Effect of Porosity on Elevated Temperature Fatigue Properties of 2¼Cr-1Mo Steel Weldments

Tests show that in 2¼Cr-1Mo steel at 1200 F, 5-8% porosity can reduce fatigue life by 98%

BY R. A. BUCHANAN AND D. M. YOUNG

ABSTRACT. The effects of porosity on the fatigue properties of 2¼Cr-1Mo steel weldments were experimentally determined. The specimens were fabricated using the shielded metal-arc welding process with ASTM A387-D base metal and matching weld metal via E9018-B3L electrodes. The shoulder and gage sections of each fatigue specimen were all weld metal. A set of nonporous specimens and two sets of specimens having approximately five and eight volume per cent (v/o) porosity were tested in fatigue at 1200 F and over a stress-amplitude range from 12 to 30 ksi. The specimens exhibiting weld-metal porosity were fabricated by injecting air and oxygen into the arc atmosphere, which yielded approximately five and eight v/o porosity respectively. Although within experimental error the fatigue lives of the five and eight v/o porosity specimens were not significantly different, the fatigue lives of the porous specimens were only approximately two per cent of the fatigue lives of the nonporous specimens at all stress amplitudes evaluated. Based on the small plastic strains at fracture and the complete absence of necking, the fatigue mechanism appeared to be the dominant failure mode. Therefore, the large effect of weld metal porosity on elevated temperature fatigue properties is discussed as much as possible with the existing data in terms of the effects of porosity on fatigue crack nucleation and propagation.

Introduction

This paper reports on a study of weld metal porosity and its effect on the elevated temperature fatigue properties of 2¼Cr-1Mo steel weldments. The investigation was performed in order to add to the growing body of knowledge concerning the effects of the various structural discontinuities that can be generated during the welding process on the resultant mechanical properties of the weldments. Until only recently most of the code and specification acceptance standards for discontinuities were based on qualitative judgments and not on the results of definitive studies (Refs. 1,2).

Porosity, produced by gas entrapment during weld metal solidification, is the most common type of structural discontinuity (Ref. 3). It is also responsible for the largest percentage of weldment rejections or repairs (Ref. 4). Consequently, weld metal porosity has been the subject of a number of investigations. Some of the earlier researchers utilized specimens with "artificial" porosity, produced for example by drilling holes. However, as shown by Norrish and Moore (Ref. 5), such specimens can produce misleading results by giving higher strength levels than specimens with "natural" porosity of the same magnitude.

Most of the work on the effect of "natural" weld metal porosity on mechanical properties has been conducted at room temperature. And the prevailing conclusion is that porosity has little effect. For example, Green et al (Ref. 6) evaluated the effects of porosity on mild steel welds at room temperature, and found that the weld cross sections could be reduced by as much as 7% by porosity without materially changing the tensile strength, ductility or Charpy impact strength. Also, as discussed by Harrison (Ref. 1), Masi and Erra (Ref. 7) tested welds with porosity levels up to 20%, and found that the tensile strength was almost constant when expressed in terms of the net cross sectional stress.

Low cycle room temperature fatigue tests of mild steel weldments appear to confirm the relatively small effect porosity has on tensile strength, and also perhaps on fatigue strength. Ishii (Ref. 8) tested weldments with porosity levels up to 5% and found that for endurances up to 10^4 cycles, the fatigue strength was nearly equal to the tensile strength. Therefore, with regard to room temperature mechanical properties, the available data verify within reason the widespread conclusion that porosity is not a significantly harmful discontinuity. However, the very limited work on high temperature properties reveals a possibly different conclusion.

Only Ishii et al (Ref. 9) in Japan

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have studied the effects of porosity on the elevated temperature mechanical properties of weldments. Tensile, creep-rupture and fatigue tests were conducted on porous welds in ASTM A302B, a Mn-Mo low alloy steel. The tensile strength at 400 C (752 F) was not affected by porosity levels up to 10%. However, the creep-rupture strength was affected, and the detrimental effect became more pronounced at longer times-to-rupture. For example, at 550 C (1022 F), 3-4% porosity reduced the creep strength for a 100 h life by 11%, but for a 1000 h life, the porosity reduced the creep strength by 50%! Only 3 specimens with 3-4% porosity were tested in fatigue. Furthermore, it is difficult to draw conclusions from the fatigue test results because no porosity-free specimens were similarly tested in order to form a basis for comparison (Ref. 1).

In summary then, information on the effect of porosity on elevated temperature properties is confined to one investigation on one material. Even this study is very limited in the data produced. Nevertheless, the results indicate that a given porosity level can be much more detrimental at elevated temperature than at or near room temperature.

Experimental Procedure

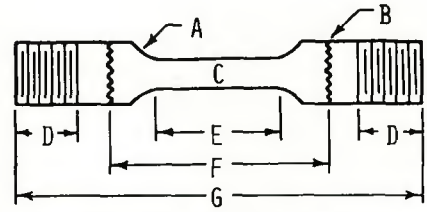
The material selected for the present investigation was ASTM A387 Grade D, which has a nominal composition of 2¼Cr-1Mo. This material was selected primarily because of its current widespread use in welded structures designed to function at elevated temperature, e.g. nuclear vessels, petroleum hydrocrackers, etc. The procedure followed during the fabrication of the weld metal fatigue specimens was as follows. First, grooves approximately one-half inch in depth with rectangular cross sections were machined into the 1 in. thick ASTM A387 Grade D base metal. Next, filler metal was deposited into the grooves utilizing the shielded metal-arc welding process and E9018-B3L electrodes. Finally, composite specimens were machined from the welded plates transverse to the longitudinal axes of the welds. As illustrated in Fig. 1, the composite specimens were all weld metal except for the gripping sections. As part of the welding process, a 400 F (204 C) preheat was used; and after welding, the plates were subjected to a stress-relief heat treatment of one hour at 1260 (682 C). The chemical compositions of the base metal and the as-deposited weld metal are presented in Table 1.

The introduction of a relatively uniform distribution of nominally spherical porosity proved to be a difficult problem. After many attempts, as summarized in Table 2, two successful methods were found. The first method involved the injection of oxygen directly into the arc atmosphere during welding, and yielded an average of approximately 8 volume per cent (v/o) porosity. The second method involved the injection of compressed air directly into the arc atmosphere during welding, and yielded an average of approximately 5 v/o porosity. It should be realized that the type of porosity developed by these laboratory techniques should be very similar to that developed during regular welding practice by such abnormalities as air currents, inadequate shielding, etc.

Porosity-free specimens were also fabricated for use as references in the fatigue testing program. All parameters in the welding procedures used to produce the porous and nonporous specimens were identical except for the injection of air and oxygen respectively in the two sets of porous specimens.

After fabrication, the welded plates were radiographed to check the average pore size and distribution, and also to verify the absence of large-scale slag inclusions and incomplete fusion. Then in order to accomplish the detailed porosity analysis, specimen blanks with square cross-sections were machined from the welded plates. Examples of the specimen blanks are shown in Fig. 2. The pore-size distribution and the v/o porosity were evaluated for each specimen by carefully analyzing photographic enlargements of two cross sections of each specimen blank. The volume fraction of porosity was evaluated as the weight of the excised pore regions of the enlargement divided by the total specimen blank enlargement weight. Although this measurement gave directly the areal fraction of porosity, it can also be considered equivalent to the volume fraction, as discussed by DeHoff and Rhines (Ref. 10). The v/o porosity associated with each fatigue specimen is presented in Table 3, and the average pore-size distributions of the air-porous and oxygen-porous specimens are shown in Fig. 3.

Inspection of the specimen blanks in Fig. 2 allows a qualitative description of the type of porosity developed. Many of the individual pores



A = 0.25 INCH (6.3 MM) RADIUS
B = WELD-METAL/BASE-METAL INTERFACE
C = 0.250 INCH (6.35 MM) DIAMETER
D = 0.50 INCH (12.7 MM)
E = 1.00 INCH (25.4 MM)
F = 1.75 INCH (44.4 MM)
G = 3.25 INCH (82.6 MM)

Fig. 1 — Fatigue specimen design

Fig. 2 — Examples of porous fatigue specimen blanks

Table 1 — Analyses of Base Plate and Deposited Weld Metal, wt. %

Material	C	Si	Mn	P	S	Cr	Mo
Plate	.12	.23	.50	.015	.020	2.37	.98
Weld metal	.01	.71	.61	.015	.004	2.06	1.06

Other trace elements: Cu, Ni, Sn, Al, W, Ti, V, Cb, Co, Pb, As

Table 2 — Trial Welding Methods and Results

Trial	Method	Results
1	Welded over Fe_3O_4 powder	No porosity generated
2	Welded over Fe_2O_3 powder	Generated extremely fine porosity
3	Welded with electrodes soaked in water 36 h	No porosity generated.
4	Welded with electrodes baked at 750 F and soaked in water 36 h	No porosity generated
5	Welded with electrodes soaked in #40 motor oil	Generated coarse porosity
6	Welded with electrodes soaked in cutting oil	Generated coarse porosity
7	Welded while injecting water into the arc atmosphere	No porosity generated
8	Welded while injecting water mist into the arc atmosphere	No porosity generated
9	Welded with long arc, low amperage (100A), and high travel speed	Generated porosity but had slag inclusions and lack of fusion
10	Welded with long arc, low amperage (100A), high travel speed and erratic manipulation	Generated porosity but had slag inclusions and lack of fusion
11	Welded using bare wire welding	Generated porosity but had lack of fusion, general unsoundness
12	Welded with partial electrode-flux removal, leading edge bare	No porosity generated
13	Welded with flux removed from electrode	Generated porosity, penetration limited, general unsoundness
14	Welded with partial electrode-flux removal, trailing edge bare	No porosity generated
15	Welded over sodium bicarbonate	Some fine porosity generated but not consistently

were not nominally spherical in shape; in fact some of them exhibited rather acute angles. Furthermore, although many of the pores were characterized by bright, clean walls, some of the pore walls were oxidized, and other pores contained large amounts of slag type material, i.e., perhaps very small slag inclusions. The characteristics of the porosity developed in this study are not too surprising since the porosity was primarily caused by the evolution of oxygen and nitrogen during weld metal solidification.

The fatigue testing program was conducted with a constant-displacement type fatigue machine. In all cases the cyclic stress variation was sinusoidal in nature, varying from zero to a maximum tensile stress. The frequency employed was 500 cycles per minute (8.33 cycles per second); the temperature was 1200 ± 15 F (648 ± 8 C); and the stress amplitudes ranged from 12.0 to 30.0 ksi (83 to 207 MPa). Since for a given test a constant stress amplitude was desired, it was necessary to monitor the load applied to the specimen and manually make periodic adjustments to compensate for specimen elongation. Load checks, and adjustments when necessary, were made every 10,000 cycles; this procedure resulted in a constant load amplitude within ± 5% of the desired value.

Results

The results of the elevated temperature fatigue tests are presented in Table 3, which for each specimen gives the v/o porosity, the stress amplitude employed, and the number of cycles to fracture. The fatigue results are also shown as S-N curves, i.e., stress amplitude vs number of cycles to fracture, in Figs. 4 and 5. In Fig. 4, the stress amplitude is based on the total cross sectional area, i.e., nonporous plus porous areas; whereas in Fig. 5, the stress amplitude is based on the net cross sectional area, i.e., nonporous area only. The net cross sectional areas were derived from the experimental v/o porosity values and not from fracture surface examinations.

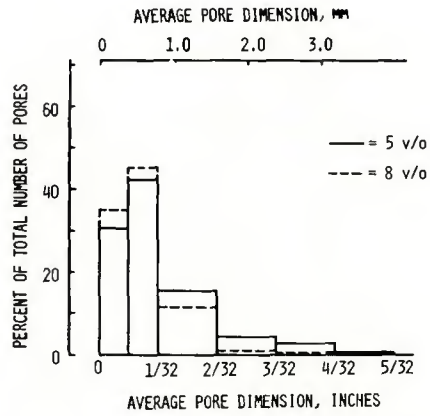


Fig. 3 — Porosity size distributions for air and oxygen injected specimens

As seen in Figs. 4 and 5, the results did not allow a differentiation between the fatigue properties of the air-porous and oxygen-porous specimens, which represented a range in porosity levels from approximately 4 to 10 v/o. However, a vast difference was found between the fatigue properties of the porous and nonporous specimens. The fatigue lives of the porous specimens were only approximately 2% of the fatigue lives of the nonporous specimens at all stress amplitudes. Moreover, this huge reduction in fatigue life could not be explained as being strictly caused by the loss in cross sectional area due to porosity, as evidenced by the data as plotted in Fig. 5.

Since the fatigue testing was performed at 1200 F (648 C), one might expect that considerable creep could have taken place and therefore influenced the failure mechanism. Surprisingly however, the specimens exhibited very little creep strain. In all cases the total plastic strain at fracture was less than 0.8%. Furthermore, no evidence of localized deformation or necking before fracture was found, and all the fracture surfaces were perpendicular to the tensile axis. These brittle, fatigue type failures occurred despite the high temperature and the fact that room temperature hardness measurements after the fractures indicated relatively ductile weld metal microstructures with hardnesses in the range R_B 85-100.

Discussion of Results

Because of the particular alloy studied and also the elevated temperature at which the tests were conducted, a number of factors in addition to the basic fatigue process could have contributed to the failures.

Probably the most important is temper enbrittlement, to which the 2¼Cr-1Mo alloy is particularly susceptible (Ref. 11). Another failure

mechanism that could have been operative is creep embrittlement (Ref. 12). However, under the conditions of the present investigation, these two factors were expected to have had equal effects on both the porous and nonporous specimens; and therefore could not have caused the large reduction in fatigue life due to porosity.

However, there are several more factors that could have had different effects on the porous and nonporous specimens. The first is of course the creep phenomenon. There is experimental evience, as cited earlier, that creep rates should be higher in porous specimens than in nonporous specimens. And yet under the experimental conditions of this work, virtually no creep strain was developed in any of the specimens. Perhaps if the cyclic frequency of the fatigue tests had been reduced, the creep mechanism would have played a larger role in the results.

The final two factors to be considered are oxygen and nitrogen embrittlement. The two techniques used to generate the weld metal porosity certainly produced the porosity per se, but also may have produced embrittled microstructures (Ref. 13). In the oxygen injected specimens, residual oxygen may have been present in the dissolved state and/or as oxide particles in the microsructure. In the air injected specimens, both oxygen and nitrogen may have been present in the dissolved state and/or as oxides and nitrides. Indeed, it is probable that embrittled regions surrounding the pores significantly enhanced the detrimental effects of the porosity itself.

Therefore, the over-all results of this study indicated that the porosity itself, as well as probably changes in the microstructures and/or chemistry of the weld metal surrounding the pores, significantly reduced the fatigue life of the weld metal by modifying the basic fatigue processes of crack initiation and propagation. In view of this conclusion, the effects of porosity on fatigue crack initiation and propagation will now be discussed in order to explain the data presented in Figs. 4 and 5.

Let the total number of cycles required to initiate and propagate a fatigue crack to critical length for a porous specimen be represented as N_f^p, similarly, let the total number of cycles for a nonporous specimen be N_f^n. These parameters can be expressed as

$$N_f^p = N_i^p + N_p^p \quad (1)$$

$$N_f^n = N_i^n + N_p^p \quad (2)$$

where N_i and N_p refer to the number of cycles required for crack initiation and propagation respectively. In order to simplify the manipulations to follow, let the following parameters be defined: $x = N_i^p/N_i^n$, which reflects the effect of porosity on the cycles required for initiation; $y = N_p^p/N_p^p$, which reflects the effect of porosity on the cycles required for propagation; $f_i = N_i^n/N_f^n$, i.e., the fraction of cycles required for crack initiation in the nonporous specimens; and finally, $R = N_f^p/N_f^n$, which reflects the effect of porosity on total fatigue life. By combining Eqs. (1) and (2) , and utilizing the previous definitions, the following equation results

$$R = xf_i + y(1-f_i) \quad (3)$$

Since the data generated in this study indicated that $R \simeq 0.02$ at all stress amplitudes, $dR/d\sigma = 0$, where σ is the cyclic stress amplitude. Therefore, differentiation of Eq. (3) with respect to σ yields

$$0 = x\delta f_i/\delta\sigma + f_i\delta x/\delta\sigma - y\delta f_i/\delta\sigma + \delta y/\delta\sigma - f_i\delta y/\delta\sigma \quad (4)$$

Equation (4) will now be analyzed in terms of the results of this investigation and other relevant information in the literature. First, it is known that the parameter f_i, the fraction of cycles required for crack initiation in a nonporous specimen, is a function of stress amplitude (Ref. 14). In fact, f_i decreases with increasing stress amplitude; and therefore the term $df_i/d\sigma$ is negative. With regard to the

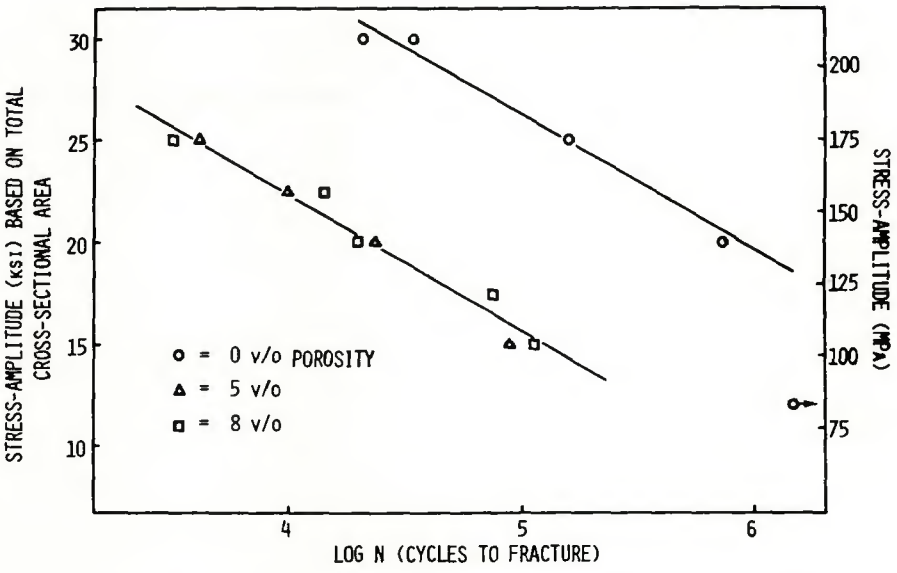


Fig. 4 — S-N curves for porous and nonporous weld metal specimens with stress amplitudes based on total cross sectional areas

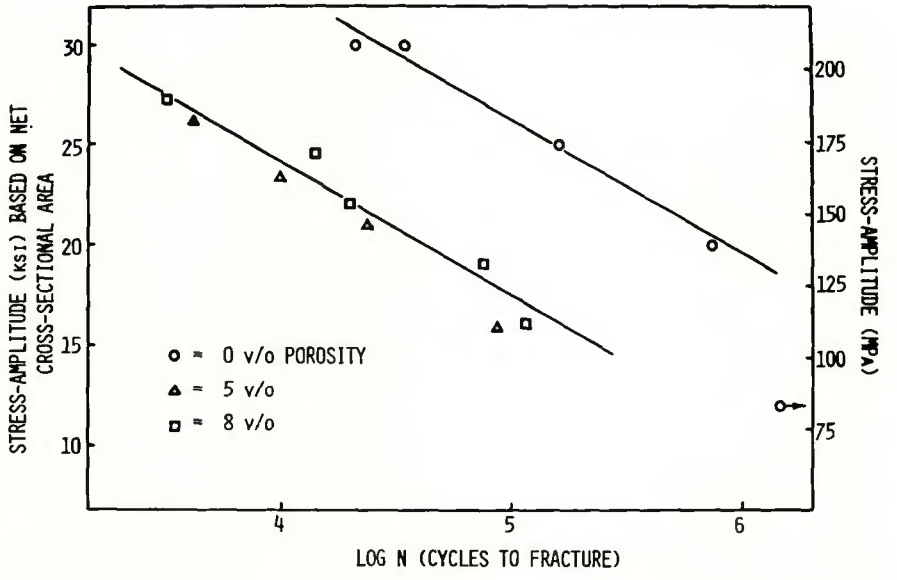


Fig. 5 — S-N curves for porous and nonporous weld metal specimens with stress amplitudes based on net cross sectional areas

Table 3 — Porosity Levels and Fatigue Properties

Specimen type and number	Avg. v/o porosity	Stress amplitude ksi	(MPa)	No. of cycles to failure
Nonporous				
2	0	25.0	(172)	164,210
3	0	12.0	(83)	1,500,000[a]
4	0	20.0	(138)	750,001
8	0	30.0	(207)	20,700
10	0	30.0	(207)	34,490
Air-injected				
1	4.74	22.5	(155)	10,001
2	4.94	20.0	(138)	24,100
5	3.64	25.0	(172)	4,200
6	5.72	15.0	(103)	88,508
11	6.42	12.5	(86)	0[b]
Oxygen-injected				
1	5.56	12.5	(86)	0[b]
3	9.49	17.5	(121)	76,684
9	6.29	15.0	(103)	0[b]
10	7.69	22.5	(155)	14,206
12	7.10	25.0	(172)	3,200
13	9.68	20.0	(138)	20,200
15	8.42	15.0	(103)	112,110

(a) Did not fracture — stopped test
(b) Fractured on application of load

parameter x, which refers to the effect of porosity on crack initiation, work by Barsom (Refs. 14,15) concerning the effect of stress concentrators on the number of cycles for initiation has verified that stress concentrators certainly reduce the cycles required for initiation. But even more important, Barsom's work has shown that the number of cycles for crack initiation in the presence of notches can be closely correlated with the parameter $\Delta K_i/\sqrt{\rho}$ where ΔK_i is the stress intensity factor range and ρ is the notch tip radius. In terms of Eq. (4), x can be regarded as the ratio of $\Delta K_i/\sqrt{\rho}$ for porous and nonporous specimens. And since (a) ΔK_i is a product of stress amplitude times geometric factors associated with the stress concentrator (Ref. 16), and (b) the pores responsible for crack initiation in all the porous specimens in this investigation were probably comparable in size and shape, in the ratio the stress amplitude should cancel. Therefore, $dx/d\sigma$ is assumed to be zero. In view of these assumptions, Eq. (4) reduces to the form

$$0 = (x-y)\,\delta f_i/\delta\sigma + (1-f_i)\,\delta y/\delta\sigma \qquad (5)$$

It is reasonable at this point to assume $x < y$, i.e., that porosity has a greater effect on crack initiation than propagation. The assumption is reasonable because a pore, probably at the surface, would influence crack initiation during the entire initiation stage; whereas during propagation, the crack propagation rate would only be altered when the crack was within the stress concentration field of influence and the embrittled region surrounding a pore. Therefore, under the previous assumptions, the only way Eq. (4) can be satisifed is for the quantity $dy/d\sigma$ to be negative, i.e., porosity must decrease the crack propagation stage by increasing amounts with increasing stress amplitude.

The conclusion regarding the effect of porosity on the fatigue crack propagation stage is not unreasonable, and can be rationalized by considering the effects of the elastic stress concentration fields surrounding the pores, the embrittled microstructural regions surrounding the pores, and finally the variation of the critical crack length with stress amplitude. The crack propagation rate will increase when the crack enters the higher stress fields and embrittled microstructures surrounding the pores, and therefore fewer cycles will be required to propagate the crack through these regions. The dimensional extent of the higher stress regions does not change with rising stress amplitude (Ref. 17). Neither, of course, does the extent of the embrittled microstructural regions. However, the critical crack length for total fracture decreases with increasing stress amplitude (Ref. 16). Consequently, the fraction of the total crack path in which the crack propagation rate is accelerated by porosity increases with increasing stress amplitude. Therefore, one would expect porosity to decrease the number of cycles required for crack propagation by increasing amounts with increasing stress amplitude.

Conclusions and Recommendations

This investigation has shown that under certain conditions porosity should not be regarded as a virtually harmless welding discontinuity. In the 2¼Cr-1Mo low alloy steel weldments examined in this study, approximately 5-8 v/o porosity *reduced* the fatigue life by 98% at 1200 F (648 C). Furthermore, the porosity generated in the laboratory was the same "porosity-type" as would be generated in production by inadequate shielding, air currents, etc. The data indicated that the fatigue mechanism per se was primarily responsible for the fatigue failures. Consequently the results were discussed in terms of the effects of porosity on fatigue crack initiation and propagation.

Other effects related to this investigation are presently under study. First, it is suspected that the injection of oxygen and compressed air into the arc atmosphere during welding not only generated porosity, but also probably modified the microstructure and/or chemial composition, and therefore the mechanical properties, of the weld metal surrounding the pores. This modification could have influenced the early fatigue failures. Consequently, studies of the porous weld metal microstructures are now underway. Another effect usually associated with high temperature mechancial behavior is also presently under study. It is suspected that the creep properties would be strongly influenced by the level of weld metal porosity. Consequently, creep tests are now being conducted on porous and nonporous 2¼Cr-1Mo weldments at 1000 F (538 C) and 1200 F (648 C).

Finally, in view of the fatigue mechanism discussion of the results of this study, it is suggested that more work be devoted to the effect of internal stress concentrators, e.g., pores, on fatigue crack propagation rates.

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"Fracture Toughness and Related Characteristics of the Cryogenic Nickel Steels"

by A. W. Pense and R. D. Stout

A large volume of data concerning the 2¼, 3½, 5 and 9% nickel steels, corresponding to ASTM specifications A203 Grades A and D, A353, A553 Type I, and A645, has been collected from the open literature and private sources. The mechanical-property data collected include tensile-test properties at ambient and cryogenic temperatures, notch toughness, fracture toughness and fatigue strength. A brief description of each of the testing methods used and their significance to cryogenic service are included. Tables and figures summarizing the data are presented. The strengths and toughnesses required by the respective ASTM steel specifications and ASTM specification A593 are normally met or exceeded by the steels. For the lower nickel steels, gains in toughness can be obtained by quenching and tempering rather than normalizing.

The effects of fabrication operations such as cold work, heat treatment and welding on the mechanical properties are considered. In general the effects of cold work and welding on the tensile properties of the cryogenic nickel steels are small. In the case of 2½ and 3½ nickel steels, the notch toughness is adversely affected by cold forming and aging but little influenced by welding operations. Subsequent thermal stress-relief treatments will restore the original toughness. Conversely, the notch and fracture toughness of the higher nickel steels are relatively insensitive to cold forming, but are reduced in weld heat-affected zones produced by high heat inputs. The fatigue properties of the cryogenic steels appear to fit the typical scatter band obtained for structural steels, and cryogenic crack-growth rates are not significantly different from those at room temperature.

An attempt has been made to evaluate the critical flaw sizes at design-stress levels for the steels as a function of temperature and stress concentration. For the lower nickel steels, it was concluded that A203 Grade A steel should be limited to service temperatures above -75 F (-59 C) and A203 Grade D should be limited to service temperatures above -110 F (-73 C). When quenched and tempered, the service temperature for these steels can be reduced by about 25 F (14 C). For the higher nickel steels, leak-before-break behavior may be predicted for most expected cryogenic applications. For example, the A353 and A553 Type I steels will meet this criterion at -320 F in 1.5 in. (38 mm) plate for a design stress of 25 ksi (172 MPa) and a stress concentration of $K_t = 2$. The A645 steel will meet these same requirements at a service temperature of -275 F (-170 C).

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