Corrosion Behavior of Seamless Pipeline C–Mn Steel by On-line Controlled Cooling Technique

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In order to obtain the low cost seamless pipe with the same mechanical properties of C–Mn–Mo steel, the application of controlled cooling technology was successfully applied to the C–Mn steel. Seamless pipeline C–Mn steel was prepared by on-line controlled cooling technique and corrosion behavior was studied experimentally by a salt spray wet/dry cyclic corrosion test. Seamless pipeline C–Mn–Mo steel by traditional quenching and tempering technique was used for comparison. Corrosion rate of C–Mn steel is higher than that of C–Mn–Mo steel in the whole test time. The rust layer of C–Mn steel contained more defects than that of C–Mn–Mo steel. Controlling cooling technique can significantly improve the strength of reduced-cost C–Mn steel, while the phenomenon can not reproduce for corrosion resistance.

KEY WORDS: seamless pipeline steel; controlled cooling technique; microstructure; corrosion.

1. Introduction

The manufacturing processes for steel tubes and pipes can be divided into welded (seamed) and seamless. Welded pipes are made by bending steel plates into the form of a tube by welding the seam, and these steel plates are usually produced by the controlled rolling and controlled cooling technique.1) However, controlled rolling and controlled cooling has not yet successfully applied in seamless pipe production at present, because seamless pipe production process and deformation process involved are too complex to control.2–4) Therefore, quenching and tempering treatment after hot-rolled are usually employed in producing high grade seamless steel pipes.

Compared with welded pipe, the demand for seamless pipe has been increased in many industries due to the outstanding homogeneity of mechanical property without welded joints.5,6) However, seamless pipe also faces many challenges in today’s marketplace. One of the major challenges is to reduce the cost with reasonable product quality. To resolve this problem, it is necessary to use advanced manufacturing and quality control techniques. Clearly, controlled cooling technology for seamless pipe may be the effective way to solve this problem. The controlled cooling process has several advantages over the traditional reheat quenching and tempering process. First, the microstructure can be diversified due to the varied cooling rate. Second, it reduces the manufacturing cost by eliminating the reheating and quenching steps. Third, it saves costly alloying elements by satisfying the mechanical properties through control of microstructure and alloy chemistry.7)

In recent years, controlled cooling technology for seamless pipe has been investigated in Baoshan Iron & steel Co. Ltd., which remarkably improves the strength and reduces the cost when residual deformation and residual heat had been used and costly alloying elements had been saved.

At present, high strength seamless pipe can be prepared by controlled cooling technique in Baoshan Iron & steel Co. Ltd. However, corrosion behavior of seamless pipe by controlled cooling technique has not been well investigated, and questions remain about its applicability. First, the costly alloying elements, such as Cr, Mo and so on, are added to promote corrosion resistance by their passivation protective effects. For seamless pipe produced by controlled cooling technique, the loss of costly alloying elements would result in loss of corrosion resistance. Second, the different pipe making techniques are corresponding to the different microstructures: controlled cooling for non-equilibrium microstructure and traditional heat-treatment for equilibrium microstructure. Third, most of the relevant work on seamless pipe has focused on its production processes and mechanical properties,5,6,8–10) little has been reported on the evolution of corrosion of seamless pipe by controlled cooling technique.

In this work, seamless pipe was produced by controlled cooling technique, and the corrosion behavior of seamless pipe was studied in a salt spray wet/dry cyclic corrosion test. The aim of this work is to evaluate the corrosion behavior and understand the corrosion mechanism by corrosion prop-
2. Experimental Procedure

2.1. Materials
All test steels were provided by Baoshan Iron & Steel Co. Ltd., with the composition as listed in Table 1. The manufacturing processes for steels and corresponding microstructures are shown in Table 2. The steels were from seamless steel pipes with sizes of 244.45 mm (outer diameter) × 11.05 mm (the thickness).

2.2. The Atmospheric Corrosion Tests
The atmospheric corrosion of iron and steel is the most common corrosion process, causing immensely financial losses. The corrosion tests were performed on the Salt Spray Tester. The corrosion testing procedure involved cycling between a dry environment and a mist composed of 3.5 wt.% NaCl and the specimens were suspended in a vertical position. Each wet/dry cycle (24 h) consisted of a wetting period (12 h at 35°C, 97±1% RH) plus a drying period (12 h at 35°C, 20% RH). The tests covered 30 cycles.

2.3. Electrochemical Measurements
The atmospheric corrosion resistance of steels depends on the properties of rust layer formed on their surfaces. Therefore, to further understand the effect of the existence of a rust layer on the subsequent corrosion process of steels, the tested steels were pre-rusted by periodic wet/dry cyclic corrosion tests to produce a rust layer on the working surface area of the mounted specimen.

The electrochemical cell consisted of three electrodes: a rusted working electrode, a Pt plate auxiliary electrode and a saturated calomel reference electrode (SCE). For potentiodynamic polarization experiments, the potential was scanned from -0.3 to +0.6 V vs. SCE at a scan rate of 2 mV s⁻¹. Corrosion potentials (Ecorr) and the corrosion current densities (icorr) were calculated by using instantaneous Tafel-type fit Gamry-DC105 corrosion analysis software.

2.4. Microstructural Characterization
Microstructural evaluations of tested samples were carried out using optical microscopy (OM), EVO MA25 scanning electron microscopy (SEM) and JEM 2100F transmission electron microscopy (TEM) equipped with energy-dispersive X-ray spectroscopy (EDX). The external appearances of corroded samples were photographed. Further morphology analysis and reaction product characterization were conducted using SEM and D8 DISCOVER X-ray diffraction (XRD) with Cu Ka radiation.

3. Results

3.1. Microstructure and Mechanical Properties
The cross-section microstructures of C–Mn–Mo steel by quenching/tempering (C–Mn–Mo–QT) and C–Mn steel by on-line controlled cooling technique (C–Mn–CC) are shown in Fig. 1. The microstructure of C–Mn–Mo–QT consisted of tempered martensite and particulate precipitates with the sizes ranging from 0.1 to 1.0 μm (see Figs. 1(a), 1(c), 1(e)). EDX microanalysis shows that the particles precipitated more or less uniformly dispersed were cementite in a dominate position and very fine spherical V-rich particles, corresponding to the red arrow in Fig. 1(e).

Table 1. Chemical composition of tested steels (wt.%).

<table>
<thead>
<tr>
<th>Seamless pipe</th>
<th>C</th>
<th>Mn</th>
<th>S</th>
<th>P</th>
<th>Mo</th>
<th>V</th>
</tr>
</thead>
<tbody>
<tr>
<td>C–Mn–CC</td>
<td>0.11</td>
<td>1.5</td>
<td>≤0.005</td>
<td>≤0.012</td>
<td>–</td>
<td>0.08</td>
</tr>
<tr>
<td>C–Mn–Mo–QT</td>
<td>0.06</td>
<td>1.5</td>
<td>≤0.005</td>
<td>≤0.012</td>
<td>0.32</td>
<td>0.08</td>
</tr>
</tbody>
</table>

Table 2. Heat treatment technique of tested steels and the corresponding microstructure.

<table>
<thead>
<tr>
<th>Seamless pipe</th>
<th>Heat treatment technique</th>
<th>Microstructure</th>
</tr>
</thead>
<tbody>
<tr>
<td>C–Mn–CC</td>
<td>Hot-rolled + controlled cooling to 550°C + air cooling</td>
<td>Bainite and Ferrite</td>
</tr>
<tr>
<td>C–Mn–Mo–QT</td>
<td>Hot-rolled + 900°C water quenched + 650°C tempered</td>
<td>Tempered martensitic</td>
</tr>
</tbody>
</table>

The working area of mounted specimen was wet-ground and cleaned according to the same method mentioned above. Periodic atmospheric corrosion test was performed to produce a rust layer on the working surface area of the mounted specimen.

The tendency of equiaxed grain morphology was significantly seen for two steels, as shown Figs. 1(b), 1(d). EDX microanalysis shows that the particles precipitated more or less uniformly dispersed were cementite in a dominate position and very fine spherical V-rich particles, corresponding to the red arrow in Fig. 1(e).

With two different processes were quite different microstructures. The microstructure of C–Mn–CC consisted of the pre-eutectoid ferrite and bainite in a dominate position, as shown in Figs. 1(b), 1(d). Details of microstructure of C–Mn–CC were further examined by TEM, as shown in Fig. 1(f). A large number of particles were observed in Figs. 1(d), 1(f) and EDX microanalysis shows that these particles were all cementite. No apparent V-rich particles by EDX were detected in the microstructure, which could be due to low V content in the steel and high cooling rate of controlled cooling technology. It is noted that the different grain morphology was significantly seen for two steels, as shown Figs. 1(a)–1(d). The tendency of equiaxed grain morphology was observed in C–Mn–Mo steel due to the completely recrystallization of the austenite grains during reheating at 900°C in quenching/tempering process. However, the tendency was not observed in C–Mn–CC. The reason is that, in controlled cooling process after hot-rolled, the austenite grains were heavily deformed in the non-recrystallization region before controlled cooling and the pancaked state were maintained to room temperature after controlled cooling.
In addition, it can be seen that there are high density tangles of dislocations in C–Mn–CC (see Fig. 1(f)), while the phenomenon are not significantly observed in C–Mn–Mo–QT (see Fig. 1(e)). Obviously, the dislocation density of C–Mn–CC was higher than that of C–Mn–Mo–QT, as exhibited in Figs. 1(e), 1(f). The high dislocation density in C–Mn–CC was inherited from the deformed austenite and was generated during bainite phase transformation. During the controlled cooling process of seamless pipe, the water from ultrafast cooling control device first cooled the outer wall, and then cooled the inner wall through heat transfer. This simple means that high stress was generated in the seamless pipe by controlled cooling technique. This is another reason for the high dislocation density in C–Mn–CC.

The mechanical properties of the two steels are shown in Table 3. As seen in Table 3, the mechanical properties of the two steels, such as the strength, the elongation and impact toughness, did not experience obvious change. The high strength of C–Mn–Mo–QT is available for precipitation strengthening of fine cementite and V-rich particles and solid solution strengthening of Mo. It is clear that a lot of bainite and separated fine ferrite have high strength and good toughness.2) Meanwhile, high dislocation density in the C–Mn–CC also can result in an increase in the strength of the steel. Clearly, the reduced strength of C–Mn steel caused by Mo decrease suitably offset by the application of controlling cooling technique, resulting in a reduced-cost C–Mn steel without sacrificing its mechanical property.

### Table 3. Mechanical properties of C–Mn–CC and C–Mn–Mo–QT steels.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Yield strength (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
<th>−40°C Impact toughness (J)</th>
</tr>
</thead>
<tbody>
<tr>
<td>C–Mn–CC</td>
<td>584</td>
<td>681</td>
<td>22.3</td>
<td>115.7</td>
</tr>
<tr>
<td>C–Mn–Mo–QT</td>
<td>591</td>
<td>683</td>
<td>23.1</td>
<td>121.0</td>
</tr>
</tbody>
</table>

3.2. Corrosion Rates from the Atmospheric Corrosion Tests

Corrosion rate was calculated by the following equation:

\[
V_{\text{corr}} = \frac{87 \times 600 \Delta m}{S \rho t}
\]

Where \(V_{\text{corr}}\) is the corrosion rate, \(\text{mm/y}\), \(\Delta m\) is the weight loss (g), \(S\) is the surface area of the specimen (cm²), \(\rho\) is the density of the test steel (g · cm⁻³) and \(t\) is the test time (h).

Figure 2 presents the results of the atmospheric corrosion tests as corrosion rates of tested steels with respect to test time (error bar was obtained based on two individual measurements). It can be seen that corrosion rate of C–Mn–CC was higher than that of C–Mn–Mo–QT in the whole test.
time. This means that controlling cooling technique can significantly improve the strength of reduced-cost C–Mn steel, while the phenomenon cannot reproduce for corrosion resistance.

3.3. Identification of the Phases in the Rust Layer

Surface product identification was carried out by XRD for two steels after 1 day and 30 days corrosion tests, as shown in Fig. 3. In addition to matrix structure of ferrite, \(\beta\)-FeOOH and \(\gamma\)-FeOOH were detected on the surface for both steels after 1 day, and then \(\alpha\)-FeOOH and Fe\(_3\)O\(_4\) were also observed in the rust layer after 30 days. It can be seen from Fig. 3 that as the test time increases for both steels, the peak intensities of \(\beta\)-FeOOH and \(\gamma\)-FeOOH also increase (see solid line boxes in Figs. 3(a), 3(b)), corresponding to the decrease of ferrite. Additionally, C–Mn–Mo–QT shows higher the peak intensities of \(\alpha\)-FeOOH and Fe\(_3\)O\(_4\) after 30 days than C–Mn–CC (see dotted boxes in Fig. 3(b)), which indicates that C–Mn–Mo–QT has higher the content of \(\alpha\)-FeOOH and Fe\(_3\)O\(_4\) in the rust layer than C–Mn–CC.

3.4. Electrochemical Corrosion Behavior

Figure 4 gives the potentiodynamic polarization curves of various pre-rusted steels in 3.5 wt% NaCl solution. No evident characteristics of passive corrosion were exhibited for these pre-rusted steels. It is clear that these samples had similar polarization curve shapes in the tested solution. Their \(E_{\text{corr}}\) (corrosion potential), \(\beta_a\) (the anodic Tafel constant), \(\beta_c\) (the cathodic Tafel constant) and \(i_{\text{corr}}\) (corrosion current density) values of samples, obtained from the polarization curves, are illustrated in Table 4. Two obvious trends can be concluded from Fig. 4 and Table 4. First, C–Mn–Mo–QT had higher \(E_{\text{corr}}\) and lower \(i_{\text{corr}}\) in 3.5 wt% NaCl solution than C–Mn–CC in whole atmospheric corrosion process and \(E_{\text{corr}}\) increases with the increase of test time. Second, at the first day of corrosion, the anodic...
Tafel constant ($\beta_a$) of C–Mn–CC was significantly larger than that of C–Mn–Mo–QT, indicating that a rust layer seem to form rapidly for C–Mn–CC. However, the anodic and cathodic Tafel constants ($\beta_a$ and $\beta_c$, respectively) of 7, 14 and 30 days did not obviously change their values, which indicates a similar mechanism for the corrosion reaction of the two steels in the later stages. It is worth noting that the magnitude relation between $E_{corr}$, $\beta_a$, $\beta_c$ and $i_{corr}$ of C–Mn–Mo–QT and C–Mn–CC after 14 days are opposite to those of other test times. The reason is not clear and needs further research.

### 3.5. Morphological Studies of the Rust Layer

Macroscopic views of the rust layer formed on C–Mn–CC after 30 days and C–Mn–Mo–QT after 1, 7, 30 days are shown in Figs. 5(a)–5(d), and the corresponding appearances of the substrate after the rust layer removal are shown in Figs. 5(e)–5(h). A lot of fresh metal on the surface was seen on the surface of C–Mn–Mo–QT after 1 day, which indicates that the rust layer has not been completely formed at the initial stage. Plenty of corrosion products were stacked on the corroded surface of C–Mn–Mo–QT during the following corrosion test, as shown in Figs. 5(b), 5(c). The uniform attack was the most common form of

<table>
<thead>
<tr>
<th>Test time (day)</th>
<th>Sample</th>
<th>$E_{corr}$ (mV)</th>
<th>$\beta_a$ (mV/dec)</th>
<th>$\beta_c$ (mV/dec)</th>
<th>$i_{corr}$ (mAcm$^{-2}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>C–Mn–CC</td>
<td>−803.7</td>
<td>498.9</td>
<td>92.5</td>
<td>0.259</td>
</tr>
<tr>
<td></td>
<td>C–Mn–Mo–QT</td>
<td>−815.1</td>
<td>406.5</td>
<td>108.8</td>
<td>0.074</td>
</tr>
<tr>
<td>7</td>
<td>C–Mn–CC</td>
<td>−790.1</td>
<td>318.7</td>
<td>102.6</td>
<td>0.131</td>
</tr>
<tr>
<td></td>
<td>C–Mn–Mo–QT</td>
<td>−722.5</td>
<td>302.1</td>
<td>136.4</td>
<td>0.137</td>
</tr>
<tr>
<td>14</td>
<td>C–Mn–CC</td>
<td>−749.0</td>
<td>254.2</td>
<td>125.8</td>
<td>0.095</td>
</tr>
<tr>
<td></td>
<td>C–Mn–Mo–QT</td>
<td>−783.5</td>
<td>298.9</td>
<td>110.1</td>
<td>0.153</td>
</tr>
<tr>
<td>30</td>
<td>C–Mn–CC</td>
<td>−740.7</td>
<td>257.7</td>
<td>143.3</td>
<td>0.212</td>
</tr>
<tr>
<td></td>
<td>C–Mn–Mo–QT</td>
<td>−607.8</td>
<td>150.8</td>
<td>167.2</td>
<td>0.116</td>
</tr>
</tbody>
</table>

Fig. 5. Macro-morphologies before (a–d) and after the rust layer removal (e–h): (a–c) and (e–g) C–Mn–Mo–QT steel after 1, 7 and 30 days, respectively; (d) and (h) C–Mn–CC steel after 30 days, respectively.

Fig. 6. SEM morphologies before (a–b) and after the rust layer removal (c–d) after 30 days: (a), (c) C–Mn–Mo–QT steel; (b), (d) C–Mn–CC steel.
corrosion of both steels. Compared to the rough surface of C–Mn–Mo–QT, C–Mn–CC surface was considerably smoother. This indicates that the corrosion of C–Mn–CC is likely a more homogenous dissolution compared to the C–Mn–Mo–QT.

Figures 6(a), 6(b) shows the surface SEM images of the rusted steels after the periodic wet/dry cyclic corrosion for 30 days. Compared with that on the rust layer of C–Mn–Mo–QT, the surface on the rust layer of C–Mn–CC was rougher and contained more cavities. In order to obtain further insight into corrosion mechanism, the surface morphologies of studied steels after removing the rust layer were characterized by SEM (see Figs. 6(c), 6(d)). It can be seen from Figs. 6(c), 6(d) that local corrosion had not been observed for both steels.

The cross section images of the rust layer formed on the steels exposed to the tested environment for 30 days were observed, as illustrated in Fig. 7. From the cross-sectional images of C–Mn–Mo–QT (Figs. 7(a), 7(c)), it is clearly seen that the rust layer consisted of outer and inner layers. Most flowery and plate-like structure, which is the typical micromorphology of $\gamma$-FeOOH and $\beta$-FeOOH crystal grains,\textsuperscript{13} were distributed in the outer layer. It can be observed that there was an open structure among these flower-shaped and plate-shaped products. Some authors have reported that the oxide layer has an open structure permitting the easy access of corrosive species to the metallic substrate.\textsuperscript{13} However, the inner layer was compact, which can protect the steel substrate from further corrosion. With respect to the C–Mn–CC exposed to the tested environment for 30 days, the micrograph of the cross section of the rust layer was similar to the C–Mn–Mo–QT but the more crakes, more cavities and less compact were generated. This means that the rust layer allows easy access of corrosive ions to the steel substrate and facilitates the occurrence of further corrosion. To determine the elements in the inner layer, EDX analysis was performed, as seen in Figs. 7(e), 7(f). A large amount of Cl and Na were detected in the inner layer of C–Mn–CC, but were not found in the inner layer of C–Mn–Mo–QT. It is worth noting that many steel substrate particles were obviously seen in the inner layer of C–Mn–CC according to EDX results, as shown Fig. 7(d).

All these finding indicates that the corrosion resistance of C–Mn–Mo–QT is better than that of C–Mn–CC in atmospheric environment. This is in close agreement with we drew from weight loss and polarization tests.
4. Discussion

The atmospheric corrosion is an electrochemical process that occurs under a thin film of liquid due to the condensation of water vapor caused by the change of temperature and/or humidity of atmosphere. During the exposure process, a transition of wet/dry cycle contributes to the formation and stability of the rust layer. During the periodic wet/dry cyclic corrosion, the following electrochemical corrosion reactions were involved at the wet stage:

Anodic reaction: \[ \text{Fe} \rightarrow \text{Fe}^{2+} + 2e^- \]  
Cathodic reaction: \[ \text{O}_2 + 2\text{H}_2\text{O} + 4e^- \rightarrow 4\text{OH}^- \]

Subsequently, \( \text{Fe}^{2+} \) reacts with \( \text{OH}^- \) and \( \text{O}_2 \) dissolved in the solution:

\[ 4\text{Fe(OH)}_2 + \text{O}_2 \rightarrow 4\text{FeOOH} + 2\text{H}_2\text{O} \]

The reactions (3) and (4) described the generation of FeOOH.

The morphology of \( \gamma \)-FeOOH and \( \beta \)-FeOOH looks like a flowery and plate-like structure, inside which many open spaces appear, making it easier for the outer corrosive ions to have access to the structure and destroy the surface. As mentioned previously, the \( \gamma \)-FeOOH and \( \beta \)-FeOOH are instable products, which can transform into \( \text{Fe}_3\text{O}_4 \) in the presence of \( \text{Fe}^{2+} \) ions in the rust. The reaction can be expressed as follows:

\[ 8\text{FeOOH} + 2\text{Fe}^{2+} + 2e^- \rightarrow 3\text{Fe}_3\text{O}_4 + 4\text{H}_2\text{O} \]

Kumar et al. reported that the non-stoichiometry \( \text{Fe}_3\text{O}_4 \) was an important factor in its corrosion protective ability. The nanocrystalline \( \text{Fe}_3\text{O}_4 \) rusts could form to fill the voids and make the inner rust layer denser and in turn be helpful to improve the corrosion resistance. It is also reported that \( \gamma \)-FeOOH can change into \( \alpha \)-FeOOH with the further corrosion of steel. This transformation of \( \gamma \)-FeOOH to \( \alpha \)-FeOOH is very important for the corrosion process of steel. In other words, the smooth development of this transformation has the benefit of improving the protective ability of the rust layer. In general, the dense and compact \( \text{Fe}_3\text{O}_4 \) and \( \alpha \)-FeOOH layer was located in the inner layer, which was attributable to protectiveness of the weathering steel.

It is well known that the atmospheric corrosion resistance of steels depends on the properties of rust layer formed on their surfaces. The properties of the rust layer mainly include the composition and structure of the rust layer. According to XRD analysis, the content of \( \alpha \)-FeOOH and \( \text{Fe}_3\text{O}_4 \) in the rust layer of C–Mn–Mo–QT was higher than that of C–Mn–CC, as shown in Fig. 3(b). With the help of SEM observation, the rust layer of C–Mn–CC exhibited the more crakes, more cavities and less compact than that of C–Mn–Mo–QT, as shown in Figs. 7(c), 7(d). According to the above description, it could be deduced that the properties of rust layer of C–Mn–Mo–QT is better than that of C–Mn–CC. This phenomenon should be associated with chemical composition, microstructure characterization and corrosion mechanism of C–Mn–CC.

In general, the corrosion occurs primarily at surface defects (such as dislocations and grain boundaries), and the oxide film or other passive film are prone to nucleate at the surface crystalline defects. Due to a large number of dislocations mentioned above, C–Mn–CC has more nucleation sites and is much easier to form oxide film than C–Mn–Mo–QT. This indicates that the anodic Tafel constant (\( \beta_a \)) of C–Mn–CC is significantly larger than that of C–Mn–Mo–QT at the first day of corrosion. Although the C–Mn–CC could form oxide film much faster than C–Mn–Mo–QT, oxide film of C–Mn–CC surface is not endowed to improved passivation ability and could not drastically prevent the penetration of chloride ions and water molecules into the steel matrix (see Fig. 7(f)). Once water molecules reach the steel matrix, the drastic corrosion reactions of steel matrix would happen because of the more corrosion activation provided by large numbers dislocations. Many defects of the rust layer, such as micro-cracks and cavities, might be caused by the high growth stress in the formation process of rust layers, due to the volume change between the formed corrosion product and the consumed Fe. Chloride ions and water molecules would penetrate into steel matrix through these defects, and corrosion reactions would happen in steel matrix.

For C–Mn–Mo–QT, Mo may enrich in the inner rust layer, as shown in Fig. 7(e). Many research reported that Mo remarkably enriches in the inner rust layer as MoO\(_3\) or FeMoO\(_4\), increase the density of rust layer and enhance the protective effect of rust layer and improve the corrosion resistance. Clearly, oxide film of C–Mn–Mo–QT surface is endowed to improved passivation ability and could prevent the penetration of chloride ions and water molecules into the steel matrix than that of C–Mn–CC. Additionally, C–Mn–Mo–QT has lower dislocation density than C–Mn–CC due to two different fabrication processes. Compared with C–Mn–CC, the lower corrosion activation of steel matrix of C–Mn–Mo–QT leads to fewer defects in the rust and less corrosion reaction rate in the steel matrix.

From those analysis above, the lower corrosion resistance of C–Mn–CC can come down to two aspects: First is the more corrosion activation of steel matrix caused by large numbers of dislocations; second is the oxide film with little protective effect in chloride sodium solution due to no costly alloy elements in the C–Mn–CC, such as Cr, Mo and so on.

During the controlled cooling process of C–Mn–CC, water first cooled the outer wall, and then cooled the inner wall, which indicates that the distribution of stress is not uniform in the microstructure. It is clear that the corrosion occurs primarily at high stress area (high dislocation area), and the volume expansion of corrosion products in these areas result in local corrosion stress which promotes defects and speeds up the corrosion process. On the contrary, the low stress area will relatively remain due to preferential dissolution of high stress area. This is probably reason that many steel substrate particles were obviously seen in the inner layer due to the non-uniform distribution of stress in the microstructure. However, the rust layer of C–Mn–CC has high porosity and many cracks, and thus provide nearly no protection against the penetration of Cl\(^-\), O\(_2\) and H\(_2\)O. This indicated that the local corrosion could not be continuous developed in the whole corrosion process, and
then the uniform attack is still the main form of corrosion of C–Mn–CC.

Uniform rust layer with few cracks tend to form on C–Mn–Mo–QT surface due to homogeneous microstructures and uniform distribution of stress or no stress for tempered martensite, and this is advantageous for the formation of a compact rust layer atmospheric corrosion. However, uniform micro-structures will result in over even interfaces between rust layers and steel matrix, which will probably lead to frequent peeling of rust layers from steel matrix because stress is induced by wet-dry alternations. Corrosion in the area of peeling of rust layers is more server than the other area. For C–Mn–CC, many steel matrix extruding into the rust layer (see Fig. 7(c)) and a large of random distribution of cementite phases play a key role in anchoring steel matrix and rust layers. Therefore, Compared with C–Mn–Mo–QT, C–Mn–CC has more flaws such as cracks and pores in the rust layer but the rust layers are bonded well to the substrates. This is probably reason that the surface of C–Mn–Mo–QT after removing the rust layer has considerably rougher than that of C–Mn–CC (see Figs. 5(g), 5(h)) due to frequent peeling of rust layers of C–Mn–Mo steel.

5. Conclusion

On-line controlled cooling technology for seamless pipe can remarkably improve the strength and reduces the cost. The microstructure, mechanical properties and corrosion resistance of C–Mn steel by controlled cooling technology was investigated and this investigation can provide some valuable information to use in commercial applications. Many conclusions are as follows:

1. High dislocation density in the C–Mn steel by on-line controlled cooling technique can result in an increase in the strength of the steel.

2. Controlling cooling technique can significantly improve the strength of reduced-cost C–Mn steel, while the phenomenon can not reproduce for corrosion resistance.

3. The lower corrosion resistance of C–Mn steel by on-line controlled cooling technique can come down to two aspects: First is the more corrosion activation of steel matrix caused by large numbers of dislocations; second is the oxide film with little protective effect in chloride sodium solution due to no costly alloy elements in the C–Mn steel, such as Cr, Mo and so on.

REFERENCES