PROPERTIES OF THICK TYPE 308 STAINLESS STEEL WELDS WITH CONTROLLED RESIDUAL ELEMENTS

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~~APPLIED TECHNOLOGY~~

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WITH CONTROLLED RESIDUAL ELEMENTS

R. T. King, J. O. Stiegler, and G. M. Goodwin

JUNE 1974

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88

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CONTENTS

	<u>Page</u>
Abstract	1
Introduction	1
Experimental Procedures	2
Mechanical Properties and Microstructural Evaluation of Weld Metal .	5
Structure and Property Variations Throughout the Weldment	9
Anisotropy of Weld Metal Deformation	12
Material Property Comparisons	13
Discussion	16
Summary	17
Acknowledgments	18
References	19

PROPERTIES OF THICK TYPE 308 STAINLESS STEEL WELDS WITH CONTROLLED RESIDUAL ELEMENTS

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ABSTRACT

The tensile and creep properties of Type 308 stainless steel shielded metal-arc welds with controlled B, P, and Ti contents have been investigated up to 1100°F. The creep ductility of these welds was generally greater than that previously reported for conventional Type 308 weld metal, while strength properties were comparable or greater than those of conventional weld metal. Systematic microstructural and mechanical property variations were found through the thickness of the welds, which may be related to each other by existing theories. Related property and microstructural variations also occurred in the base metal near the weld. Anisotropic deformation occurred in the weld metal.

INTRODUCTION

Until recently, nuclear reactor primary containment systems have been designed to Sect. III of the ASME Boiler and Pressure Vessel Code.¹ The code restricts the use of austenitic and ferritic steels to temperatures below 800 and 700°F, respectively. However, some pressure vessels operate at temperatures above those limits. The combinations of operating temperatures and stresses will cause plastic deformation and creep to occur during the anticipated 20- to 30-year service life of a vessel. Consequently, new design data are needed for conditions that have not previously been explored in depth. Experience has shown that the weld joints are critical regions of a pressure vessel and the present demand for high performance reliability can only be satisfied by careful investigation of weld properties for inclusion in design considerations.

The Welding and Brazing and Engineering Mechanical Properties laboratories of Oak Ridge National Laboratory have collaborated for several years on programs to improve and understand the properties of stainless steel welds for reactor service.²⁻⁶ A major problem was recognized for shielded metal-arc Type 308 stainless steel welds: that under uniaxial creep conditions the strain at failure could be as small as 1% (refs. 2,7). Furthermore, the trend of the data showed that rupture ductility tended to decrease with decreasing stress or increasing rupture time. Extrapolations of the maximum total creep ductility usually predicted less than 1% ductility for 10,000-hr rupture times from about 1050 to 1200°F. In an attempt to solve the problem by

controlling the amounts of minor elements in the weld deposits, a series of shielded metal-arc electrodes was produced by an industrial firm for screening tests. The screening experiment began with standard lime-, lime-titania, and titania-coated electrodes and proceeded through a series of chemical additions to the fluxes, in an effort to vary the weld deposit chemical analyses; however, these deposit analyses were all within the specifications for Type 308 stainless steel weld metal.⁸ Screening tests at 1200°F showed that weld metal containing higher than average amounts of P, B, and Ti (about 0.04% P, 0.06% Ti, and 0.007% B) had better ductility and higher creep strength than normal Type 308 stainless steel weld deposits. Type 308 stainless steel weld metal with the optimum composition is denoted CRE (for Controlled Residual Element) weld metal.

A set of weld test plates with CRE Type 308 stainless steel welds was produced by manual shielded metal-arc welding. Mechanical properties tests have been performed on weld specimens from these plates to verify the improved ductility of the CRE weld metal and to obtain design data. Creep, tensile, and stress-relaxation tests are being performed at the Oak Ridge National Laboratory, while other laboratories are performing fatigue and stress-relaxation studies on them. Concurrent microstructural investigations have given insight into the significance of the results, and provided invaluable aids in interpreting and extrapolating data.

It is not the purpose of this paper to report all of the test results that have been obtained to date. Rather, a scoping view of the test results will be presented together with interpretations and comparisons of the data for CRE weld metal with those of conventional Type 308 stainless steel weld metal. The heterogeneity of properties in weld metal and the anisotropic deformation characteristics of weld metal are special features that are certain to complicate the analysis of stresses and strains in weldments.

EXPERIMENTAL PROCEDURES

The manual shielded metal-arc test welds joined $6 \times 8 \times 2 \frac{3}{8}$ in.-thick sections of stainless steel base plate. The base plates were from two heats of Type 304 stainless steel, 600414-1A and 300380-1A, for which the vendor's chemical analyses are given in Table 1. Combustion Engineering of Chattanooga, Tennessee, joined the weld test plates in a double-U joint design with 0.25-in.-diam CRE Type 308 stainless steel weld metal electrodes. After a root pass had been deposited, excess base metal was back-chipped from the joint grooves and the grooves were filled with weld metal to achieve a cross section that was nearly symmetrical about the center of the plate. The finished joint is two passes wide except in the root. Four different batches of flux-coated electrodes were used and the chemical compositions of deposits made with these electrodes are given in Table 2.

Two types of mechanical properties specimens were used in the testing program, a "standard" specimen and a smaller "buttonhead" specimen. The locations of the standard specimen blanks within the weld are shown in Fig. 1(a). It is important that the specimens be

Table 1. Chemical Composition of Type 304 Stainless Steel
Plates Used in FFTF Vessel Test Weld Study

Element	Amount Present, wt %, in Heat	
	600414-1A	300380-1A
C	0.061	0.058
Mn	1.48	1.60
P	0.016	0.016
S	0.014	0.011
Si	0.64	0.55
Cr	18.88	18.68
Ni	9.56	8.44

Table 2. Chemical Composition of CRE Type 308 Stainless Steel
Weld Deposits

Element	Chemical Composition, wt %, of Weld Metal From Batch Number			
	2	4	6	9
Si	0.6	0.52	0.56	0.49
S	0.009	0.016	0.005	0.004
P	0.042	0.043	0.040	0.044
Mn	1.87	1.95	1.83	1.91
C	0.061	0.066	0.064	0.061
Cr	20.10	19.73	19.70	19.70
Ni	10.14	9.88	9.61	9.63
Mo	0.24	0.25	0.07	0.24
Cb + Ta	<0.01	0.01	<0.01	<0.01
Ti	0.06	0.06	0.06	0.06
Co	0.07	0.07	0.03	0.07
Cu	0.17	0.16	0.04	0.04
B	0.007	0.007	0.007	0.007
V	0.10	0.09	0.07	0.09
N ₂	0.042	0.039	0.030	0.042
Fe	bal	bal	bal	bal

Fig. 1. Mechanical Properties Test Specimens. (a) Standard and (b) "buttonhead" test specimens: Hourglass-shaped regions represent weld metal.

grouped according to the distance of the specimen center line from the nearest surface. Specimens of the L1, L2, and L3 types and specimens of the T1, T2, and T3 types are located with their center lines approximately 0.19, 0.56, and 0.94 in. below the nearest surface of the weld, to an accuracy of approximately ± 0.10 in., as a result of the uncertainty in normal saw-cutting and machining operations. The finished standard specimens had 0.25 in. gage diameters and 1.25-in. gage lengths.

Both tensile and creep tests were run for the standard specimens. Nearly all of the standard creep specimens were fitted with averaging extensometers. The entire strain-time curve was monitored and recorded incrementally using dial gages, but a continuous record of strain versus time was obtained for the early part of the test using linear voltage differential transformers (LVDT). Similarly, the standard tensile specimens were fitted with averaging extensometers and LVDT's to record the strains below 2%. Load measurement was accomplished through a load cell in the pull train. The crosshead deflection-load chart from the Instron tensile testing machine was used for strains from 0.2% to fracture, since it agreed well with the extensometer over this range. Tensile tests were performed at a nominal strain rate of 0.004 min^{-1} (the tests were not strain-controlled) at temperatures between room temperature and 1200°F (649°C). Creep tests were performed at 900, 1050, and 1100°F . The extensometer systems were capable of 10 $\mu\text{in.}$ sensitivity, and temperatures were controlled to $\pm 3^\circ\text{C}$.

The "buttonhead" specimens had 0.125-in. gage diameters and 1.25-in. gage lengths. They were cut parallel to the welding direction, but gage

lengths included base metal, weld metal, or weld-and-base metal as indicated in Fig. 1(b). Those specimens indicated by shading were tested in creep at 1100°F under a uniaxial stress of 35,000 psi.

In addition to the tensile and creep specimens, optical and electron microscopy specimens were cut from selected regions of the weld and base metal. Sections were cut parallel and perpendicular to the welding direction. Wafers for electron microscopy were thinned electrolytically and examined in the transmission mode to determine the amount of variation that occurred in the microstructure from region to region. Quantitative measurements of dislocation and precipitate densities were made using stereo techniques to measure foil thickness and dislocation densities in three dimensions. The correction for invisible dislocations was made by assuming all Burgers vectors to be equally probable, an assumption that was verified by examining two or more reflections in some regions.

MECHANICAL PROPERTIES AND MICROSTRUCTURAL EVALUATION OF WELD METAL

Heterogeneity of microstructure and mechanical properties is a prominent feature of welds in general. The heterogeneous nature of these welds is summarized in Fig. 2, which illustrates both the gross mechanical property and microstructural variations that were observed in the weld metal proper.

The optical microstructures were typical of Type 308 stainless steel welds in that they consisted of a matrix of austenite containing long, narrow islands of ferrite distributed in a cellular dendritic pattern. The substructure visible in the etched microstructure of Fig. 2 consists of columnar austenite subgrains, many of which clearly resulted from epitaxial growth of base metal austenite grains into the melt and later across inter-pass boundaries. On a macroscale, the microstructures were inhomogeneous in the sense that the amount of ferrite and the orientation of its greatest dimensions varied from place to place within the weld.

Finer details of the microstructure such as the dislocation and precipitate structures were reasonably homogeneous over a local scale although long-range systematic variations, qualitative as well as quantitative, were observed. At each location in Fig. 2, the uppermost electron micrograph illustrates the dislocation structure and the lowermost micrograph shows the austenite-ferrite interface. Near the weld center line [1.2 in. beneath the surface, Fig. 2(d) and 2(g)], the dislocations were grouped in a densely packed cell structure: the dislocations were grouped into walls surrounding material relatively free of dislocations. In addition to dislocations, the dense cell walls contained a high concentration of small, barely resolvable dislocation loops, about $5 \times 10^{15}/\text{cm}^3$ averaged over the area of the photomicrograph. The austenite-ferrite phase boundaries in this portion of the weld were decorated with massive M_{23}C_6 precipitate particles. These face-centered cubic precipitate particles were crystallographically oriented parallel to the austenite lattice.

Quantitative measurements show that the dislocation density decreased toward the weld surface (Fig. 2). At a location 0.56 in. from the surface [Fig. 2(c) and 2(f)], a cell structure was present, but it was considerably

Fig. 2. Variation of Microstructure and Mechanical Properties Through the Thickness of a Shielded Metal-Arc Type 308 Stainless Steel Weld with Controlled Residual Elements.

less well-developed than in the region of the earliest weld passes. Loops and straight dislocations were found in the areas of relatively low dislocation density. The austenite-ferrite interfaces again were covered with $M_{23}C_6$ precipitate particles, but these were more plate-like and less massive than those found deeper within the weld.

Near the weld surface [0.2 in. below the surface, Fig. 2(b) and 2(e)] the dislocation density was greatly reduced and the microstructure was dominated by relatively straight dislocations. A hint of a crude cell structure can be seen in this region in that areas of high dislocation and loop density surround regions of low density. In this part of the weld the austenite-ferrite interfaces were completely free of any precipitate particles.

The yield stress results shown in Fig. 2 were all obtained from standard specimens out of one test plate. The stress increase of 14,000 psi between specimens L1 and L3 at 1050°F is slightly higher than the average for specimens from the other three plates; a 9000- or 10,000-psi increase is more common. Yield strength increases of about this magnitude were observed over the range of test temperatures from room temperature to 1100°F. Systematic variations in strain hardening behavior were also observed.

The rupture time for standard creep specimens tested at 1050°F under 37,000-psi stress also increased with increasing distance beneath the weld surface, as shown in Fig. 2. The strain-time curves for these specimens are shown in Fig. 3. Quite clearly, the amount of strain that occurs on loading, first-stage creep strain, and the minimum creep rate are all greater for the specimen from near the surface than for the specimens from nearer the center line of the weld. Similar behavior has been observed for all temperatures and stresses that have been investigated to date.

These systematic variations in microstructure and properties are more than casually related. If we assume that the deformation of the continuous austenite phase controls the yielding phenomenon, then rough estimates of the effects of the observed microstructural changes on that property are possible from existing theories. The strength increment due to the presence of $N = 5.5 \times 10^{10}$ loops/cm³ may be estimated as⁹

$$\Delta\sigma_L = \frac{2Gb}{3.7} (Nd)^{1/2} , \quad (1)$$

where the shear modulus $G = 5 \times 10^{11}$ dynes/cm², the Burgers vector $b = 2.5 \times 10^{-8}$ cm, the loop diameter $d = 10^{-6}$ cm. At the center of the weldment, the loops cause $\Delta\sigma_L^G = 22.5$ psi (1.58×10^6 dynes/cm²). The increase in the yield stress due to a forest of dislocations is used to estimate the effect of straight dislocation lines and dislocations in the cell walls, ignoring the effect of dislocation array geometry¹⁰:

$$\Delta\sigma_d = \alpha Gb (\gamma)^{1/2} , \quad (2)$$

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Fig. 3. Comparison of Creep-Rupture of L1, L2, and L3 Type 308 Stainless Steel FFTF Vessel Test Weld Specimens at 1050°F (565°C) Tested at 37,000 psi.

where α is a constant about unity, and the dislocation density is γ . Near the weld center line, $\gamma = 2.8 \times 10^{10}/\text{cm}^2$ and $\Delta\sigma_d^c = 30,000$ psi (2.1×10^9 dynes/cm²). Near the surface of the weld, $\gamma = 1 \times 10^{10}/\text{cm}^2$, and $\Delta\sigma_d^c = 17,000$ psi (1.2×10^9 dynes/cm²).

The yield strength of a structure consisting of a matrix of yield strength σ_m and obstacles to dislocation motion of similar strength and concentration is

$$\sigma_y = \sigma_m + (\Sigma \alpha \Delta\sigma_d^2)^{1/2}, \quad (3a)$$

or

$$\sigma_y = \sigma_m + \sum_i \Delta\sigma_i, \quad (3b)$$

where the summation is performed for all obstacles present. Since $\Delta\sigma_l$ is small, the difference in strength from the center to the surface of the weld due to dislocations and loops is approximately:

$$\Delta\sigma = \Delta\sigma_d^c - \Delta\sigma_d^s, \quad (4)$$

or about 13,000 psi. This is very close to the observed variations. The theoretical predictions made above do provide a close approximation to the observed variations in strength through the weld thickness. Whether they are truly meaningful is difficult to assess at this stage, since they fail to account for all of the features of a complex microstructure and do not as yet account for possible variations in local chemical composition.

There is no good basis for quantitatively assessing the effect of initial microstructure on creep behavior throughout a creep test. However, the observation that specimens from near the surface deform more on initial loading and during first-stage creep than do specimens from near the center is reasonable in view of the predicted effects of structure on yield behavior. If microstructural features control creep, the more rapid second-stage creep and earlier failure for specimens near the surface are indications that initial microstructure either is retained or otherwise affects the creep deformation structure that develops with time and strain.

STRUCTURE AND PROPERTY VARIATIONS THROUGHOUT THE WELDMENT

Systematic variations of microstructure and mechanical properties were found in the base metal as well as the weld metal. In the base metal adjacent to the weld, the dislocation density decreased with increasing distance from the weld. Figure 4 illustrates the microstructure of the base metal at distances of 0.04 and 0.4 in. (0.1 and 1.0 cm) from the weld metal-base metal interface at a depth of 0.6 in. (1.5 cm) below the weld surface. Sections taken to include the fusion line showed no discontinuous change in the dislocation structure. In fact, no clearly defined fusion line could be detected; the gradual appearance of ferrite was the only change that occurred in crossing the fusion zone. This is in agreement with optical metallographic observations (Fig. 2). In the base metal section 0.04 in. (0.1 cm) from the weld there was a tendency for tangling and cell formation to occur similar to that found within the weld. Further from the weld, only straight dislocations were found typical of a very light, low-temperature deformation, as in a straightening operation, for example.

This variation in structure was reflected in a variation of creep behavior for the buttonhead creep specimens at 1100°F (593°C) for 35,000-psi stress. Figure 5(a) shows the behavior of weld metal.

Fig. 4. Dislocation Structure in Base Metal Approximately (a) 0.04 in. (0.1 cm) and (b) 0.4 in. (1.0 cm) from the Fusion Line.

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Fig. 5. Creep Behavior of Buttonhead Specimens at 1100°F (593°C) and 37,000-psi Stress. (a) Weld metal. (b) Base metal and fusion line ("fusion line" specimens contained both base metal from the heat-affected zone, the fusion boundary, and weld metal).

Figure 5(b) shows the behavior of base metal and specimens that straddled the fusion line. The two base metal specimens distant from the fusion line strained nearly 20% in approximately 500 hr (specimens 301 and 305). Their behavior is typical of this heat of stainless steel in the as-received condition. The higher dislocation content material (specimens 319 and 324) near and straddling the fusion line strained less on loading than the material in the as-received condition; it deformed a total of about 8% strain in about 1200 and 900 hr, respectively. By contrast, all of the weld metal specimens that were tested were relatively weak and ductile, straining over 20% in less than 100 hr.

ANISOTROPY OF WELD METAL DEFORMATION

In addition to the systematic variations of properties and structure that occur from point to point in the weld metal, there is a pronounced anisotropy of the deformation characteristics of the weld metal. Room temperature acoustic measurements of the elastic shear modulus taken about 0.5 in. (1.25 cm) below the weld surface showed that the modulus was about 6×10^{11} dynes/cm² parallel to the solidification substructure and about 10×10^{11} dynes/cm² perpendicular to the solidification substructure.¹¹ The shear waves were propagated normal to the surface shown in Fig. 2.

Longitudinal, all-weld metal specimens always exhibited greater diametral strain in the direction parallel to the solidification substructure than in the direction perpendicular to the solidification substructure. In room temperature tests, the ratio of the maximum to minimum diametral strain in the uniformly strained regions was found to remain constant during plastic deformation in a periodically interrupted test. Values of the ratio ranged from about 1.5 to about 4 for all tensile test temperatures. A similar range was found for ruptured creep specimens.

The cross section of a transverse specimen (T1) tested at room temperature is shown in Fig. 6. It is clear from the macroetched region that anisotropic deformation occurred, and that the orientation of the principal axes of anisotropy varies across the weld cross section. The maximum and minimum diametral strains are indicated for corresponding sections in the directions perpendicular to the surface of the original weld and parallel to the original weld surface. In regions where the solidification substructure is perpendicular to the original weld surface, deformation occurs most readily in the direction of the solidification substructure. In all other regions, deformation occurs more readily parallel to the welding direction than perpendicular to the surface. Similar behavior is observed for transverse specimens tested at all temperatures.

Because of the point-to-point differences in anisotropic deformation, the stress-strain curves for transverse specimens such as that shown in Fig. 6 cannot be readily interpreted to answer the question, "Is the weld metal stronger in the longitudinal or transverse direction?" However, the available strength values for transverse specimens are generally as large as or larger than those for longitudinal specimens from similar locations in the weld and for comparable test conditions.

Fig. 6. Two Views of the Gage Section of a Transverse Weld Specimen Taken 0.19 in. Below the Weld Surface, Pulled at Room Temperature at 0.04 min^{-1} . Left side has been cut, polished, and macroetched (2 parts HNO_3 , 1 part H_2SO_4). Left side is rotated about the original specimen axis by 180° relative to right side.

MATERIAL PROPERTY COMPARISONS

The strength and ductility of the CRE Type 308 stainless steel weld metal may be compared with available properties for conventional Type 308 stainless steel weld metal to determine whether it is an improved weld metal in some sense. The same properties may be compared with the properties of the base metal that it joins to verify that the properties are generally compatible. From the earlier discussion of heterogeneity, it is obvious that the CRE weld metal data should be grouped according to distance from the weld surface in data presentations and comparisons.

The 0.2% offset yield strengths for specimens made from all four batches of CRE Type 308 stainless steel electrodes are plotted in Fig. 7. Available data at room temperature, 900 and 1100°F show the range of values obtained for L1, L2, and L3 specimens. The range of strengths for the three types of specimens nearly spans the width of the entire scatter band for Type 308 stainless steel literature data.^{7,12} The literature data are for a variety of joint configurations and include weld metal deposited by the gas-tungsten arc, shielded metal-arc, and submerged-arc processes. No characterization of specimen position in the weld is usually given in the literature. Both conventional and CRE weld metal exhibit strengths above or in the upper edge of the scatter band for Type 304 stainless steel base metal. However, it is clearly possible for

Fig. 7. The 0.2% Offset Yield Strength of Type 308 Stainless Steel Weld Metal and Type 304 Stainless Steel Base Metal.

a particularly strong base metal to be stronger than part or all of a weld made with standard Type 308 stainless steel, and for such base metal to be stronger than at least some of CRE Type 308 stainless steel weld metal.

The creep rupture times for CRE Type 308 stainless steel at 1050°F are compared with the available literature data for standard Type 308 stainless steel weld metal^{7,12} and Type 304 stainless steel base metal¹³ in Fig. 8. The data for L1-CRE weld metal specimens lie within the scatter band for standard Type 308 stainless steel weld metal. However, the L2- and L3-CRE weld metal specimens exhibit much longer rupture times than standard weld metal. The creep rupture times for both L1-CRE weld metal and standard weld metal lie within the 95% confidence limit range ± 1.645 standard deviations of average values for Type 304 stainless steel base metal.¹³ Similar results have been obtained for creep tests performed at 900 and 1100°F.¹⁴

Improved creep ductility in Type 308 stainless steel weld metal was a major goal of the original weld development program that produced the CRE-Type 308 stainless steel weld metal. The total elongations for uniaxially tested CRE weld metal are compared with data for standard Type 308 stainless steel weld metal in Fig. 9. The scatter band for

Fig. 8. The 1050°F Creep-Rupture Properties of Type 308 Stainless Steel (CRE) Test Weld, Compared with Conventional Weld Metal and Type 304 Stainless Steel.

Fig. 9. Ductility in Creep-Rupture Tests of Type 308 Stainless Steel Welds.

more conventional weld metal⁴ at 1200°F (650°C) and the literature data⁷ are all enveloped by an upper limit (excluding one point at about 3500 hr), which corresponds roughly with the lower limit of the scatter band for CRE weld metal. No total creep elongation for CRE weld metal is smaller than 12.2% and a CRE weld metal specimen that failed after 7417.9 hr under 35,000-psi stress at 1050°F strained 13.8%. Thus, the CRE Type 308 stainless steel weld metal possesses the improved creep elongation behavior that was originally sought.

DISCUSSION

From a design viewpoint, the CRE Type 308 stainless steel welds reported here have strength properties comparable to or better than those already reported for Type 308 stainless steel weld metal. The CRE weld metal enjoys a substantial advantage in long-term creep-rupture ductility. Use of this weld metal for elevated-temperature service appears reasonable at this time from the standpoint of tensile and creep properties.

The systematic property and structural variations and anisotropy that were found in this study are not necessarily expected to occur in all welds. Some guidance in anticipating whether these features might occur in other welds is obtained from considering the physical causes of these features. In this regard, the apparent relationship between microstructure and properties is quite important.

Local dislocation densities and configurations are important since they have been linked to mechanical properties, although the effects of other microstructural features or variation in local chemical composition on properties are not yet accounted for.

The dislocation densities in the CRE weld metal are well over an order of magnitude higher than those found in a recent study of Inconel by Yoshima and Winterton.¹⁵

The dislocation distributions are distinctly different from the uniform distributions found in Inconel and the creep-like microstructure reported by Downs¹⁶ in stainless steel. They are different too from the microstructures produced by tensile or creep tests. This probably results from the numerous heating and cooling cycles required in these large, multipass welds, which result in cyclic deformation processes with accompanying thermal treatments. The high concentrations of dislocation loops found in the cell walls may arise from this. During deformation cycles, loops can be generated within the cells, possibly from jogs in moving screw dislocations. During the reverse part of the cycle these may be annihilated or if dislocations are generated on another slip system that repel the loops, the loops may be pushed ahead of the moving dislocations into the cell walls. The higher concentration of loops and dislocations and the better definition of the cell structure near the weld center line are due to increased number of thermal and stress cycles seen by the initial part of the weld. The heavy carbide precipitation at the austenite-ferrite interface and the absence of it at the free surface are a result of the larger number of thermal cycles that are seen by the initial weld passes. Local microstructure and properties

thus depend strongly on welding parameters and joint configurations that determine the precise thermomechanical history of a given weld. The physical causes of anisotropic deformation are not clear at this time. At least two possible reasons deserve considerations. Preferred orientation in welds has already been observed.^{17,18} In any small region of the weld metal, subgrains of similar orientation would tend to be present, and they might then be expected to deform in a manner analogous to the deformation of a single crystal. In view of the variation in anisotropy observed along the length of weld metal in a transverse specimen, this mechanism could only be possible if the net orientation of subgrains changes in moving from one side of the weld to the other. A second possible explanation is that the finely distributed second phases, particularly ferrite, are important in determining the elastic and plastic deformation characteristics of the weld metal. Because ferrite is distributed in a cellular dendritic pattern that is related to the local solidification processes, local rotations of the axes of anisotropy are easily explained. However, clear evidence for distinguishing between these mechanisms is not yet available.

The systematic variation of the structure and properties of base metal near the fusion line is clearly due to the thermal and, particularly, the stress cycling that occurred during welding. Except at the last pass, base metal adjacent to any given pass is stressed by thermal expansion and solidification shrinkage effects not only by that pass, but by succeeding passes. The increase in dislocation density in the base metal as the fusion line is approached provides an explanation for the differences in creep properties observed for base metal from different locations (Fig. 5).

This work substantiates the position that weld metal, heat-affected zones, and base metal in stainless steel may have substantially different properties from each other. Given that fact, predictions of the behavior and service life of weldments from base metal properties are not yet possible. Because of property variations at welds, the presence of a weld may in itself affect the local stresses and strains in a component. Numerical analysis work has already been done to assess the significance of differences in properties on the stress and strain distributions in and near certain weldments. Chan et al.¹⁹ have analyzed creep deformation for a weld in an austenitic steel pipe, and Walters and Cockcroft²⁰ have treated a butt weld. Both of these analyses treated the weld metal, heat affected zone, and base metal as homogeneous, isotropic regions. Such assumptions may have been justified for the cases considered in those studies, but quite clearly a complete analysis of the class of welds in this study requires that property variations and anisotropic deformation be considered. Analysis may help to establish the service conditions under which these features are important.

SUMMARY

Experimental work is in progress to evaluate Type 308 stainless steel with Controlled Residual Elements for elevated-temperature service. Tensile and creep data are reported for this weld metal, which demonstrate

that it possesses improved creep ductility compared to conventional Type 308 stainless steel weld metal. Tensile and creep testing up to 1100°F in conjunction with optical metallography and electron microscopy have demonstrated the following points:

(1) For the test conditions explored, the tensile and creep strengths of the CRE Type 308 stainless steel weld metal are as great as or greater than those previously reported for standard Type 308 stainless steel weld metal. These weld metals may be either stronger or weaker than Type 304 stainless steel base metal, depending upon the base metal heats and welds which are compared.

(2) The CRE weld metal has higher total creep elongations than standard weld metal for test times approaching 10,000 hr.

(3) The strength and ductility of the weld metal vary through the thickness of the weld. Generally, weld metal from near the surface of the weld is weaker but more ductile than that from near the center of the weld.

(4) Systematic variations in dislocation density and configuration, carbide precipitates on austenite-ferrite boundaries, and dislocation loop density occur through the thickness of the weld. Close agreement is obtained between observed yield strength variations and theoretical predictions of the variation based on dislocation density in the austenite phase. Dislocation loops have a relatively small effect, but the effects of other structural features and possible variations in local chemical composition have not been assessed yet.

(5) The base metal immediately adjacent to the fusion line has a higher dislocation density than elsewhere. In the region of high dislocation density, longer creep rupture times and lower ductility are found than elsewhere at 1100°F for 35,000-psi creep tests.

(6) The weld metal deforms anisotropically: the direction parallel to the principal growth direction of the solidification substructure is relatively weak. The orientation of the substructure, and hence the orientation of the principal axes of anisotropy, changes throughout the weld.

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