Filler Metal 16-8-2 for Structural Welds on 304H and 347H Stainless Steels for High-Temperature Service

ABSTRACT

The use of Type 16-8-2 filler metal was examined for application in structural welds on 304H and 347H stainless steels for high-temperature service applications and compared to welds with matching filler metals 308H and 347, respectively. Microstructural stability during elevated temperature exposure, weld metal impact properties, and susceptibility to stress-relief cracking were examined. It was found that the lean composition and low ferrite (~2 Ferrite Number [FN]) in 16-8-2 weld metal provides high resistance to intermetallic phase formation. No hot cracking was observed despite the low ferrite level. The 16-8-2 weld metals displayed superior toughness as compared to the matching filler metal welds, especially after longer elevated-temperature exposure. Experimental evidence for some martensite transformation in aged 16-8-2 weld metal upon cooling to ambient temperature was presented and explained an increase in magnetic response (as FN) after postweld heat treatment at 1300°F (705°C). None of the tested weld metals failed by stress-relief cracking mechanisms under the applied test conditions. The 16-8-2 filler metal welds exhibited significantly lower levels of stress relief during high-temperature exposure and significantly higher tensile strength after high-temperature hold as compared to the matching filler metal welds.

KEYWORDS

- Stainless Steel • 16-8-2 Filler Metal
- Impact Properties • Microstructure Stability
- Stress Relief Cracking (SRC)

Introduction

Type 16-8-2 (16 wt-% Cr, 8 wt-% Ni, 2 wt-% Mo, nominally) filler metal is the leanest of the austenitic stainless steel weld metal specifications. This weld metal composition was originally developed in the mid-1950s for welding of austenitic pressure vessel and piping components for high-temperature service applications (Refs. 1, 2). Composition limits for the solid wire (American Welding Society [AWS] A5.9, Specification for Bare Stainless Steel Welding Electrodes and Rods) and the shielded metal arc (SMA) weld metal deposit (AWS A5.4, Specification for Stainless Steel Electrodes for Shielded Metal Arc Welding) are given in Table 1.

Austenitic stainless steel weld metals are typically susceptible to embrittlement by the formation of intermetallic phases (Ref. 3). Sigma (σ) and chi (χ) phase precipitate during long-term elevated-temperature exposure. The presence of ferrite in the weld metal microstructure significantly accelerates intermetallic phase formation. The low total level of Cr + Mo and low ferrite (<5 Ferrite [FN]) found in Type 16-8-2 weld metals restrict the formation of brittle intermetallic phases (Refs. 4–7). Ferrite levels above 3 FN are usually required in austenitic stainless steel weld metals to avoid weld solidification cracking (Ref. 3). Despite its low ferrite content, Type 16-8-2 weld metal has been shown to be very resistant to solidification cracking (Refs. 8–10).

Figure 1 plots the nominal chemical composition range from Table 1 on the WRC-1992 diagram modified with extended axes and the martensite boundary for 1% Mn (Ref. 11). A nitrogen content of 0.05 wt-% and a typical minimum carbon content of 0.04 wt-% were assumed and are close to the actual composition of the filler metals used in this study (Table 2). Several Type 16-8-2 weld metals were included in the original WRC weld metal database (Refs. 12, 13). However, the leanest compositions extend beyond the iso-ferrite lines of the diagram. It can be seen that primary ferrite (FA) solidification is predicted for Type 16-8-2 weld metals even at less than 2 FN, which explains the high resistance to solidification cracking at very low ferrite levels. The leanest corner of the 16-8-2 composition box touches the lower limit of the martensite boundary. Martensite is predicted for weld metal compositions below and to the left of this martensite boundary. Compositions within the shaded region of the martensite boundary are unpredictable (Ref. 11). As-deposited martensite has not been reported in Type 16-
8-2 weld metal and is considered unlikely because dilution effects would lower the Mn content, and the final weld metal composition would shift to the left and up, farther away from the martensite boundary. However, strain-induced martensite formation upon bending has been previously observed in Type 16-8-2 weld metals (Ref. 10).

The excellent microstructural stability and good weldability of Type 16-8-2 compositions make it a candidate filler metal for structural high-temperature applications in the oil and gas downstream industry. Catalytic cracking units used in the petroleum refining process experience long-term high-temperature exposure, typically in the range of 1200˚-1400˚F (650˚-760˚C), and require materials that balance creep resistance, oxidation resistance, and overall cost. Type 304H and 347H stainless steels are predominantly used for this application. Welds using matching filler metals 308H and 347, respectively, have been associated with loss of high-temperature ductility and creep property degradation due to intermetallic phase formation (Refs. 14, 15).

The purpose of this investigation was to explore the potential application of Type 16-8-2 filler metal for high-temperature structural welds in oil and gas downstream applications. A comparative evaluation of the microstructural stability during elevated-temperature exposure, impact toughness, and susceptibility to stress-relief cracking of 304H and 347 steel welds produced with Type 16-8-2 and with matching filler metals was conducted.

**Experimental Procedures**

**Base and Filler Metals**

The base metals used in the present work are commercial 304H and 347H stainless steel. Welds on each base material were made using Type 16-8-2 filler metal, and as a reference,

| Specification | C   | Cr  | Ni  | Mo  | Mn  | Cu  | Si  | S  |
|---------------|-----|-----|-----|-----|-----|-----|-----|----|
| ER16-8-2      | < 0.10 | 14.5-16.5 | 7-9.5 | 1-2 | 1-2 | < 0.75 | 0.3-0.65 | < 0.03 | < 0.03 |
| AWS A5.9      |     |     |     |     |     |     |     |    |
| E16-8-2-15    | < 0.10 | 14.5-16.5 | 7-9.5 | 1-2 | 0.5-2.5 | < 0.75 | < 0.60 | < 0.03 | < 0.03 |
| AWS A5.4      |     |     |     |     |     |     |     |    |

| Specification | C   | Cr  | Ni  | Mo  | Mn  | Cu  | Si  | P  | S  |
|---------------|-----|-----|-----|-----|-----|-----|-----|----|----|
| ER308H        | 0.044 | 19.65 | 9.61 | 0.09 | 2.12 | 0.038 | 0.16 | —  | 0.55 | 0.024 | 0.016 |
| E308H         | 0.060 | 18.70 | 10.00 | 0.03 | 0.60 | —    | 0.15 | —  | 0.70 | 0.032 | 0.021 |
| ER347         | 0.050 | 18.40 | 9.40 | 0.01 | 1.81 | 0.016 | 0.01 | 0.570 | 0.33 | 0.019 | 0.004 |
| E347          | 0.078 | 18.60 | 9.78 | 0.02 | 1.70 | 0.074 | 0.04 | 1.000 | 0.13 | 0.027 | 0.004 |
| ER16-8-2      | 0.048 | 14.97 | 7.79 | 1.25 | 1.36 | 0.058 | 0.14 | —  | 0.38 | 0.026 | 0.011 |
| E16-8-2-15    | 0.050 | 14.80 | 7.80 | 1.00 | 1.40 | —    | 0.08 | —  | 0.37 | 0.020 | 0.004 |

**Table 1 — Weld Metal Specifications for Lean Austenitic Type 16-8-2 Stainless Steel (wt-%)**

**Table 2 — Chemical Composition (wt-%) of Base Metals and Filler Metals**
using matching filler metal 308H or 347, respectively. The chemical composition of the base and filler metals is presented in Table 2. Notice the lean composition of the Type 16-8-2 filler metal as compared to the 308H and 347 filler metals.

Welding Conditions

Two welded joints were fabricated for each base material to compare the effect of using Type 16-8-2 filler metal over the corresponding matching filler metal. Multipass welding was conducted in accordance with the ASME Boiler and Pressure Vessel Code, Section IX, using 12.7-mm- (1/2-in.-) thick plates with a 75-deg Y-groove and a root face of 0.8–2.4 mm (1/32–3/32 in.). The first (root) passes (4–6 passes) of each weld were fabricated using gas tungsten arc welding (GTAW) followed by subsequent passes (3–4 passes) utilizing the shielded metal arc welding (SMAW) process to complete the weld. The filler metal rods and covered electrodes were 3.2 mm (1/8 in.) in diameter. The welding parameters are given in Table 3.

High-Temperature Exposure

Upon completion of welding, welded samples were extracted for characterization of weld metal and heat-affected zone (HAZ) microstructures, impact toughness and high-temperature tensile testing, and stress-relief cracking (SRC) testing. Part of the samples were subjected to shorter and longer time heat-treatment (HT) procedures; others remained in the as-welded condition. The heat-treatment temperatures of 1650˚F (900˚C) and 1300˚F (705˚C) corresponded with stabilization heat treatment and high-temperature service conditions for 347H and 304H structural welds, respectively. The holding time was 4 and 168 h with subsequent air cooling. Table 4 presents an overview of the applied heat-treatment (aging) procedures.

Metallurgical Characterization

The welds were sectioned in as-welded and heat-treated conditions and prepared for light optical microscopy, scanning electron microscopy, and energy dispersive spectroscopy using standard metallography procedures. For observation of the weld metal microstructure, electrolytic etching was performed with a 10% aqueous oxalic (C₂H₂O₄) solution at 5 V for 10–20 s. To reveal the presence of sigma phase in the weld metal, a solution of 45-g KOH and 60-mL water was used for electrolytic etching at 2.5 V for 6-s. This etchant turns sigma and chi phase into a yellow color, ferrite appears gray to blue-gray, carbides are barely etched, and austenite is not etched. Selected welds were dipped in a solution of 6-g CuCl₂, 60-mL HCl, and 66 mL of distilled water (Kane’s etch, diluted with an equal volume of water). The samples were immersed for 10 s. Kane’s etch darkens martensite, outlines ferrite, and leaves austenite unetched (Ref. 11).

Ferrite Number Measurement

All welds were analyzed for their ferrite content in as-welded and heat-treated conditions. Magnetic measurements of FN were performed using a Magne-Gage (tear-off force) calibrated according to AWS A4.2M:2006, Standard Procedures for Calibrating Magnetic Instruments to Measure the Delta Ferrite Content of Austenitic and Duplex Ferritic-Austenitic Stainless Steel Weld Metal. The Fischer FeritScope™, which measures...
the ferrite content based on magnetic permeability, was also utilized. Measurements were taken on cross sections in the weld metal spaced from weld root to weld cap to obtain a total of five measurements per weld. Note: This is a departure from AWS A4.2M:2006, which specifies measurement on the smoothed top weld surface of a weld.

**Impact Toughness Testing**

For evaluation of weld metal toughness, a total number of 36 samples in the as-welded condition were machined transverse to the weld direction with dimensions of 11 x 11 x 56 mm. Part of the samples were subjected to heat-treatment procedures, as previously described (Table 4). All samples were then milled into Charpy V-notch size test bars (10 x 10 x 55 mm) with a 2-mm notch (45-deg angle) in accordance with ASTM E23, *Standard Test Methods for Notched Bar Impact Testing of Metallic Materials*. The notch was...
placed at the weld centerline parallel to the surface. Testing was performed at a temperature of 0˚F (–18˚C). For all welds, a total number of three samples were tested in the as-welded condition and three samples for each heat-treatment condition (4 and 168 h).

### High-Temperature Tensile Test

Limited high-temperature tensile tests were performed, testing one sample transverse to the welding direction for each weld. Samples were heated to the heat-treatment temperature (1650˚ and 1300˚F, respectively) at 200 K/h and pulled to failure at a stroke rate of approximately 0.008 mm/s. The results are shown in Table 5. The high-temperature yield strength and tensile strength were used as input parameters for the SRC tests.

### Stress-Relief Cracking Test

Evaluation for SRC susceptibility at heat-treatment temperatures (1650˚ and 1300˚F) was performed utilizing a Gleeble®-based SRC test developed at The Ohio State University (OSU) (Refs. 16, 17). This test was designed to replicate postweld heat treatment (PWHT) in highly restrained welds with high levels of residual welding stress. An example of the SRC testing procedure, as performed in this work, is shown in Fig. 2.

Test samples were extracted transverse to the welding direction with the weld located in the center of the gauge section, as shown in Fig. 3. A minimum of two samples were tested for each base metal/filler metal combination. The high-temperature mechanical properties used for development of the SRC testing procedure are listed in Table 5.

### Results and Discussion

#### Microstructural Characterization in the As-Welded Condition

The modified WRC-1992 diagram (Ref. 11) shown in Fig.
was used to predict the weld metal microstructures and ferrite levels in the as-welded condition. Note that both Type 16-8-2 filler metals extend beyond the iso-ferrite lines of the diagram. However, the prediction of microstructures using the WRC-1992 diagram has been reported to remain reasonably accurate as long as the actual weld metal compositions, considering dilution effects, fall within the original iso-ferrite bounds (Ref. 3). The dilution of the welds was assumed to be approximately 30%. It was noted that the final weld passes were performed with the SMAW process. The predicted solidification mode and FN for each weld are presented in Table 6. All weld metal compositions are predicted to fall within the austenite + ferrite region with primary ferrite solidification (FA), thus greatly reducing the susceptibility to weld solidification cracking (Ref. 3). However, the 16-8-2 filler metals, especially the lean composition of the SMAW electrode (E16-8-2), clearly extend the predictions beyond the limits of the iso-ferrite lines in the WRC-1992 diagram. This leanest composition touches the lower end of the martensite boundary at 1% Mn (shaded region in Fig. 4). However, given the associated base metals in this study, the formation of martensite in the welds is unlikely. The dilution with the base metal will shift the final weld metal composition up and to the right, farther away from the martensite boundary. More than 70% dilution would be required before the as-welded martensite would be present in the actual Type 16-8-2 weld metal microstructure.

Table 6 shows results for predicted and measured FN. The two measurement techniques used align reasonably well and were in good correlation to the prediction of the modified WRC-1992 diagram. The 347H weld with matching filler metal was found to have the highest ferrite level of all welds, although slightly lower than what was predicted. The use of Type 16-8-2 filler metals resulted for both base materials in much lower ferrite in the final weld metal microstructure as compared to their respective matching filler metal welds.

The as-welded microstructures were examined light optically and compared in terms of ferrite content and morphology. In general, the micrographs in Fig. 5 are in good qualitative agreement with the measured ferrite levels in the four weld metals. Both Type 16-8-2 welds contain less ferrite than their respective matching filler metal welds, and the 347H weld with matching filler metal has the highest weld metal ferrite content. In general, the ferrite morphologies for all welds were found to be similar. The examples of the GTAW portion of the weld microstructures in Fig. 5 show a skeletal ferrite morphology that is indicative of the FA solidification mode (Ref. 3). In the higher ferrite matching filler metal welds, the ferrite is more interconnected than for the low-ferrite Type 16-8-2 welds. Evidence of a partial lath ferrite morphology was found only in the higher ferrite...
The weld metal ferrite level is shown in Fig. 6 for all welds as FN measured from weld root to weld cap using the Magne-Gage. Again, it can be seen that the Type 16-8-2 welds contain much less ferrite than their respective matching filler metal welds. A difference in FN between the GTA weld root and the SMA weld cap was apparent for the welds with matching filler metal. The higher ferrite level in the weld root is somewhat unexpected, at least for the 347H weld, since both Type 347 covered electrode and wire have very similar Creq/Nieq — Fig. 4. A higher base metal dilution in the root pass, and dissolution of ferrite resulting from thermal cycles during subsequent weld passes, would be expected to result in slightly lower ferrite in the GTA weld root. The lower heat input and associated higher cooling rates as compared to the SMAW process used for the fill and cap passes might be an explanation for the instead higher ferrite level.

Microstructural Stability During High-Temperature Exposure

The microstructural stability of the stainless steel weld metals was studied by monitoring the ferrite level remaining after high-temperature exposure for 4 and 168 h. The results of FN measurements using the Magne-Gage are presented in Fig. 7. Upon high-temperature exposure, the FN in the 347H and 304H welds with matching filler metals decreased with increasing holding time at 1650˚ and 1300˚F (900˚ and 705˚C), respectively. In both welds, the ferrite level decreased within 168 h to less than 50%.

The decrease in ferrite level is a result of two mechanisms: 1) The ferrite phase partially dissolves upon high-temperature exposure. Vitek and David (Ref. 18) reported that for aging of Type 308 weld metals in the range of 1200˚ to 1580˚F (650˚ to 850˚C), the initial M23C6 carbide formation along the austenite/ferrite interface is followed by the dissolution of the ferrite phase. The ferrite dissolution occurs relatively quickly. In their welds, the ferrite level decreased by nearly 40% within the first hour of aging. 2) With additional holding time, any further depletion in weld metal ferrite occurs by the transformation of ferrite into sigma phase. Nucleation is the rate-limiting step of the ferrite-to-sigma transformation. Once nucleation occurs, growth of sigma in the ferrite phase proceeds quite rapidly (Ref. 18). For Type 308 weld metals, sigma phase formed within less than 100 h in the temperature range of 1200˚ to 1580˚F (650˚ to 850˚C) (Refs. 18, 19).

In the present study, examination of etched weld metal microstructures after 168 h of high-temperature exposure revealed the occurrence of sigma phase in both matching filler metal welds. Figure 8C and D shows that a large number of ferrite grains (gray) transformed partially to sigma phase (yellow). Sigma phase formation seemed more pronounced in the Type 347 weld metal after 168 h — Fig. 8C. The faster ferrite-to-sigma transformation might be due, in part, to the absence of M23C6 carbides (Ref. 18). Niobium and titanium in 347 are more potent carbide formers than chromium and minimize M23C6 formation. In Type 308H weld metal, chromium-rich M23C6 carbides compete with the sigma for the chromium available in the ferrite phase so that less ferrite transforms to sigma phase (Ref. 18). M23C6 carbides cannot be observed by light optical microscopy — Fig. 8D. Electron microscopy is necessary to identify these precipitates at the austenite/ferrite interface. Vitek and David (Ref. 18) have shown for Type 308 weld metal that M23C6 carbides form an extensive network throughout the austenite matrix after aging at 1200˚ to 1580˚F (650˚ to 850˚C) within less than 1 h.
For the Type 16-8-2 weld metals, Fig. 7 shows that the weld on Type 347H base metal exhibits a decrease in measured FN with increasing holding time similar to the behavior observed for the matching filler metal welds. The ferrite level decreases to less than 1 FN after aging for 168 h at 1650˚F (900˚C). The micrographs confirm the partial dissolution of the ferrite phase (Fig. 8A) as compared to the as-welded condition. No evidence of sigma phase formation was found in the etched weld metal microstructure.

The opposite observation in terms of apparent ferrite level was made for the aged 16-8-2 filler metal weld on 304H base metal. Figure 7 shows an unexpected increase in the measured FN with increasing holding time at 1300˚F (705˚C). The measured magnetic response after aging for 168 h is more than four times higher as compared to the as-welded condition. This apparent increase in ferrite content, however, is not reflected in micrographs of the aged weld metal microstructure. Figure 8B shows a very low amount of ferrite phase throughout the matrix. Small, chain-shaped M_{23}C_{6} carbides formed along the austenite/ferrite interface showing the prior ferrite grain structure. Leitnaker (Ref. 6) showed that aging of Type 16-8-2 weld metal at 1200˚ and 1350˚F (649˚ and 732°C) results in M_{23}C_{6} carbide precipitation. No evidence of sigma phase formation was found in the etched weld metal microstructure, with the exception of small amounts of sigma phase in regions highly diluted with the 304H base metal on both sides of the GTAW root pass.

An increase in measured FN in Type 16-8-2 weld metal after high-temperature exposure at 1382˚F (750˚C) has previously been reported in a study by Marshall and Farrar (Ref. 20). This unexpected behavior was explained with the formation of martensite in the weld metal upon cooling to ambient temperature. During high-temperature exposure, the precipitation of carbides consumes chromium and carbon from the matrix, thus destabilizing a portion of the austenite phase. This compositional change raises the martensite start (M_{s}) temperature of the austenite, leading to some transformation to martensite during cooling (Ref. 21). The high measured FN in the aged weld metal was observed because both martensite and ferrite are ferromagnetic and trigger a magnetic response. Marshall and Farrar (Ref. 20) did not show further experimental evidence of martensite formation after high-temperature exposure in Type 16-8-2 weld metal. They noted, however, that this behavior is actually predicted by the WRC-1992 diagram. The removal of chromium and carbon from the austenite phase shifts the composition box of the aged Type 16-8-2 weld metal to the left and down toward the martensitic zone — Fig. 9A.

In the present study, a special etchant (Kane’s etch) was used to reveal the presence of martensite in the 16-8-2 filler metal weld on 304H base metal. Figure 9B shows patches of dark-etching martensite in an austenite matrix, along with numerous precipitates and little residual ferrite. X-ray diffraction data of this weld in the as-welded condition shows face-centered cubic peaks (111, 200, 220) for the austenitic weld metal — Fig. 10A. The as-welded amount of ferrite was too low (< 2 FN) to be observed in x-ray diffraction. After high-temperature exposure for 168 h, a significant body-centered cubic/body-centered tetragonal peak (110) correlates to the martensite in the weld metal microstructure, which triggered the high-measured FN (9 FN) in this weld metal.

The 16-8-2 filler metal weld on 347H base metal did not experience an increase in measured ferrite upon high-temperature exposure, thus not indicating any weld metal martensite after aging. The reason for this is the higher aging temperature of 1650˚F (900˚C) used for this weld. The phase equilibrium in Fig. 10B was calculated using Thermo-Calc®. The diagram shows that an aging temperature of 1650˚F (900˚C) is above the temperature range for M_{23}C_{6} carbide precipitation in the Type 16-8-2 composition. Hence, carbon and chromium are not removed from the matrix during high-temperature exposure, and the austenite phase remains stable and does not experience a martensite transformation during cooling to ambient temperature.
It should be noted that no published reports could be found on martensite in Type 16-8-2 weld metals after high-temperature service or postweld heat treatment for commercial welds. As Marshall and Farrar (Ref. 20) pointed out, this effect in 16-8-2 weld metal is likely to not impose a problem for high-temperature service applications, which are typically above 1004°F (540°C). The formed weld metal martensite will have a low A1 temperature so that it will easily reverse to austenite upon high-temperature service conditions or postweld heat treatment.

Effect of High-Temperature Exposure on Impact Toughness

The Charpy-impact properties of the stainless steel weld metals in the as-welded condition and after high-temperature exposure for 4 and 168 h are shown in Figs. 11 and 12. Both Type 16-8-2 filler metal welds showed high toughness and ductility in the as-welded condition. All welds experienced a reduction in toughness after high-temperature exposure. Impact toughness decreased in all welds with increasing holding time at 1650˚ and 1300˚F (900˚ and 705˚C), respectively. In both matching filler metal welds, sigma phase formation led to a drop in impact properties. This is more pronounced in the 308H filler metal weld, which might be due to the M23C6 carbides in this weld metal, providing an easy path for crack propagation when forming a continuous network along the austenite/ferrite interface.

For the welds on 347H base metal, the use of Type 16-8-2 filler metal significantly increased toughness as compared to the matching filler metal weld in the as-welded condition and after high-temperature exposure. For the welds on 304H base metal, the matching filler metal weld obtained better toughness than the Type 16-8-2 filler metal weld in the as-welded condition. However, with increasing holding time at 1300˚F (900˚C), the properties of the 16-8-2 weld dropped significantly less than the impact toughness of the matching filler metal weld, so that the use of 16-8-2 filler metal provides superior toughness when welding 304H stainless steel, especially at longer high-temperature exposure. The drop in impact properties of the 16-8-2 filler metal weld was due to the martensite formation upon cooling to ambient temperature coupled with carbide precipitation, as described earlier.

Stress-Relief Cracking Susceptibility at Elevated Temperatures

None of the tested Type 16-8-2 and matching filler metal welds on 347H and 304H base metal failed in stress-relief cracking mode under the applied SRC test conditions. However, the tested welds exhibited notable differences in terms of stress relaxation behavior and mechanical properties. Examples of stress and strain curves from SRC testing are shown in Fig. 13. Tables 7 and 8 summarize the test results in terms of room and test temperature preload ($\sigma_{RT}$, $\sigma_{HT}$, and $\sigma_{PW}$).
ε_HT), stress relaxation and strain accumulation at the end of the 8-h high-temperature hold (σ_PW and ε_PW), and high-temperature ultimate tensile strength (UTS) and elongation when pulled to failure after the 8-h hold (UTSSRC and ε_failure).

Preloading at the test temperature of 1650°F (900°C) resulted in slightly higher stress and strain values (σ_HT and ε_HT) for the 347H steel weld with matching Type 347 filler metal as compared to the Type 16-8-2 weld (see Table 7). This preloading with stress close to the high-temperature UTS was followed by a significant stress relaxation during the 8-h high-temperature hold under fixed displacement (from σ_HT to σ_PW) for both filler metal welds. The Type 347 weld experienced a larger stress reduction as compared to the Type 16-8-2 weld, 127 vs. 89 MPa on average, respectively. When pulled to failure after the 8-h high-temperature hold, the Type 347 filler metal weld displayed a lower UTSSRC as compared to the Type 16-8-2 weld. Both welds exhibited a significant reduction in tensile strength (UTSSRC), as compared to the high-temperature UTS HT that was obtained in the as-welded condition (see Table 5).

The 304H steel welds with both filler metals exhibited similar stress and strain values (σ_HT and ε_HT) when preloaded at the test temperature of 1300°F (705°C) (see Table 8). The matching Type 308H filler metal weld displayed a higher amount of stress relaxation during the 8-h high-temperature hold (from σ_HT to σ_PW) as compared to the Type 16-8-2 weld, 32 vs. 10 MPa on average, respectively. When pulled to failure after the 8-h high-temperature hold, both filler metal welds displayed a tensile strength (UTSSRC) that was equal to the high-temperature strength (UTS HT) obtained in the as-welded condition (see Table 5).

**Conclusion**

Type 16-8-2 filler metal was examined for application in structural welds on 304H and 347H stainless steels for high-temperature service applications and compared to welds with matching filler metals 308H and 347, respectively.
stability of the weld metal microstructure during elevated-temperature exposure, impact toughness of the welds, and susceptibility to stress-relief cracking were studied.

Based on these investigations, the following conclusions can be drawn:

1. Low ferrite levels (≤ 2 FN) were obtained in Type 16-8-2 filler metal welds on 347H and 304H stainless steels as compared to much higher ferrite (up to 9 FN) when using matching filler metals 347 and 308H. Skeletal (and partly lathy) ferrite morphology was observed in the stainless steel weld metals.

2. No evidence of weld solidification cracking or liquation cracking in the reheated weld metal was observed in the Type 16-8-2 filler metal welds.

3. All stainless steel weld metals experienced a reduction in toughness after high-temperature exposure. However, both Type 16-8-2 filler metal welds showed superior toughness as compared to the matching filler metals welds, especially after longer high-temperature exposure (168 h).

4. An unexpected increase in magnetic response (FN) for the aged Type 16-8-2 filler metal weld on 304H base metal was shown to be due to martensite formation upon cooling to am-

5. Sigma phase formation in aged Type 16-8-2 weld metals occurred only locally in highly diluted regions with the base metal (GTA root), while significant amount of sigma phase formed in the matching filler metal welds. High resistance to intermetallic phase formation of Type 16-8-2 weld metal is due to the low Cr + Mo level, controlled carbon, and low ferrite content.

6. None of the tested weld metals failed by stress-relief cracking during simulated PWHT at 1650°F or 1300°F for the 347H and 304H steel welds, respectively, tensile stress testing at 90% of the PWHT temperature yield strength or fixed displacement replicating high weld restraint. The Type 16-8-2 weld metals exhibited significantly lower levels of stress relief during PWHT and significantly higher UTS at the respective PWHT temperature compared to the 347 and 308H weld metals.

7. Filler metal 16-8-2 provides a viable alternative for high-temperature service welds on Type 347H and 304H stainless steels.

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References

1. Carpenter, O. R., and Wylie, R. D. 1965. 16-8-2 Cr-Ni-Mo for the welding electrode. Metal Progress 70(5): 65–73.

2. Hoke, J., Eberle, F., and Wylie, R. D. 1957. Embrittling tendencies of austenitic superheater materials at elevated temperatures. ASTM Proceedings, 821–832.

3. Lipold, J. C., and Kotecki, D. J. 2005. Welding Metallurgy and Weldability of Stainless Steels. Hoboken, N.J.: John Wiley.

4. Schulten, D., and Gerlach, H. 1971. Beitrag zum Schweifen hochwarmfester unstabilisierter austenitischer Stähle. Schweissen und Schneiden 23(5): 167–171.

5. Smith, J. J., and Farrar, R. A. 1993. Influence of microstructure and composition on mechanical properties of some AISI 300 series weld metals. International Materials Reviews 38(1): 25–51. DOI: 10.1179/imr.1993.38.1.25

6. Leitnaker, J. M. 1982. Prevention of chi and sigma phases formation in aged 1648-2 weld metal. Welding Journal 61(1): 9-s to 12-s.

7. Marshall, A. W., and Farrar, J. C. M. 1999. Type 304H austenitic stainless steel weld metals for high temperature service. Stainless Steel World ’99 Conference, 147–162.

8. King, J. F., Sullivan, M. D., and Slaughter, G. M. 1977. Development of an improved stainless steel to ferritic steel transition joint. Welding Journal 56(11): 354-s to 358-s.

9. Lohrmann, G. R., and Ohrt, E. F. 1982. Qualifizierung eines Schweißzusatzwerkstoffes des Typs CrNiMo 16 8 2 fuer unstabilisierte austenitische Stahle im Hochtemperatureinsatz. DVS-Berichte 75: 33–38.

10. Lundin, C. D., Delong, W. T., and Spond, D. F. 1975. Ferrite-fusing relationship in austenitic stainless steel weld metals. Welding Journal 54(8): 241-s to 246-s.

11. Kotecki, D. J. 1999. A martensite boundary on the WRC-1992 diagram. Welding Journal 78(5): 180-s to 192-s.

12. Siewert, T. A., McCowan, C. N., and Olson, D. L. 1988. Ferrite Number prediction to 100 FN in stainless steel weld metal. Welding Journal 67(12): 289-s to 298-s.

13. Siewert, T. A., McCowan, C. N., and Olson, D. L. 1989. Stainless steel weld metal: Prediction of ferrite content. WRC Bulletin 342. Welding Research Council: New York.

14. Boniszewski, T. 1978. The correct choice of 'stainless' electrode for high temperature (creep) applications. Trends in Steel and Consumables for Welding: An International Conference, 461–472.

15. Hau, J. L., and Seijas, A. 2006. Sigma phase embrittlement of stainless steel in FCC service. Corrosion, Paper No. 06578.

16. Norton, S. J., and Lippold, J. C. 2003. Development of a Gleeble-based test for postweld heat treatment cracking susceptibility. 6th International Trends in Welding Research Conference Proceedings, 609–614.

17. Strader, K., Alexandrov, B. T., and Lippold, J. C. 2016. Stress-relief cracking in simulated-coarse-grained heat affected zone of a creep-resistant steel. Cracking Phenomena in Welds IV. Eds. T. Boellinghaus, J. C. Lippold, and C. Cross. Springer, pp. 475–493. DOI: 10.1007/978-3-319-28434-7_21

18. Vitek, J. M., and David, S. A. 1984. The solidification and aging behavior of types 308 and 308CRE stainless steel welds. Welding Journal 63(8): 246-s to 253-s.

19. Vitek, J. M., and David, S. A. 1986. The sigma phase transformation in austenitic stainless steels. Welding Journal 65(4): 106-s to 111-s.

20. Marshall, A. W., and Farrar, J. C. M. 2001. Lean austenitic type 16.8.2 stainless steel weld metal. Stainless Steel World 2001 Conference, 216–222.

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