Creep of austenitic stainless steel welds

S L MANNAN* and M D MATHEW

Materials Development Division, Indira Gandhi Centre for Atomic Research, Kalpakkam 603 102, India

Abstract. Austenitic stainless steels (SS) find extensive application in power, petrochemical and nuclear industries in view of their excellent elevated temperature mechanical properties, corrosion resistance, formability and weldability. However, they are susceptible to hot cracking during fusion welding. To avoid this problem, chemical composition of the welding consumable is generally adjusted to promote primary ferrite mode of solidification and retain about 3 to 10% δ -ferrite in the as-welded condition. The duplex microstructure of the weld metal undergoes transformation to carbides and a variety of intermetallic phases during elevated temperature service and causes deterioration in the mechanical properties.

This paper presents a comprehensive review of the current understanding of the solidification microstructures, ageing processes and their influence on the creep behaviour of types 308 and 316 SS weld metals. The effects of varying chemical composition, δ -ferrite content, electrode coating and welding processes on creep strength and ductility are examined. Current trends in the design of welded components for creep application are also discussed.

Keywords. Creep; stainless steel; weld metal; δ -ferrite; hot cracking.

1. Introduction

Austenitic stainless steels (SS) form an important class of materials that are used extensively in the power, petrochemical and nuclear industries. These steels possess relatively good elevated temperature mechanical properties, corrosion resistance, formability and weldability. Among the various grades of austenitic stainless steels, types 304 and 316 SS are more commonly used; type 316 SS being preferred for higher temperature of service. Type 316 SS is also the most widely used material for critical structural components of fast breeder reactors (FBRs) operating under creep conditions (i.e. temperature up to 823 K and service life of ~ 30 years). Fabrication of components invariably involves fusion welding. Austenitic stainless steels are susceptible to hot cracking during welding. To minimize this problem, the amounts of elements stabilizing the austenite and the ferrite phases are suitably adjusted in the welding consumable so that the weld metal has a duplex microstructure consisting of about 3 to 10% δ -ferrite in the austenitic matrix. Delta-ferrite is unstable on elevated temperature exposure and transforms to carbides and several intermetallic phases.

Despite significant advances made in recent years in producing wrought steels of better quality in terms of low inclusion content, and close control of alloying and residual elements resulting in better and consistent mechanical properties, the mechanical behaviour of welds continues to be a cause of concern for the structural integrity of components. At high temperatures, welds are weak links and creep failures are known to occur in the weld metal. Welds are regions of geometric discontinuity, contain residual stresses and exhibit a gradient in microstructure and hence in the mechanical properties. Till 1987, creep design at high temperatures has been carried

*Author for correspondence

out using properties of the base metal due to lack of adequate information on the creep properties of welds. Nuclear codes now incorporate strength reduction factors based on properties of the weld metal (ASME Nuclear Code Case 1987). However, these design rules are still considered inadequate since they do not consider the actual behaviour of weldments. Understanding the creep deformation and fracture behaviour of base metal, weld metal and weld joints is essential for improving the reliability of design of welded components for high temperature service.

This paper presents a comprehensive review of the current understanding of the solidification behaviour, ageing processes and their influence on the creep properties of welds of types 304 (welded with type 308 SS) and 316 SS, the two most extensively used austenitic stainless steels. Effects of varying chemical composition, δ -ferrite content, electrode coating and welding processes on the creep strength and ductility are discussed. The need to develop complete understanding of the mechanical behaviour of welds for safe and economic design of components, and for realistic assessment of remanent life of components in service is highlighted.

2. Microstructure of welds

2.1 Solidification behaviour

Austenitic stainless steels are susceptible to hot cracking (i.e. interdendritic cracking in the weld area that occurs before the weld cools to room temperature) due to presence of impurity elements like phosphorous and sulphur which form low melting eutectic phases. Hot cracking can be avoided if the amounts of austenite and ferrite stabilizing elements in the welding consumable are adjusted such that the weld metal has a duplex microstructure consisting of about 3 to 10% δ -ferrite, and solidification takes place in the primary ferrite mode (Vitek and David 1987; Brooks and Thompson 1991; Smith and Farrar 1993). Table 1 gives the chemical composition of types 304 and 316 SS base metals and various types of electrodes used in welding these steels (i.e. 308, 316, 19Cr-12Ni-3Mo, 18Cr-12Ni-2Mo, 17Cr-8Ni-2Mo, 18Cr-11Ni and 16Cr-8Ni-2Mo).

The solidification behaviour of austenitic weld metals has been a subject of extensive investigations (Kujanpaa *et al* 1979; Lippold and Savage 1980). A comprehensive

Table 1. Chemical composition of some austenitic SS base metal and weld metal.

Grade	C	Si	Mn	Cr	Ni	Mo	P	S
Base metal								
304	0.08	1.00	2.00	18-20	8-12			
316	0.08	1.00	2.00	16-18	10-14	2-3		
Electrode								
308	0.08	0.90	0.5-2.5	18-21	9-11	0.75	0.04	0.03
316	0.08	0.90	0.5-2.5	17-20	11-14	2-3	0.04	0.03
19Cr-12Ni-3Mo	0.08	1.00	0.5-2.5	17-20	10-14	2-2.5	0.04	0.035
18Cr-12Ni-2Mo	0.045-0.065	0.4-0.8	1-2	18-19	11-13	1.8-2.2	0.03	0.025
17Cr-8Ni-2Mo	0.06-0.1	1.0	0.5-2.5	16.5-18.5	8-9.5	1.5-2.5	0.03	0.04
16Cr-8Ni-2Mo	0.1	0.6	0.5-2.5	14.5-16.5	7.5-9.5	1-2	0.03	0.03
18Cr-11Ni	0.03-0.07	0.75	2.0	17.5-19.5	9-11	0.05	0.025	0.015

Figure 1. Vertical section of the Fe-Cr-Ni ternary phase diagram at 70 wt% iron. The dashed line represents the approximate composition of type 308 stainless steel.

review on this subject has been carried out recently by Brooks and Thompson (1991). Initial solidification occurs as primary austenite or as primary δ -ferrite depending on the position of the alloy composition with respect to the phase diagram as shown in figure 1. Alloys with compositions which are located on the Ni-rich side of the eutectic trough solidify as primary austenite; compositions which are located on the chromium rich side solidify as primary δ -ferrite. Over a wide range of solidification conditions, the critical $Cr_{eq.}/Ni_{eq.}$ ratio that separates primary austenite formation from primary ferrite formation is ≈ 1.5 (figure 2) (Vitek and David 1993). The solidification mechanisms for these two are quite different. Welds having $Cr_{eq.}/Ni_{eq.}$ ratio less than 1.5 solidify primarily as austenite and the δ -ferrite if any is formed from the remaining melt at the dendrite boundaries (figures 3a, b). Welds having $Cr_{eq.}/Ni_{eq.}$ ratio higher than 1.5 solidify primarily as ferrite. Austenite nucleates between ferrite dendrites from liquid enriched in austenite stabilizing elements and grows into the ferrite. This solid state transformation to austenite is not complete at the cooling rates obtained during welding so that the as-welded structure contains some residual ferrite located on the dendrite axis (figures 3c,d) (Lundin *et al* 1975; Gill *et al* 1986). Type 308 SS solidifies as primary ferrite whereas type 316 SS has a composition that lies in the three-phase region ($\gamma + \delta + \text{liquid}$) so that the solidification mechanism is very sensitive to small compositional variations (Rodriguez and Gill 1988).

A number of reasons have been suggested for the beneficial effects of ferrite and primary ferrite mode of solidification in reducing hot cracking (Vitek and David 1987, 1993). At high temperatures, ferrite has better ductility than austenite. This facilitates relaxation of weld thermal stresses. The thermal expansion coefficient of ferrite is lower than that of austenite. Hence residual stress is lower. Solubility of phosphorus and sulphur is more in ferrite than in austenite. Hence less amount of these elements

Figure 2. A plot of solidification mode versus dendrite growth rate showing that over an extensive range of solidification conditions, the critical Cr_{eq}/Ni_{eq} separating primary ferrite solidification from primary austenite solidification is essentially independent of growth rate.

Figure 3. Schematic model summarizing solidification and transformation behaviour of fully austenitic and duplex microstructures.

segregate to the liquid solidifying last in primary ferrite solidification. Ferrite, which is distributed in the austenite matrix, provides greater interfacial area and serves as crack arrester. It is suggested that in primary ferrite mode of solidification, the residual elements sulphur and phosphorus are not partitioned to the liquid as much as in primary austenite mode so that the overall segregation of these elements is less, thereby reducing the tendency to form low melting eutectics. Primary ferrite solidification leaves ferrite along the axis of dendrites. As solidification proceeds, the austenite boundaries move and get incorporated in the δ -ferrite thereby impeding the propagation of solidification cracks. Thus a crack would propagate only a short distance along the γ/γ boundary before it is arrested by the irregular δ/γ boundary.

Figure 4. Morphology of δ -ferrite in the as-welded condition in a type 316 SS weld metal.

Figure 5. Time-temperature-precipitation diagram for type 316 base metal (wrought) and 19-12-3 weld metal (cast).

2.2 Ageing behaviour

Duplex microstructure containing δ -ferrite and austenite is unstable on ageing at elevated temperatures. Figure 4 shows vermicular morphology of δ -ferrite in the as-welded condition. At elevated temperatures, δ -ferrite undergoes transformation to carbides and various intermetallic phases. The single most important factor that influences the mechanical behaviour of the weld metal in comparison to the base metal is the ageing behaviour of δ -ferrite. Numerous studies have been reported on the microstructural stability of austenitic SS base and weld metals. Figure 5 shows a time-temperature precipitation diagram for type 316 SS wrought and weld metals (Horton *et al* 1982). Precipitation starts early, and also at a lower temperature in the weld metal compared to the base metal. A compilation of the various types of precipitates observed in austenitic SS weld metals is given in table 2 (Smith and Farrar 1993).

The most generally accepted mechanism for transformation of δ -ferrite during ageing is illustrated schematically in figure 6 (Thomas 1983). Transformation begins with the formation of $M_{23}C_6$ carbides, preferentially along the δ/τ interface.

Table 2. Precipitates observed on ageing type 316 SS weld metal.

Composition, wt. %							Ferrite content(%)	Temp(°C)	Aging time (max)h	Phases identified during aging
C	Cr	Ni	Mo	Si	Nb					
0.088	24.9	8.0	—	0.42	—	—	40	700	25	M ₂₃ C ₆ + σ
0.03	17.9	10.9	2.5	0.3	—	—	2-4	900	200	M ₂₃ C ₆
								600	—	M ₂₃ C ₆
								750	—	M ₂₃ C ₆ + σ
								800	—	M ₂₃ C ₆ + σ + χ
0.04	18.1	11.7	2.5	0.27	—	—	5	600	3000	M ₂₃ C ₆
								700	300	M ₂₃ C ₆ + R + σ + χ
								800	15	M ₂₃ C ₆ + σ + χ
								600	3000	M ₂₃ C ₆
0.06	17.6	8.9	1.75	0.32	—	—	4	700	300	M ₂₃ C ₆
								800	15	M ₂₃ C ₆ + σ + χ
0.04	17.2	10.8	2.3	0.44	—	—	6	850	—	M ₂₃ C ₆ + σ + χ
								625	1000	M ₂₃ C ₆ + σ + χ
0.07	18.0	8.9	1.9	0.35	—	—	6	850	6	M ₂₃ C ₆
								625	1000	M ₂₃ C ₆ + σ + χ
0.05	20.5	10.1	3.1	0.41	—	—	18	625	6	M ₂₃ C ₆ + σ + χ
								850	1000	M ₂₃ C ₆ + σ + χ
0.04	16.4	8.7	1.6	0.33	—	—	4.5	649	5000	M ₂₃ C ₆ + σ + χ
								732	10000	M ₂₃ C ₆ + χ
0.073	15.2	8.4	1.3	0.42	—	—	3.7	649	5000	M ₂₃ C ₆
								732	10000	M ₂₃ C ₆
0.086	16.9	9.3	2.1	0.63	—	—	11	600	22000	M ₂₃ C ₆ + η
								500-850	200	M ₂₃ C ₆ + NbC + σ + χ
0.029	18.7	12.4	2.7	0.91	0.6	—	5	600	1000	M ₂₃ C ₆ + χ
								500-850	1000	M ₂₃ C ₆ + σ
0.11	15.2	9.7	2.1	0.38	—	—	0	500-850	200	M ₂₃ C ₆ + σ
								500-850	1000	M ₂₃ C ₆ + σ
0.1	18.8	8.1	3.4	0.34	—	—	18	500-850	200	M ₂₃ C ₆ + σ
								500-850	1000	M ₂₃ C ₆ + σ
0.018	18.5	11.8	1.94	0.69	—	—	10	500-850	200	M ₂₃ C ₆ + σ
								500-850	1000	M ₂₃ C ₆ + σ
0.032	19.5	11.2	2.25	0.75	—	—	20	500-850	200	M ₂₃ C ₆ + σ
								500-850	1000	M ₂₃ C ₆ + σ
0.033	18.6	12.4	2.43	0.36	—	—	6	500-850	200	M ₂₃ C ₆ + σ
								500-850	1000	M ₂₃ C ₆ + σ
0.040	19.3	12.4	2.40	0.66	—	—	8	500-850	200	M ₂₃ C ₆ + σ
								500-850	1000	M ₂₃ C ₆ + σ
0.021	18.7	12.5	2.32	0.51	—	—	6	500-850	200	M ₂₃ C ₆ + σ
								500-850	1000	M ₂₃ C ₆ + σ
0.026	19.5	12.0	2.27	0.57	—	—	12	500-850	200	M ₂₃ C ₆ + σ
								500-850	1000	M ₂₃ C ₆ + σ
0.021	18.7	12.5	2.32	0.51	—	—	6	500-850	200	M ₂₃ C ₆ + σ
								500-850	1000	M ₂₃ C ₆ + σ

0.026	19.5	12.0	2.27	0.57	—	12	700	$M_{23}C_6 + \sigma$
0.076	17.2	8.4	1.5	0.29	—	4.6	700–850	$M_{23}C_6$
0.012	17.3	10.0	1.6	0.34	—	6	700–850	$M_{23}C_6$
0.015	19.1	13.8	2.59	0.31	—	6	500	$M_{23}C_6 + \sigma + \chi$
							600	$M_{23}C_6 + \sigma$
							700	$M_{23}C_6 + \sigma$
0.07	16.9	9.1	1.6	0.28	—	5	600	$M_{23}C_6 + \sigma$
0.07	17.6	8.8	1.7	0.32	—	8	700	$M_{23}C_6 + \chi$
							800	$M_{23}C_6 + \sigma + \chi$
							850	$M_{23}C_6 + \sigma + \chi$
							900	$M_{23}C_6 + \sigma$
0.015	16.6	11.5	1.29	0.39	0.21	2–5	500	$M_{23}C_6 + R + \eta$
							600	$M_{23}C_6 + R + NbC + \eta$
							700	$M_{23}C_6 + NbC + M_6C + \eta + \sigma + \chi$
0.053	20.2	9.4	—	0.46	—	8.5	550	$M_{23}C_6 + \sigma$
0.036	19.8	9.9	—	0.61	—	13	650	$M_{23}C_6 + \sigma$
							750	$M_{23}C_6 + \sigma$
							850	$M_{23}C_6 + \sigma$
0.080	17.4	8.5	1.74	0.29	—	4.7	600	$M_{23}C_6 + \sigma + \eta$
							700	$M_{23}C_6 + \sigma$
0.092	18.2	8.7	1.84	0.51	—	5.9	850	$M_{23}C_6 + \sigma$
							600	$M_{23}C_6 + \sigma$
							700	$M_{23}C_6 + \sigma + \eta$
0.077	18.6	9.7	1.80	0.44	—	11.4	800	$M_{23}C_6 + \sigma$
							600	$M_{23}C_6 + \sigma$
							700	$M_{23}C_6 + \sigma$
0.079	16.4	8.4	1.80	0.68	—	9.8	850	$M_{23}C_6 + \sigma + \chi$
							600	$M_{23}C_6 + \sigma + \eta$
							700	$M_{23}C_6 + \sigma$
0.050	18.5	11.4	2.17	0.24	—	5.3	850	$M_{23}C_6 + \sigma$
							600	$M_{23}C_6 + \sigma$
							700	$M_{23}C_6 + \sigma$
0.075	17.8	8.6	1.10	0.41	—	4.2	850	$M_{23}C_6 + \sigma$
							700	$M_{23}C_6 + \sigma$
							850	$M_{23}C_6 + \sigma$

Nucleation and growth of carbides results in depletion of the ferrite stabilizing elements in δ -ferrite and favours its transformation to austenite in regions adjacent to the carbides. During ageing, the δ/γ interface advances into the ferrite leaving $M_{23}C_6$ particles isolated in the austenite as shown in figure 6. The remaining region of δ -ferrite continues to get enriched in ferrite stabilizing elements, and attains the chemical

Figure 6. Precipitation sequence at δ -ferrite.

Figure 7. a, b.

Figure 7. Microstructure of type 316 weld metal (a), as-welded (b), (c) and (d) after creep testing at 923 K for 200 h, 1100 h and 4300 h respectively.

composition of σ phase at which it undergoes transformation to σ phase. The σ particles grow and agglomerate at the expense of $M_{23}C_6$. Figure 7 shows the various stages of transformation of δ -ferrite into carbides and σ phase as observed in type 316 SS weld metal. Agglomerated σ particles were observed after exposure to 923 K for 6300 h while carbides which were seen during the early stages of ageing were absent (Mathew *et al* 1993). Transformation to intermetallic phase is reported to be more rapid in 19Cr–12Ni–3Mo than in the 17Cr–8Ni–2Mo deposit, which is relatively leaner in ferrite stabilizing elements (Horton *et al* 1982).

An alternate mechanism for transformation of δ -ferrite has been suggested by Vitek and David (1986) on the basis of extensive studies carried out on types 308 and 308 CRE SS welds. They have disputed the importance of compositional factors in controlling the transformation and have suggested that it is essentially a nucleation controlled process. Once nucleation of σ phase begins, its growth is very fast and it is for this reason that partially transformed ferrite grains are never observed.

Figure 8. Variation of weight of extracted precipitates as a function of creep time.

Compared to type 316 SS, welds of type 308 SS are used at lower temperatures at which the kinetics of transformation is slow. Vitek and David (1987) studied the ageing behaviour of types 308 and 308 CRE SS in the as-welded and homogenized conditions over a range of temperatures (748–1123 K) and time durations (up to 10000 h). Above 823 K, precipitation of carbides as well as transformation of ferrite to σ phase (when ferrite was present), or formation of ferrite (in ferrite-free material) was observed in type 308 SS welds. Type 308 CRE welds precipitated titanium rich carbides and nitrides apart from σ phase. Below 823 K, precipitation of G-phase within the ferrite, as well as spinodal decomposition of ferrite into α (iron rich) and α' (chromium rich) phases took place in type 308 SS welds. G-phase (a silicon rich phase based on the stoichiometric formula $\text{Ti}_6\text{Ni}_{16}\text{Si}_7$) was observed above 823 K for short duration ageing in type 308 CRE weld metal. The spinodal decomposition is similar to the α - α' embrittling transformation observed in ferritic steels.

Besides time and temperature of aging, transformation behaviour of the weld metal is closely related to microcompositional factors. Carbon has a strong influence on precipitation of intermetallic phases. This is due to the formation of carbides, which takes away Cr and Mo, and, thus, suppresses/reduces formation of intermetallic phases. Hence, low carbon variants of 304 and 316 SS are more susceptible to

Figure 9. Weight of carbides extracted as a function of creep time.

formation of intermetallic σ phase. Deposits containing molybdenum have the greatest tendency to form σ phase, followed by silicon, tungsten, vanadium and niobium (Wegrzyn and Klimpel 1981; Smith and Farrar 1993).

During creep, ageing takes place under stress. The influence of stress on the precipitation kinetics of the weld metal has been analysed using the gauge section and the shoulder region of creep specimens tested at 823, 873 and 923 K (Mathew *et al* 1994). Detailed microstructural investigation consisting of optical metallography and X-ray diffraction analysis of precipitates extracted from the gauge section and the shoulder region was carried out. The amount of precipitates was appreciably higher in the gauge section than in the shoulder region at 873 and 923 K (figures 8b and c) whereas it was comparable at 823 K (figure 8a). Precipitation of carbides was found under all the test conditions and the amount of carbides was independent of temperature, time of exposure and presence of stress. Figure 9 shows a plot of the weight of extracted carbide after exposure to 823, 873 and 923 K for durations up to 10,000 h. Precipitation of σ phase was strongly influenced by the presence of stress.

Another factor which would influence creep behaviour of weldments is the presence of residual stresses. In thick section components, these stresses may approach the yield stress due to low thermal conductivity and high coefficient of thermal expansion of these steels and may extend up to 50 mm from the fusion line into the base metal (Fidler 1979). Post-weld heat treatments are not generally specified for austenitic stainless steel components for relieving these stresses as it may cause distortion of the component and precipitation of undesirable phases. The residual stresses would relax under creep conditions during service. Due to different strengths in various regions of weldment, a nonuniform stress field would develop with time due to creep. Further, these stresses may be enhanced by the weld geometry and the loading conditions and influence creep behaviour.

3. High temperature design of welds

Although there is a growing recognition of the fact that service performance of a welded component is largely dependent upon the properties of the weld metal, rules for the design of welded components were based on the properties of the base metal till 1987. The nuclear design codes have been now modified taking into account the creep

Figure 10. Expected minimum creep rupture strength curves (ASME) for type 316 SS base metal and for weld metal obtained by applying strength reduction factors.

rupture strength of the weld metal (ASME Nuclear Code Case 1987). For the base metal, the time and temperature dependent allowable stress, S_t , is less than or equal to $2/3 S_r$, where S_r is the minimum creep rupture strength of the base metal. At weldments, the allowable stress, S_t , is specified to be the lower of the tabulated values for the base metal or $4/5 S_r \times R$, where R is the appropriate ratio of creep rupture strength of the weld metal to creep rupture strength of the base metal. The quantity $S_r \times R$ is essentially the minimum creep rupture strength of the weld metal. Thus, the intent of the code is to substitute the weld metal creep rupture strength for the base metal rupture strength. A lower safety factor of 1.25 is employed for the weld metal compared to a value of 1.5 for the base metal since use of a safety factor of 1.5 already includes some allowance for reduced strength of welds and further reduction in S_t is intended only if the reduction in the strength of the weld metal is more pronounced (Corum 1989, 1990). Figure 10 shows the expected minimum creep rupture strength curves for type 316SS base and weld metals according to the nuclear code case after applying the strength reduction factors to the stress rupture life for the base metal. The difference in the rupture strengths increases with increase in life.

The tendency for low ductility rupture in type 316SS weld metal has led to design penalties on the allowable deformation. The ASME code case recommends that welds are not used in high strain regions and that when inelastic strains are significant, the strain in weldments should be limited to 50% of that permitted for the parent material.

It must however be pointed out that the integrity of a welded structure depends on the performance of the complete weldment, not only of the weld metal. A more realistic approach involves experimentally characterizing the creep properties of the constituent regions of the weldment, followed by numerical simulation, to give the overall behaviour of the welded structure. A procedure suggested by Shan-Tung Tu and Rolf

Figure 11. Procedure for the characterization of weldments at elevated temperatures.

Sandstorm (1994) is shown in figure 11. The weld metal is not of the same composition as the parent metal. Moreover, when components are welded, the thermal effects can significantly alter the metallurgical structure of the parent metal to produce heat affected zone. The different parts of the weldment thus can have different constitutive behaviours. During elevated temperature service, this mismatch in the material properties will lead to stress enhancement in certain parts of the weldment which reduces creep life. So, the ideal way to test weldments is to test welded components, although it will be highly expensive and time consuming. Numerical simulation is a solution to account for the stress multiaxiality.

4. Creep properties

There have been several studies on the creep rupture properties of austenitic weld metals (Wood 1983; Thomas *et al* 1984; Vitek and David 1986; Senior 1990; Mathew *et al* 1993, 1994). Mathew *et al* (1993) have carried out long term creep tests on type 316 SS base metal and weld metal in the temperature range 823–923 K. As shown in figure 12 weld metal showed higher minimum creep rates at 823 and 873 K, whereas at 923 K the creep rates were equal. This difference in creep rates was attributed to the absence of intragranular carbide strengthening in the weld metal at lower temperatures unlike in the case of the base metal and also to the presence of δ -ferrite.

Figures 13a–c compare the rupture life of the weld and base metals at 823, 873 and 923 K respectively. Rupture life of the weld metal was always lower than that of the base metal; the factor of reduction varied between five to ten depending upon the test conditions. It is seen that rupture strength of the weld metal is lower than the minimum stress to rupture specified for the base metal by the ASME nuclear code. However, the observed rupture strength was comparable or even better when the strength reduction factors for the weld metal introduced in the codes are taken into account. The marked

Figure 12. Variation of minimum creep rate with applied stress.

difference in the strength of the base and the weld metal has been attributed to the following: (i) the base metal has an equiaxed grain structure whereas the weld metal has a columnar grain structure and (ii) the base metal is strengthened by the precipitation of carbides on matrix dislocations, especially at 823 and 873 K. In the weld metal, precipitation of carbides is mostly confined to δ/γ interphase boundaries. The matrix is relatively free of precipitates. At 823 and 873 K, transformation of ferrite to brittle σ phase takes place, which is detrimental to the strength of the weld metal.

Figure 14a shows the variation of rupture elongation with applied stress for the base and weld metals at 923 K. The weld metal ductility was considerably low when compared to that of the base metal. The difference in ductility increased with decrease in stress level. At 823 and 873 K, ductility of the weld metal was more than that of the base metal at high stresses but less at low stress levels (figures 14b and c).

Weld metal specimens tested at 923 K showed extensive creep cavitation associated with coarse precipitates of intermetallic phase σ (figure 15a). The growth and linkage of these cavities gave rise to low ductility failure of the specimen (figure 15b). At 823 and 873 K, transformation of δ -ferrite was partial. The cracks which formed on the high

Figure 13. Variation of rupture life with applied stress and comparison with the ASME design curve.

Figure 14. Variation of rupture elongation with applied stress for the base and weld metals.

angle grain boundaries were impeded by the presence of δ -ferrite network, and, hence larger strains are accumulated before the cracks grew to a critical size causing failure. Moreover, stress concentrations that develop at the grain boundary triple points due to sliding are expected to be relaxed to a greater extent in the welds since the matrix is relatively weak, being devoid of carbide precipitates at dislocations.

There are several factors which influence the creep properties of austenitic SS welds. These include chemical composition, type of electrode coating and the welding process. The role of each of these factors is discussed briefly in the following sections.

4.1 Chemical composition

A compilation of the published data from different countries on rupture strength of several molybdenum free (18Cr-11Ni) and molybdenum containing (18Cr-12Ni-2Mo, 17Cr-8Ni-2Mo etc) welds by Wood (1983) has established the superior rupture strength of molybdenum containing welds by as much as 50%. Titanium, boron and

Figure 15. (a) Creep damage associated with σ particles at 923 K, and (b) brittle fracture surface of the specimen in (a).

phosphorous are found to improve the strength and ductility of type 308 SS weld metal (Klueh and Edmonds 1986; Vitek and David 1987) (figure 16). This has led to the development of controlled residual element (CRE) electrodes that yield weld metal deposits containing 0.5% Ti, 0.04% P and 0.06% B. The role of residual elements on the creep properties has been the subject of several studies. Comparable quantities of δ -ferrite with similar morphologies were observed in 308 and 308 CRE weld metals. The principal effect of titanium was to change the type of precipitates formed and also the distribution within the duplex structure. Rather than forming chromium-rich $M_{23}C_6$ carbides along the δ/τ interface, Ti-rich carbides, nitrides and sulphides formed readily throughout the matrix. Addition of titanium did not retard σ formation. Creep tested 308 CRE weld specimens did not contain internal cracks in the vicinity of σ phase/austenite boundaries. But such cracks were common in type 308 SS. The elimination of the continuous network of $M_{23}C_6$ carbides (present in high ferrite content 308 weld) is mainly responsible for the improved elevated temperature creep

Figure 16. Creep rupture properties of GTA welds of large commercial heats of 308 CRE (data points) tested at 923 K and compared to properties of types 304 and 308 GTA welds.

Figure 17. Rupture life as a function of applied stress for type 316 SS with and without nitrogen.

rupture properties of the CRE weld metal. Boron and phosphorous promote formation of finer $M_{23}C_6$ precipitates. Precipitation on dislocations during creep is enhanced thereby improving the strength, and continuous grain boundary precipitation is eliminated thereby improving the ductility.

Studies on the effect of nitrogen on the creep properties of type 316 SS weld metal (Sasikala *et al* 1994) has shown that addition of 0.06 wt% nitrogen improves the strength at 923 K appreciably as shown in figure 17. Rupture ductility was found to be lowered by the presence of nitrogen (figure 18). Improvement in the creep strength of type 316 SS weld metal by addition of carbon and nitrogen has been reported by Lefebvre *et al* (1985) in the temperature range 823–973 K (figure 19).

4.2 Delta-ferrite content

The amount of δ -ferrite in the weld metal is dependent on the chemical composition of consumables used and welding process employed. Its effect on the creep properties

Figure 18. Variation of rupture elongation with rupture life for type 316 SS with and without nitrogen.

Figure 19. Effect of total carbon and nitrogen content on 10,000 h creep rupture strength of type 316 SS weld metals.

depend on the test temperature. Berggren *et al* (1978) studied the effect of δ -ferrite content in the range 0 to 15% on the creep properties of 308 SS submerged arc welds. Increasing the δ -ferrite content resulted in an order of magnitude decrease in the minimum creep rate. Thomas (1978) reported that the optimum creep properties are obtained at a δ -ferrite content of ≈ 5 FN. Fully austenitic weld metals are inherently creep-brittle due to their large columnar grain size. Crack nucleation occurs at grain boundary triple points. Due to the largely uninterrupted crack path, crack propagation is easy and requires little strain accumulation to cause failure. In the material containing δ -ferrite, the failure mode is cavitation at γ/γ and δ/γ boundaries. δ -Ferrite when present in a discontinuous form can impede crack propagation leading to better

Figure 20. Comparison of 10,000 h rupture life of 316 SS weld metals prepared by employing different welding processes.

strength and ductility. However, when the δ -ferrite content is high and is present as a continuous network, larger cracks can form more readily. It is argued that there is no link between low ductility failures and the presence of a certain phase. The salient controlling features are amount and morphology of the precipitates on phase boundaries, their effect on interface properties and the relative strength of the matrix.

4.3 Welding processes

There are various fusion welding processes, the choice of which is based upon the size and shape of the component to be welded. The most popular among them are manual metal arc (MMA), gas metal arc (GMA) and submerged arc (SA) welding. There are significant differences between them which influence weld metal chemistry as well as the metallurgical structures of the weld metal and the heat affected zone. The heat input and thermal cycling experienced by the weldment are different for these processes resulting in different metallurgical structures. MMA and GMA welding are relatively low heat input processes whereas SA welding is a high heat input process. Ingress of oxygen and nitrogen during welding and flux coating in the case of MMA welding electrodes result in compositional variations between welds produced by different processes. Dilution of the filler metal by the base metal to different extents for different processes is another factor which significantly influences the behaviour of the weldments. Typical dilutions are 20–40% for GMA, 30–50% for MMA and about 60% for SA welding. Presence of impurity elements in different levels due to different processes lead to differences in grain boundary behaviour.

From the above discussion it is clear that the welding process employed can have a significant effect on the creep behaviour of the weldment. However, systematic investigations of this topic have not been undertaken in any great detail. Limited studies by Van der Schaaf (1990) on SA, MMA, GMA and electron beam (EB) welded type 316 SS have shown that the 10,000 h creep strength is the highest for GMA weld and the lowest for SA weld (figure 20). Also Smith and Farrar (1993) concluded that the SA weldments are significantly weaker than MMA weldments, from a compilation of short term stress rupture data available in the literature. The differences in the

properties may be due to changes in the dislocation density and substructure, grain size and/or δ ferrite content and morphology. From the earlier discussion it is obvious that these can arise from different heat inputs and chemistry variations resulting from the use of different electrodes (with and without coating) or different extents of dilution by the base metal.

4.4 Electrode coating

It was mentioned in the previous section that in MMA welding the electrodes are always coated with an extruded flux. The functions of these coatings are to stabilize the arc, to flux away any impurities, to form protective gas shields and slag over the weld and to add any metallic element needed to adjust the final composition to that required. Various types of coatings like acidic, basic and rutile are popularly used depending upon the application. These coatings contain different proportions of lime (CaCO_3), fluorspar (CaF_2), silica (SiO_2) and titania (TiO_2). Generally, two types of electrode coatings, namely, basic and acid types, are used. Basic coatings have higher proportions of CaCO_3 and CaF_2 while the acidic coatings have higher proportions of SiO_2 and TiO_2 . Relative quantities of different oxides influence the arc characteristics, slag viscosity, shape of the solidifying weld pool as well as composition of the weld metal. Chemical nature of the flux plays an important role in determining weld metal composition. Therefore, the welding consumable is designed to achieve the required composition by suitable alloying additions to compensate for flux chemistry. Highly basic fluxes have a low oxygen activity corresponding to lower supersaturation of oxygen in the weld metal, since the basic oxides are not easily dissociated. Silicon activity in basic fluxes is also higher compared to acidic or rutile fluxes. But they produce more fluid slag and are therefore less tolerant to all positional welding. They are also more susceptible to moisture pick-up and difficult to deslag. However, they produce cleaner weld deposits with lower inclusion content. Hence better toughness is achieved in weld metal obtained using basic coated electrodes (Kenneth 1992; Smith and Farrar 1993). The more acid fluxes tend to be less susceptible to moisture pick-up, produce higher oxygen and silicon contents and form glassy slags. Rutile fluxes tend to improve the arc characteristics.

Variations in chemical composition with respect to alloying elements, residual elements, impurities and inclusion content resulting from the use of different electrode coatings are expected to produce variations in the creep properties. Binkley *et al* (1973) studied the effect of electrode coating on the short-term creep rupture properties of type 308 SS weld metal at 923 K. They reported that the rupture strengths for rutile deposits were higher by a factor of 2 to 3 compared with those for basic deposits. They did not investigate the cause of improvement in rupture strength. However, from a later study (Binkley *et al* 1974) on the effect of controlled additions of residual elements, it was implied that the improvement in rupture properties may be related to differences in residual element contents due to different electrode coatings. Marshall and Farrar (1985) also reported that the rutile coated Nb-doped deposits had superior rupture strength compared with the basic coated V-doped deposits. The highest rupture strengths were observed in acid rutile deposits, which also exhibited lower ductilities. They correlated the higher rupture life with the residual Nb content as well as Si content, which resulted from different basicities of the coatings. More recently,

Figure 21. Stress rupture results for basic (A) and rutile (B) weld deposits.

Senior (1986, 1988, 1990) suggested that the improved rupture strength of rutile deposits compared to the basic ones is related to the difference in inclusion and silicon contents. In this study, two deposits of 17-8-2 weld metal with basic and rutile coatings were examined. Figure 21 shows the stress rupture results obtained. Cavitation due to decohesion at σ/γ , carbide/ γ and inclusion/ γ interfaces as well as cracking of spinel type inclusions was observed. Since the inclusion density in weld A (basic) in figure 21 was found to be almost twice as much as that in weld B (rutile), the author concluded that cavitation at inclusions and easier coalescence would have resulted in the reduced rupture life of weld A. He also considered the higher silicon content of weld B which gives substantial solid solution strengthening to be responsible for the better rupture life. However, the higher inclusion content in the basic deposit is unexpected because basic fluxes are known to have a low oxygen content and hence low inclusion content. This discrepancy probably arises from limited studies on only two weld metals.

5. Concluding remarks

Creep failure of austenitic stainless steel weldments takes place in the weld metal. The duplex microstructure of the weld metal and the transformation of δ -ferrite to brittle intermetallic phases like σ during elevated temperature ageing have a strong influence on the creep behaviour of the weld metal. Even minor changes in the chemical composition and δ -ferrite content from one type of electrode to another affect the transformation behaviour. For a given type of electrode, the nature of electrode coating and the welding process employed also lead to variations in creep strength and ductility. Performance of welded components during service depends on the response of the weldments to elevated temperature loading. Detailed understanding of the creep response of weldments is required in order to achieve safe design and realistic life assessment of high temperature components.

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