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Hardness and microstructural gradients in the heat affected zone of welded low-carbon quenched and tempered steels

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Keywords
affected, heat, steels, gradients, tempered, microstructural, hardness, quenched, carbon, low, welded, zone

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Hardness and microstructural gradients in the heat affected zone of welded low-carbon quenched and tempered steels

W. Pang 1, N. Ahmed 2 and D. Dunne 3

This paper concentrates on the form of the hardness gradient in the heat affected zones (HAZs) produced by submerged arc welding of two low-carbon Q & T steels. The results show unequivocally that the gradient differs from that found in steels of lower carbon equivalent in that the peak HAZ hardness is displaced from the grain coarsened heat affected zone (GCHAZ) into the grain refined heat affected zone (GRHAZ). Weld thermal cycle simulation has been used to confirm the results obtained from actual welds and to clarify the cause of this unexpected phenomenon.

Keywords
Submerged arc welding (SAW), multiple-wire SAW, low-carbon quenched and tempered steels, heat affected zone (HAZ), HAZ hardness gradients.

Introduction
Low-carbon, weldable, quenched and tempered (Q&T) steels are used as high strength constructional steels in applications such as pressure vessels, building columns, mining plant, road tankers and submarine hulls. These steels are effectively refinements of the USS “T1” type A, low-carbon martensitic steel 1 that are based on alloy-lean compositions, particularly reduced carbon content, to improve weldability and decrease the susceptibility to hydrogen embrittlement. Boron additions are also commonly used to maintain hardenability.

As supplied Q&T steel plate typically has a structure of well tempered martensite/bainite with a minimum yield strength of 690 MPa and a hardness higher than about 220 HV. Welding causes local hardening in the heat affected zone (HAZ) of the parent plate due to formation of martensite and/or bainite that, in the absence of post-weld heat treatment (PWHT), is effectively untempered.

During the weld thermal cycle, regions of the HAZ that experience temperatures above AC3 are fully re-austenitised and can subsequently transform to martensite/bainite on cooling. An austenite grain size gradient is also established with the coarsest grains forming adjacent to the fusion boundary (the grain coarsened heat affected zone, GCHAZ). For temperatures between the solidus temperature (about 1500°C) and AC3 (about 875°C) the austenite grain size decreases as the peak temperature falls, with an accompanying decline in hardenability. Thus the hardness gradient across the parent metal HAZ typically shows a maximum hardness immediately adjacent to the fusion boundary with a progressive decrease across the GCHAZ, the GRHAZ (about 1100°C to AC3), and the intercritical heat affected zone (ICHAZ), which is bounded by AC3 and AC1 and is only partially re-austenitised by the weld thermal cycle. The actual form of the hardness gradient across the weld metal, the HAZ and the parent metal depends on the welding conditions and, in particular, the selected weld metal. Schematic hardness gradients are shown in Figure 1 for the cases where the weld metal is, relative to the steel base, (a) over-strength; (b) strength-matching; and (c) under-strength. Figure 1(d) also depicts strength-matching weld metal, but shows a decrease in hardness below that of the base plate in the ICHAZ (case (i)), in the adjacent parent steel (case (ii)) and a continuous decrease in hardness across the HAZ (case (iii)). Case (i) has been reported for low-carbon steels that depend on precipitation hardening to achieve the specified base plate strength, e.g. ASTM A710 type steels that are precipitation hardened by Cu 2. In case (ii) softening occurs in a region bounded approximately by AC1 and about 600°C due to re-tempering of a tempered martensitic/bainitic base plate structure; or by restoration of a cold worked steel base. Figures 2(a) and 2(b) respectively show actual hardness gradients that are similar to those of case (iii) in Figure 1(d) and to Figure 1(a) for the welding conditions and steel compositions specified in the caption.

Figure 1. Schematic hardness profiles across the HAZ for weld metal that is (a) over-strength; (b) strength-matching; (c) under-strength; and (d) strength-matching with softening in the parent metal (case (i)), in the ICHAZ (case (ii)) and over the whole HAZ (case (iii)).
Table 1. Compositions (wt%) of 20 mm and 50 mm BIS80 and 12 mm EM812 plate steels.

<table>
<thead>
<tr>
<th>Steel</th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>Si</th>
<th>S</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>Cu</th>
<th>Al</th>
<th>Ti</th>
<th>B</th>
<th>V</th>
<th>CE\textsubscript{HW}</th>
<th>P\textsubscript{cm}</th>
</tr>
</thead>
<tbody>
<tr>
<td>BIS80 20 mm</td>
<td>.17</td>
<td>1.21</td>
<td>.018</td>
<td>.47</td>
<td>.003</td>
<td>.025</td>
<td>.85</td>
<td>.20</td>
<td>.032</td>
<td>.031</td>
<td>.023</td>
<td>.0018</td>
<td>–</td>
<td>.585</td>
<td>.313</td>
</tr>
<tr>
<td>BIS80 50 mm</td>
<td>.17</td>
<td>1.18</td>
<td>.016</td>
<td>.41</td>
<td>.002</td>
<td>.025</td>
<td>.86</td>
<td>.20</td>
<td>.015</td>
<td>.028</td>
<td>.020</td>
<td>.0018</td>
<td>–</td>
<td>.580</td>
<td>.309</td>
</tr>
<tr>
<td>EM812 12 mm</td>
<td>.12</td>
<td>.90</td>
<td>.015</td>
<td>.25</td>
<td>.002</td>
<td>1.25</td>
<td>.51</td>
<td>.36</td>
<td>.19</td>
<td>.06</td>
<td>.005</td>
<td>.0015</td>
<td>.017</td>
<td>.532</td>
<td>.238</td>
</tr>
</tbody>
</table>

Table 2. Steel Q&T treatments, basic mechanical properties of the 20 mm and 50 mm BIS80 and 12 mm EM812 plates and the welding conditions employed.

<table>
<thead>
<tr>
<th>Steel</th>
<th>Heat Treatment</th>
<th>YS MPa</th>
<th>HV</th>
<th>Welding Process</th>
<th>Heat input kJ/mm</th>
<th>Passes</th>
<th>Prep.</th>
<th>Filler Wire</th>
<th>Flux</th>
</tr>
</thead>
<tbody>
<tr>
<td>BIS80 20 mm</td>
<td>900°C(WQ);640°C(T)</td>
<td>725</td>
<td>270</td>
<td>BOP SAW 4 wire SAW</td>
<td>2.4, 7 2.4</td>
<td>1* 2-4**</td>
<td>None</td>
<td>LAC M2 2.4 mm</td>
<td>880 M</td>
</tr>
<tr>
<td>BIS80 50 mm</td>
<td>900°C(WQ);600°C(T)</td>
<td>730</td>
<td>298</td>
<td>BOP SAW 4 wire SAW</td>
<td>2.4, 7 2.4</td>
<td>1* 5-10^</td>
<td>None</td>
<td>LAC M2 2.4 mm</td>
<td>880 M</td>
</tr>
<tr>
<td>EM812 12 mm</td>
<td>925°C(WQ);690°C(T)</td>
<td>690</td>
<td>250</td>
<td>BOP SAW 2.58</td>
<td>1^^</td>
<td>None</td>
<td>LA 100 2.4 mm</td>
<td>880 M</td>
<td></td>
</tr>
</tbody>
</table>

* No pre-heat; ** No pre-heat, 150°C interpass; ^ 100-150°C pre-heat, 200°C interpass; ^^ 20 and 200°C pre-heats.

Figure 2. Measured hardness gradients for 4-wire SAW welds of (a) 20 mm thick BIS60 – 550 MPa (0.09C, 1.3Mn, 0.6Cr), 5 kJ/mm, 2000 mm/min; and (b) 50 mm thick AS1204 – 250 MPa (low carbon structural steel), 2.5 kJ/mm, 1500 mm/min. The average base plate hardness remote from the weld was 222 HV5 for the 20 mm BIS60 and 135 HV5 for the 50 mm AS1204.
A weldment introduces a structural heterogeneity with a corresponding change in hardness and other mechanical properties relative to the base steel. In effect, a structural notch is introduced and other notches can be present due to weld defects such as sharp profile changes, undercut, weld metal porosity and lack of fusion. These notches concentrate stress, which is inevitably present due to the weld configuration, and thermal and transformation stresses. Therefore, the fatigue and creep resistance of a weld fabricated component can be compromised by the weldment. A more serious issue for carbon steels is the potential hardening of the HAZ by formation of martensitic/bainitic structures. The presence of even minute amounts of hydrogen, an unwanted impurity of most arc welding processes, can result in hydrogen assisted cold cracking (HACC) at points of stress concentration. Consequently, a significant effort has been mounted to define welding procedures that result in sufficiently slow cooling rates to ensure the absence of hard martensitic and/or bainitic structures, which are prone to HACC. The weld heat input, the plate thickness, the pre-heat and the interpass temperature for multi-pass welding are the key factors that control the weld thermal cycle, while the steel composition and the austenite grain size determine its transformation characteristics and propensity to form martensite on cooling.

The schematic hardness gradients in Figure 1 are not exhaustive and other variants are possible depending on the strength levels of the weld and the parent metals and their relative hardenabilities for the weld thermal cycle experienced. However, for the cases shown in Figures 1 and 2, and for most steel weldments, the maximum HAZ hardness occurs in the GCHAZ. This trend applies particularly to welded low carbon equivalent (CE), which is effectively a welding-hardenability factor. In this case, CEHW is below about 0.40, ensuring that pre-heat is unnecessary provided the cooling rate is sufficiently slow. The weld heat input and the weld configuration are appropriate to prevent rapid cooling.

The higher hardenability typical of Q&T constructional steels (CEHW > 0.5) usually requires pre-heat for normal arc welding conditions, in order to reduce the cooling rate and thereby decrease HAZ hardening and susceptibility to hydrogen assisted cold cracking.

This paper concentrates on the form of the HAZ hardness gradient produced by submerged arc welding of two Australian-manufactured Q & T steels.

Experimental methods

The compositions of the 690 MPa Q&T steels investigated are shown in Table 1 and the heat treatments applied to these steels are given in Table 2.

As part of an extensive project on high productivity welding of Australian plate steels 4, 20 mm and 50 mm Q&T BIS80 steel was subjected to bead-on-plate (BOP) and 4-wire submerged arc butt welding, using the heat inputs (HI) indicated in Table 2 and a welding speed of 1000 mm/min. The BIS80 plate was subjected to 4-wire BOP welding and to 4-wire butt welding using the prepared joints indicated in Table 2. The heat inputs for both the BOP and butt welding of BIS80 represent the nominal total heat input per pass. A neutral flux, Lincolnweld 880 M, was used in the SAW process and welding was performed along the rolling direction of the steel plate. Details of the alignment, separation and polarity of the electrodes are documented elsewhere 3, 4.

The multi-pass 4-wire SAW of 20 mm plate involved a single vee preparation and consisted of 6 passes for 2 kJ/mm heat input and 4 passes for 4 kJ/mm heat input. No pre-heat was employed but the interpass temperature was maintained at 150°C.

For welding of 50 mm plate a double vee preparation was used with 10 passes/side for 2kJ/mm and 5 passes/side for 4 kJ/mm. The pre-heat was 100-150°C and the interpass temperature was maintained at 200°C.

Four-wire SAW BOP welds were produced for the 20 and 50 mm plates at 2, 4 and 7 kJ/mm heat inputs, with no pre-heat.

Samples of EM812 plate, 300 mm long x 200 mm wide x 12 mm thick, were subjected to single-wire BOP SAW. The plate was welded without pre-heat (20°C) and after pre-heating to 80, 120 and 200°C. Only the 20°C and 200°C pre-treatments are reported in this paper. The equilibrated oxy-acetylene method was used to set the pre-heating temperature. Welding was conducted using 2.4 mm diameter LA 100 filler wire with neutral 880 M flux. The arc voltage was 38 V, the current was 1135 A and the welding speed was 1000 mm/min, giving a nominal heat input of 2.58 kJ/mm.

Weld samples were sectioned for macro- and micro-structural examination and hardness testing. Hardness and structural gradients were recorded. Prior austenite grain sizes in the HAZ were measured after etching in a picric acid solution.

A thermal simulator was used for EM812 samples to simulate the actual BOP weld cooling cycle in the centre of the test bar, 110 mm x 11 mm x 11 mm, which was supported horizontally.
between water cooled cast iron grips. The test bars were resistance heated at a rate of 250°C/s, consistent with the actual weld thermal cycle, to the peak temperature and then cooled under controlled conditions. The high heating rate made it difficult to control the peak temperature, which varied from sample to sample, for example, over a range of about 50°C, from 1346-1399°C, for an aim peak temperature of 1400°C. The simulator consisted of a computer control system, temperature measuring and data logging equipment and a transducer to measure sample dilatation. A Pt/Pt-13% Rh thermocouple was spot-welded to the surface of the bar at mid-length, with readings for monitoring and control of the thermal profile being taken every 0.01 s. The heating rate, peak temperature and the cooling rate were computer controlled by controlling the power cycle. The temperature gradient along the bar was estimated using Widgery’s equation, which is based on an inverted parabolic relationship with the vertex being at the bar centre where the thermocouple is located. The details of the equipment are described elsewhere.

Peak temperatures were selected that corresponded to the GCHAZ (1400°C) and the upper GRHAZ (1100°C) and the cooling rate was set on the basis of data obtained from embedded thermocouples for single wire BOP SAW. \( \Delta t_{800-500} \) values were measured for specific welds and used to calculate cooling curves for other plate thicknesses/peak temperatures and heat inputs by means of the Rosenthal equations as modified by Easterling. For this analytical method, the 20 mm BIS80 and 12 mm EM812 plates qualified as “thin plates” which are characterised by direct proportionality between \( \Delta t_{800-500} \) and \( (HI)^2 \), whereas the 50 mm BIS80 plate was considered to be a “thick plate” for which \( \Delta t_{800-500} \) is directly proportional to \( HI \).

A combination of a transducer-based dilatometric method during simulation and thermo-mechanical analysis (Mettler TMA 40) was employed to determine a partial continuous cooling transformation diagram for BIS80, and dilatometry was subsequently used to determine the CCT diagram for an alloy variant still classified as BIS80. A reported CCT diagram for a steel of closely similar composition was applied in the case of EM812 to rationalize structural evolution during the weld thermal cycle.

**Results and discussion**

**BIS80: BOP welds**

Macrographs of the 2, 4 and 7 kJ/mm BOP welds on 20 and 50 mm thick plate are shown in Figures 3 and 4. The weld bead profile shows an “inverted hat” shape for the 2 kJ/mm weld on 20 mm plate and a “lens-shape” for the 50 mm plate.

For the higher heat inputs of 4 and 7 kJ/mm, the volume of the weld bead increased, as expected, and the weld bead shape was inverted hat for both the 20 mm and 50 mm plates. It is inferred that for either high heat input or limited thickness, heat extraction perpendicular to the plate surface (in the root region) is restricted, resulting locally in greater penetration of the molten zone into the base plate. The profile of the welds for the 50 mm thick plate is quantified in Figure 5 which plots the HAZ width perpendicular to the fusion line with distance around the fusion line for BOP welds on 50 mm plate for the indicated heat inputs. (0 represents the weld root.)

**Figure 5. HAZ width perpendicular to the fusion line with distance around the fusion line for BOP welds on 50 mm plate for the indicated heat inputs. (0 represents the weld root).**

**Figure 6. Hardness (HV10) as a function of distance from the weld fusion line (FL). The 80 of the identifying code represents BIS80, the next number the heat input (2, 4 or 7) and the last two, the plate thickness (20 or 50 mm).**

The steepest gradient occurs at the root and the shallowest gradient at the shoulder, where dissipation of the thermal load into the base plate is most restricted. An important consequence of this observation, which is not explored in detail in this paper, is that \( \Delta t_{800-500} \) can be different in different regions of the same weld bead.
Hardness traverses across the weld metal and HAZ at the weld root are shown in Figure 6 for the BOP welds. The hardness profiles show skewing of the HAZ peak hardnesses away from the GCHAZ adjacent to the fusion boundary, despite the fact that the prior austenite grain size, and therefore the hardenability, continuously decreases with distance from the fusion boundary. An example of the grain size gradient is given in Figure 7 for case of the 4 kJ/mm BOP weld on 50 mm plate. The HAZ peak hardness occurred about 1.5 mm from the fusion boundary, corresponding to a prior austenite grain size of about 10 μm, compared with 55 μm in the GCHAZ adjacent to the fusion boundary and an average of 19 μm for the both the 20 and 50 mm base plates.

Figure 8 shows micrographs of samples etched to reveal the prior austenite grains in the HAZ of 50 mm plate welded at 2 kJ/mm. This weld-shape was anomalous...
Figure 10. Hardness (HV10) in the GCHAZ at the root and shoulder of the 20 mm and 50 mm BIS80 plates after BOP welding at the indicated heat inputs. The average hardness of the base plate was 270 HV10 for 20 mm plate and 298 HV10 for the 50 mm plate (Table 2).

Figure 11. Microstructure of the GCHAZ of BOP welded, 50 mm BIS80 for a heat input of 4 kJ/mm ($\Delta t_{800-500} = 17$ s (~18°C/s)).

Figure 12. Continuous cooling transformation (CCT) diagram for a low CE BIS80 steel determined by dilatometry (full lines) and for the high CE BIS80 alloy used in the current investigation (dashed lines), which were determined during weld thermal cycle simulation and by TMA3.
The corresponding average hardneses in the GCHAZ region at the root and shoulder are shown in Figure 10. In general, the hardness was higher for 50 mm plate and higher in the root than in the shoulder. The first trend is consistent with the expected difference in cooling rate with plate thickness and the second, with the coarser prior austenite grain size, the higher HAZ width and, by implication, the slower cooling rate in the shoulder region. Slower cooling promotes austenite transformation over a higher temperature range with a higher degree of auto-tempering during cooling. As a consequence, the hardness decreases.

The microstructures of the GCHAZ of both the 20 mm and 50 mm plates, in both the root and shoulder positions, did not show that martensite had been produced under the weld cooling conditions. The GCHAZ microstructure for the 4 kJ/mm BOP weld on 50 mm plate is shown in Figure 11. The cooling times, \( \Delta t_{800-500} \), were established for the BOP welds using embedded thermocouples\(^3,4\). The values 8, 17 and 34 s were determined for 2, 4 and 7 kJ/mm welding of 50 mm plate and, due to the inferior heat sink, the values for 20 mm plate were higher: 11, 41 and 108 s.

Figure 12 is a CCT diagram for a low CE variant of BIS80 (CE IIW = 0.37), which was determined by dilatometry (full lines), using linear cooling rates from 900°C of 10-150°C/s. Also included are data obtained during simulation and by TMA for the actual BIS80 variant investigated in the current work (dashed lines). The three lowest dilatometer cooling rates correspond to \( \Delta t_{800-500} \) values of 30 s (10°C/s), 15 s (20°C/s) and 10 s (30°C/s), which overlap the actual \( \Delta t_{800-500} \) values for all three of the 50 mm welds and the 2 and 4 kJ/mm welds for 20 mm plate. Bearing in mind that the dilatometer result is for a leaner alloy chemistry and a relatively fine (900°C) austenite grain size, the actual weld CCT diagram applicable to the GCHAZ might be expected to be shifted to the right, increasing the likelihood of formation of martensite. However, the partial CCT data for the actual alloy under simulated thermal cycles do not show any significant shift in that direction. Moreover, none of weld HAZs exhibited a martensitic structure, in general agreement with the structural evolution predicted by the CCT diagram. Therefore, formation of a substantially martensitic structure requires cooling times less than the minimum of 8 s measured for the actual welds. It is inferred that cooling rates higher than about 50°C/s are required to form martensite in BIS80 and although these rates are achievable by roller water quenching in industrial processing, they are unlikely to apply to weld cooling conditions. Figure 11 shows the GCHAZ microstructure of the 50 mm plate, BOP welded at 4 kJ/mm. Bainitic ferrite, \( \alpha''_B \), is clearly present, consistent with the structure predicted by the CCT diagram for the corresponding cooling rate of about 18°C/s. Even for the fastest cooling rate (2 kJ/mm on 50 mm plate, \( \Delta t_{800-500}=8 \) s, \(~40°C/s\)) at least some bainitic ferrite was present at the expense of martensite. The formation of martensite is even less likely for the slower cooling rates corresponding to the higher heat input welds, the 20 mm plate and the 4-wire butt welds performed with multi-pass weld runs and elevated interpass temperatures.

Figure 13. Macrographs of the 2 kJ/mm (left) and 4 kJ/mm 4-wire SAW welds on 50 mm BIS80 plate.

Figure 14. Photomicrographs of the GCHAZ of 2kJ/mm 4-wire SAW butt welds on (a) 20 mm plate and (b) 50 mm plate.
BIS80: Four-wire SAW butt welds

Macrographs of the 2 and 4 kJ/mm 4-wire SAW butt welds are shown in Figure 13 for the 50 mm plates. The microstructures of the GCHAZ for the 2 kJ/mm 4-wire SAW welds on the 20 and 50 mm plates are shown in Figure 14 and indicate that bainitic ferrite is the dominant constituent.

Hardness traverses 2 mm below the plate surface were conducted in each case, Figure 15. The traverses included weld metal, the HAZ and the base plate, which had a mean hardness of 270 HV10 for 20 mm plate and 298 HV10 for 50 mm both plate. The hardness gradients again clearly show that the peak HAZ hardness does not occur in the CGHAZ.

There is evidence in Figure 15 of slight softening near the termination of the HAZ for the multi-pass 4 wire SAW welds on 20 and 50 mm plates for both 2 and 4 kJ/mm. This softening is likely to be due to precipitate coarsening in the parent metal and/or in the intercritical region subjected to peak temperatures just above \( A_{C1} \).

![Hardness profiles across 4-wire SAW butt welds, measured 2mm below plate surface, for heat inputs of 2kJ/mm (a) and 4 kJ/mm (c) on 20 mm plate; and 2 kJ/mm (b) and 4 kJ/mm (d) on 50 mm plate.](image-url)
The effect of simulated post-weld heat treatment (PWHT) on the hardness gradient was examined and Figure 16 is a typical result: major softening of the HAZ and minor softening in the weld metal and base plate. The overall effect is significant attenuation of the hardness variation across the weldment.

It was decided to conduct BOP welding trials on another 690 MPa Q&T steel to confirm that the “displaced hardness peak effect” was not a unique feature of the composition and structure of BIS80. EM812 was selected for this purpose and BOP SAW was carried out for 12 mm thick samples of this steel (Table 2). Further, the weld thermal cycle corresponding to the BOP welds was simulated for EM812 samples in an attempt to more closely identify the HAZ region associated with peak hardness.

**EM812: BOP welds**

Figure 17 shows hardness traverses across the roots of BOP SAW welds for pre-heats of 20°C and 200°C. The peak hardness is clearly displaced from the fusion boundary in both cases and is about 60 HV points lower for the 200°C pre-heat (320 HV). This observation is consistent with a ∆_{500-500} value of about 80 s for the higher pre-heat compared with 20 s for no pre-heat (20°C).
The approximate cooling curves for the two pre-heats have been superimposed on the CCT diagram for a steel of similar composition and CE to EM812, Figure 18. Slower cooling allows transformation to start at a higher temperature and proceed over a longer time period with more significant auto-tempering and restoration of the defect structure induced by transformation. It should be noted that the CCT diagram predicts that an entirely bainitic structure forms for 20°C pre-heat, with a mixed bainite/martensite structure for 200°C. However, the CCT diagram was produced for austenitising at 930°C, whereas the weld thermal cycle results in a wide range of austenitising temperatures, and the coarser austenite grain size for peak temperatures > 1350°C would be expected to shift the CCT curves to the right, effectively increasing the steel hardenability. Nevertheless, although this type of steel can be rapidly quenched to produce martensite during manufacture (e.g. by roller quenching), the cooling rates typical of arc welding are likely to generate at least partially bainitic structure.

The optical micrographs of the GCHAZ structures for the two pre-heats, Figure 19, show bainitic ferrite – ferrite plates with elongated interlath islands of MA (martensite-austenite) constituent. The prior austenite grain size is clearly larger for the higher pre-heat temperature. The grain size gradients are shown in Figure 20 for the 2.58 kJ/mm welds produced at the indicated pre-heats. The maximum HAZ grain size increased with pre-heat, as expected on the basis of an increased dwell time at high temperature. The austenite grain size was difficult to measure in the ICHAZ and the graph beyond about 2 mm from the fusion boundary reflects structural refinement rather than prior austenite grain size. The average grain size of the base plate was 11 μm and therefore welding resulted in a smaller austenite grain size in the GRHAZ. The hardness peaks in Figure 17 occurred at about 1.5 mm from the fusion boundary, where the austenite grain size was about 5 μm, Figure 20. This result is similar to that obtained for the BOP welds on BIS80.

**Simulated weld thermal cycles for EM812**

Weld thermal cycle simulation was based on thermal profiles calculated for the actual welds based on cooling data obtained from embedded thermocouples. Data for no pre-heat (20°C) and 200°C pre-heat are reported here for actual BOP welds and simulations based on peak temperatures of about 1360°C (GCHAZ) and about 1100°C (upper GRHAZ).

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*Figure 18. CCT diagram for a steel of similar composition and CE to EM812, with superimposed cooling curves associated with BOP SAW with pre-heats of 20°C and 200°C.*

*Figure 19. GCHAZ microstructures for BOP SAW welds with 20°C (left) and 200°C pre-heats. The bar represents 15 μm. Etched in 2% Nital.*
Hardness as a function of distance on either side of the sample centerline are reported in Figures 21 and 22 for simulated thermal cycles corresponding to 2.58 kJ/mm heat input, peak temperatures of about 1100°C and 1360°C, and pre-heats of 20°C and 200°C. The hardness profile shows symmetry about the centerline, with the hardness peaks displaced by several millimetres. Lack of complete symmetry in some cases probably reflects different cooling efficiencies by conduction through the water-cooled grips. For peak temperatures of both 1100°C and 1360°C, the hardness peaks were sharper for 200°C pre-heat, but peak position was displaced a greater distance from the centerline for a peak temperature of 1360°C (5-6 mm for 1360°C compared with 4-5 mm for 1100°C).

The grain size gradients shown in Figure 23 clearly illustrate that the peak hardness is associated with a small prior austenite grain size. For example, the peak hardness in Figure 21(a)
for the 1108°C peak corresponds to a distance of about 4.25 mm from the sample centre and the peak hardness shown in Figure 22(a) for a 1358°C peak temperature is 5.5 mm from the centre. The corresponding prior austenite grain sizes indicated by Figure 23 are less than 10 μm.

Figure 24 shows hardness as a function of temperature along the bar, as predicted by Widgery’s equation using a peak temperatures of 1108°C and 1124°C, and pre-heats of 20°C and 200°C. The hardness measurements were higher for no pre-heat, but the peak hardness occurred at about 900-950°C for both pre-heats. The corresponding austenite grain sizes are < 10 μm (Figure 23). The microstructures of the simulated samples at the centre of the bar are shown in Figure 25. The dominant constituent is bainitic ferrite, in agreement with the GCHAZ of the actual welds (Figure 19) and the predicted structure based on Figure 18.

The results unambiguously show that the peak HAZ hardness for welded 690 MPa steels occurs in a fine grained region significantly displaced from the fusion boundary. Possible explanations for this phenomenon are considered below.

Mechanism of displaced hardness peak effect

The strength of mixed martensite/bainite structures has been shown to peak at about 20% bainite, despite the fact that bainite generally has a lower strength than martensite. Young and Bhadeshia ascribed this effect in part to carbon partitioning into the remnant austenite from which the bainite forms, but mainly to a constraint effect exerted by the harder martensite (the brazing joint effect). Although this mixed structure strengthening/hardening effect might appear to provide a plausible explanation for the higher hardness of the fine-grained region of the HAZ, the displaced hardness peak effect is exhibited for structures in which little or no martensite is present.

Although the types of steels investigated are sufficiently hardenable to produce a minimum yield strength of 690 MPa, they are hardenable with respect to formation of both martensite and bainitic ferrite. Both structures consist of fine ferritic laths with high dislocation densities. The strengthening obtained is a sensitive function of temperature range of formation and the cooling rate through that range. Rapid cooling (quenching) lowers the temperature of formation to the $M_S-M_F$ range,
produces a more carbon saturated ferrite and limits dislocation annihilation and auto-tempering. Bainitic ferrite formation in relatively low carbon steels, such as those investigated, occurs over a slightly higher temperature range at slower cooling rates, with carbon partitioning to remnant austenite which becomes isolated in small islands between laths. This austenite can partially transform to a higher carbon martensite at lower temperatures, producing martensite-austenite (MA) constituent. The bainitic hardenability of the steels investigated is relatively high, with bainite start temperatures ($B_S$) of about 500°C. Further, $B_S$ is intercepted for cooling rates typical of arc welding, without austenite transformation to softer/lower strength ferritic products such as polygonal ferrite $\alpha_P$ and quasi-polygonal ferrite $\alpha_Q$\textsuperscript{13,14}. Therefore, despite a lower austenite grain size in the GRHAZ, bainitic ferrite still forms and is characterized by a higher hardness than the GCHAZ (which also consists of bainitic ferrite) because of a Hall-Petch grain-size contribution. The refined prior austenite grain size remains a structural entity within the bainitic structure (i.e. austenite grain boundaries are not obliterated as they are when $\alpha_P$ and/or $\alpha_Q$ are formed). Therefore, it is proposed that a Hall-Petch effect occurs, as demonstrated by Grange\textsuperscript{15} for higher alloy martensitic steels. The increased surface area per unit volume of austenite grain boundary also promotes refinement of the bainitic lath size, adding a further increment of strengthening.

Honeycombe and Bhadeshia\textsuperscript{16} proposed that the strength of martensite and bainite can be estimated by an equation that sums the effects of the following barriers to dislocation motion: the intrinsic strength of iron ($\sigma_Fe$), the sum of substitutional alloy strengthening ($\Sigma x_i (\sigma_{ss})_i$), the inverse of the lath/plate width ($K_{L} L^{-1}$) and the dislocation density ($K_{D} \rho$. The subscripts L and D refer to constants associated with lath size and dislocation density and $x_i$ is the concentration of the $i$th element that contributes a solid solution strengthening increment, ($\sigma_{ss})_i$. However, the present results and those of Grange suggest that this equation is incomplete because the effect of austenite grain size ($d_A$) is significant and is not included as a Hall-Petch term, $K_A d_A^{-0.5}$, where $K_A$ is a constant.

**Conclusions**

The results show unequivocally that the HAZ hardness gradients produced in the investigated Q & T steels differ markedly from those found in steels of lower carbon equivalent. The peak HAZ hardness is displaced from the GCHAZ to the GRHAZ. Although the present work was confined to SAW and two different 690 MPa steels, it is concluded that this displaced hardness peak effect is general for steels of this kind and is not dependent on the type of arc welding process. The origin of this phenomenon is considered to be an austenite grain size hardening effect, in concert with a bainitic hardenability that is high enough to ensure the formation of bainitic ferrite over the range of cooling rates and austenite grain sizes associated with the arc welding process.

Another surprising outcome of this investigation is that although these steels are classified as martensitic, bainitic ferrite was found to be the dominant decomposition product of austenite in the HAZ for the welding conditions examined.

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