Evaluation of creep properties and fracture behaviour of 1Cr-1Mo-$\frac{1}{4}$V cast steel welded with 2$\frac{1}{4}$ Cr-1Mo electrodes

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Introduction

Low alloy ferritic steel 1Cr-1Mo-$\frac{1}{4}$V in cast condition finds extensive application because of good castability, reasonable cost and high temperature strength for components such as stop and throttle valve chests, interceptor chests, inlet branches, stub pipes and turbine casings in power plant equipment. This steel is designed for temperatures of continuous service upto 565°C. At a stress of 9 kgf/mm^2 and temperature of 565°C, it has a proven rupture life of about 100,000 hrs. In many cases this steel is required to be welded either during manufacturing or in situ at power plants and the welds are generally made manually using 2$\frac{1}{4}$Cr-1Mo or matching composition electrodes. The usual practice is to stress relieve the welds at around 700°C before putting into service. One of the serious problems associated with the welding of this steel is formation, growth and coalescence of cavities eventually leading to cracking at grain boundaries in the coarse grained region of the heat affected zone (HAZ) immediately adjacent to the fusion boundary. This problem of 'reheat' or 'stress-relief cracking' was first detected in 1950s in austenitic stainless steels and in low alloy steels in 1960s and soon assumed a major research status to study and understand the mechanisms responsible to such type of cracking.[1–6] It has been reported that the reheat cracking has been the cause of several costly plant outages[7]. An explanation, that is perhaps widely accepted, to reheat cracking is significant reduction in grain-boundary ductility during stress relief cycle or service due to either segregation of trace impurities or the precipitation processes e.g., grain boundary carbide precipitation. The reduction in ductility, perhaps, is to such an extent that it is insufficient to accomodate the plastic deformation associated with stress relaxation. It is generally believed that a 2$\frac{1}{4}$Cr 1Mo deposit is used for welding 1Cr1Mo$\frac{1}{4}$V steel since such a deposit though weaker than the parent metal in stress-rupture tests, has better rupture ductility than stronger weld metals.[2] A wide variation in the use of electrodes for welding 1Cr1Mo$\frac{1}{4}$V steel has been observed in that 2$\frac{1}{4}$Cr 1Mo finds extensive application in the USA and UK while matching composition electrodes in East European countries and Soviet Union. There have been instances also when high nickel containing electrodes are used for in situ welding of 1Cr1Mo$\frac{1}{4}$V steel as these electrodes do away with the cumbersome processes of pre-heating and post weld heat treatment. However, systematic grouping of results dealing with welding of 1CrMoV steel with electrodes of varying compositions, stress rupture and structural characteristics of such welds and studies concerning their cracking susceptibilities is not readily available in literature.

This paper describes part of an investigation into differences in stress-rupture characteristics and fractographic features of the zones representing parent metal, weld metal and weld joint from a commercial pure 1Cr1Mo$\frac{1}{4}$V cast steel welded with 2$\frac{1}{4}$Cr1Mo electrodes.

Materials

The material used in this programme was taken from a keel block of commercial purity 1Cr1Mo $\frac{1}{4}$V cast steel and the three stages of heat treatment given to the casting originally was as follows :

(a) Homogenise annealing at 1020°C for 8 hours, cooling in furnace upto 500°C and thereafter in air.
(b) Austenitising at 935°C for 14 hours, quenching in oil to a temperature of 500°C and then cooling in air.
(c) Tempering at 710°C for 13 hrs. and cooling in air.

The casting received in this heat-treatment had prior austenite grain size of about 0.030 mm.

The welding electrodes used were 2$\frac{1}{4}$ Cr 1Mo variety of Indian origin.

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Experiments and Results

The chemical analyses of the casting and electrodes are given in Table 1.

Welding

The casting after edge preparation as shown in Fig. 1, was welded with 2¼Cr1Mo electrodes by manual metal arc (MMA) process as per shopfloor practice. The welding parameters adopted are given in Table 2. The preheat and interpass temperature was 300-350°C. After the welding was completed, the casting was cooled slowly to room temperature under the cover of asbestos cloth. The welded casting was then subjected to stress-relieving treatment by heating at a rate of 50°C/hr to 680°C, holding at this temperature for 6 hrs. and then cooling in furnace.

The entire area of the welded casting was subjected to ultrasonic testing and was found to be free from defects of size 1mm or more.

Mechanical Properties

The room temperature mechanical properties of the parent metal, weld metal and the weld joint are given in Table 3.

Stress rupture testing

Stress rupture specimens were machined from the welded casting whose geometries are as shown in Fig. 2. The weld joint specimens consisted of parent metal, weld metal and HAZ in the gauge length. Stress rupture tests were run in air at constant load corresponding to initial stress of 23, 21, 16, 13.5 and 10 kgf/mm² at 550°C in automatic lever adjusting type machines with a temperature control of ±2°C. The results are given in Table 4.

The plots of log time to rupture (log tr) Vs stress (σ), rupture elongation (%E) and reduction in area (%RA) are given in Fig. 3.

The stress levels for expected rupture lives of 1000, 5000, 10000 and 20000 hrs were computed by regression analysis by fitting experimental values of tr and σ in the parabolic equation of second order,

$$Y=AX^2+BX+C \quad \ldots \quad (1)$$

These results are shown in Table 5. The values of multiple correlation coefficients (MCC) for each section of the results from the computer output is also shown in the Table.

Table 2 : Welding Parameters

Electrode Dia mm	3.15	4.00	5.00
Arc voltage, V	20	25	25
Current, Amps	120	180	200
Number of passes	16	140	64

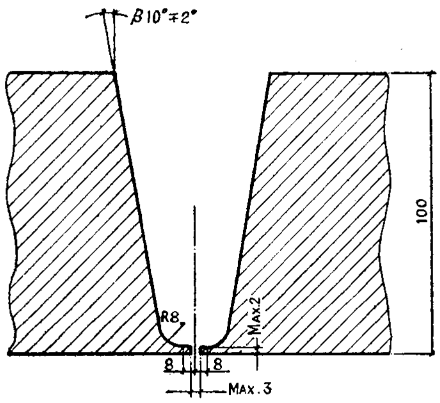

Fig. 1. Details of edge preparation

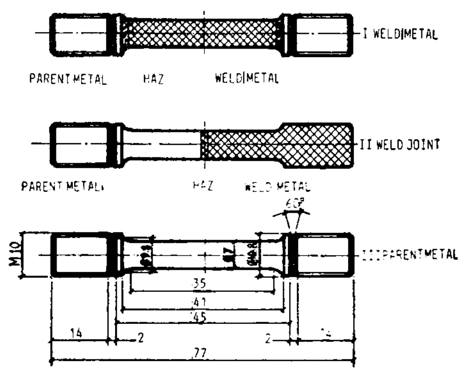

Fig. 2. Specimens for stress rupture testing

Table 1 : Chemical Analyses of Casting And Electrodes

Element, wt. %

	Ni	C	Si	Mn	P	S	Cu
Casting	0.08	0.17	0.42	0.95	0.016	0.010	0.04
Electrodes	—	0.06	0.25	0.63	0.017	0.011	0.044
	Cr	Mo	V	Al	As	Sb	Sn
Castings	1.30	0.93	0.23	0.005	0.003	0.0019	0.003
Electrodes	2.40	1.00	—	0.01	0.001	0.008	0.02

Table 3 : Mechanical Properties

Zone of Testing	0.2% Proof strength N mm²	Tensile strength N mm²	% Elongation (lo=25mm)	% Reduction in area	Charpy'V' Notch Impact energy joules
Parent metal	531	681	19	58	33
Weld metal	563.1	637.1	20.6	52.9	52.8
Weld joint	516.8	656.1	5.2*	46.2	20.0

*Fractured outside gauge length in the weld metal section.

Hardness Tests

Microhardness testing applying a load of 20 pounds was done in the bainitic regions of as received and ruptured specimens in the stress unaffected zone. The results are shown in Fig. 4. The results of Vickers Hardness testing at a load of 30 kg (Hv30) are given in Fig. 5.

Metallographic Examination

Metallographic specimens etched in Villella were examined under optical microscope. The microstructure of as received casting as shown in Fig. 6 consisted of a uniformly distributed admixture of upper bainite, matrix and grain boundary carbides. The microstructure of weld metal shown in Fig. 7 consisted of bainite laths and acicular ferrite. The size of carbides in weld metal was observed to be less than in parent metal while the grain size similar to each other indicating some refinement in the structure of weld metal during the course of welding. The structure of HAZ shown in Fig. 8 consisted of coarse prior austenite grains within which coarse bainite laths with little ferrite was observed. The prior austenite grain size in HAZ was found to vary from 0.115 to 0.125 mm.

Table 4 : Results of Stress Rupture Tests at 550°C in Air

	Rupture life. hrs.			Rupture Elogn., %			Reduction in area %		
Stress kgf/mm²	Parent metal	Weld metal	Weld joint	Parent metal	Weld metal	Weld joint	Parent metal	Weld metal	Weld joint
23	6.5	8	6.4	24	18	17	81	74	64
21	123	346	148	18	18	24	83	76	80
	194	234	—	26	24	—	78	75	—
	—	236	—	—	14	—	—	73	—
16	1536	2069	1724	27	13	21	83	56	76
	1106	1633	1586	39	14	23	83	62	82
13.5	5167	4494	4791	28	17	14	67	59	36
	5867	6924	5013	19	18	9	51	68	39
10	13805	15194	12987	29	14	5	60	53	16

Table 5 : Results of Regression Analysis

Expected Life Hrs	Values of computed stress in kg/mm² Parent metal	Weld metal	Weld Joint
1,000	17.27	18.10	17.40
5,000	13.30	13.72	13.20
10,000	11.27	11.41	11.04
20,000	9.05	8.86	8.65
M.C.C.	0.9981	0.9920	0.9988

Metallographic examination of stress ruptured specimens has been conducted and it was found that the structure in weld metal (Fig. 9) had recrystallized and carbides had coarsened, while the ferrite content and extent of carbides in parent metal had increased (Fig. 10). In the HAZ, heavy precipitation and coarsening of carbides occurred and a few recrystallized ferrite grains started appearing (Fig. 11). These factors indicate that there is a general trend of softening of the structure with time.

Metallographic examination of the weld joint specimens in the gauge length after rupture indicated recrystallization of HAZ, bainite to ferrite colonies in weld metal and parent metal. A tendency for carbides to go into solution in ferrite has been observed as a result of less population of carbides in the ferrite of gauge portion than in the stress unaffected region (specimen head). No cracks have been observed in HAZ regions or near fusion boundaries indicating that the weld joint, in question, did not become susceptible to reheat cracking either during stress-relief treatment or during the course of testing at 550°C.

Fractographic Examination

Scanning electron microscopic examination of fracture surfaces of stress ruptured specimens indicated that the weld metal failed by dimple mode with fine dimples nucleating around reprecipitated fine carbides but the parent metal developed a few smooth regions alongwith ductile dimple areas with increasing rupture life. The fractographic analysis, in general, indicated that the weld metal samples fractured invariably in a ductile mode while parent metal started showing signs of mixed mode in the long duration

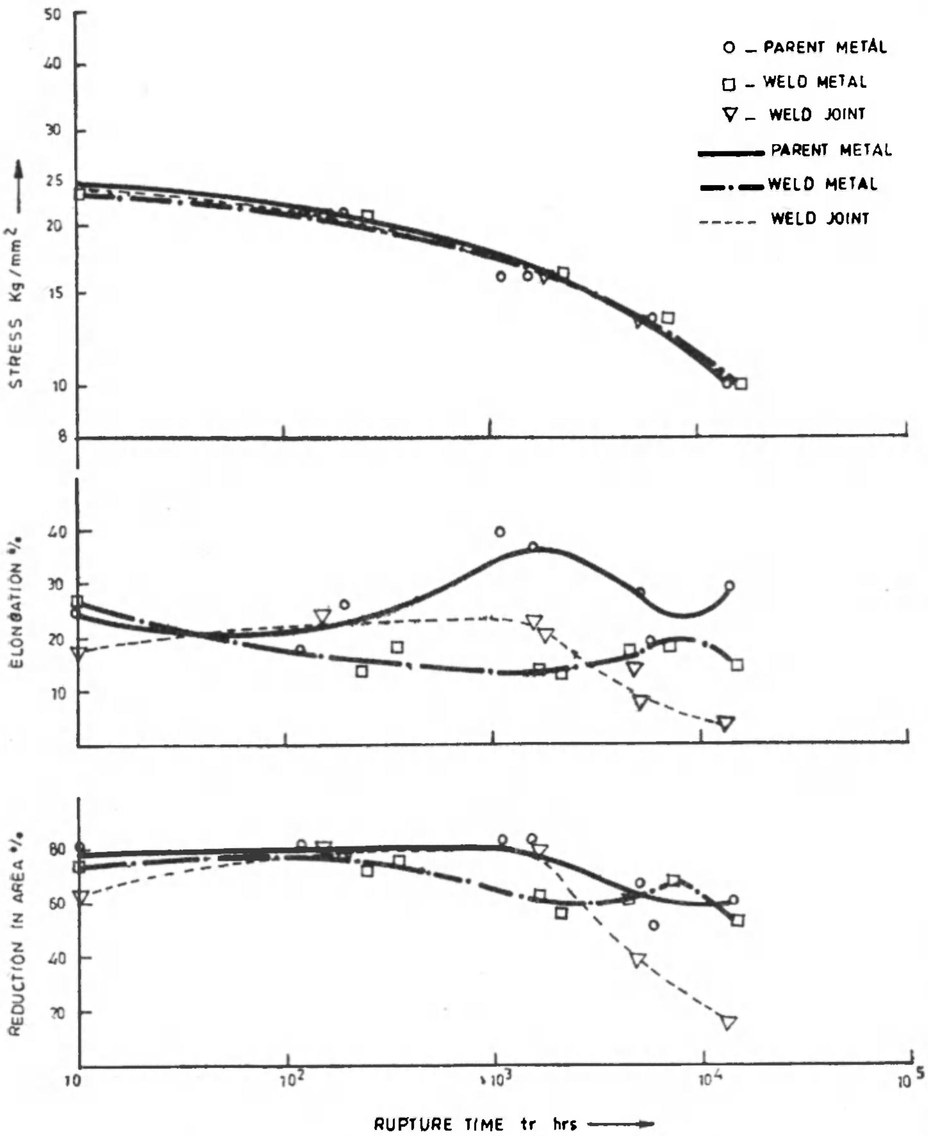

Fig 3. Rupture life as a function of stress and ductility

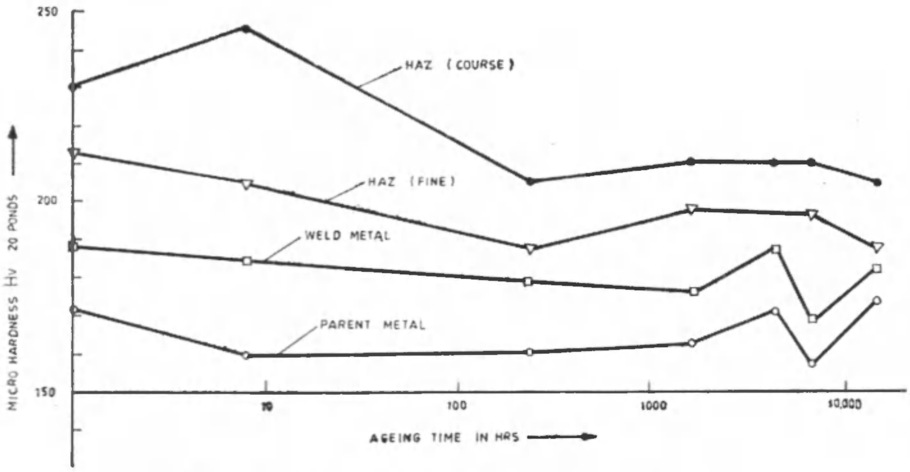

Fig 4. Variation of Micro-Hardness with ageing time

test. Typical fractures of weld and parent metals in stress rupture testing are shown in Figs. 12, 13, and 14.

Fracture of weld joint at low magnification is shown in Fig. 15, exhibiting very little reduction in gauge diameter. The fracture seen at higher magnificaton showed a combination of fine dimples and smooth regions in the weld joint.

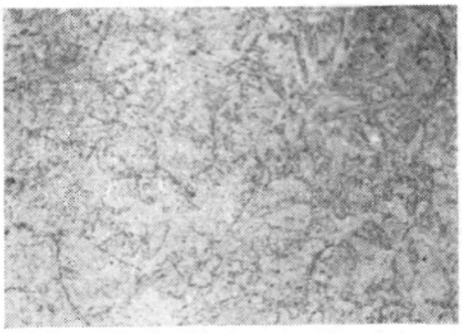

Fig 6. Keel block 1CrMoV steel casting, as received, vilella etchant X 250

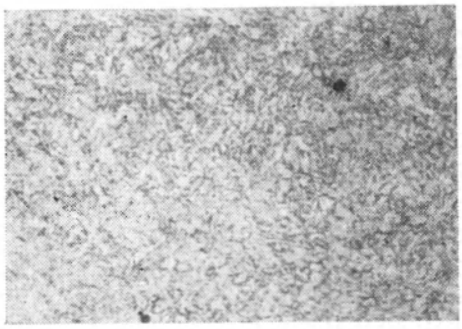

Fig 7. Weld-metal deposit, 2¼Cr1Mo steel, vilella etchant X 250

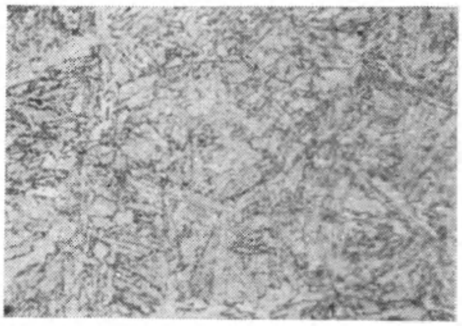

Fig 8. Coarse HAZ near fusion zone boundary, vilella etchant X 250

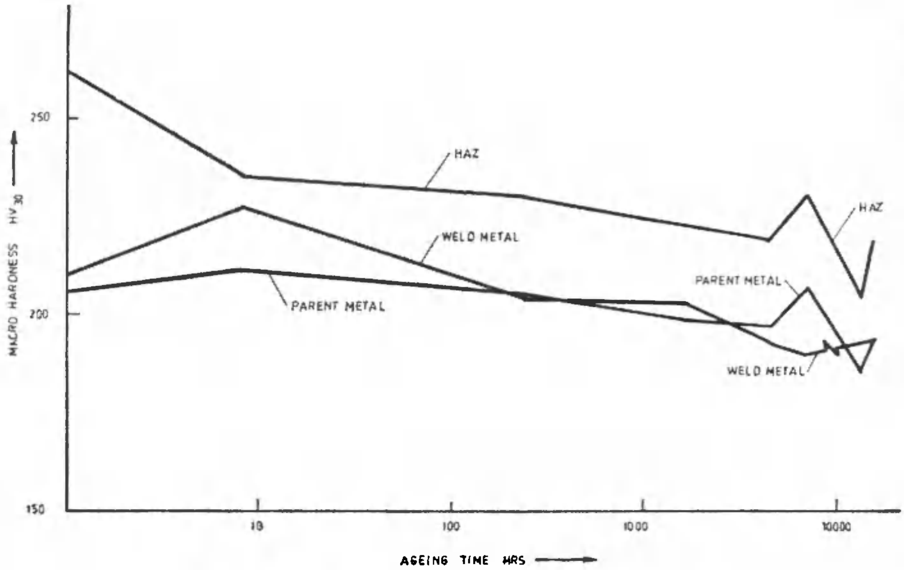

Fig 5. Variation of Macro-Hardness with ageing time

This feature, perhaps, explains the lower ductilities obtained in weld joint specimens in stress rupture testing.

Discussions

The stress rupture results of parent metal, weld metal and weld joint have not shown any significant variation. This observation indicates that the weld joint has compatible strength under the weld parameters followed in shop floor. The published rupture data show that $2\frac{1}{4}$Cr 1Mo weld metal has a lower creep strength than the $\frac{1}{2}$CrMoV parent metal for lives upto 10,000 hrs.[8]

It is expected that the variation in the strength between the two steels is due to variation in bainite content and the strengthening of bainite by fine distribution of V_4C_3 in CrMoV steel enhances its creep strength more than in $2\frac{1}{4}$Cr1Mo steel. The refinement in grain size and favourable precipitation sequence occurring during the course of welding in the weld metal, has, perhaps, caused

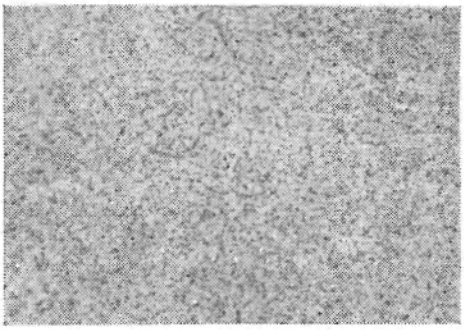

Fig 9.—X 250

Fig 10.—X 250

Fig 11.—X 100

Micro-structural changes in the stress unaffected regions of specimen stress ruptured at 550°C, 10 kg/mm² tr= 15194 hrs. (9) Weld deposit, (10) Parent metal (11) HAZ
Etchant : Vilella.

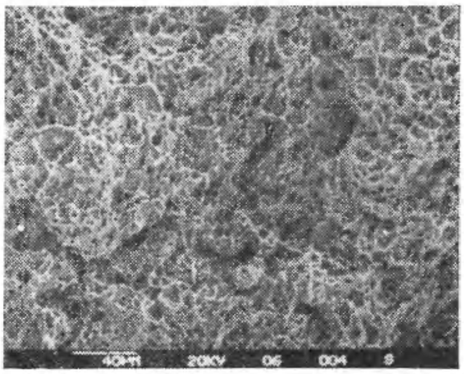

Fig 12. Fracture of weld-metal X 500
Fractured at 550°C, 10 kg/mm², tr= 15194 hrs.

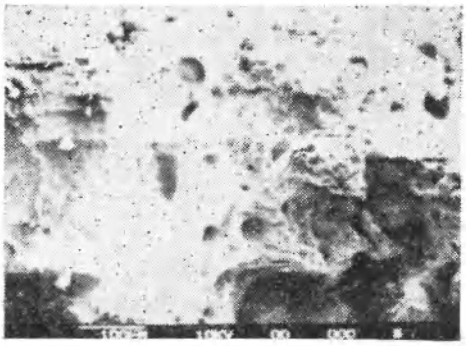

Fig 13. Fracture of Parent Metal X 200
Fractured at 550°C, 13.5 kg/mm², tr= 5867 hrs.

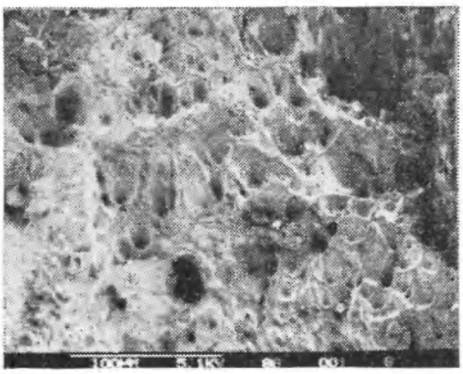

Fig 14. Fracture of Parent Metal X 450
Fractured at 55°C, 10 kg/mm², tr= 13807 hrs.

$2\frac{1}{4}$Cr1Mo to exhibit comparable rupture strength with that of 1CrMoV steel. In longer duration tests than those conceived in the present work, the weld metal may perhaps show incompatible rupture strength since this steel does not have the advantage of V_4C_3 precipitate and a high matrix dislocation density because of this carbide.

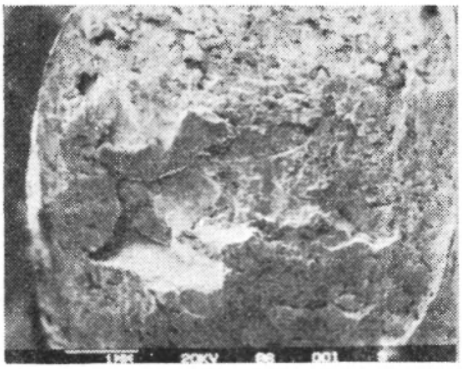

Fig 15. Fracture of weld joint in weld-metal region
Fractured at 550°C, 10 kg/mm², tr= 12987 hrs.

The observation made that in the weld joint specimens, the rupture occurred always in weld metal zone indicates that when the three sections are exposed to stress under similar conditions, weld metal (2¼Cr 1Mo) is the weakest. The individual ductilities exhibited by the parent and weld metals were higher than in the weld joint specimens. It appears that the width of the HAZ in the gauge length of weld joint specimens and the grain boundary chemistry in the weld metal near the fusion zone boundaries are contributing to lower rupture ductilities. However, a more detailed investigation in the fracture region of weld joints in relation to precipitation sequences may possibly explain the lower ductilities observed.

Conclusions

1. For a given set of welding parameters as per shop floor practice, the rupture lives of parent metal, weld metal and the weld joint of 1CrMoV cast steel welded with 2¼Cr 1Mo electrodes and tested at 550°C did not vary significantly.

2. The rupture ductilities of parent and weld metal are higher than those of the weld joint.

3. The microstructures of weld metal, parent metal and HAZ deteriorated at long-durations of testing, thus showing signs of softening.

4. Metallographic examination of stress ruptured samples from weld joint did not reveal cracking in HAZ or fusion zone boundary. Weld joint specimens fractured invariably in the weld metal regions indicating that the weld metal is the weakest in the weld joint.

5. Fractographic examination revealed that while parent metal and weld metal failed in a ductile manner exhibiting mostly a "dimple" mode of fracture, the weld joint fractured in a mixed mode with a few "smooth" regions alongwith fine dimples.

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References

1. F. E. Asbury, B. Mitchell and L. H. Toft, "Cracking in welded joints of Austenitic steel in CE-GB power stations", Brit, Welding J., November 1960, pp. 667-678.
2. F. Mullery and R. O. L. Cadman, "Cracking of welded joints in Ferritic Heat-Resisting Steels", Brit. Welding J., April 1962, pp. 212-220.
3. K. P. Bentley, "Precipitation During Stress Relief of Welds in Cr-Mo-V Steels", Brit. Welding J., Oct. 1964, pp. 507-515.
4. W. Barr and M. J. Siverns, "Hot Tensile properties of ½Cr-½Mo-¼V and 2¼Cr-1Mo steam pipe material", Brit. Welding J., May 1967, pp. 238-243.
5. C. F. Meitzner and A. W. Pense, "Stress-Relief Cracking in low-alloy steel weldments", Welding Research Supplement, Oct. 1969, pp. 4315-4405.
6. R. McPherson, "Research on Stress-Relief heat treatment of Weldments", Australian Weldding Journal, July/August 1974, pp. 25-29.
7. L. H. Toft and D. E. Yeldham, International Conference on Welding Research related to power plant, CEGB, March Wood, Southampton, 1972, pp. 5-19.
8. A. D. Batte and M. C. Murphy, "Creep rupture properties of 2¼Cr1Mo weld deposits", Welding J. Res. Supp., June 1973, pp. 261-267.

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The National Seminar on Welding for this year will be held at Madras during 16-18 October, 1986. First announcement and Call for Papers have been sent to all members. Contact address : Dr. Placid Rodriguez, Head-Metallurgy Programme, Indira Gandhi Centre for Atomic Research (IGCAR) Tamilnadu-603102