc energy of less than 0.5 kJ/mm causes insufficient side wall fusion. In addition to arc energy, filler metal ER 2209 with enriched nickel content plays a major role in the stabilization of the austenite phases in the weldment.

Formation of coarser ferrite grains near the fusion line (HTHAZ) in all the welded samples usually gives ferrite austenite ratio as 80:20. There is no significant variation in the percentage of the austenite phases and there was no precipitation of sigma phases in the LTHAZ by increasing arc energy during welding.

The change in the ferrite austenite ratio does not contribute to the hardness and shows nearly the same hardness in all three samples of weldment. It shows that the partition of alloying elements do not have sufficient time for diffusion into the ferrite and austenite phases, respectively. But significant variations were observed in the hardness between the base metal, the weld and HAZ on the Vickers scale. High hardness value was experienced in the weldment. Also HTHAZ leads to higher hardness due to the formation of coarser ferrite grains.

Charpy impact test result shows that there was no reduction in the toughness of the base metal from room temperature to - 40°C. However, the weld zone and HAZ exhibit lower values of toughness at - 40°C. The formation of coarser ferrite grains near the fusion line may be the key reason for reduction in low temperature toughness.

Shot peening introduces a high strength plastic skin on the DSS welded plates. The surface roughness ($R_a$) increases after peening. The roughness values are however not very sensitive to peening time and remains almost constant. The chloride induced SCC test shows that the shot peening was effective in reducing the attack on the grains attack in a chloride environment. Unpeened specimen was corroded in chloride environment whereas peened specimen was not attacked by corrosion during the time period of testing.

Acknowledgements

The authors acknowledge the support given by Diamond Heat Treaters Pvt Ltd, Ambattur, Chennai, India by providing shot peening machine for experiments.

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