Comparative Evaluations on Stress Corrosion Cracking Mechanism of Low Alloy Ultra-High Strength Steels in Neutral 3.5 wt% NaCl Solution

Qiang GUO, Jian-hua LIU*, Mei YU, Song-mei LI

Key Laboratory of Aerospace Advanced Materials and Performance (Ministry of Education), School of Materials Science and Engineering, Beihang University, Beijing 100191, P. R. China

*E-mail: liujh@buaa.edu.cn

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The stress corrosion cracking (SCC) mechanism of low alloy ultra-high strength steels 30CrMnSiNi2A and 40CrNi2Si2MoVA in neutral 3.5 wt% NaCl solution is investigated by slow strain rate technique (SSRT), surface analysis technique and electrochemical measurements. At anodic polarization, 30CrMnSiNi2A shows excellent SCC resistance; on the contrary, 40CrNi2Si2MoVA displays much higher SCC susceptibility than 30CrMnSiNi2A. At cathodic polarization potential, the steels are both susceptible to hydrogen induced cracking (HIC). At OCP, SCC mechanism of 30CrMnSiNi2A is mainly under control of HIC; whereas, SCC process of 40CrNi2Si2MoVA at OCP is controlled by a mixed mode of HIC and anodic dissolution.

Keywords: Stress corrosion cracking; Ultra-high strength steel; Cl−; HIC; Anodic dissolution

1. INTRODUCTION

In aerospace industry, corrosion problems are inevitable for structural materials in some outstandingly rigorous environment, such as coastal and marine atmosphere environments. For a long time, SCC has been generally acknowledged as a serious problem in marine atmospheric environment for structural metals, particularly for high strength steels [1-4].

Usually, SCC occurs unexpectedly and results in catastrophic accidence. The SCC mechanism of high strength steels is widely accepted as anodic dissolution and HIC [5-6]. The occurrence of SCC behaviour of steels in a certain environment is related with metallurgical microstructure, environmental factors and stress condition [7].
Liu [8] suggested that the SCC mechanism of the high strength steel X70 in the acidic solution was mix-controlled by both anodic dissolution and HIC. Prawoto [9] found that austenite was the most prone to SCC in the different microstructures including austenite, ferrite-pearlite and martensite, respectively. Contreras [10] found that the diffusion of atomic hydrogen was related to the fracture forms, and hydrogen permeation flux increased with the increasing of temperature. Ramamurthy [11] supposed that SCC velocity was independent of steels composition. Saruiglu [12] found that the crack growth behaviour and fracture mode was sensitive to the presence of inclusions and to the microstructure of steels. Hui [13] found that the alloying element Mo could improve the delayed fracture behaviour of high strength steels, and the secondary hardening carbides of V and Mo could also improve the SCC behaviour. Chu [14-15] found that hydrogen could induce delayed plastic deformation and was the necessary and sufficient condition of SCC or hydrogen induced delayed cracking.

30CrMnSiNi2A and 40CrNi2Si2MoVA are two kinds of low alloy ultra-high strength steels, which are widely applied in aerospace and other industries, due to their excellent comprehensive performance, such as ultra-high strength, excellent corrosion resistance, excellent fracture toughness and low manufacturing cost [16-20].

In the present work, SCC behaviour of 30CrMnSiNi2A and 40CrNi2Si2MoVA are comparatively investigated by SSRT with different corrosion potentials, surface analysis technique and electrochemical measurements. It is anticipated that the present research work could provide an essential insight into the SCC mechanism of low alloy ultra-high strength steels in neutral Cl⁻ environment.

2. EXPERIMENTAL

2.1 Materials and solution

Table 1. The main chemical composition and mechanical properties of 30CrMnSiNi2A and 40CrNi2Si2MoVA

<table>
<thead>
<tr>
<th>Steel type</th>
<th>C</th>
<th>Ni</th>
<th>Cr</th>
<th>Si</th>
<th>Mn</th>
<th>Mo</th>
<th>V</th>
<th>σ₀₂ / MPa</th>
<th>σₐ /MPa</th>
</tr>
</thead>
<tbody>
<tr>
<td>30CrMnSiNi2A</td>
<td>0.2</td>
<td>1.40</td>
<td>1.20</td>
<td>1.20</td>
<td>1.30</td>
<td>—</td>
<td>—</td>
<td>1450</td>
<td>1800</td>
</tr>
<tr>
<td>40CrNi2Si2MoVA</td>
<td>0.40</td>
<td>1.90</td>
<td>0.71</td>
<td>1.66</td>
<td>0.64</td>
<td>0.37</td>
<td>0.088</td>
<td>1640</td>
<td>1980</td>
</tr>
</tbody>
</table>

The main chemical composition and mechanical properties of two kinds of low alloy ultra-high strength steels 30CrMnSiNi2A and 40CrNi2Si2MoVA are shown in Tab.1. The heat treatment system is as follows: 30CrMnSiNi2A was austenitized at 890°C for 1 h and quenched into oil, then tempered at 280°C for 2~3 h; 40CrNi2Si2MoVA was austenitized at 870°C for 1 h, and subsequently quenched in oil, followed by doubly tempering at 300°C for 2 h with air-cooling to ambient temperature. The steels are machined into flat-plate tensile specimens and the schematic diagrams are shown in Fig.1. The tensile specimens were sealed with epoxy resin without the gauge section. Prior to each SCC test,
the gauge section was ground through successive silicon carbide papers to 800 grits, and subsequently
degreased with acetone, washed with deionized water and dehydrated ethanol by an ultrasonic cleaner,
respectively, and then dried in air. The experimental solution in the work was neutral 3.5 wt% NaCl
solution at room temperature (25±1°C, pH≈6.9).

![Figure 1. Schematic diagram of the flat plate tensile specimen for 30CrMnSiNi2A and 40CrNi2Si2MoVA](image)

2.2 Slow strain rate test (SSRT)

SSRT was performed on the tensile specimens by a tensile testing system at a strain rate of
$1 \times 10^{-6}$/s in air and in neutral 3.5 wt% NaCl solution at 25±1°C with different polarization potentials,
such as OCP, −550 mV, −700 mV, −850 mV and −1000 mV with respect to a KCl saturated calomel
electrode (SCE). Prior to each test, the specimens were all pre-loaded at 100 N. Each SCC test was
repeated at least three times in order to ensure the reliability and accuracy of the experimental data.

2.3 Electrochemical measurements

The electrochemical tests were performed in a conventional 3-electrode cell on a
electrochemical working station (VersaSTAT MC-4, Ametek, USA), with a KCl saturated calomel
electrode (SCE) as a reference electrode, a grand square platinum plate as counter electrode and the
polished steels as the working electrodes. The steel electrodes with exposing area 1 cm² were all
polished to 2000 grit silicon carbide papers, degreased in acetone, washed in deionized water and
derhydrated ethanol before the measurements, respectively. Prior to polarization tests, open circuit
potential (OCP) was recorded for at least 1 h until the potential was steady within 10 mV in 10 min.
Potentiodynamic polarization curves of the steels were carried out with a potential sweep rate of 0.5
and 50 mV/s.

2.4 Surface characterization

The fracture and the side surface adjacent to the fracture of the SSRT specimens were observed
by Field Emission Scanning Electron Microscope (FE-SEM, Apollo 300, CamScan, UK).
3. RESULTS

3.1 SSRT

![Graph](image)

**Figure 2.** Stress-strain curves of 30CrMnSiNi2A (a) and 40CrNi2Si2MoVA (b) at different polarization potentials in neutral 3.5 wt% NaCl solution

Fig.2 (a-b) are stress-strain curves of the SSRT 30CrMnSiNi2A and 40CrNi2Si2MoVA specimens in neutral 3.5wt% NaCl solution at OCP, -550 mV<sub>SCE</sub>, -700 mV<sub>SCE</sub>, -850 mV<sub>SCE</sub> and -1000 mV<sub>SCE</sub>, respectively. The specimens tensile in air were selected as the referenced ones in an inert environment without SCC tendency.

The tensile strength loss ($I_{\sigma}$), elongation loss ($I_{\delta}$) and reduction-in-area loss ($I_{\Psi}$) are used to evaluate the SCC susceptibility, which are calculated by Eq. (1-3).
\[
\begin{align*}
I_a &= \left[1 - \frac{\sigma_a}{\sigma_{a\text{,air}}}\right] \times 100 \quad (1) \\
I_{\delta} &= \left[1 - \frac{\delta_a}{\delta_{a\text{,air}}}\right] \times 100 \quad (2) \\
I_{\Psi} &= \left[1 - \frac{\Psi_a}{\Psi_{a\text{,air}}}\right] \times 100 \quad (3)
\end{align*}
\]

Where \(\sigma_{a\text{,air}}, \delta_{a\text{,air}},\) and \(\Psi_{a\text{,air}}\) are tensile strength, elongation ratio and reduction-in-area ratio of the specimens tensile in air, respectively, and \(\sigma_{\text{sol}}, \delta_{\text{sol}},\) and \(\Psi_{\text{sol}}\) are tensile strength, elongation ratio and reduction-in-area ratio of the specimens tensile in neutral 3.5 wt% NaCl solution.

**Figure 3.** Fracture time \((t_f)\) and tensile strength loss \((I_\sigma)\), elongation loss \((I_\delta)\) and reduction-in-area loss \((I_{\Psi})\) of 30CrMnSiNi2A and 40CrNi2Si2MoVA at different polarization potentials in neutral 3.5 wt% NaCl solution.

Fig.3 (a-d) show fracture time \((t_f)\) and tensile strength loss \((I_\sigma)\), elongation loss \((I_\delta)\) and reduction-in-area loss \((I_{\Psi})\) at different applied potentials, respectively. For 30CrMnSiNi2A, \(t_f\) is the shortest at \(-1000\ \text{mV}_{\text{SCE}}\) in all applied potentials, secondly at \(-550\ \text{mV}_{\text{SCE}}\), and then at OCP; however, \(t_f\) at \(-850\) and \(-700\ \text{mV}_{\text{SCE}}\) is longer than that at OCP. For 40CrNi2Si2MoVA, \(t_f\) is the shortest at \(-550\ \text{mV}_{\text{SCE}}\), secondly at \(-1000\ \text{mV}_{\text{SCE}}\) and then at OCP; \(t_f\) at \(-700\) and \(-850\ \text{mV}_{\text{SCE}}\) is much longer than that of OCP. Moreover, \(t_f\) at \(-700\ \text{mV}_{\text{SCE}}\) is the longest in all polarization potential, even more than that in air.

As shown in Fig.3 (b), for both 30CrMnSiNi2A and 40CrNi2Si2MoVA, \(I_\sigma\) at \(-550\ \text{mV}_{\text{SCE}}\) is the highest in all applied potentials, respectively, secondly at \(-1000\ \text{mV}_{\text{SCE}}\); however, \(I_\sigma\) at \(-700\)
mV\text{SCE} is the lowest. \(I_\sigma\) at -700 and -850 mV\text{SCE} of 30CrMnSiNi2A shows more than that at OCP. Moreover, \(I_\sigma\) of 40CrNi2Si2MoVA at -700 and -850 mV\text{SCE} is much higher than that at OCP.

In Fig.3 (c), \(I_\delta\) at OCP and -700 mV\text{SCE} of 30CrMnSiNi2A is lower than those at other potentials; the value at -1000 mV\text{SCE} shows the highest value. \(I_\delta\) at -700 and -850 mV\text{SCE} of 40CrNi2Si2MoVA is lower than others. \(I_\delta\) at -550 mV\text{SCE} of 40CrNi2Si2MoVA shows the highest value in all the applied potentials. \(I_\delta\) at OCP and -1000 mV\text{SCE} of 40CrNi2Si2MoVA approached to each other. However, the value of \(I_\delta\) at -850 and -1000 mV\text{SCE} of 30CrMnSiNi2A is much higher than those of 40CrNi2Si2MoVA.

In Fig.3 (d), \(I_\Psi\) of 30CrMnSiNi2A and 40CrNi2Si2MoVA shows negative value at -700 mV\text{SCE}, which increases with applied potentials increasing or decreasing from -700 mV\text{SCE}. The value of \(I_\Psi\) of 40CrNi2Si2MoVA at OCP is much than that of 30CrMnSiNi2A at OCP. The values of \(I_\Psi\) of 40CrNi2Si2MoVA at -850, -1000 mV\text{SCE} is close with those of 30CrMnSiNi2A, respectively.

In all, 40CrNi2Si2MoVA shows much higher SCC susceptibility than 30CrMnSiNi2A at OCP and -550 mV\text{SCE}. 40CrNi2Si2MoVA and 30CrMnSiNi2A both display the similar SCC susceptibility at -1000 mV\text{SCE}.

3.2 Fractography

![Fracture morphology of the SSRT 30CrMnSiNi2A specimens applied at different potentials](image)

Figure 4. Fracture morphology of the SSRT 30CrMnSiNi2A specimens applied at different potentials (a-e: OCP; -550 mV\text{SCE}; -700 mV\text{SCE}; -850 mV\text{SCE}; -1000 mV\text{SCE})

Fig.4 (a-e) shows that the fracture morphology of 30CrMnSiNi2A at different applied potentials. At OCP and -1000 mV\text{SCE} of 30CrMnSiNi2A, the brittle fracture surface is composed of intergranular fracture and secondary cracks; at -700 mV\text{SCE}, there are little cleavage regions and dimples in the fracture; at -850 mV\text{SCE}, there are several small brittle fracture in a small area distributed
along edge, mainly composed of intergranular fracture and secondary cracks. At $-550 \text{ mV}_{\text{SCE}}$, there are intergranular fracture characteristic and some secondary cracks with obvious corrosion mark in the brittle fracture.

**Figure 5.** Side surface morphology close to the fracture of the SSRT 30CrMnSiNi2A specimens applied at different potentials (a-e: OCP; $-550 \text{ mV}_{\text{SCE}}$; $-700 \text{ mV}_{\text{SCE}}$; $-850 \text{ mV}_{\text{SCE}}$; $-1000 \text{ mV}_{\text{SCE}}$)

**Figure 6.** Fracture morphology of the SSRT 40CrNi2Si2MoVA specimens applied at different potentials (a-e: OCP; $-550 \text{ mV}_{\text{SCE}}$; $-700 \text{ mV}_{\text{SCE}}$; $-850 \text{ mV}_{\text{SCE}}$; $-1000 \text{ mV}_{\text{SCE}}$)
Fig. 5 (a–e) shows that the side morphology close to fracture of 30CrMnSiNi2A at different applied potentials. In side morphology, there are a few cracks and corrosion pits on the side surface at OCP. A lot of pits and minor broad cracks are present on the side surface at $-550 \text{ mV}_{\text{SCE}}$. At $-700 \text{ mV}_{\text{SCE}}$, a lot of shallow cracks on the surface of the steel is initiated; at $-850 \text{ mV}_{\text{SCE}}$, more secondary cracks appear on the surface of the steel without any pits. At $-1000 \text{ mV}_{\text{SCE}}$, minor secondary cracks are found on the side surface.

Fig. 6 (a–e) exhibits the fracture morphology of the SSRT 40CrNi2Si2MoVA specimens applied at different potentials. At OCP, fracture surface of 300M is composed of intergranular, transgranular and secondary cracks fracture mode. At $-550 \text{ mV}_{\text{SCE}}$, there are mainly cleavage facets, with corrosion mark in the fracture, which shows characteristics of anodic dissolution. At $-700 \text{ mV}_{\text{SCE}}$, the fracture surface consists of quasi-cleavage and cleavage region. At $-850 \text{ mV}_{\text{SCE}}$, the brittle area is very small, and mainly contains quasi-cleavage and secondary cracks. At $-1000 \text{ mV}_{\text{SCE}}$, the brittle area is very broad, and mainly consists of cleavage and intergranular fracture modes.

![Images of side and fracture morphology](image)

**Figure 7.** The side surface morphology close to the fracture of the SSRT 40CrNi2Si2MoVA specimens applied at different potentials (a–e: OCP; $-550 \text{ mV}_{\text{SCE}}$; $-700 \text{ mV}_{\text{SCE}}$; $-850 \text{ mV}_{\text{SCE}}$; $-1000 \text{ mV}_{\text{SCE}}$)

Fig. 7 (a–e) shows side surface morphology close to fracture of the SSRT 40CrNi2Si2MoVA specimens applied at different potentials. At OCP, the side surface of 40CrNi2Si2MoVA shows fragmented cracks and corrosion marks. At $-550 \text{ mV}_{\text{SCE}}$, the side surface of the steel is seriously corroded with many shallow pits. The side surface at $-700 \text{ mV}_{\text{SCE}}$ shows remarkable deformation morphology with little cracks. At $-850\text{mV}$, there are a few broad cracks and a lot of small cracks. At $-1000 \text{ mV}_{\text{SCE}}$, there are only few long secondary cracks in the side surface beneath the fracture.
3.3 Potentiodynamic polarization curves

Fig. 8 (a-b) show potentiodynamic polarization curves of 30CrMnSiNi2A and 40CrNi2Si2MoVA with low and fast sweep rates of 0.5 and 50 mV/s, respectively. Corrosion potential ($E_{\text{corr}}$), corrosion current density ($I_{\text{corr}}$), anodic slope ($\beta_a$) and cathodic slope ($\beta_c$) of 30CrMnSiNi2A and 40CrNi2Si2MoVA at a sweep rate of 0.5 mV/s are shown in Tab. 2. At 50 mV/s, the corrosion potentials of 30CrMnSiNi2A and 40CrNi2Si2MoVA are $-900$ and $-975$ mV$_{SCE}$, respectively. In Fig. 8 (a) and (b), they show no passivity of the steels with different sweep rates in neutral 3.5 wt% NaCl solution.

![Potentiodynamic polarization curves](image)

**Figure 8.** Potentiodynamic polarization curves of 30CrMnSiNi2A (a) and 40CrNi2Si2MoVA (b) with fast and slow sweep rates 50 and 0.5 mV/s in neutral 3.5 wt% NaCl solution

**Table 2.** The corrosion potential ($E_{\text{corr}}$), corrosion current density ($I_{\text{corr}}$), anodic slope ($\beta_a$) and cathodic slope ($\beta_c$) of 30CrMnSiNi2A and 40CrNi2Si2MoVA at a sweep rate of 0.5 mV/s

<table>
<thead>
<tr>
<th></th>
<th>$E_{\text{corr}}$ /mV$_{SCE}$</th>
<th>$I_{\text{corr}}$ /mA.cm$^{-2}$</th>
<th>$\beta_a$ / mV.decade$^{-1}$</th>
<th>$\beta_c$ / mV.decade$^{-1}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>30CrMnSiNi2A</td>
<td>$-560$</td>
<td>0.0046</td>
<td>48.85</td>
<td>434.78</td>
</tr>
<tr>
<td>40CrNi2Si2MoVA</td>
<td>$-580$</td>
<td>0.014</td>
<td>84.21</td>
<td>750</td>
</tr>
</tbody>
</table>
4. DISCUSSION

The potentiodynamic polarization measurements of the steels 30CrMnSiNi2A and 40CrNi2Si2MoVA with different sweep rates of 0.5 and 50 mV/s confirms that the steels show little occurrence of spontaneous passivation in neutral 3.5 wt% NaCl solution. Due to lower content of Cr, 40CrNi2Si2MoVA shows lower corrosion resistance than 30CrMnSiNi2A in the initial periods, according to the results of $E_{\text{corr}}$, $I_{\text{corr}}$, $\beta_a$ and $\beta_c$ in Tab.2. According to Park’s work [21-22], during slow sweeping, there is sufficient time for the formation of rust on the steel surface, which could be used to simulate the quasi-stable electrochemistry of the crack wall. With sweep rate increasing, the quasi-equilibrium state was disturbed and changed into non-equilibrium condition. At fast sweeping, the rust formation is inhibited, which generates a potential range where the steel is under active dissolution state to simulate the crack tip electrochemistry.

According to the corrosion potentials obtained from Fig.8, the potentiodynamic polarization curves of 30CrMnSiNi2A and 40CrNi2Si2MoVA with fast and slow sweep rates of 0.5 and 50 mV/s were separated into different potential regions, as shown in Fig.8 (a) and (b). To be clearly discussed, the applied potentials were also marked in the polarization curves in Fig.8 (a) and (b), respectively.

For 30CrMnSiNi2A, the polarization curves could be divided into three sections for the discussion of SCC mechanism, such as region I: $E \geq -560$ mV$_{SCE}$, the crack tip and crack wall are both in anodic polarization state; region II: $-900$ mV$_{SCE} < E < -560$ mV$_{SCE}$, the crack tip and crack wall are in anodic and cathodic polarization, respectively; region III: $E \leq -900$ mV$_{SCE}$, the crack tip and crack wall are both in cathodic polarization state.

In region I, i.e. at $-550$ mV$_{SCE}$, 30CrMnSiNi2A shows excellent ductile characteristic, such as low elongation loss. It confirms that the steel shows excellent SCC resistance at anodic polarization. The high reduction-in-area loss of the steel at $-550$ mV$_{SCE}$ is due to serious electrochemical dissolution.

In region II, i.e. at $-700$ and $-850$ mV$_{SCE}$, the crack tip and crack wall of 30CrMnSiNi2A stay in anodic dissolution and cathodic polarization mode, respectively. The hydrogen evolution potential in neutral NaCl solution occurs at $-584.7$ mV$_{SCE}$, which is calculated according to Eq. (4) [23-24]. With polarization potential decreasing, the hydrogen evolution increases. The elongation loss and reduction-in-area loss increases with polarization potential increasing.

$$E_h (\text{mV}_{SCE}) = -59 \times \text{pH} - 177.6$$

(4)

In region III, i.e. at $-1000$ mV$_{SCE}$, the reduction of hydrogen at the crack tip and crack wall plays as dominant cathodic reaction. The hydrogen atoms in the steel reaches the critical amount responsible for HIC. At the condition, the SCC behaviour is prohibited, and in fact, hydrogen embrittlement is responsible for the failure of the steel at the polarization potential. 30CrMnSiNi2A shows high susceptibility to hydrogen embrittlement at $-1000$ mV$_{SCE}$.

$$\text{H}_2\text{O} + e^- \rightarrow \text{H} + \text{OH}^-$$

(5)

At OCP, 30CrMnSiNi2A shows good mechanical properties, such as low values of fracture strength loss, elongation loss and reduction-in-area loss. It means that the steel has excellent SCC resistance in neutral 3.5wt% NaCl solution. The fracture mode of 30CrMnSiNi2A at OCP shows
intergranular and transgranular fracture morphology. Therefore, the SCC mechanism of 30CrMnSiNi2A is mainly under control of HIC.

For 40CrNi2Si2MoVA, region I: \( E \geq -580 \text{ mV}_{\text{SCE}} \), the crack tip and wall is in anodic polarization; region II: \(-975 \text{ mV}_{\text{SCE}} < E < -580 \text{ mV}_{\text{SCE}} \), the crack tip is in anodic dissolution state and the crack wall is in cathodic polarization status; region III: \( E \leq -975 \text{ mV}_{\text{SCE}} \), the crack tip and crack wall are both in cathodic polarization state.

In region I, i.e. at \(-550 \text{ mV}_{\text{SCE}} \), anodic dissolution of the steel both occurs at crack tip and wall. The steel shows low fracture strength and much higher values of reduction-in-area loss and elongation loss than 30CrMnSiNi2A at \(-550 \text{ mV}_{\text{SCE}} \). It indicates that the steel shows high susceptibility to SCC induced by anodic dissolution.

In region II, such as, at \(-700, -850 \text{ mV}_{\text{SCE}} \), the crack wall is in cathodic polarization state and the crack tip is under anodic dissolution mode. The effect of HIC on the SCC behaviour increases and the effect of anodic dissolution decreases with polarization potential decreasing from \(-700 \) to \(-850 \text{ mV}_{\text{SCE}} \).

In region III, i.e. at \(-1000 \text{ mV}_{\text{SCE}} \), as polarization potential decreases to \(-1000 \text{ mV}_{\text{SCE}} \), hydrogen evolution plays dominant cathodic reaction (Eq. 5), and the hydrogen evolution increases to a critical amount for hydrogen embrittlement. It results in a large ductile loss at the polarization potential. 40CrNi2Si2MoVA shows higher ductile loss than 30CrMnSiNi2A at \(-1000 \text{ mV}_{\text{SCE}} \), such as fracture strength loss and reduction-in-area loss. It indicates that 40CrNi2Si2MoVA is more susceptible to hydrogen embrittlement than 30CrMnSiNi2A.

At OCP, 40CrNi2Si2MoVA shows high value of ductility loss. 40CrNi2Si2MoVA shows much more susceptibility to SCC behaviour than 30CrMnSiNi2A at OCP. Due to high SCC susceptibility induced by anodic dissolution of 40CrNi2Si2MoVA, the SCC mechanism at OCP is mainly controlled by a mixed mode of anodic dissolution and HIC.

5. CONCLUSIONS

(1) At anodic polarization potential \(-550 \text{ mV}_{\text{SCE}} \), the low alloy ultra-high strength steel 30CrMnSiNi2A shows excellent resistance to SCC. At cathodic polarization \(-1000 \text{ mV}_{\text{SCE}} \), the steel shows high susceptibility to SCC induced by HIC. The SCC mechanism at OCP is mainly controlled by HIC.

(2) At anodic polarization potential \(-550 \text{ mV}_{\text{SCE}} \), 40CrNi2Si2MoVA shows high susceptibility to SCC, which is controlled by anodic dissolution. At \(-1000 \text{ mV}_{\text{SCE}} \), the steel shows high susceptibility of SCC behaviour. The SCC mechanism at OCP is under control of a mixed mode of anodic dissolution and HIC, which shows higher SCC susceptibility than 30CrMnSiNi2A.

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References

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