CHAPTER

15

Welding-associated failures in power boilers

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The function of a boiler is to produce superheated steam by transferring the heat produced by the combustion of oil or natural gas to water and steam flowing through the tubes that form the boiler. The super-heaters consist of banks of tubes suspended in the combustion gas flow path in the upper parts of the boiler. The steam flows from the super-heaters through the main steam header to the main turbine, where it drives the turbine-generator. The service conditions involved in boilers, which are mainly governed by alternating high temperatures and induced thermal stresses; residual stresses in welds; as well as the harsh corrosive environments, promotes failure causes such as thermal fatigue; stress-corrosion cracking (SCC); corrosion; and premature creep failures due to stress raisers at welding defects.

In high-temperature service (345-815 °C), and in the design of pressure vessels and piping, engineers and designers are confronted with the problem of selecting materials for a wide range of service conditions. The chromium molybdenum ferritic and austenitic stainless steels (SSs) are generally used for design temperatures above 425 °C [1]. In addition to service temperature, corrosion resistance, and fabricability, the following conditions should be considered in high-temperature applications: allowable maximum temperature and stress, type and value of load, expected life of the structure, and cost.

Austenitic SSs are generally used for design temperatures above 425 °C, the ASME code, section VIII, division 1, shows the SS grades that are suitable for the normal temperature range of usage (345-815 °C). Also some creep-resistant martensitic and ferritic SSs are used for high-temperature applications. The duplex stainless steels (DSSs) with both ferrite and austenite phases have gained increasing interest in recent years in power generation applications in chloride-containing solutions due to their better resistance to localized corrosion compared to single-phase austenitic SSs [2–4]. Welded DSSs are used at temperatures up to 250 °C. The increase in the thermal efficiency of fossil fuel-fired steam power plants, that can be achieved by increasing the steam temperature and pressure, has provided the incentive for the development of heat-resistant steels with excellent creep properties, as well as superior oxidation and corrosion resistance properties [5]. In the last two decades and to face up these requirements; several new Cr-Mo and 9-12% chromium steels were developed ranging from P11 (1Cr-0.5Mo) to P122 (12Cr-1Mo) [6].

Table 15.1 summarizes the main types of steels used in pressure vessels with their working conditions.

<table>
<thead>
<tr>
<th>Steel Type</th>
<th>Working Conditions</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ferritic stainless steels</td>
<td>430-510 °C</td>
</tr>
<tr>
<td>Austenitic stainless steels</td>
<td>595-815 °C</td>
</tr>
<tr>
<td>Duplex stainless steels</td>
<td>Temperatures up to 250 °C, with chloride-containing solutions</td>
</tr>
<tr>
<td>Creep-resistant martensitic</td>
<td>510-650 °C, with oxidation environments</td>
</tr>
</tbody>
</table>

Table 15.1 The Main Types of Steels Used in Pressure Vessels with Their Working Conditions
Austenite is the predominant phase in austenitic SSs, while in DSS the chemical composition is tailored to produce a structure consisting of ferrite and austenite. Unfortunately, during welding of SSs many discontinuities may be produced. These discontinuities act as stress raisers that can lead to a decrease in the life of the weld. Moreover, many of the involved structures include straight runs of tubes, near bends, and welded tubes. High-strength creep-resistant steels are usually subjected to normalizing and tempering heat treatment prior to service and therefore, have microstructures of tempered martensite. Creep strength of these steels is improved by the martensitic lath structure, precipitation strengthening effects of M23C6 carbide and MX (M = Nb, V, Cr and X = C, N) carbo-nitrides, and the solid-solution strengthening effect of Mo and W in the matrix. The precipitation strengthening effect of MX carbo-nitrides is especially important, because its coarsening rate is small and the fine particle size is maintained for long-term service [7].

2 STAINLESS STEELS (SSs) AND THEIR WELDING CHARACTERISTICS

SSs are the most important family of corrosion-resisting alloys. These alloys acquire their SS properties from the presence of the chromium element. The corrosion resistance of these alloys allows them to find numerous applications in chemical plants, high-temperature applications, food processing equipment, and others.

The effect of chromium lays in the formation of a tight, impervious layer of chromium oxide, which protects the metal form the environment. The essential level of chromium in SS is more than 11%. At this level, the chromium oxide layer is continuous and provides full protection. The other alloying additions in SSs, such as nickel, provide additional desirable properties such as low-temperature toughness, improved ductility, and strength.

2.1 CLASSIFICATION OF STAINLESS STEELS (SSs)

The structure of SS is determined by the stability of phases, such as ferrite, austenite, and martensite in microstructure, and hence by alloy composition leading to these phase stabilities. In this respect, it is worth mentioning that Cr, Mo, Si, Ti, Nb, V, and Al are ferrite stabilizers, while Ni, Mn, C, N, Co, and Cu are austenite stabilizers. Based on the effect of these alloying additions, SSs are composed of four basic families of alloys: ferritic, martensitic, austenitic, and DSSs. In addition, there is a family of alloys relying on precipitation hardening to improve mechanical properties. The duplex alloys are designed to have nominally 50% ferrite and 50% austenite. They provide superior toughness, strength, and corrosion resistance than ferrite alloys and better resistance to SCC than the austenitic alloys, and thus they find increasing applications in oil, natural gas, and chemical industries. However, their use is limited to 250 (for welded structures) and to 300 °C for other structures in order to prevent precipitation of different undesirable phases.
Table 15.2 summarizes the main families of steels with their main microstructural characteristics.

### 2.2 WELDING CONSIDERATIONS

Welding of SSs may cause a change in phase stability, precipitation of undesirable phases, or a delay in phase transformation due to welding heat cycle, in addition to welding residual stresses, which may cause failures in-service. The basic considerations of welding SSs as compared to carbon steels are discussed in the following sections.

#### 2.2.1 Thermal properties

The melting temperature and the thermal conductivity are lower for austenitic SSs than carbon steels, while the coefficient of thermal expansion is higher in austenitic SSs than carbon steels. For example: for 304 stainless and carbon steels, these properties are 1425 °C, 16.2 W/(m. K), and $17.6 \times 10^{-6} \text{°C}^{-1}$, and 1510 °C, 46.0 W/(m. K), and $11.7 \times 10^{-6} \text{°C}^{-1}$, respectively. Thus, less heat is required to melt austenitic SS than carbon steel. However, due to the lower thermal conductivity, narrower heat-affected zone (HAZ), and slower cooling rates are encountered compared to carbon steels. This may cause distortion and buckling especially with the larger coefficients of thermal expansion of SSs [8,9]. More tack welding is required when welding austenitic SS to support the structure and decrease distortion. Heat input for welding DSSs should be as low as required for good welding quality and for avoiding harmful precipitates in the weld and HAZ.

#### 2.2.2 Electrical resistivity

Electrical resistivities of SSs are generally higher than that of carbon steels, for example, four to five times larger than that of carbon steel at room temperature. This implies that higher heat generation is obtained in resistance welding and higher deposition rate in arc welding, that is, higher wire feed speed in submerged arc welding (SAW) and gas metal arc welding (GMAW).
2.2.3 Oxidation and depletion of chromium

Chromium is highly susceptible to oxidation and depletion due to the high chemical affinity of chromium to oxygen and carbon. This may lead to oxide inclusion defects and depletion of chromium from the weld metal and the HAZ, which deteriorates the corrosion resistance of welded joints. Therefore, there are some restrictions related to oxygen for welding SSs. Addition of stabilizing gas as oxygen or CO₂ in GMAW should be restricted to 2% of the shielding gas to avoid loss of alloying elements. The fluxes of SAW and shielded metal arc welding (SMAW) should have low oxygen levels to reduce oxidation and fluoride additions to facilitate chromium oxide fluxing. During the root pass welding of pipes, it is essential to protect the unshielded side by inert gas purging, this helps to avoid the loss of corrosion resistance at the inner side of the weld [8,10]. High heat inputs should be avoided to prevent chromium depletion in HAZ during slow cooling after welding of austenitic SSs.

After this concise review on some basic considerations of welding austenitic and DSSs, it might be useful to discuss some of the welding parameters and the potential welding problems and their remedies.

2.3 REVIEW OF WELDING PARAMETERS RELEVANT TO STAINLESS STEELS

The combination effect of heat input and heat treatment on the microstructure and mechanical properties of weld metal and HAZ in SSs and Cr steels has a major role on material performance. This combined effect shows on the resulted microstructure and formed phases and carbides as well as their stability. Accordingly, the final creep-resistant prosperities and SCC resistance are affected. Stringent control of welding conditions as per process qualification records requirements is an essential process for controlling the weld quality.

However, daily failures during service are reported and investigated from the chemical and energy plants [11,12], many of which prove to be originating from welded joints as they act as critical places where high residual stresses combined with heterogeneous structures exist so they act as potential sites for failures by stress-driven or environmental-driven mechanisms[13].

Though welding of austenitic SS types has reached a mature state of art, it has been shown that two ferrite morphologies precipitate at the austenite matrix of AISI/SAE 304 SS welded with an electrode type 308L using SMAW, GMAW, and flux-cored arc welding (FCAW) processes (skeletal and lathy type, the latter being characteristic of higher ferrite content). It has been also shown that (unlike FCAW) SMAW and GMAW cause a variation in grain size, as well as the appearance of exogenous phases [14].

DSS have a two-phase microstructure (austenite and ferrite) resulting their higher strength, higher resistance to intergranular corrosion, and lower cost compared to austenitic SSs. However, their difficult thermomechanical processing conditions limit their applications. Also, they are sensitive to heating and cooling conditions during welding and deleterious phases such as (σ phase and Cr₂N) may appear during welding of these steels. The Cr₂N phase is reported [2] to be the main precipitate found in the HAZ of...
welded joints. The conditions that favor the formation of austenite, such as high nitrogen content and low cooling rate, minimize the formation of Cr₂N precipitates, and increase the volume fraction of austenite. During cooling after welding of DSS the δ-ferrite transforms to austenite; the greater the cooling rate after welding the lower will be the austenite content, which deteriorates the tensile and corrosion properties of weld metal and HAZ. In order to increase austenite content in the weld metal, Ni-enriched filler metals (9% Ni) are used. The structure in the HAZ would accordingly exhibit increased δ-ferrite/austenite ratio compared to the base metal due to either fast cooling rates from the solution annealing temperature 1050 °C or to decomposition of the austenite to ferrite during holding at the temperature ≈600 °C [15]. Moreover, DSS are prone to age hardening and embrittlement at virtually all temperatures above 300 °C. At higher temperatures above 475 °C rapid aging results from the precipitation of various intermetallic phases (α, χ, R, and G phases) [16].

### 2.4 WELDING PROBLEMS

The common welding problems encountered in SSs are sensitization, SCC, hot cracking, HAZ cracking, weld metal solidification cracking, and fissuring. These problems are discussed briefly in the following sections.

#### 2.4.1 Sensitization

This problem can cause a rapid attack at the welding position. When the steel is heated at the temperature range of 500-850 °C, the carbon combines with chromium and forms carbides. These carbide particles form preferentially at grain boundaries, causing the depletion of chromium at the surrounding area. This results in a local loss of corrosion resistance. This process is called sensitization. This kind of damage takes place during welding at the HAZ, where the peak temperature reaches 600-850 °C, and therefore a localized damage called weld decay occurs in the HAZ, see Figure 15.1. The severity of sensitization depends on the carbon content of the base metal, the temperature, and the time of heating.

To avoid sensitization, three approaches are useful: (i) use of low carbon alloys such as 304L, 316L, 317L, these alloys do not have sufficient carbon (<0.03%) to

![Sensitization position, where the peak temperature in HAZ reaches 600-800°C](image.png)

**FIGURE 15.1**

Sensitization in the heat-affected zone [8].
react with chromium; (ii) use of stabilized SS grades (e.g., 321 and 347), which contain alloying additions with higher affinity to carbon than chromium, such as titanium and niobium; and (iii) carrying out a solution treatment for the welded joint at a temperature range of 1050-1100 °C, this treatment helps to dissolve any chromium carbide in the joint and restores the corrosion resistance [8,17].

### 2.4.2 Stress-corrosion cracking

This damage takes place as a result of tensile stresses combined with corrosive conditions [18]. Environments with halogen ions such as chlorides are commonly attributed to these failures. The welding residual stresses play major role in these failures. As a consequence, stress relieving may constitute a remedy for this problem.

### 2.4.3 Cracking

There are several causes of cracking in SSs welds. Hydrogen cracking is common in ferritic and martensitic grades. Stabilized austenitic SSs are prone to HAZ cracking after welding and reheat cracking in-service. The first is due to the formation of low melting point compounds at grain boundaries, while the second mechanism is due to precipitation hardening inside the grains. Both mechanisms are promoted by the stabilizing element [8].

### 2.4.4 Solidification cracking

Typical solidification cracking appears at the centerline of weld metal. This problem is caused by two factors: the presence of sulfur and phosphorus impurities, and the rejection of these elements to grain boundaries, where it forms a low melting phase, which may tear out due to tensile stresses of shrinkage late during solidification. The mode of solidification also affects this phenomenon, so that crack-free weld can be obtained when the metal solidifies as ferrite-austenite rather than autenite-ferrite, since ferrite has high solubility of sulfur and thus, little rejection to grain boundaries occurs[19].

### 2.5 POST-WELD HEAT TREATMENT

With regard to boiler steels, the final heat treatment for T/P91, T/P92, and T/P911 consists of normalizing and tempering, an austenitising temperature of about 1060 °C is adopted for hardening. A fully martensitic structure is obtained upon cooling to room temperature over a wide range of cooling rates. Welding of these steels is similar to welding hardenable alloy steels, since complete transformation to martensite during air cooling after welding occurs, therefore, the specification of preheat and post-weld heat treating is a major requirement for these steels. A detailed investigation on this steel is given in Ref. [20]. P91 can be welded satisfactorily by many processes including manual metal arc, submerged arc, and gas tungsten arc welding. Postweld heat treatment (PWHT) is necessary for tempering the martensite formed during welding, and many investigations have highlighted the need for optimizing the PWHT temperature and time as well as filler material composition [21,22].
On the other hand, due to the heat input during welding, not only different microstructures are obtained within the weld but also microstructural changes take place in a small area besides the fusion line. It has been shown that there is a great influence of heat input and heat treatment on the mechanical properties of P91 weld joints, where the best combination of properties was obtained in low to medium range of heat input, between 1 and 2.8 kJ mm\(^{-1}\) for both treatments, and a 90% increase in time to rupture was observed for normalizing/tempering treatment compared to subcritical post-weld treatment [23,24].

3 STRESS-CORROSION INDUCED FAILURES

3.1 Pitting corrosion of reboiler 304L tubes resulting from wrong material selection

Some case studies, with special relevance to failure by stress-corrosion induced failures will be discussed hereafter. Since recently a detailed study on failures associated with ferritic steels has been presented [25], this work will focus on failure cases on austenitic and DSS grades (304, 316 and duplex steel) that failed due to combined environmental and stressing conditions by corrosion- and/or mechanical-induced mechanisms.

A number of the reported failures on parts made from SSs have been attributed to either pure corrosion or combined stress-corrosion causes [11,26–28]. However, failures may be attributed to pure mechanical causes only; the failure of two centrifugal pump shafts (one made from 316 austenitic and the other from DSS) during use in a Brazilian petroleum refinery were attributed to fatigue [13], where in both cases, the failure started with crack nucleation and was promoted by the presence of nonmetallic inclusion particles, which appeared in the austenitic SS as a result of the early stages of steelmaking process (in the 316 SS shaft), and could not be attributed to the material properties in the case of DSS (where the failure started at an edge, where a nut locked the rotor in the shaft).

Pitting corrosion and/or SCC were found to be the cause for a number of the reported failures in SS parts in both welded or none-welded parts, in various structures made of 304 SSs working in aggressive environmental conditions [29–32].

Pitting corrosion, or pitting, is a form of extremely localized corrosion that leads to the creation of small holes in the metal. The driving power for pitting corrosion is the lack of oxygen around a small area. This area becomes anodic while the area with excess of oxygen becomes cathodic, leading to very localized galvanic corrosion. The corrosion penetrates the mass of the metal, with limited diffusion of ions, further pronouncing the localized lack of oxygen. The mechanism of pitting corrosion is probably the same as crevice corrosion [33]. This kind of corrosion is considered more dangerous than uniform corrosion as it causes little loss of material with small effect on its surface, while it damages the deep structures of the metal. The pits on the surface are often covered by corrosion products and the pits penetrate through the
thickness at faster rates than general corrosion. Pitting corrosion is usually initiated by a number of factors depending on the working environment. The pits usually initiate at surface defects (scratches, local changes in composition, or damage to protective coating). Pitting corrosion may also be caused by bacteria in promoting environments such as the oil industry. Microbiologically influenced corrosion caused by sulfate-reducing bacteria has been found to be a serious threat to pipelines. The presence of oxygen (even at low levels 0.1-0.2 ppm in pipeline water) promotes bacterial corrosion. The presence of CO₂ also causes severe corrosion, which is sometimes called “mesa” corrosion [34].

A failure study on two reboiler 304L tubes from two trains at Obaiyed Field was conducted at the Faculty of Engineering Cairo University [35]. The steam reboilers consist of a large number of tubes bent in U shape supported by tube sheets, all being contained in a large shell. The tubes carry hot oil at about 260 °C under pressure of 12 bar and the outer diameter of the tubes (shell side) is in contact with water. The temperature and the pressure of the water were 150 °C and 5 bar, respectively. The water analysis obtained from BAPETCO reports showed that the chloride content was in the range of 10-70 ppm for train E2613 and from 8 to 62 ppm in train E2623. This range of chloride content is relatively high for austenitic SS 304, especially at high working temperature of this equipment. The data indicated that the chloride content in the steam generators feed water was around 15 ± 5 ppm and around 60 ± 5 ppm at the outlet. The data also showed that the pH ranged from 5.8 to 10.4.

Table 15.3 shows the chemical analysis of the tubes’ steel from which it is confirmed to be accepted as 304L. This steel is known to have the lowest pitting resistance among austenitic SSs. It may be used in environments containing very low chloride contents especially at relatively high temperatures. High chloride content can cause pitting. High chloride content and high tensile stresses can cause SCC in this type of steel. The tensile strength as shown in Table 15.4 also confirms that the steel complies with standard requirements.

The two tubes were investigated visually; where one tube (E2613) showed pitting to a greater degree than the other one (E2623). The surfaces of both tubes were

<table>
<thead>
<tr>
<th>Sample</th>
<th>C</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Cu</th>
<th>Mn</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>V</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>E2613</td>
<td>0.03</td>
<td>0.49</td>
<td>0.009</td>
<td>0.036</td>
<td>0.11</td>
<td>1.69</td>
<td>20.1</td>
<td>10.37</td>
<td>0.27</td>
<td>0.07</td>
<td>0.004</td>
</tr>
<tr>
<td>E2623</td>
<td>0.027</td>
<td>0.47</td>
<td>0.007</td>
<td>0.034</td>
<td>0.11</td>
<td>1.75</td>
<td>19.8</td>
<td>10.22</td>
<td>0.21</td>
<td>0.06</td>
<td>0.006</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Specimen Code</th>
<th>0.2% Yield Strength (N/mm²)</th>
<th>Ultimate Tensile Strength (N/mm²)</th>
<th>%Total Elongation</th>
</tr>
</thead>
<tbody>
<tr>
<td>E2623</td>
<td>271</td>
<td>574</td>
<td>47</td>
</tr>
</tbody>
</table>
covered with black scale. No cracks were observed visually. For visual and penetrant tests, specimens were taken from the straight areas of the tubes as well as from the bent areas and the areas of the contact between the tubes and support plates. The penetrant test was carried out using solvent removable visible penetrant. No cracks were observed as seen from Figure 15.2.

Visual and macroscopic study of the unetched surfaces of the failed tubes (Figure 15.3) revealed the presence of various forms of subsurface and horizontal pits, where some were very deep and penetrated more than 75% of the thickness in one of the tubes, while others were narrow at the surface and wide under the surface. The pits were filled with a black dense residue. No cracks were found in 25 investigated cross-sections taken from the tubes even below the pits. Etched specimens showed normal austenitic microstructure under and around the pits (Figure 15.3). The chemical analysis of the scale is shown in Table 15.5, from which it is seen that the scale contains mainly oxides of iron, vanadium, and chromium. Vanadium and chromium oxides probably come from the process conditions.

The absence of cracks (visually or by penetrant test) indicated that SCC was not the reason of failure. This may be attributed to the very low stresses acting on the tubes at the working pressure of 12 bars (about 6.0 N/mm²), which is below the threshold stress for SCC. The presence of the dense black layer on some of the pits suggested possibility of the occurrence of crevice corrosion initiating at the contact areas between the tubes and the support plates. Specimens from contact areas between the tubes and support plates were examined using liquid penetrant test, but no evidence for such pits was revealed at the tube-plate contact areas. Accordingly, the failure reason was suggested to be pitting corrosion. The high chloride content combined with stagnant conditions induced pitting of the SSs. Once the pits were initiated, the liquid trapped inside the pit became denser and more acidic than the bulk, and corrosion was accelerated, extending partly toward the inner surface of the tube and partly in a downward direction. However, in some locations evidence for crevice corrosion was found beneath the dense oxide scale that covers the outer surface, this suggests the importance of periodic cleaning of the tubes. A previous study on corrosion of steel weldments in water desalination plants has revealed that left-over slag on weld surfaces can produce an initiation point for crevice corrosion [26].
FIGURE 15.3

Unetched specimens from E2613, (a–f), and etched specimens from E-2613 at two different magnifications (g, h), 50 × [35].
Pitting corrosion in steels is usually promoted by surface deposits that set up local concentration cells and dissolved halides that produce local anodes by rupture of the protective oxide film. Anodic corrosion inhibitors, such as chromates, can cause rapid pitting if present in concentrations below a minimum value that depends on the metal environment combination, temperature, and other factors. Pitting also occurs at mechanical ruptures in protective organic coatings if the external environment is aggressive or if a galvanic cell is active. Pitting normally occurs in stagnant environments. With corrosion-resistant alloys, such as SSs, the most common cause of pitting corrosion is highly localized destruction of passivity by contact with moisture that contains halide ions, particularly chlorides. Chloride-induced pitting of SSs usually results in undercutting, producing enlarged subsurface cavities, or caverns [36].

Since decreasing the chloride content below the current levels is impractical for such applications, it has been recommended to consider replacing the austenitic SS grade 304L by a SS of a higher pitting resistance number (e.g., 316L or 904L or even duplex steel grades).

### 3.2 FAILURE OF A SUPER-AUSTENITIC STAINLESS STEEL FIREFIGHTING WELDED PIPE LINE RESULTING FROM WRONG WELDING PARAMETERS

A failure investigation on super-austenitic stainless steel (SASS: UNS S31254) firefighting pipe line was reported recently [37]. SASSs are a special class of austenitic SSs with higher Mo and Cr content for high chloride-containing environments. The firefighting pipe was welded by GTAW process using a nickel base filler metal (ER NiCrMo3). The firefighting network was established in the oil and gas field to provide the fire protection through piping seawater as firefighting water. Leakage of the firefighting was observed at the weld line and/or adjacent to the weld zone (i.e., at HAZ) after the firefighting network was pressurized and put in-service as shown in Figures 15.4 and 15.5. The failure mechanism was identified to be pitting corrosion along the weld joint. Microstructural examination of the failed area showed that the failure started at the fusion zone between the weld metal and the HAZ as shown in Figure 15.6. This area is called “unmixed zone” (UMZ) and has a microstructure of dendrites of austenite similar to that of an autogenous weld [38].

EDX analysis showed that Mo and Cr levels varied across the joint. Mo has a partitioning coefficient less than 1; hence, it tends to segregate during solidification to the grain boundaries. Cr, also, segregates during solidification in a manner similar to that of Mo but with lesser degree. Thus, the core of the dendrites within the UMZ

### Table 15.5 Chemical Analysis of Elements of the Scale on the Tube (E2623)

<table>
<thead>
<tr>
<th>C%</th>
<th>V%</th>
<th>Mn%</th>
<th>Cr%</th>
<th>Ni%</th>
<th>Fe%</th>
<th>Mg%</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.4</td>
<td>28.2</td>
<td>0.54</td>
<td>6.7</td>
<td>2.4</td>
<td>20.64</td>
<td>0.06</td>
</tr>
</tbody>
</table>
becomes depleted in Mo. The molybdenum content in the dendrite cores can be as low as 4.2 wt.% leading to occurrence of localized attack in harsh environment [39]. This problem arises when SASS are welded using Ni-base filler metal such as ER NiCrMo3. The root cause of the failure case was the use of improper welding condition.

Accordingly, further investigation was conducted on the proper welding conditions for welding SASS with high Cr and Mo contents. GTAW with two different fillers were compared; namely: bead-on-plate using Ni-base filler metal (ER NiCrMo3) (similar welds) and fusion welding using Incoloy 28 filler metal (dissimilar welds).
For both cases, welding heat input was varied within the range from 1 to 5 kJ/mm. The effect of welding heat input on the microstructure and microsegregation in the UMZ, and the corrosion properties was investigated for both cases. The increase in heat input to 5 kJ/mm led to slight decrease in the Mo microsegregation in the UMZ and appreciable amount of NbC precipitation in the weld metal accompanied by slight increase in the weight loss due to corrosion, for the similar weld joints [40]; whereas, for dissimilar welding using Incoloy 28 the increase in heat input to 5 kJ/mm led to slight decrease in the Mo microsegregation in the UMZ accompanied by improvement in the corrosion resistance of the weld joint [41]. Thus, it was recommended to keep the heat input for similar welds of SASS at 1 kJ/mm and at 5 kJ/mm for dissimilar welding.

FIGURE 15.6
The microstructure at the failed area. The failure starts at the fusion zone (UMZ) between the weld metal and the HAZ [37].

A study has linked failure types, in the presence of mechanical loads, in weld austenitic SS joints to the mechanical and microstructural properties resulting from different welding conditions; Andres et. al [14] showed that some welding conditions promote the appearance of ferrite, formation of nonmetallic inclusions, and grain size and morphology variations of both austenite and ferrite phases. The behavior of the different structures under stress conditions varied and different causes for crack nucleation and propagation occurred to the GMAW and SMAW processes. Three different fracture modes were reported in this study at the welding joints; the FCAW process was influenced mainly by the first failure mode, while the other two had a mixture of the three different failure modes. The first failure mode was a geometrical stress concentrator at the weld root, due to a variation of the cross-section area of the weld that acted as the nucleation site of the crack, and further propagated through the weld or the HAZ. The second failure mode was generated at the HAZ, where the crack nucleated due to a variation in the grain size produced by the welding process, and further propagated through the edge or the center of the weld. The third failure mode appeared due to the presence of exogenous inclusions or slag generated by the welding process, which acted as stress raisers in the weld and produced the initiation and further propagation of the crack.
3.3 ROLE OF MICROSTRUCTURE ON SCC FAILURES OF AUSTENITIC SS

Corrosion attack is usually attributed to the difference in the composition and microstructures of the weld and parent metals caused by the high temperature and cooling rate; these in turn are being affected by the welding procedure used [42]. The welding defects which cause corrosion include poor adhesion, slag inclusions, crevice formation, macro and microfissures, liquid embrittlement, galvanic coupling, etc. Any of these defects can act as initiation points for corrosion attack on weld [26]. Under conditions of high chloride concentration and stagnancy, pits are developed at the vicinity of weld zone forming crevices which act subsequently as initiation centers for corrosion. It was reported that autogeneous welding has a detrimental effect on pitting resistance, as the pitting potential and critical pitting temperatures were lower for welded than unwelded steel [43]. The role of nonmetallic inclusions and left-over slag should be also emphasized. For deeper understanding of pitting corrosion, the reader may be referred to Frankel [44].

Almost, all austenitic SS grades, especially types 304 and 316, are susceptible to SCC. Other localized forms such as pitting and crevice corrosion probably precedes SCC and may act as origins where the cracks can nucleate and then propagate into the metal. These stresses may be originating from external sources (applied) or internal sources (residual stresses). Nearly, all metals are susceptible to SCC in the presence of tensile stresses in specific environments. There are three necessary conditions for SCC to occur; one factor is susceptible microstructure, the second is stress conditions either resulting from intrinsic or extrinsic sources and the third is the presence of a corrosant environment. Complete lists of environments that may cause SCC in many common metals are available in the literature [45].

In many austenitic SSs, the heat of welding causes sensitization (depletion of chromium from the matrix due to the precipitation of chromium carbides along the grain boundaries) [46]. Stress-corrosion cracks are reported to extend in the chromium-depleted regions. This has been explained in view of Cr being the element responsible for giving SS its corrosion resistance [45]. Other microstructural changes associated with sensitization are sigma phase precipitation and grain growth. Sensitization promotes intergranular corrosion [47].

SCC has been found to be related to microstructure. A study [48] showed that SCC happens in almost all microstructures: austenitic, martensitic, and ferritic-pearlitic, in discerningly order, where each microstructure produced a unique shape for SCC. The initiation point was pitting type cracks for austenite, corroded pearlite for ferrite-pearlite and corroded carbides for martensite structures. In all cases, propagation was intergranular, associated with grain boundary thickening in austenite. A recent approach has been handled in literature [49] where correlations are made between intergranular stress-corrosion cracking (IGSCC) resistance and grain boundary character and crystallographic texture for pipeline steels, however, no work could be found related to SSs. These studies have linked SSC susceptibility to crystallographic texture and grain boundary features resulting from the recrystalization and thermomechanical effects during rolling of these steels.
A failure investigation on secondary super-heater tubes in a power boiler indicated that SCC failure started at a crack originating from a highly stressed area and progressed outwards, with presence of branched cracks and pits, following an intergranular microstructural path associated with other characteristics of an IGSCC mechanism [30]. The macroscopic examination of the external and internal surfaces of in-service failed 304H tubes of a boiler (of the composition shown in Table 15.6) that exhibited SCC during service in a boiler has revealed that the tubes have experienced the occurrence of thin cracking originating at the internal surface of the tube and progressing outward circumferentially, as shown in Figure 15.7, showing the progression of the

**Table 15.6** Chemical Composition of Studied Material (wt.%)

<table>
<thead>
<tr>
<th>C</th>
<th>Cr</th>
<th>Ni</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Al</th>
<th>Ti</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.089</td>
<td>18.05</td>
<td>9.48</td>
<td>1.66</td>
<td>0.455</td>
<td>0.005</td>
<td>0.025</td>
<td>0.00359</td>
<td>0.00402</td>
<td>Bal</td>
</tr>
</tbody>
</table>

**FIGURE 15.7**

Full section crack [30].
crack from the internal surface from which it is revealed that in addition to the fine branched cracks, pitting also occurred as could be observed in association within central portions of stains. The microscopic investigations showed cracks propagating nearly through the wall as shown in Figure 15.8. The principle crack exhibited some branching near the inner surface (Figure 15.8). The cracks propagated for a distance of less than one-quarter of the tube wall and exhibited some branching. Results of microstructural examination (Figure 15.9) revealed austenitic grain structure with continuous precipitation of chromium carbides along grain boundaries, where the hardness reached 78-79 HRB at locations of continuous precipitation of chromium carbides along the grain boundaries. Figure 15.10 reveals the cracking around the grain boundaries. The recorded observations are consistent with (IGSCC) [50]. The study also showed that local heating resulting from bend solution treatment or heat input during welding of SS parts create the necessary conditions developing susceptible microstructures to SCC, the most important being the sensitization effect resulting from the precipitation of chromium carbides along the grain boundaries. Stress-corrosion cracks originate at regions of high stresses and propagate outward following the Cr-depleted zones (passes) to less stressed parts till they stop. Pitting and branching of finer cracks was associated with Cr-depleted zones. The study also showed a good correlation between SCC behavior and sensitization effects in 304H SS.

3.4 FAILURE OF A DSS AIR COOLING TUBE PLATE BY SULFIDE STRESS-CORROSION CRACKING

A failure case of air cooling tube-plate connections was reported in Slovakia in 2013 [51]. The failure happened after 3 weeks of service, following a welding maintenance stop. The air cooler consists of a plug header with dissimilar welded joints; the
The internal part of the plug header is made of the 15Mo3 steel, while the tubes Ø 25 mm were made of duplex steel DIN 1.4462 welded to the tube-plate \( h = 40 \text{ mm} \) (see the scheme of the header in Figure 15.11a and b). The welds were fabricated by TIG process using a special automatic welding machine with ER 309Mo wire Ø 0.6 mm, which is generally recommended for welding the given combination of steels. The chemical composition of the materials used is given in Table 15.7. The working medium for this air cooler consisted of hydrogen (up to 70%), hydrocarbons (up to 28%), and hydrogen sulfide, as well as water and other admixtures. The working pressure was 13.48 MPa and working temperature 50-122 °C. The metallurgical analysis of the case by macrostructure, metallographic, microhardness, and microanalysis (Figure 15.12) showed a defected welded joint. The weld metal hardness attained values of HV 435, the chemical microanalysis (EDX) of the weld metal revealed its considerable dilution by the tube-plate material: 10.03 wt.% Cr, 6.42 wt.% Ni, 1.55 wt.% Mo.

The analysis of the collected data revealed that failure occurred as a result of single pass welding as a consequence of very small thickness of the tube-plate [51]. The failure cause was identified to be sulfide stress-corrosion cracking (SSCC). From the viewpoint of SSCC, the hardness should not exceed the values of HV 248 according to NACE standard for low-alloy steels (for duplex steels the hardness limit is higher, namely, HV 285). The dilution of the weld metal to the previous composition was suitable for martensitic transformation upon fast cooling of welding process, as show
FIGURE 15.10

Microstructure adjacent to through-wall crack, etched electrolytically using oxalic acid solution [200 x] [30].
in Figure 15.13. This has provided the conditions required to form SSCC ((1) suitable microstructure (martensite + high hardness), (2) high percentage of sulfide compounds, and (3) stresses (residual stresses from welding process)). The combination of these factors (existing after welding of the previous components) led to the failure in the tube-plate by SSCC, as shown in Figures 15.14 and 15.15.
FIGURE 15.13
Martensitic microstructure of the WM [51].

FIGURE 15.14
SSCC in the weld metal [51].

FIGURE 15.15
SEM of SCC in the weld metal [51].
CONCLUSION

It should be kept in the mind of design and material engineers that the service conditions for power boilers provides the three necessary conditions (susceptible microstructure, state of stress, and corroding environment) causing SCC to end the life of parts. If stresses are absent, other forms of corrosion (pitting) may appear. During installation and assembly, it should be remembered that welding of attachments provides two out of three parameters essential for SCC existence. The wrong selection of the material or welding process may lead to design or metallurgical conditions promoting SCC.

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Welding-associated failures in power boilers


Non-Print Items

Abstract
This chapter provides a review on some important welding considerations that need to be considered to minimize chances for in-service failures of three of the most important steels (austenitic, duplex, and creep-resistant steels) used in manufacturing parts of power boilers. The service conditions for power boilers provide the three necessary conditions (susceptible microstructure, state of stress, and corroding environment) causing stress-corrosion cracking to end the life of parts. The review shows how failure is an end result of wrong material selection, or wrong welding conditions that do not count for possible metallurgical changes. In addition to proper material selection, the study highlights the significance of considering and understanding metallurgical related issues before adopting the welding route and the selection of the welding parameters.

Keywords: Power boilers, Stress-corrosion cracking, Austenitic stainless steels, Duplex stainless steels, Pitting corrosion