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

Studying the stability of the weld metal microstructure during elevated temperature exposure, impact toughness of the welds, and susceptibility to stress-relief cracking

BY C. FINK, H. WANG, B. T. ALEXANDROV, AND J. PENSO

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 provide 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-

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Fig. 1 — Modified WRC-1992 diagram indicating Type 16-8-2 composition range. Reproduced from Kotecki (Ref. 11).

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

Fig. 2 — Temperature and force profile during the OSU SRC test.

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,

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

Specification	C	Cr	Ni	Mo	Mn	Cu	Si	P	S
ER16-8-2 AWS A5.9	< 0.10	14.5–16.5	7.5–9.5	1.0–2.0	1.0–2.0	< 0.75	0.30–0.65	< 0.03	< 0.03
E16-8-2-15 AWS A5.4	< 0.10	14.5–16.5	7.5–9.5	1.0–2.0	0.5–2.5	< 0.75	< 0.60	< 0.03	< 0.03

Table 2 — Chemical Composition (wt-%) of Base Metals and Filler Metals

	C	Cr	Ni	Mo	Mn	N	Cu	Nb + Ta	Si	P	S
Base Metals											
304H	0.050	18.45	8.17	0.32	1.46	0.060	0.46	0.034	0.55	0.034	0.001
347H	0.040	17.35	9.08	0.38	1.63	0.040	0.42	0.730	0.40	0.036	0.001
Filler Metals											
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	19.40	9.40	0.01	1.81	0.016	0.11	0.570	0.33	0.019	0.004
E347	0.018	19.60	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

Fig. 3 — Stress-relief cracking test sample design with the weld location (red) in the gauge section.

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) complied 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.

Fig. 4 — Modified WRC-1992 diagram with location of 304H and 347H base metals, matching Type 308H and 347 filler metals, and lean composition Type 16-8-2 filler metals. Shaded region indicates martensite boundary at 1% Mn. Reproduced from Kotecki (Ref. 11).

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 ($C_2H_2O_4$) 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 $CuCl_2$, 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

Table 3 — Parameters for the GTAW and SMAW Process

Welding Process	Arc Current	Arc Voltage	Welding Speed	Calculated Heat Input
GTAW	95–155 A	11–12.5 V	0.72–1.35 mm/s	1.3–2.2 kJ/mm
SMAW	110–125 A	22.5–24 V	0.8–1.4 mm/s	1.7–3.4 kJ/mm

Fig. 5 — As-welded microstructures of gas tungsten arc weld metal: A — 347H/16-8-2; B — 304H/16-8-2; C — 347H/347; D — 304H/308H (base metal/filler metal).

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 × 11 × 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 × 10 × 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

Table 4 — Heat-Treatment (HT) Procedures

Base Metal	Filler Metal	HT 1 (4 h)	HT 2 (168 h)
347H	347	1650°F (900°C)	1650°F (900°C)
347H	16-8-2	1650°F (900°C)	1650°F (900°C)
304H	308H	1300°F (705°C)	1300°F (705°C)
304H	16-8-2	1300°F (705°C)	1300°F (705°C)

Table 5 — High-Temperature Tensile Test Results

Base Metal	Filler Metal	Test Temperature	Yield Strength (MPa)	Tensile Strength (MPa)
347H	347	1650°F (900°C)	110	140
347H	16-8-2	1650°F (900°C)	115	132
347H*	—	RT	205	515
304H	308H	1300°F (705°C)	153	255
304H	16-8-2	1300°F (705°C)	170	268
304H*	—	RT	205	515

*Specified minimum room temperature (RT) properties for the base metals taken from ASTM A240/A240M-19 are shown as reference.

Fig. 6 — Plots of FN as measured (Magne-Gage) in the weld toe (GTAW) and weld cap (SMAW).

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.

Fig. 7 — Effect of exposure at high temperature on FN (Magne-Gage) in stainless steel weld metals. Error bars represent minimum and maximum values obtained.

To simulate high-level welding residual stresses, samples were initially loaded with 90–95% of the 304H and 347H steel and specified minimum room temperature yield strength (YS_{RT}) (Table 5). After loading, the displacement was fixed to simulate high weld restraint. The samples were then heated to the heat-treatment temperature with a heating rate of 200 K/h. The heat-treatment temperature was 1300°F (705°C) and 1650°F (900°C) for the 304H and 347H weld samples, respectively. Due to the high thermal expansion coefficient of the tested materials, full stress relaxation occurred on heating to the heat-treatment temperature for all tested samples, as schematically shown in Fig. 2. Once the heat-treatment temperature was reached, the 347H and 304H stainless steel weld samples were reloaded with 90–95% of the high-temperature yield strength (YS_{HT}) (Table 5). After reloading, the samples were held at fixed displacement for 8 h. The level of stress relaxation and strain evolution were monitored during this stage of the SRC test. If no failure occurred during the 8-h, high-temperature hold, the samples were pulled to failure at a stroke rate of 0.008 mm/s and the ultimate tensile strength (UTS_{SRC}) and elongation ($\epsilon_{@failure}$) were determined. The strain over the uniformly heated gauge section was monitored using a strain gauge and recorded throughout the entire test duration.

Results and Discussion

Microstructural Characterization in the As-Welded Condition

The modified WRC-1992 diagram (Ref. 11) shown in Fig.

Table 6 — Solidification Mode and Ferrite Number (FN) as Predicted by the Modified WRC-1992 Diagram and Results of FN Measurements (average values reported)

Base metal	Filler Metal	Solidification Mode	Modified WRC-1992	Fischer FeritScope®	Magne-Gage
347H	347	FA	8	5.4	6.7
347H	16-8-2	FA	1.5	2.2	2.0
304H	308H	FA	4	2.4	4.5
304H	16-8-2	FA	2	2.2	1.7

Fig. 8 — Aged (168 h) weld metal microstructures: A — 347H/16-8-2; B — 304H/16-8-2; C — 347H/347; and D — 304H/308H (base metal/filler metal). Note sigma (yellow) in matching filler metal welds (C, D).

4 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 lathy ferrite morphology was found only in the higher ferrite

Fig. 9 — A — Modified WRC-1992 diagram indicating shifted composition range of aged 16-8-2 weld metal (carbon and chromium were removed from the matrix as $M_{23}C_6$ carbides), reproduced from Marshall and Farrar (Ref. 21); B — light optical micrograph showing martensite (dark regions) in aged Type 16-8-2 weld (304H base metal) after aging at 1300°F (705°C) for 168 h (Kane's etch).

347H weld with matching filler metal — Fig. 5C. The SMAW portion of all welds exhibits the same FA solidification mode with a skeletal ferrite morphology but contains more porosity. No evidence of solidification or liquation cracking in the base metal HAZ or in the reheated weld metal was found in any of the welds.

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 Cr_{eq}/Ni_{eq} — 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 $M_{23}C_6$ 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 $M_{23}C_6$ carbides (Ref. 18). Niobium and titanium in 347 are more potent carbide formers than chromium and minimize $M_{23}C_6$ formation. In Type 308H weld metal, chromium-rich $M_{23}C_6$ carbides compete with the sigma for the chromium available in the ferrite phase so that less ferrite transforms to sigma phase (Ref. 18). $M_{23}C_6$ 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 $M_{23}C_6$ 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.

Fig. 10 — A — X-ray diffraction data of Type 16-8-2 weld metal in as-welded and aged conditions; B — equilibrium property diagram from Thermo-Calc® for Type 16-8-2 composition.

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 Type $M_{23}C_6$ 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.

Fig. 11 — Effect of exposure at high temperature on Charpy-impact energy of stainless steel weld metals. Error bars represent minimum and maximum values obtained.

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 $M_{23}C_6$ carbides in this weld metal, providing an easy path for crack propagation when forming a continuous network along the austenite/ferrite interface.

Fig. 12 — Relationship between Charpy-impact energy and lateral expansion for stainless steel weld metals in as-welded and aged conditions at 0°F (–18°C).

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 (σ_{RT} , σ_{HT} , and

Table 7 — SRC Test Results on 347H Base Metal Welds with Matching and 16-8-2 Filler Metals

# Sample	Base Metal	Filler Metal	σ_{RT} (MPa)	σ_{HT} (MPa)	ϵ_{HT}	σ_{PW} (MPa)	ϵ_{PW}	UTS _{SRC} (MPa)	$\epsilon_{@failure}$
1	347H	347	205	149	0.030	12.1	0.130	22.0	0.196
2*	347H	347	186	145	0.019	27.9	0.086	49.4	0.539
3	347H	347	193	133	0.020	7.1	0.129	8.2	0.135
1	347H	16-8-2	171	124	0.022	30.3	0.052	53.9	0.261
2*	347H	16-8-2	193	128	0.022	46.2	0.045	84.4	0.251
3	347H	16-8-2	194	126	0.022	33.3	0.038	54.1	0.044

*Run at lower peak RT load (11 kN) and HT load (5.8 kN) to attempt to compensate for load overshooting

Fig. 13 — Examples of stress and strain curves from SRC testing of stainless steel welds: A — 347H/347; B — 347H/16-8-2; C — 304H/308H; D — 304H/16-8-2 (base metal/filler metal).

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 (UTS_{SRC} and $\epsilon_{@failure}$).

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 UTS_{SRC} as compared to the Type 16-8-2 weld. Both welds exhibited a significant reduction in tensile strength (UTS_{SRC}), 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 (UTS_{SRC}) 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. The

Table 8 — SRC Test Results on 304H Base Metal Welds with Matching and 16-8-2 Filler Metals

#Sample	Base metal	Filler metal	σ_{RT} (MPa)	σ_{HT} (MPa)	ϵ_{HT}	σ_{PW} (MPa)	ϵ_{PW}	UTS_{SRC} (MPa)	$\epsilon_{@failure}$
1	304H	308H	187	164	0.015	137	0.023	260	0.621
2	304H	308H	188	169	0.015	132	0.028	258	0.673
1	304H	16-8-2	189	172	0.016	169	0.019	298	0.817
2	304H	16-8-2	195	163	0.016	146	0.021	264	0.800

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 ambient temperature after high-temperature exposure at 1300°F (705°C). $M_{23}C_6$ carbide precipitation raises the M_s temperature of the austenite phase by removing carbon and chromium from the matrix. The transformed martensite will have a low A_{c1} temperature and will likely reverse to austenite under high-temperature service conditions.

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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CAROLIN FINK (fink.242@osu.edu), HUIMIN WANG, and BOIAN T. ALEXANDROV are with the Department of Materials Science and Engineering, The Ohio State University, Columbus, Ohio. WANG is also with the National Center for Materials Service Safety, University of Science and Technology Beijing, Beijing, China. JORGE PENSO is with Shell Global Solutions (US) Inc., Houston, Tex.