Effect of Solution Treatment Temperature on Microstructural Evolution, Precipitation Behavior, and Comprehensive Properties in UNS S32750 Super Duplex Stainless Steel

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Abstract: The changes of microstructures, element distribution, and comprehensive properties were studied to explore their interactions with each other, induced by solution treatment of UNS S32750 super duplex stainless steel. The results showed that the ferrite content improved, while the austenite content declined as the temperature increased. From 900 to 1000 °C, the σ phase existing at α/γ grain boundaries and in ferrite grains gradually dissolved. At 1050 °C, the microstructures consisted of only ferrite and austenite. From 1050 to 1300 °C, the Cr₂N precipitated in ferrite and gradually grew and coarsened. The impact energy and pitting potential of UNS S32750 first improved and then weakened, while the hardness is the opposite, owing to the combined effects of element distribution, microstructures, and precipitates. In the presence of the σ phase, the corrosion resistance and mechanical properties of UNS S32750 correspond directly to the σ phase fraction. Subsequently, the rise in temperature promoted γ → α phase transformation, and the elements partitioning ratios of Cr and Mo declined, resulting in reduced toughness and corrosion resistance and a rise in hardness. Consequently, when the solution treatment temperature is 1050 °C, the α/γ ratio of UNS S32750 approached 1:1, with excellent overall properties.

Keywords: UNS S32750 super duplex stainless steel; solution treatment temperature; microstructure; precipitates; mechanical property; corrosion resistance

1. Introduction

Super duplex stainless steel (SDSS) has a double-phase ferritic-austenitic microstructure in approximately equal proportions, which provides it with excellent properties that cannot be achieved with austenitic or ferritic stainless steels [1–3]. In addition, its high resistance to pitting corrosion, with a pitting resistance equivalent number (PREN = wt% Cr + 3.3 wt% Mo + 20 wt% N) greater than 40 [4,5], renders it a high corrosion-resistant stainless steel. Compared with ordinary duplex stainless steel, super duplex stainless steel has higher strength and superior corrosion resistance, and is widely used in harsh environments such as petrochemicals and seawater desalination [6–8].

The high alloying of super duplex stainless steels increases the risk of secondary phase precipitation, and the corrosion resistance and mechanical properties of super duplex stainless steels are limited during the thermal processing in the range of 400–1000 °C, owing to the precipitation of sigma (σ), chi (χ), chromium nitrides (Cr₂N), G phases, and so on [9–11]. Among them, the σ phase is commonly considered to be the most hazardous component and has been extensively studied by scholars [12–14].
It is a brittle intermetallic phase rich in Cr and Mo with a tetragonal crystallographic structure, which mainly precipitates at 600–1000 °C [15–19]. Its precipitation seriously affects the plasticity, toughness, and corrosion resistance of materials [20]. The study indicates that the precipitation of a 5% σ phase leads to an 80% drop in impact energy and a 25 °C reduction in critical pitting temperature [21–24]. The σ phase is able to re-dissolve after the solution treatment temperature rises to a certain value, but during the rapid cooling process after solution treatment above 1200 °C, chromium nitride precipitates within the ferrite grains, forming a high-density Cr2N zone and reducing the material properties [25–27]. Therefore, proper control of the solution treatment temperature seems to be the simplest solution to obtain the desired performance.

However, the solution treatment temperature also affects the ratio of ferritic (α) and austenitic (γ) phases, the distribution of alloying elements between the two phases, and the corrosion resistance of each phase. It has been noted that alloying elements have different partition coefficients in the two phases [28]; chromium and molybdenum are enriched in ferrite, while nickel and nitrogen are concentrated in austenite. The reduction of Cr and Mo content in ferrite leads to a drop in the precipitation dynamics of the σ phase, which effectively inhibits σ-phase precipitation and enhances the toughness and corrosion resistance of steel [9,12]. In the process of rapid cooling at a high temperature, the supersaturated N in ferrite grains easily combines with Cr to form Cr2N [29,30]. The corrosion resistance of duplex stainless steels is mainly attributed to the action of the elements Cr, Mo, and N. The different content of alloying elements in the austenitic and ferritic phases leads to variations in the corrosion resistance of the two phases [31–33]. In this regard, it is of great significance to study the correlation between the solution treatment parameters and the microstructural characteristics.

In the current studies, however, few works have provided a complete and detailed description of the precipitation phase, biphasic ratio, alloying element content, mechanical properties, and pitting resistance of UNS S32750 as a function of variable thermal treatment temperature. At present, few scholars have investigated elemental partitioning ratios as a function of σ phase content and biphasic ratios, revealing mechanical properties as a function of σ phase content, biphasic ratios, and elemental partitioning ratios. In the present work, various microstructural states were obtained by varying the treatment temperature between 900 °C and 1300 °C to investigate the variation of microstructure characteristics and overall properties, such as α/γ ratio, precipitates behavior, element partitioning ratios, hardness, impact toughness, and corrosion resistance of UNS S32750 duplex stainless steel. The aim of this paper is to provide a sufficient theoretical basis for establishing the reasonable heat treatment process for UNS S32750 duplex stainless steel and analyze the interaction among the microstructure transition, precipitation phase, and alloying element partitioning ratios, as well as their influence on the mechanical properties and corrosion resistance.

2. Materials and Methods

The material used for this experiment was a 75 mm diameter UNS S32750 super duplex stainless steel bar, the main chemical composition of which is shown in Table 1. The size of the solid solution treated specimen was 10 mm × 10 mm × 12 mm, the size of the electrochemical test specimen was 10 mm × 10 mm × 5 mm, and the impact specimen was 10 mm × 10 mm × 55 mm in the standard form of a Charpy V-impact specimen.

The thermodynamic equilibrium phase diagram of UNS S32750 super duplex stainless steel was calculated by Thermo-Calc thermodynamic software, as shown in Figure 1. The solution treatment temperature range of experimental steel was determined according to the equilibrium phase diagram. The specimens were held at solution treatment temperatures of 900, 950, 1000, 1050, 1100, 1150, 1200, 1250, and 1300 °C for 40 min using a muffle furnace, the holding time including the time required to reach the target temperature, and then water quenched.

After mechanical grinding and polishing, the specimens were electrolytically etched with a 20% NaOH solution at a voltage of 10 V and an electrolysis time of 10–20 s. The microstructure was observed by the MX6R optical microscope (OM, SDPTOP, Shanghai, China), and the area fractions of ferrite and
austenite were counted using ImageJ image analysis software. Each sample was measured at 20 fields of view and the average value was taken. The structure morphology and chemical composition of ferrite and austenite were studied by JSM-6701F scanning electron microscopy (SEM, JEOL, Tokyo, Japan) and NS7 energy dispersive spectroscopy (EDS, Thermo, Waltham, MA, USA), and nitrides were observed and analyzed by JEM-2200FS transmission electron microscopy (TEM, JEOL, Tokyo, Japan). To analyze the microstructural changes of the precipitates, the specimens were electrolytically eroded with a 10% oxalic acid solution. The specimens were etched in the reagent of 1 g K2S2O5 + 15 mL HCl + 85 mL H2O for 10–20 s before counting the area ratio of the σ phase.

Table 1. Composition analysis of UNS S32750 super duplex stainless steel (mass fraction, wt%).

| C  | Si  | Mn  | P   | S   | Cr  | Ni  | Mo  | N  | Fe  |
|----|-----|-----|-----|-----|-----|-----|-----|----|-----|
| 0.024 | 0.39 | 0.67 | 0.022 | 0.002 | 25.49 | 6.18 | 3.47 | 0.288 | Bal. |

Figure 1. Equilibrium phase diagram of the UNS S32750 duplex stainless steel calculated by Thermo-calc software.

The Rockwell hardness (HRC) of each heat-treated sample was measured with 150 kg load for 10 s by TH300 Rockwell hardness tester (Beijing Time High Technology Co Ltd., Beijing, China), with at least 10 indentations per specimen was determined and the average hardness value was calculated. The Charpy impact test was carried out at room temperature with JB-30B impact tester (Wuzhong material testing machine Co., Ltd., Ningxia, China), and data were the average values of three standard impact samples.

Electrochemical tests were carried out with AUTOLAB PGSTAT302N electrochemical workstation (Metrohm, Beijing, China) in a typical three-electrode system, with the specimen as the working electrode, the saturated calomel electrode (SCE) as the reference electrode, and the platinum electrode as the auxiliary electrode. The 3.5% NaCl solution simulating seawater was used as the test medium at room temperature. The working electrode was first poled at −1 V for 10 min to remove the oxide film formed on the surface of the sample in air, and then the open circuit potential was measured for 50 min. After the open circuit potential was stabilized, the electrochemical impedance spectroscopy and potentiodynamic polarization curves were tested. Electrochemical impedance spectroscopy (EIS) was carried out at open-circuit potential, the test frequency range was 100 mHz–100 kHz, and the amplitude of the AC excitation signal was 10 mV. The scanning range of the potential polarization curve is −0.3–1.4 V (relative to the open circuit potential), and the scanning frequency is 1 mV/s.

3. Results and Discussion

3.1. Effect of Solution Treatment Temperature on Microstructural Evolution

The proportion of ferrite (α) to austenite (γ) has an important influence on the mechanical properties and corrosion resistance of duplex stainless steel. The solution treatment temperature plays a key role in the ratio of α/γ at a given chemical composition [24]. Figure 2 illustrates the optical
microstructures for UNS S32750 specimens as solution treated at 900–1300 °C. The ferrite gradually increases and the austenite gradually decreases with the rise in solution treatment temperature; the austenite and ferrite structures recover and recrystallize and grow, and the austenite structure gradually changes from long strips to islands distributed on the ferrite matrix. The σ phase distributes at the grain boundary of austenite and ferrite and within ferrite grain at 900–1000 °C. When the specimen is treated at 1050 °C, the microstructure of the specimen consists of ferrite and austenite phases, with no σ phase. It can be appreciated that the two main phases are found to be in approximately equal proportions.

Figure 3 presents the area fractions of ferrite and austenite and phase ratio after solution treatment at different temperatures. As can be observed, the ratio of α/γ improves rapidly, indicating the sharp increase in the content of ferrite during solution treatment at 900–1000 °C. At solution treatment temperatures ranging from 1000 °C to 1100 °C, the austenite transforms into ferrite slowly, the phase ratio is balanced, and the σ/γ ratio is 1.07–1.28. When the solution treatment temperature exceeds 1100 °C, the transformation rate from austenite to ferrite is accelerated, and the ratio of two phases (α/γ) rises rapidly. The reason is that the eutectoid reaction, α → γ$_2$ + σ, is reversible, and the σ phase and austenite phase rapidly transform into the ferrite phase, leading to a significant rise in ferrite content. The nitrogen is a strong austenitic forming element, which can expand and stabilize the austenitic zone. In the temperature range of 1000–1100 °C, the N element inhibits the transformation from austenite to ferrite. However, the stabilizing effect of the N element on austenite begins to weaken with the rise in temperature, resulting in the accelerated transformation of austenite to ferrite.

![Figure 2. Microstructures of the UNS S32750 super duplex stainless steel treated at different solution treatment temperatures. (a) 900 °C; (b) 1000 °C; (c) 1050 °C; (d) 1100 °C; (e) 1200 °C; and (f) 1300 °C.](image-url)
3.2. Effect of Solution Treatment Temperature on Alloying Elements Distribution in Two-Phase Structures

The energy spectroscopic analysis of the ferrite and austenite phases is performed under scanning electron microscopy and the results are shown in Figure 4, where the N element content is calculated by Thermo-Calc thermodynamic software. The Cr and Mo elements are mainly enriched in the ferrite phase, and the content of N and Ni in the austenitic phase is significantly greater than that in the ferritic phase. At 900–1050 °C, the content of Cr and Mo in ferrite rises significantly, while the Cr and Mo content drops gradually at 1050–1300 °C. With the rise in temperature, the content of Mo and N in austenite increases significantly, while the content of Ni reduces slightly; the content of Ni and N in ferrite continuously improves.

![Figure 3](image1.png)

Figure 3. Relationship between solution treatment temperature and (a) area fraction of α and γ phase and (b) phase ratio (α/γ) (error bars represent standard deviation).

![Figure 4](image2.png)

Figure 4. Chemical composition of austenite and ferrite phase in the UNS S32750 steel varies with solution treatment temperature. (a) Cr content; (b) Mo content; (c) Ni content; and (d) N content (error bars represent standard deviation).

The σ phase precipitates at the ferrite area along the phase boundaries or grain boundaries of γ/α and α/α in the 900–1000 °C range. Analysis of the relationship between the σ phase proportion and the content of Cr and Mo in ferrite, as shown in the Figure 5, shows that the rise in the precipitation of the σ-phase leads to the reduction of the content of Cr and Mo in ferrite. This is because, during
the precipitation of the \( \sigma \) phase, the Cr and Mo elements in the ferrite diffuse to the \( \sigma \) phase and are enriched, resulting in a decline in the content of Cr and Mo in the ferrite. With the increase of the two-phase ratio, the Cr and Mo partitioning ratio gradually drops, while the Ni partitioning ratio continuously rises, as shown in Figure 6. The increase of solution treatment temperature promotes the transformation of austenite phase to ferrite phase, the ferrite content rises, and the austenite content decreases, so that the concentration of Cr and Mo in ferrite improves, while the concentration of Cr and Mo in austenite declines. The Ni in the austenite diffuses into ferrite, and the Ni content in the austenite reduces slightly, while the Ni content in the ferrite increases.

![Figure 5](image5.png)

**Figure 5.** The relationship between the area ratio of the \( \sigma \) phase and the mass fraction of Cr and Mo in ferrite.

The N content in austenite is far higher than that in ferrite. The capacity of the face-centered cubic structure of austenite to accommodate interstitial atoms is much stronger than that of the body-centered cubic structure of ferrite, and the maximum solubility of N in the ferrite phase is about 0.05%, which makes N almost completely soluble in the austenitic phase. As the solution treatment temperature increases, the N content in austenite increases rapidly, while the N content in ferrite rises slightly [9,12]. In addition, the binding ability between N and Cr is strong, which hinders the dissolution of Cr in ferrite phase during phase transformation [4]. Therefore, Cr is more easily soluble in austenite phase during the phase transformation of austenite and ferrite.

3.3. Effect of Solution Treatment Temperature on the Precipitation Phase

Figure 7 shows the microstructure of the precipitated phase after solution at different temperatures. It is noticed that a large amount of \( \sigma \) phase precipitates in the matrix, after solution treatment at 900 °C. With the increase of solution treatment temperature, the precipitation of the \( \sigma \) phase gradually decreases, and as the solution treatment temperature rises to 1050 °C, the \( \sigma \) phase is completely dissolved. When the solution treatment temperature is between 1100 °C and 1300 °C, the black nitrides precipitates at the ferrite grain boundaries as well as inside the grains. With the increase of solution treatment temperature, these fine and dispersed precipitates begin to grow into the ferrite grains, and gradually grow coarse to form high-density regions.
The σ phase is formed during the eutectoid reaction, $\alpha \rightarrow \gamma_2 + \sigma$, and it preferentially precipitates at the interface between the austenite and ferrite phase and the grain boundary of the ferrite phase. The precipitation temperature range of the σ phase is 600–1000 °C [34,35], as indicated in Figure 1. Figure 8 shows the SEM image of the σ phase, and Figure 9 indicates the variation of the σ phase precipitation at different solution treatment temperatures. It is clear that the σ phase is widely distributed within the ferrite grains and at the $\alpha/\gamma$ phase boundaries, after solution treatment at 900 °C. With the increase of solution treatment temperature, the precipitation of the σ phase decreases, and the

Figure 7. Metallographic picture of precipitated phase morphology at different solution treatment temperatures. (a) 900 °C; (b) 950 °C; (c) 1000 °C; (d) 1050 °C; (e) 1100 °C; (f) 1200 °C; and (g) 1300 °C.
σ phase changes from concentrated distribution to dispersed existence in ferrite. At 1000 °C, the σ phase distributed in the ferrite is completely dissolved; only a small amount of the σ phase exists at the grain boundaries of the two phases, which is only 0.8%. After solid-solution treatment at 1050 °C, no σ phase is present in the matrix. There are two reasons for the dissolution of the σ phase. First, with the increase of solution treatment temperature, the N element content in ferrite increases, which reduces the activity of Cr and inhibits the precipitation of the σ phase [14,28]. Second, the austenite structure decreases gradually; subsequently, the interface between ferrite and austenite decreases and the nucleation site of the σ phase decreases [16,18]. Therefore, the solution treatment temperature should be at least higher than 1000 °C in order to eliminate the σ phase in UNS S32750 duplex stainless steel.

Figure 4d shows that the N element content in the ferrite increases, as the solution treatment temperature increases. However, the solubility of the N element in the ferrite phase is relatively low; it easily reaches the saturation state. Therefore, the N element and Cr element are combined to precipitate Cr₂N, during the rapid cooling process after dissolution [29,30]. Figure 10 exhibits the morphology of Cr₂N analyzed by SEM and TEM. The Cr₂N appears short and rod-like, and the Cr₂N precipitates in ferrite grains are fine or coarse. The different size of Cr₂N is probably due to sub grain boundaries and inhomogeneous nucleation at defects [27].

![Figure 8. Scanning electron microscopy (SEM) micrograph of σ phase. (a) 900 °C; (b) 950 °C.](image)

![Figure 9. Area fraction of σ phase at different solution treatment temperatures (error bars represent standard deviation).](image)
3.4. Effect of Solution Treatment Temperature on Mechanical Properties

The mechanical properties of UNS S32750 change curvilinearly with the increase of solution treatment temperature, as shown in Figure 11. The hardness of UNS S32750 decreases from 900 °C to 1050 °C and increases in the 1050–1300 °C range. The hardness of steel is the lowest at 1050 °C, while the impact energy is the opposite, reaching the highest value. The change of mechanical properties is consistent with that of microstructure. At 1050 °C, no σ-phase precipitation is present, and the ratio of ferrite to austenite is well balanced, approaching 1:1, resulting in the best performance.

When the solution treatment temperature is in the range of 900–1000 °C, the σ phase precipitates at the boundaries of ferrite and austenite and inside the ferrite grains. The σ phase is hard and brittle, whose precipitation has a great effect on the impact toughness and hardness of UNS S32750 duplex stainless steel. When the specimen is exposed to an impact load, the σ phase takes the lead in cracking, causing the grain boundary to become brittle and fracture along the grain, thereby reducing the impact toughness of the material. Figure 12 shows that the hardness of UNS S32750 corresponds directly to the proportion of σ phase. It is noteworthy that a small amount of the σ phase significantly reduces impact toughness, with only about 0.8% of σ phase decreasing the impact energy from 315.87 J to 100.93 J.
As the solution treatment temperature rises, the σ phase dissolves completely and the structural transformation proceeds slowly, the complete degree of recrystallization of the structure of the steel gradually increases, its constituent phases gradually grow larger, and the strength of the steel gradually weakens. Therefore, when the solution treatment temperature reaches 1050 °C, UNS S32750 has the highest impact toughness and the lowest hardness. Figure 13a shows the effect of \( \alpha/\gamma \) phase ratio on impact energy and hardness, and it is clear that the impact toughness and hardness directly correspond to the biphasic ratio. At room temperature, the strength of the ferrite with a body-centered cubic structure is higher than that of the austenite with a face-centered cubic structure [36]. The toughness of UNS S32750 is mainly derived from the austenitic tissue and the strength is primarily reflected in ferrite structure. As a result, the impact toughness weakens, while the hardness increases with the raise of \( \alpha/\gamma \) ratio. In addition, the influence of alloying elements, especially N elements, needs to be considered. With the rise in solution treatment temperature, the N partitioning ratio increases, and the hardness of the material improves, while the toughness weakens, as shown in Figure 13b. Nitrogen exists as interstitial atoms in the octahedral interstices of the matrix, forming an interstitial solid solution reinforcement that is about two orders of magnitude larger than the substitution solid solution reinforcement formed by chromium, nickel, and molybdenum [29]. Furthermore, it should be noted that the solubility of N in ferrite is low, and during the water-cooling process, supersaturated N combines with Cr to form Cr\(_2\)N, which enhances the hardness of steel and reduces the impact toughness.

![Figure 12. Effect of σ phase on impact energy and hardness.](image)

**Figure 12.** Effect of σ phase on impact energy and hardness.

3.5. Effect of Solution Treatment Temperature on the Corrosion Resistance

It is evident from Figure 14a that the Nyquist diagrams at different solution treatment temperatures are in the form of the capacitive arc, indicating that a more complete passivation film is formed on the surface of the material and the corrosion mechanism of UNS S32750 steel does not change. The radius of the capacitive loop increases first and then decreases, and reaches the maximum at 1050 °C. A larger radius of the capacitive arc indicates a stronger resistance to charge transfer at the metal–solution interface, which means better corrosion resistance of the metal [37]. Therefore, from the changing trend of the capacitive arc radius, it can be observed that the increase of solution treatment temperature

![Figure 13.](image)

**Figure 13.** (a) Effect of \( \alpha/\gamma \) phase ratio on impact energy and hardness; (b) effect of N partitioning ratio on impact energy and hardness.
makes the corrosion resistance of UNS S32750 steel first stronger and then weaker, and the corrosion resistance performance is optimal at 1050 °C.

Figure 15 shows the dynamic potential polarization curves of UNS S32750 at different solid solution treatment temperatures, and it can be seen that all the anodic polarization curves show a passivation zone, which is related to the formation of the passive film on the metal surfaces. Based on the analysis of the polarization curves, Table 2 presents the main electrochemical parameters. As the solution treatment temperature rises, the pitting potential \( (E_p) \) of UNS S32750 duplex stainless steel first improves and then weakens, and the pitting potential reaches the highest at 1050 °C, which indicates that the tendency of pitting corrosion at this solution treatment temperature is inferior. Because of the precipitation of the \( \sigma \) phase between 900 and 1000 °C, chromium and molybdenum depleted zones are easily formed in the vicinity, which leads to pitting corrosion. The corrosion resistance of duplex stainless steel directly corresponds to the proportion of the \( \sigma \) precipitation phase. The two-phase structure (ferritic and austenitic phase) in UNS S32750 at 1050 °C is more uniformly distributed, which is conducive to the stability of the passivation film on the steel surface and has a better inhibitory effect on pitting corrosion, manifested by a higher pitting potential \( (E_p) \). When the temperature of solution treatment rises above 1100 °C, the distribution of elements in the two phases is uneven. With the rise in solution treatment temperature, the ferrite phase ratio improves, and the concentration of Cr and Mo, the key elements for pitting corrosion resistance, decreases in ferrite, causing the corrosion rate to rise. The element N, an austenite-forming element, has a high solubility in the austenitic phase [26], leading to preferential corrosion in the ferrite phase due to the lack of N, which makes an outstanding contribution to pitting resistance. At the same time, with the increase of the solution treatment temperature, the \( \text{Cr}_2\text{N} \) precipitates in the ferrite during the water-cooling process, resulting in the reduction of pitting resistance of the ferrite phase. Therefore, the pitting potential \( (E_p) \) of UNS S32750 shows a downward trend at 1050–1300 °C.

| Temperature/°C | \( I_{\text{coor}}/\text{A cm}^{-2} \) | \( E_{\text{coor}} \) vs. SCE/V | \( E_p \) vs. SCE/V | \( (E_p-E_{\text{coor}})/V \) |
|----------------|---------------------------------|-------------------------------|-------------------|------------------|
| 900            | \( 9.8877 \times 10^{-8} \)     | \(-0.2686\)                   | 0.8460            | 1.1146           |
| 950            | \( 7.4826 \times 10^{-8} \)     | \(-0.2831\)                   | 0.8772            | 1.1603           |
| 1000           | \( 6.1035 \times 10^{-8} \)     | \(-0.2956\)                   | 0.9634            | 1.2590           |
| 1050           | \( 4.9744 \times 10^{-8} \)     | \(-0.2812\)                   | 1.0019            | 1.2831           |
| 1100           | \( 5.7373 \times 10^{-8} \)     | \(-0.2928\)                   | 0.9451            | 1.2380           |
| 1150           | \( 5.1575 \times 10^{-8} \)     | \(-0.2898\)                   | 0.9694            | 1.2592           |
| 1200           | \( 5.8994 \times 10^{-8} \)     | \(-0.2884\)                   | 0.9616            | 1.2500           |
| 1250           | \( 5.8948 \times 10^{-8} \)     | \(-0.2719\)                   | 0.9425            | 1.2144           |
| 1300           | \( 6.1156 \times 10^{-8} \)     | \(-0.2634\)                   | 0.9317            | 1.1951           |

In addition, the \( (E_p-E_{\text{coor}}) \) values listed in Table 2 show the same trend as the \( (E_p) \) values, with a tendency to rise and then fall. Because the \( (E_p-E_{\text{coor}}) \) value represents the resistance of the nucleus [15], the resistance of the pitting nucleus also increases and then reduces, i.e., the pitting resistance of the steel improves and then weakens as the solid solution temperature increases. Furthermore, as the solution temperature increases, the self-corrosion current density \( (I_{\text{coor}}) \) first becomes smaller and then larger. At 1050 °C, the lowest self-corrosion current density is \( 4.9744 \times 10^{-8} \text{ A cm}^{-2} \), indicating that UNS S32750 has quite good corrosion resistance.
The solution treatment temperature affects the alloying element content of austenite and ferrite in the steel, resulting in a variation in the corrosion resistance of both phases. The main alloying elements that affect pitting corrosion in chloride environment are Cr, Mo, and N. The pitting resistance equivalent number of the austenitic and ferritic phases can be calculated using the elemental contents of the austenitic and ferritic phases at different solution treatment temperatures in Figure 4. PREN is an empirical value for predicting the corrosion resistance of an alloy or phase; a larger PREN value represents a better corrosion resistance of the alloy or phase \[1\]. From Figure 16, it is evident that the PREN\(_\alpha\) of the ferrite phase first increases and then decreases with the rise in the solution treatment temperature, and the PREN\(_\gamma\) of the austenite phase gradually increases, which is caused by the change of the chemical composition in each phase, and this change is consistent with the variations of the composition of each phase in Figure 4. When the solution treatment temperature exceeds 1000 °C, no \(\sigma\) phase precipitates in UNS S32750, and the PREN value of both ferrite and austenite is greater than 40, which complies with the standard of super duplex stainless steel. It should be pointed out that the formula only considers the role of alloying elements, and does not take into account the effects of tissue inhomogeneity and precipitation phases. It is not appropriate to use only the PREN value to assess the pitting corrosion resistance of duplex stainless steel, because the distribution of the decisive Cr, Mo, and N alloying elements between the two phases is not balanced. The poor zones of these elements are prone to pitting corrosion, and the actual pitting resistance of the steel is lower than what is represented by the PREN value.

Figure 14. Electrochemical impedance spectroscopy (EIS) diagrams of UNS S32750 duplex stainless steel treated at different solution treatment temperatures. (a) Nyquist plot; (b) Bode plot.

Figure 15. Polarization curves of UNS S32750 solution treated at different temperatures.
4. Conclusions

(1) As the solution treatment temperature increases, the ferrite content rises, while the austenite content declines. At 1050 °C, the dual-phase ratio approaches 1:1. From 900 to 1000 °C, the σ phase fraction drops sharply from 19.33% to 0.80%. When the temperature exceeds 1050 °C, Cr$_2$N precipitates in the ferrite grain, and continuously grows and coarsens to form a dense Cr$_2$N region.

(2) The Cr and Mo are enriched in the ferrite, while Ni and N are concentrated in the austenite. From 900 to 1050 °C, the Cr and Mo content in the ferrite phase rises with the drop of σ phase precipitation. From 1050 to 1300 °C, the elements partitioning ratios of Cr and Mo decline with the increase of the α/γ ratio, while those of Ni and N continue to rise.

(3) With the rising solution treatment temperature, the mechanical properties of UNS S32750 vary in a curve. From 900 to 1050 °C, the hardness drops with σ phase dissolution, and the impact toughness improves. From 1050 to 1300 °C, as the α/γ ratio rises, the strength and hardness of the material rise, while the impact toughness weakens. At 1050 °C, the two-phase structure equilibrates, and UNS S32750 has a minimum hardness of 25.74 HRC and a maximum impact energy of 315.87 J.

(4) As the solution treatment temperature rises, the $E_p$ and $(E_p - E_{coor})$ values of UNS S32750 show a trend of increasing and then decreasing, while $I_{coor}$ shows the opposite trend. UNS S32750 has the best resistance to pitting corrosion treated at 1050 °C. The AC impedance test indicates that the capacitive arc radius and maximum phase angle vary in a curve, which is consistent with the conclusions obtained from the potentiodynamic polarization curve.

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