Influence of Austenitizing Temperature on the Microstructure and Corrosion Resistance of 55Cr18Mo1VN High-Nitrogen Plastic Mould Steel

Hua-Bing Li^{1,2} · Wei-Chao Jiao¹ · Hao Feng¹ · Zhou-Hua Jiang¹ · Cui-Dong Ren¹

Received: 15 August 2016 / Revised: 8 October 2016 / Published online: 31 October 2016
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Abstract The influence of austenitizing temperature on the microstructure and corrosion resistance of 55Cr18Mo1VN high-nitrogen plastic mould steel was investigated. The microstructure, elemental distribution and Cr-depleted zone of different heat-treated samples were investigated by X-ray diffraction, electron probe microanalyzer analysis, and transmission electron microscopy. The corrosion resistance was evaluated using electrochemical measurements, and the analysis of passive film was carried out by X-ray photoelectron spectroscopy. The results indicated that the volume fraction of precipitates decreased, and the homogeneity of elements was improved with increasing austenitizing temperature. The degree of Cr-depleted zone around coarse $M_{23}C_6$ was severer than that around M_2N , and pitting corrosion initiated preferentially around $M_{23}C_6$. The corrosion resistance of the samples increased with the austenitizing temperature. With the increase in austenitizing temperature, the passive film was thickened and $Cr(III)_{Cr_2O_3}$ in the inner layer of passive film was enriched, which enhanced the corrosion resistance of the steel. The higher content of nitrogen in solid solution at higher austenitizing temperature contributed to the increased intensity of CrN and NH_3 , leading to the increase in pH value in the pit, and promoting the repassivation of 55Cr18Mo1VN steel.

KEY WORDS: High-nitrogen plastic mould steel; Austenitizing temperature; Microstructure; Corrosion resistance; Cr-depleted zone

1 Introduction

The continuous and rapid development of plastic industry has increased the demands for plastic mould steel with high hardness and wear resistance, good polishability and

outstanding corrosion resistance [1]. Corrosive medium, such as HCl, HF and SO_2 , was released during the manufacturing of plastic products [2], which were aggressive to plastic moulds and accelerated their failure. As an important alloying element, nitrogen is normally added into mould steel to improve its corrosion resistance and hardness [3]. In addition, nitrogen in steel contributes to inhibiting the formation of δ -ferrite and chromium-rich coarse eutectic carbides, avoiding the negative impacts on polish ability, fatigue capabilities and corrosion resistance [4]. However, the content of nitrogen in martensitic stainless steel is limited because of the low-nitrogen solubility under atmospheric pressure. Pressurized metallurgy is an effective method for the production of high-nitrogen martensitic stainless steel, such as 30Cr15Mo1N, 30Cr13N and 55Cr18Mo1VN with uniform composition and

Available online at <http://link.springer.com/journal/40195>

✉ Hua-Bing Li
huabing_li@163.com

✉ Hao Feng
fenghao241@163.com

¹ School of Metallurgy, Northeastern University, Shenyang 110819, China

² Key Laboratory for Ecological Metallurgy of Multimetallurgical Mineral (Ministry of Education), Northeastern University, Shenyang 110819, China

compact structure [3]. The 55Cr18Mo1VN high-nitrogen plastic mould steel with about 0.25 wt% N has been developed recently using pressured induction melting in Northeastern University, China. With higher hardness and better corrosion resistance, the 55Cr18Mo1VN steel could meet more rigorous requirements for plastic mould steel.

For plastic mould steel, austenitization is a transition but indispensable process for the heat treatment, which decides the effective grain size, distribution of insoluble precipitates and retained austenite, and exerts an important influence on the strength and corrosion resistance [5]. Choi *et al.* [6] and Lu *et al.* [7] investigated the effect of austenitizing temperature on the corrosion performance of the 3Cr13 plastic mould steel in chloride aqueous solution and found that the corrosion resistance increased with increasing austenitizing temperature. As reported, pitting corrosion always initiates in Cr-depleted zone adjacent to precipitates, which is closely related to austenitizing temperature [8]. So the undissolved precipitates after quenching would affect the corrosion resistance of the plastic mould steel. Additionally, improper austenitization treatment may induce decarburization, grain coarsening and formation of δ -ferrite [9]. Therefore, austenitizing temperature is very important to obtain high-quality plastic mould steel. However, little literature can be found about the heat treatment and corrosion properties of newly developed 55Cr18Mo1VN high-nitrogen plastic mould steel.

The aim of the present research is to reveal the effect of austenitizing temperature on the microstructure and corrosion resistance of 55Cr18Mo1VN high-nitrogen plastic mould steel. The phase constitution, precipitates, elemental distribution and passive film in samples austenitized at different temperatures were investigated by X-ray diffraction (XRD), electron probe microanalyzer (EPMA), transmission electron microscopy (TEM) and X-ray photoelectron spectroscopy (XPS). The corrosion resistance of the high-nitrogen plastic mould steel was investigated using open-circuit potential (OCP), potentiodynamic polarization and electrochemical impedance spectrum (EIS). Meanwhile, the pitting corrosion mechanisms of the samples austenitized at different temperatures were discussed.

2 Experimental Procedures

2.1 Material and Heat Treatment

The investigated 55Cr18Mo1VN high-nitrogen plastic mould steel was melted using a 25-kg pressurized induction furnace, and its chemical composition is shown in Table 1. The ingot was forged into 35-mm square slab in the temperature range of 1100–1150 °C. Then, the slab was kept at 1200 °C for 2 h and hot rolled into 7-mm-thick

Table 1 Chemical composition of 55Cr18Mo1VN high-nitrogen plastic mould steel (wt%)

C	Si	Mn	Cr	Mo	N	V	P	S
0.54	0.45	0.40	17.30	0.98	0.26	0.10	0.023	0.0021

plates for the subsequent testing. The plates were annealed, then fully austenitized at different temperatures (940, 980, 1020, 1060 and 1100 °C) for 0.5 h, followed by oil quenching. The heat treatment process curve is shown in Fig. 1.

2.2 Microstructure and Localized Chemical Analysis

XRD analysis of the annealed and austenitized specimens was performed on a Rigaku/SMART Lab X-ray diffractometer with CuK_α radiation. To observe the microstructure and distribution of elements (C, N and Cr), the specimens were etched for 5 s in the etchant consisting of 1 g K_2SO_3 , 15 mL HCl and 50 mL distilled water, and then examined by EPMA (JXA-8530F). TEM (JEM-2100F) observation was carried out to identify the precipitates and analyse the micro chemical composition of the precipitates. The samples for TEM analysis were grinded to 0.05 mm in thickness with a diameter of 3 mm, and then twin-jet electro-polished in electrolyte consisting of 8% perchloric acid, 10% ethylene glycol monobutyl ether and 82% alcohol solution at -20 °C. The composition of the passive film was characterized by XPS (ESCALAB 250) with a monochromatic AlK_α (1486.6 eV). The C 1s peak at 284.6 eV was used as a reference to correct the binding energies. The peak identification was performed using XPS peak 4.1 by referencing to a database of XPS.

2.3 Electrochemical Measurements

The samples used for electrochemical measurements were cut from the plates under different heat-treated conditions and then mounted in cured epoxy resin, leaving an exposed area of 1 cm². Prior to electrochemical tests, the specimens were wet grounded to 2000 grit SiC paper. All the electrochemical measurements were performed in aerated 3.5 wt% NaCl aqueous solution at 30 °C. And the tests were carried out using a Gamry Reference 600 potentiostat, comprising a thin platinum plate as the counter electrode (CE), a saturated calomel electrode (SCE) as the reference electrode and the sample as the working electrode (WE). In order to remove the original oxide film on the surface of the working electrode, cathodic polarization at -1 V/SCE was performed for 300 s. Then, the OCP was conducted for 1200 s to approach stable condition. The potentiodynamic

Fig. 1 Heat treatment process curves of the specimens

polarization was swept at a rate of 0.333 mV/s from -0.3 V/OCP to the anodic direction and finished when the current density exceeded 0.1 mA/cm². The EIS experiments were performed with 10 mV amplitude perturbation in the frequency ranging from 100 kHz to 10 mHz. The experimental results were analysed using the equivalent circuit embedded in Gamry Instruments Echem Analyst software.

2.4 Surface Observation of Pit Initiation Sites

To observe the pit initiation sites, the electrochemical samples were polished using $2.5\ \mu\text{m}$ diamond paste. Potentiodynamic polarization was carried out to induce the pitting corrosion. When the potential arrived at 50 mV below the pitting potential (E_{pit}), the experiments ended. The pit initiation sites were observed by field emission scanning electron microscopy (FE-SEM, Carl Zeiss Ultra Plus) equipped with energy-dispersive spectroscopy (EDS).

3 Results

3.1 Microstructure Evolution of 55Cr18Mo1VN

Figure 2a exhibits the XRD results of the samples under annealed and austenitized condition. The diffraction peaks of α phase with body-centred cubic (bcc) structure can be obviously observed in all of the samples. But, it is difficult to distinguish between martensite and ferrite by XRD analysis because of the same axial ratio. The hardness of the annealed sample was 17.2 HRC, while those of the austenitized samples were higher than 36.0 HRC. Based on the hardness results, α phases in annealed and austenitized samples were ferrite and martensite, respectively. The γ phase can be clearly found in the samples austenitized at 1020 and 1100 °C, which was regarded as retained austenite. The intensity of γ diffraction peaks increased and

α peaks decreased with increasing austenitizing temperature. The high C and N content in matrix due to dissolution of precipitates at high austenitizing temperatures (1020 and 1100 °C) enhanced the stability of austenite [7]. Additionally, the peaks for M_2N were detected in samples annealed and austenitized at 940 and 1020 °C.

The detailed information of the weak peaks in Fig. 2a within the 2θ range of $47\text{--}54^\circ$ is shown in Fig. 2b. The diffraction angles matched by the software were 48.44° , 50.84° and 51.60° for samples annealed and austenitized at 940 °C, which were considered as the diffraction peaks of $M_{23}C_6$, corresponding to the plane of (440), (531) and (442). The content of Cr-rich $M_{23}C_6$ and M_2N decreased with the increase in austenitizing temperature. And $M_{23}C_6$ and M_2N totally dissolved into the matrix above 1020 °C, or the content was below the detection level of XRD.

The microstructure evolution of the heat-treated samples observed via EPMA is shown in Fig. 3. Combined with the analysis of XRD, the microstructure of annealed sample consisted of ferrite and mixture of spheroidized Cr-rich $M_{23}C_6$ and M_2N (Fig. 3a). It could be evidently seen that the size and density of precipitates in annealed sample was much larger than those of the austenitized samples. With the increase in austenitizing temperature, the amount of precipitates deceased obviously (Fig. 3b–d). And coarse lath martensite with few precipitates could be observed in the sample austenitized at 1100 °C (Fig. 3d). The microstructure of 55Cr18Mo1VN was partially or entirely transformed into martensite after quenching, and the typical lath martensite was easily observed at relatively high austenitizing temperature. Besides, the morphologies showed that with increasing austenitizing temperature, the volume fraction of retained austenite also increased.

3.2 Characterization of Precipitates

The elemental distribution maps for C, N and Cr via EPMA in samples austenitized at 940, 1020 and 1100 °C are

Fig. 2 XRD patterns of the 55Cr18Mo1VN samples after different heat treatments: **a** diffraction angles from 40° to 100°; **b** detail information between 47° and 54°

Fig. 3 Microstructure of the heat-treated 55Cr18Mo1VN: **a** annealed and austenitized at **b** 940 °C; **c** 1020 °C; **d** 1100 °C

exhibited in Fig. 4. It can be seen that the C, N and Cr were enriched in precipitates (Fig. 4a, b). The volume fraction of carbides was much higher than that of nitrides, due to higher atomic friction of nitrogen in nitrides compared with carbon in carbides [10]. The volume fraction of carbides and nitrides decreased with increasing austenitizing temperature, and the elemental distribution became more uniformly. The content of C, N and Cr in matrix increased with the increase in austenitizing temperature. Analysis of C distribution via EPMA confirmed that carbon in samples austenitized at 940 and 1020 °C and sample austenitized at

1100 °C mainly concentrated in carbides and matrix, respectively.

To clearly identify the precipitates in 55Cr18Mo1VN, TEM were used to characterize the type of carbides and nitrides in the sample austenitized at 940 °C (Fig. 5). Plate precipitates of Cr_{23}C_6 with cubic structure and Cr_2N with hexagonal structure were observed. The Cr_{23}C_6 and Cr_2N were also obtained by XRD in the sample austenitized at 940 °C (Fig. 2). Similar residual precipitates were observed by Toro *et al.* [11] and Sun *et al.* [12]. In addition, the precipitation of Cr_2N was coherent with the alloy

Fig. 5 TEM results of precipitates in the sample austenitized at 940 °C: **a** Cr_{23}C_6 ; **b** Cr_2N

Fig. 6 Evolution of OCP of 55Cr18Mo1VN steels in 3.5 wt% NaCl after cathodic polarization

the tendency was consistent with that of OCP. It can be seen that all the samples showed distinct passivation region and the current density increased significantly above the pitting potential (E_{pit}) when stable pitting formed. The corrosion current density (I_{corr}) and passive current density (I_p) decreased rapidly with increasing austenitizing temperature, representing that the anodic dissolution of the steel was slower at higher austenitizing temperature and the protective passive film became more stable. Meanwhile, the E_{pit} increased with increasing the austenitizing temperature, indicating the enhanced corrosion resistance of the samples austenitized at higher temperature. Similar results were also reported by other researchers [6, 7].

In order to have a good understanding of the passive films formed on the samples austenitized at different temperatures, EIS was carried out under OCP in 3.5 wt% NaCl solution. All Nyquist plots (Fig. 8) exhibit semicircle arcs, and the diameter of semicircle arcs increased with the

Fig. 7 Potentiodynamic polarization curves of 55Cr18Mo1VN steels in 3.5 wt% NaCl

increase in austenitizing temperature, relating to the thickening of protective passive film and homogenization of the composition [7, 15]. The results showed the passive film formed at higher austenitizing temperature had better protective properties in 3.5 wt% NaCl solution [16].

The Nyquist plots in Fig. 8 were all well fitted with the model R_s - Q - R_p , which was proposed to simulate the behaviour of Cr13 martensitic stainless steel in NaCl solution [7, 17]. The equivalent electric circuit is displayed in Fig. 8, where R_s represents the solution resistance, R_p indicates the charge transfer resistance, and Q (CPE) denotes capacitive behaviour of the passive film. The electric circuit element Q was formulated as Eq. (1):

$$Z(\text{CPE}) = Q_0^{-1}(j\omega)^{-n}. \quad (1)$$

The parameter n is the dispersion coefficient relating to surface inhomogeneity [18]. Fitting results in Table 2 show that n and R_p values of the austenitized samples were

Fig. 8 Nyquist plots of heat-treated 55Cr18Mo1VN steels in 3.5 wt% NaCl

higher than those of the annealed one and increased with the increase in austenitizing temperature. It is reported that the thickness of passive film is inversely proportional to capacitance [19, 20]. The decrease in Q_0 indicates that higher austenitizing temperature resulted in the thickening of the passive films. Therefore, the samples austenitized at higher temperatures could form better protective passive film in the Cl^- containing environment, which is well consistent with the results of potentiodynamic polarization.

4 Discussion

4.1 Effect of Austenitizing Temperature on the Passive Film

XPS can provide more detailed information about the passive films on the samples austenitized at different temperatures, such as the chemical composition and the depth profile. Therefore, the samples were cathodically polarized at -1 V/SCE for 300 s to remove original oxide

films and then passivated at 0 V/SCE for 1 h to form passive films. The typical XPS spectra of Cr $2p_{3/2}$ and Fe $2p_{3/2}$ for the passive film after etching for 10 and 80 s are displayed in Fig. 9. It is clearly seen that the peaks representing CrOOH (576.8 eV), Cr_2O_3 (575.6 eV) and metallic Cr (573.6 eV) are identified in the passive film after etching for 10 s on the samples austenitized at different temperatures (Fig. 9a). The peak for CrOOH disappeared in the passive film after etching for 80 s (Fig. 9b). Similarly, deconvolution of Fe $2p_{3/2}$ spectra reveals the existence of peaks representing FeOOH (710.7 eV), FeO (709.4 eV), Fe_3O_4 (708.1 eV) and metallic Fe (706.7 eV) in the passive film after etching for 10 s (Fig. 9c). And the peaks for FeOOH and FeO disappeared after etching for 80 s (Fig. 9d). As mentioned above, the outer layer of the passive film was composed of the oxides (Cr_2O_3 , FeO and Fe_3O_4) and hydroxides (CrOOH and FeOOH), and the inner layer mainly contained Cr_2O_3 and Fe_3O_4 . Similar surface analysis results have also concluded that the passive film formed on Ni–Cr–Fe alloy surface is a bilayer film that contains an inner layer rich in Cr_2O_3 and an outer layer rich in $\text{Cr}(\text{OH})_3$ [21].

As shown in Fig. 9a, the intensity of CrOOH in the sputtering surface of the passive film for 10 s increased with increasing austenitizing temperature, which meant that the outer layer of the passive film had less protective effect. In detail, the contents of $\text{Cr}(\text{III})_{\text{CrOOH}}$ and $\text{Cr}(\text{III})_{\text{Cr}_2\text{O}_3}$ in the samples austenitized at different temperatures were calculated according to Fig. 9a, b. The ratio of $\text{Cr}(\text{III})_{\text{CrOOH}}$ and $\text{Cr}(\text{III})_{\text{Cr}_2\text{O}_3}$ in the outer layer (sputtering for 10 s) of the passive film was 1.6, 0.8 and 0.6 corresponding to the samples austenitized at 940, 1020 and 1100 °C, respectively. However, the CrOOH disappeared in the inner layer of the passive film, and the intensity of Cr_2O_3 increased with increasing austenitizing temperature. The percentage of $\text{Cr}(\text{III})_{\text{Cr}_2\text{O}_3}$ in the inner layer (sputtering for 80 s) of the passive film were 1.38, 1.41 and 2.65 wt% corresponding to the samples austenitized at 940, 1020 and 1100 °C, respectively. In summary, the amount of

Table 2 Fitting parameters of 55Cr18Mo1VN steels obtained from EIS spectra

Specimens (°C)	R_s (Ω cm ²)	CPE		R_p (Ω cm ²)
		$n(0-1)$	$Q_0/(\Omega^{-1} \text{ cm}^{-2} \text{ s}^n)$	
Annealed	2.986	0.7536	2.126×10^{-4}	2.382×10^4
940	4.399	0.7764	1.710×10^{-4}	5.613×10^4
980	1.382	0.8726	1.000×10^{-4}	7.248×10^4
1020	3.702	0.8730	7.832×10^{-5}	9.721×10^4
1060	3.021	0.8744	7.620×10^{-5}	1.209×10^5
1100	2.158	0.8865	6.330×10^{-5}	1.670×10^5

Fig. 9 Detailed XPS spectra of **a, b** Cr 2p_{3/2}, **c, d** Fe 2p_{3/2} recorded from sputtering surface of passive film for 10 and 80 s

Cr(III)_{CrOOH} in the outer layer increased with increasing austenitizing temperature, but the content of Cr(III)_{Cr₂O₃} in the inner layer increased obviously, which means that the increment of Cr(III)_{Cr₂O₃} in the inner layer would play an important role in the improved corrosion resistance with increasing austenitizing temperature. The intensity of Cr metal decreased with the increasing of austenitizing temperature, which indicated that the passive film became thicker at higher austenitizing temperature [22]. In addition, the Cr(III)/(Fe(II) + Fe(III)) ratio in the inner layer

increased with austenitizing temperature (Fig. 10), indicating the enrichment of Cr(III) and thickening of passive film at higher austenitizing temperature. These contributed to higher stability in 3.5 wt% NaCl solution, which was revealed by the increased values of OCP, E_{corr} and R_p (Figs. 6, 7; Table 2).

The N 1s spectra for the outermost of the passive films were decomposed into NH₃ (399.8 eV) and CrN (396.6 eV), as shown in Fig. 11a. After etching for 10 s, the passive films contained only one peak representing

Fig. 10 Variation of Cr(III) to (Fe(II)+Fe(III)) ratio in passive film with etching time for the samples austenitized at different temperatures

CrN. In the pitting corrosion process, CrN was generated in the passive film because of anodic segregation [23]. During the propagation of the pits, CrN may act as a precursor to passive film formation and form NH_3 in dissolution process through Eq. (2):

Besides, NH_3 can also be formed by the dissolution of atomic nitrogen in matrix as Eq. (3):

As a result, H^+ produced by self-catalytic mechanism [24] could be partially consumed, and the local acidification in the pits can be effectively inhibited, hence promoting the repassivation of pits. Many studies by XPS demonstrated that the alloyed N in the matrix was released into the solution in a form of NH_3 (NH_4^+) during the matrix dissolution [22, 25]. With increasing austenitizing temperature, the content of nitrogen in solid solution increased (Fig. 4), which improved the content of CrN and NH_3 . Therefore, the repassivation ability of 55Cr18Mo1VN steel was increased with austenitizing temperature.

4.2 Effect of Austenitizing Temperature on Pitting Corrosion Mechanism

The pit initiation sites in the sample austenitized at 940 °C were investigated by FE-SEM as shown in Fig. 12a. It can be seen that several metastable pits around precipitates could be observed. Metastable pits were considered as the initiation stage of pitting corrosion, and they might develop to stable pits [26, 27]. In order to clearly recognize the location of the metastable pits, the magnified morphology of the metastable pit zone is exhibited in Fig. 12b. The

Fig. 11 Detailed XPS spectra of N 1 s recorded from the outmost **a** and sputtering surface **b** of the passive film for 10 s on samples austenitized at different temperatures

Fig. 12 SEM images of the pit initiation sites on the sample austenitized at 940 °C: **a** metastable pits; **b** the local magnification of the metastable pits; **c** EDS analysis of precipitates exhibited in **b**

metastable pits were only observed around the precipitates with larger size. The composition analysis via EDS (Fig. 12c) indicated that precipitate with larger size (point a) was carbide, and precipitate with smaller size (point b) was nitride. This indicated that the pitting corrosion occurred preferentially around coarse carbides. Besides, the volume fraction of precipitates decreased with increasing austenitizing temperature (Fig. 3), which means the decrease in the pit initiation sites, thus the increase in the corrosion resistance.

TEM images of the precipitates in the sample austenitized at 940 °C are shown in Fig. 13a, c. The Cr content profiles from $M_{23}C_6/M_2N$ to the matrix along the lines marked in Fig. 13a, c are shown in Fig. 13b, d, and the Cr depletion was observed at the interface of precipitates and the matrix. Similar Cr-depleted zones around Cr-rich precipitates, such as Cr_2N , $Cr_{23}C_6$, σ and χ , were reported previously [28, 29]. Due to the lower chromium concentration, the Cr-depleted zone has less stable passive film [30], and acts as narrow anodic zone [13] during corrosion

process, inducing higher corrosion sensitivity. Therefore, the Cr-depleted zone was regarded as one of the factors responsible for the corrosion of the stainless steel [27]. In addition, the width and minimum Cr concentration of the Cr-depleted zone (Fig. 13) caused by $Cr_{23}C_6$ was severer than those by Cr_2N , which resulted from the smaller size of Cr_2N than $Cr_{23}C_6$. The Cr-depleted zone results are in reasonable agreement with the preferential initiation sites around the carbides (Fig. 12).

As is well known, the Cr-rich carbides and nitrides acting as preferential sites for pitting corrosion decrease the corrosion resistance of the steel [31]. With increasing austenitizing temperature, more precipitates dissolved in the matrix. Less Cr-depleted zones and more homogeneous distribution of chromium (Fig. 4c) in the matrix at higher austenitizing temperature were beneficial to improving values of OCP, E_{corr} , E_{pit} and R_p (Figs. 6, 7; Table 2) and decreasing values of I_r and I_p (Fig. 7). Therefore, the corrosion resistance of 55Cr18Mo1VN was enhanced with the increase in austenitizing temperature.

Fig. 13 a, c TEM images of precipitates in the sample austenitized at 940 °C: a Cr_{23}C_6 ; c Cr_2N ; b corresponding Cr-depleted zone of a; d corresponding Cr-depleted zone of c

Based on the results of microstructure and corrosion resistance, the schematic illustration of pitting corrosion mechanism of 55Cr18Mo1VN austenitized at low and high temperature is shown in Fig. 14. A large amount of undissolved precipitates existed in the sample austenitized at low temperature (Fig. 14a), and most of the precipitates were dissolved into the matrix at high austenitizing temperature (Fig. 14b), resulting in higher content of C, N and Cr in the matrix. Cr-rich precipitates induced the existence of Cr-depleted zone, and the degree of Cr-depleted zone around $M_{23}\text{C}_6$ was severer than that around $M_2\text{N}$. And metastable pitting corrosion initiated preferentially around the coarse $M_{23}\text{C}_6$ carbides. The less stable passive film around the precipitates corroded with Cl^- ions, forming $M\text{Cl}_n$ compound. Then, the $M\text{Cl}_n$ reacted with H_2O in the solution to

generate $M(\text{OH})_n$, Cl^- and H^+ ions, inducing low pH value and accelerating the local dissolution rate by self-catalytic mechanism. The high content of nitrogen in solid solution at high austenitizing temperature was beneficial to the formation of CrN and NH_3 (Fig. 14b), leading to the increase in pH value in the pit and promoting the repassivation of the 55Cr18Mo1VN steel.

5 Conclusions

1. With increasing austenitizing temperature, the volume fraction of Cr-rich precipitates decreased and retained

Fig. 14 Schematic illustration of pitting corrosion mechanism for 55Cr18Mo1VN steel austenitized at **a** low, **b** high temperatures

austenite increased, inducing more uniformly elemental distribution and enhancing content of C, N and Cr in the matrix. The precipitates were identified as $Cr_{23}C_6$ and Cr_2N , and the degree of Cr -depleted zone around $Cr_{23}C_6$ was severer than that around Cr_2N .

2. The increase in austenitizing temperature enhanced the values of OCP, E_{corr} , E_{pit} and R_p and decreased the values of I_r and I_p . Corrosion resistance of 55Cr18Mo1VN was improved with increasing austenitizing temperature. Pitting corrosion initiated preferentially around the coarse $M_{23}C_6$ carbides.
3. With the increase in austenitizing temperature, the passive film was thickened and $Cr(III)_{Cr_2O_3}$ in the inner layer of passive film was enriched, which enhanced the corrosion resistance of steel. The higher content of nitrogen in solid solution at higher austenitizing temperature contributed to the increased intensities of

CrN and NH_3 , leading to the increase in pH value in the pit and promoting the repassivation of 55Cr18Mo1VN steel.

Acknowledgments The present research was financially supported by National Natural Science Foundation of China (Grant Nos. 51304041, 51434004 and U1435205) and Fundamental Research Funds for the Central Universities (Grant No. N150204007).

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