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MECHANICAL PROPERTIES OF HASTELLOY X AND INCONEL 617 AFTER AGING 53,000 HOURS IN HTGR-He

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H. E. McCoy

NOTICE: This document contains information of a preliminary nature. It is subject to revision or correction and therefore does not represent a final report.

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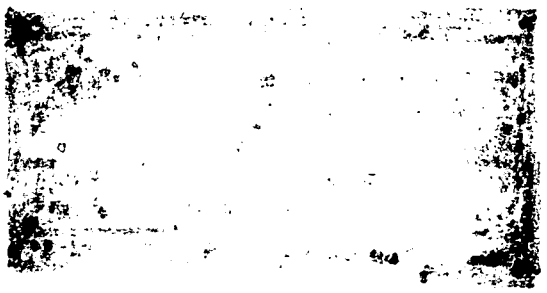

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MECHANICAL PROPERTIES OF HASTELLOY X AND INCONEL 617 AFTER AGING 53,000 HOURS IN HTGR-He*

H. E. McCoy

ABSTRACT

Twenty-four creep and tensile samples of Hastelloy X, Inconel 617, and each material containing a transverse weldment were aged 53,000 h at 593, 704, and 871°C in impure helium that simulated the coolant gas in a high-temperature gas-cooled reactor. Half of the samples were subjected to short-term tensile tests at 25°C. Large changes in properties were noted at an aging temperature of 593°C, and some overaging occurred at 704 and 871°C. The other half of the samples is being creep tested, with each sample tested at its aging temperature. This portion of the test program is in progress, and the trend thus far seems to be that the aged material is weaker than unaged material. Metallographic studies showed that large amounts of precipitate had formed and that these (1) formed continuous intergranular layers that caused brittle fracture and (2) depleted the matrix in solid-solution strengthening elements and thereby reduced the creep strength.

1. INTRODUCTION

Hastelloy X and Inconel 617 are candidate materials for high-temperature gas-cooled reactor (HTGR) applications. We have studied the tensile and creep properties of these materials for several years, and some of our findings were reported previously.[1-3] Both of these alloys are unstable at elevated temperatures, and long periods of heating can lead to significant changes in mechanical properties. In the gas-cooled reactor application a further complication is that the small amounts of methane and carbon monoxide in the service environment carburize Inconel 617 and Hastelloy X. This leads to increased precipitate formation within the alloys and can modify the properties even further.

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The property changes with time (often referred to as aging) need to be characterized to evaluate their effect on the service life of components made of these materials. We established facilities for aging test samples in HTGR-He (impure helium that simulates the coolant in an HTGR) from 593 to 817°C. Samples were removed at various time intervals for evaluation; the exposure time of the samples last evaluated was 53,000 h. The results of tests on these aged samples is discussed in this report.

2. EXPERIMENTAL DETAILS

2.1 MATERIALS

The test materials involved in this program are listed, with their chemical compositions and product forms, in Table 1. Two heats of Hastelloy X base metal and one heat of Hastelloy X weld wire were studied. Typical photomicrographs of the two heats of Hastelloy X base metal are shown in Figs. 1 and 2. Both heats contained carbide stringers, which are characteristic of this alloy. A weld was made by the gas tungsten arc (GTA) welding process, and samples were prepared so that the gage section contained base metal, filler metal, and heat-affected zone.

Table 1. Characterization of test materials
(All materials except weld wire were solution annealed by vendor)

Element	Chemical analysis (%)					
	Heat 4936[a,b]	Heat 4284[a,b]	Heat 4345[b,c]	Heat XX01A3US[a,d]	Heat XX14A6UK[a,d]	Heat XX09A9UK[d,e]
Nickel	Balance	Balance	Balance	Balance	Balance	Balance
Chromium	21.82	21.79	21.92	20.30	21.75	22.14
Cobalt	1.68	2.40	2.09	11.72	12.32	12.66
Molybdenum	9.42	8.82	8.85	8.58	8.91	8.79
Iron	19.09	19.06	18.81	1.01	0.53	0.39
Aluminum	0	0	0	0.76	1.11	1.37
Silicon	0.44	0.35	0.40	0.16	0.18	0.17
Carbon	0.07	0.06	0.08	0.07	0.06	0.08
Manganese	0.58	0.59	0.70	0.05	0.02	0.02
Sulfur	<0.005	<0.005	<0.005	0.004	0.002	0.001
Tungsten	0.63	0.63	0.42			

[a]Plate, 12.7 mm (1/2 in.).

[b]Hastelloy X.

[c]Wire, 8 mm (5/16 in.).

[d]Inconel 617.

[e]Wire, 1.1 mm (0.045 in.).

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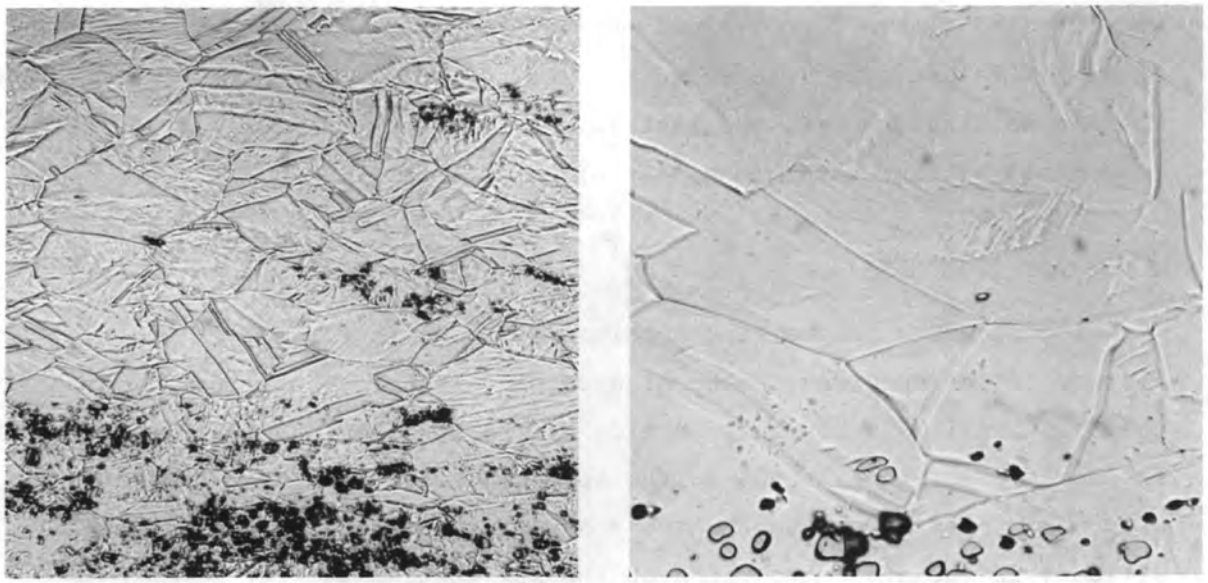

(*a*) 200 μm (*b*) 40 μm

Fig. 1. Longitudinal view of Hastelloy X (heat 4936) as received.

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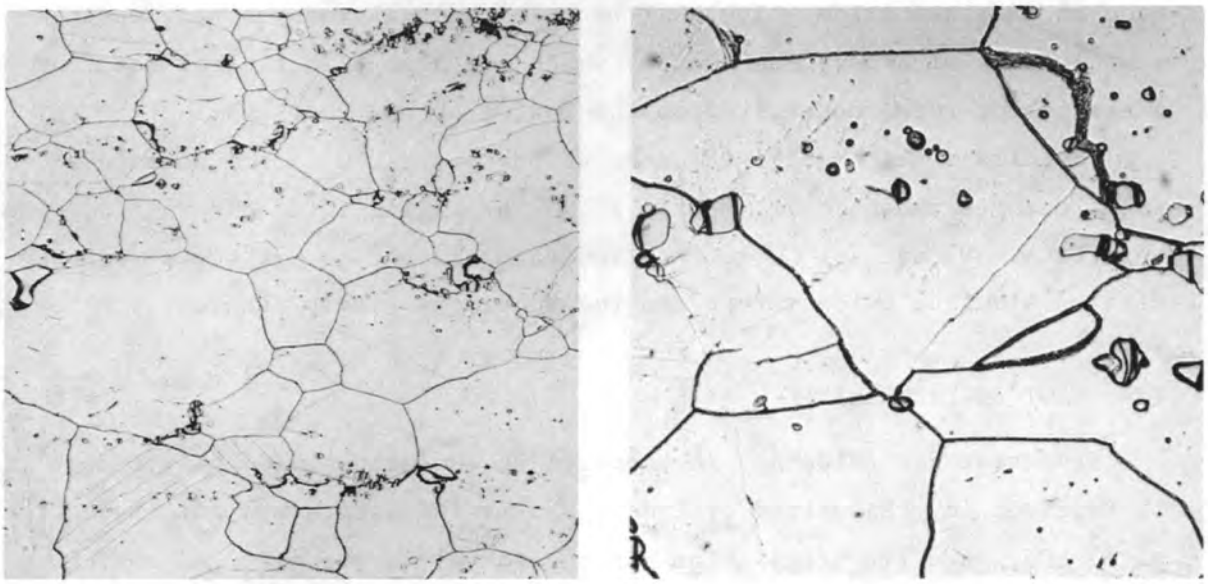

(*a*) 200 μm (*b*) 40 μm

Fig. 2. Longitudinal view of Hastelloy X (heat 4284) as received.

Two heats of Inconel 617 base metal and one heat of filler metal were studied (Table 1). Typical photomicrographs of the two heats of Inconel 617 base metal are shown in Figs. 3 and 4. Welds were prepared by the GTA welding process, and test samples crossed the weld and included base metal, filler metal, and heat-affected zone.

2.2 TEST SPECIMENS

The geometry of the test specimen is shown in Fig. 5. A good feature of the specimen is the absence of threads. The grips align on the 6.35-mm-diam (0.250-in.) portion of the specimen and pull against the surface inclined at 60° to the specimen axis. The extensometer attaches by set screws to the two grooves (which have an included angle of 100°) in the 6.35-mm portion of the specimen.

2.3 TESTING EQUIPMENT

The creep testing chambers used in this study were described in detail previously.[1] Most test retorts are made of 51-mm (2-in.) schedule 40 type 304L stainless steel pipe, but a few retorts, operated at 871°C, are made of aluminum oxide. The seal between the pull rods and the environment is made with neoprene U-cups, which are also used to seal the four extensometer rods that exit from the bottom of the test chamber.

Specimens were aged in HTGR-He at 593 and 871°C in the stainless steel retorts shown in Fig. 6 for 20,000 h. Aging is continuing in these retorts at 593 and 704°C, but the retort operating at 871°C was replaced with an aluminum oxide retort and the specimens simply stacked in place.

2.4 TEST ENVIRONMENT

The test gas (HTGR-He) is premixed by the Matheson Company, Inc. It is received in pressurized cylinders and is fed into a manifold maintained at 83 kPa, gage (12 psig). The piping and valves are arranged so that parallel streams of gas can be passed through each test retort and either returned to a sample manifold or vented. The gas stream can be passed

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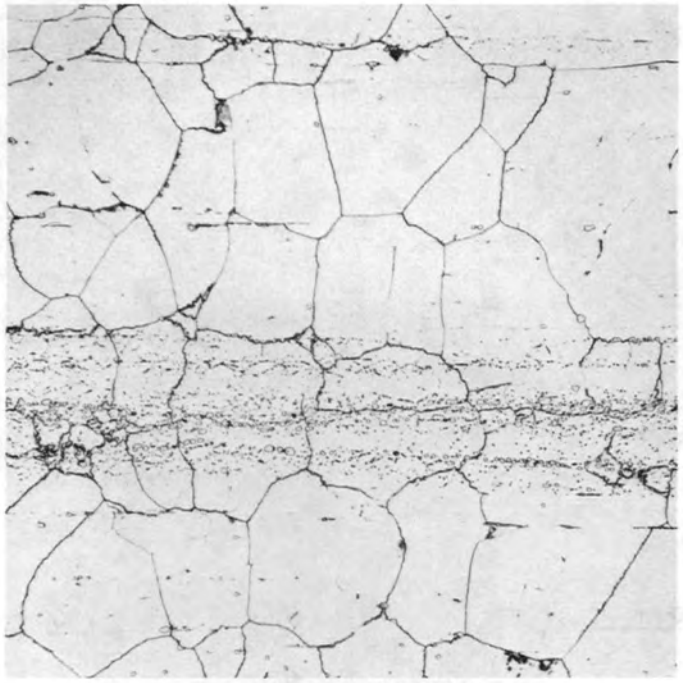

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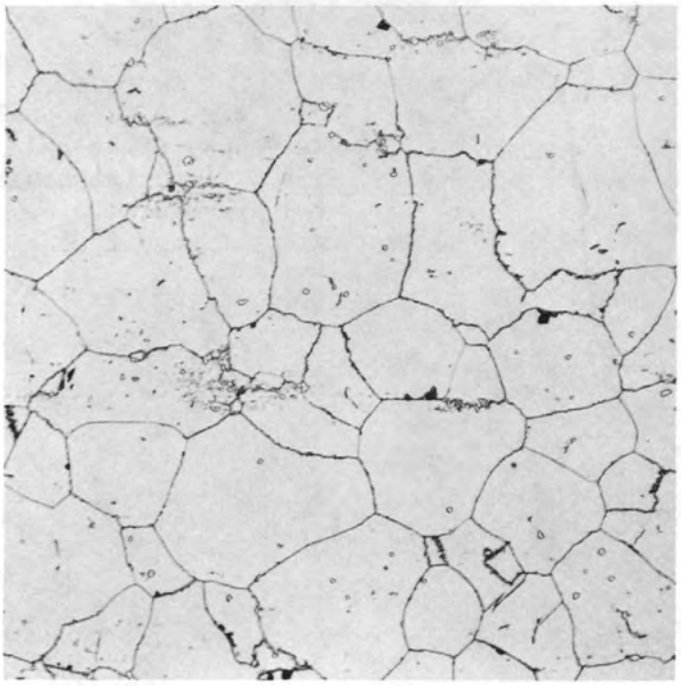

(*b*) 200 μm

Fig. 3. Inconel 617 (heat XX01A3US) as received. (*a*) Longitudinal view. (*b*) Transverse view.

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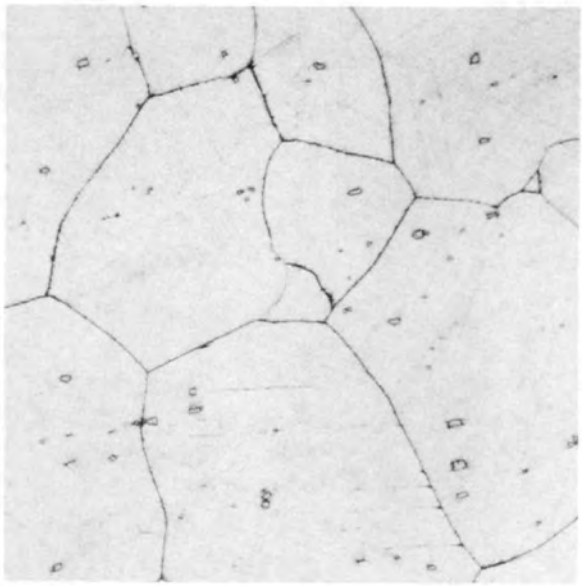

(*a*) 200 μm

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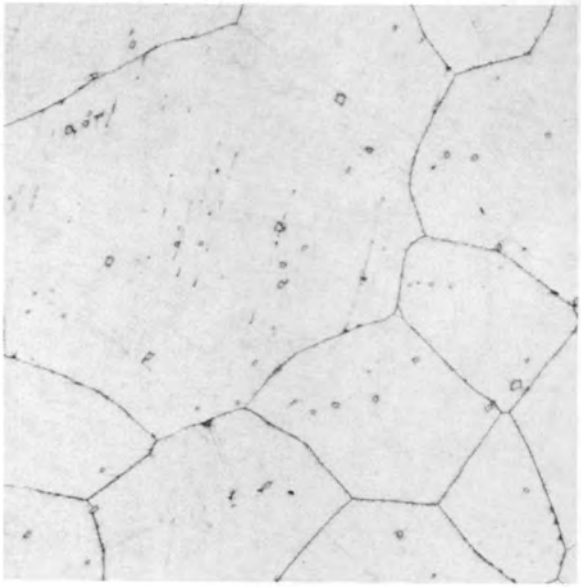

(*b*) 200 μm

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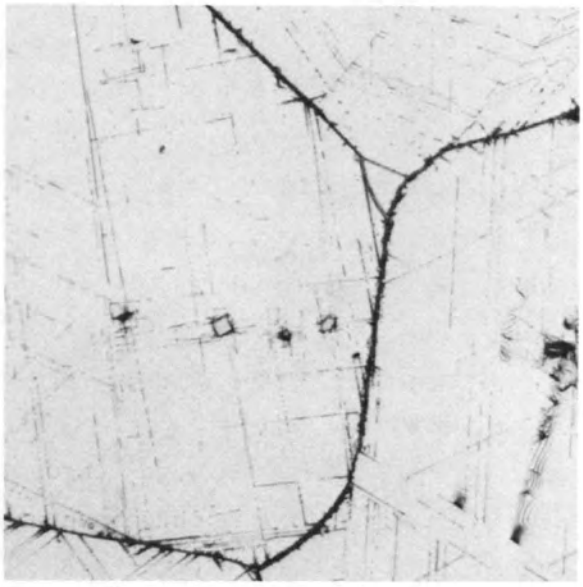

(*c*) 40 μm

Fig. 4. Inconel 617 (heat XX14A6UK) after aging 10,000 h at 593°C. (*a*) Longitudinal view. (*b*) Transverse view. (*c*) Transverse view at higher magnification.

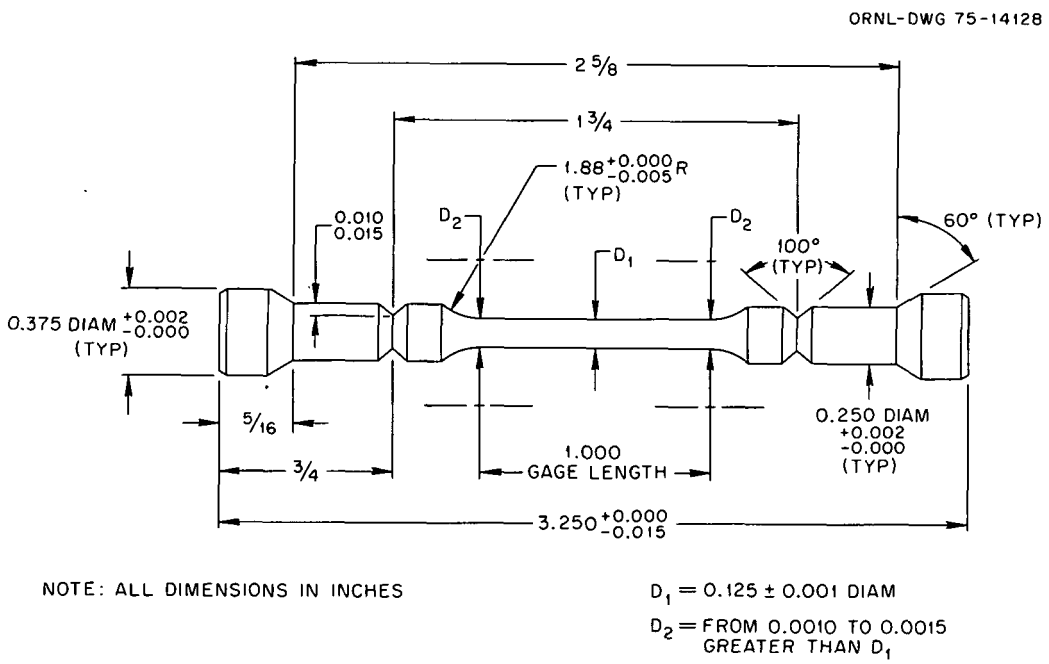

Fig. 5. Details of creep specimens for environmental tests. To convert dimensions to millimeters, multiply by 25.4.

through a gas chromatograph and the moisture monitored either before or after it passes through the test retort. The gas flow to each retort is maintained at 15 to 100 cm^3/min, with the criterion being to use a high enough flow rate to keep the gas stream from being detectably depleted in reactants. The helium composition in the test chamber is [in pascals (microatmospheres)] 34.0 (337) H_2, 3.2 (32) CH_4, 1.9 (19) CO, 0.2 (2) H_2, and less than 0.05 (<0.5) N_2. Oxygen is removed by reaction with the H_2 as the gas passes through a furnace at 500°C.

The on-line analytical system consists of a 2110 Bendix gas chromatograph and a model 2100 Panametrics moisture analyzer. The chromatograph is calibrated by use of a certified standard prepared by the Matheson Company, Inc., and weekly readings are made routinely on each machine. Manufacturer's calibrations for the moisture monitors were used without verification. On several occasions monitors were placed in series and agreed within a factor of 2. Equilibrating the moisture probes (and the remainder of the system) to read low moisture values is so difficult that about 24 h is required to obtain reliable values.

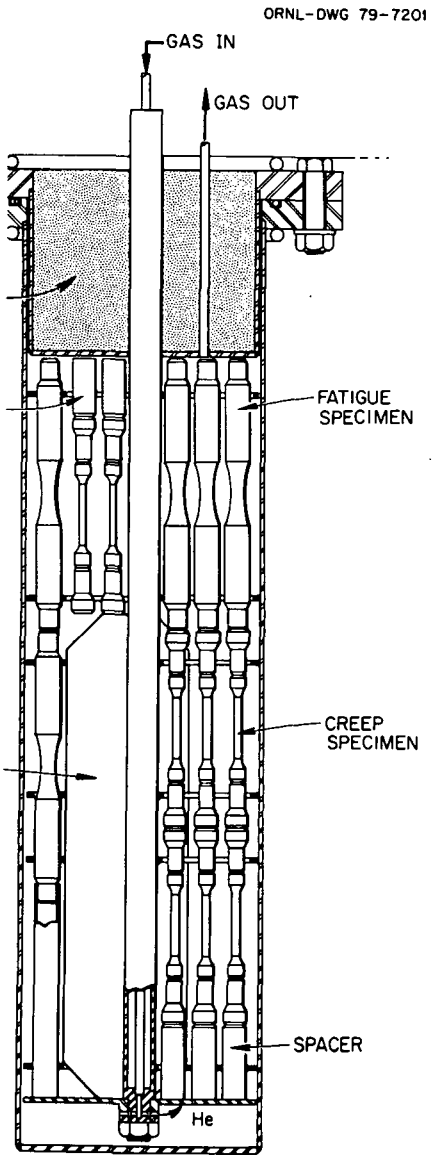

Fig. 6. Environmental exposure chamber.

The standard test start-up consisted of assembling the new test specimen in the chamber; evacuating air from the chamber; pressuring the chamber with test gas and establishing flow at 83 kPa, gage (12 psig); leak checking with a helium sniffer; heating to 400°C and holding for at least 24 h until moisture content was less than 10 ppm; heating to test temperature; and applying load to start the test. We found that all impurities were initially high but that the outgassing period at 400°C

allowed the gas composition to reach the desired operating level. Some elevation in impurity levels (less than a factor of 2) occurred during heating to the test temperature, but the gas reached the desired operating levels within 24 h.

3. RESULTS OF POST-AGING CARBON ANALYSES

The results of carbon analyses on the aged Hastelloy X samples are given in Tables 2 and 3. In all results presented, the analytical samples were duplicate wafers from the gage section and were totally consumed in the analysis. Thus, the results are averages of the entire 3-mm-diam cross section. The surface area in the aging vessels is quite large, so some depletion of the reactive gases likely occurs, particularly at 871°C. Data are shown in Fig. 7 from single creep samples for comparison.[1] The data in Fig. 7 extended only to 7500 h, but they show no detectable carburization at 649, 704, and 760°C, slight carburization at 816°C, and carburization to levels of 0.15% after 7500 h at 871°C. By comparison, the data in Table 2 for the aged samples show no consistent evidence of carburization at 593°C, slight carburization at 704°C, and significant carburization at 871°C. The depletion effect at 871°C is apparent from

Table 2. Carbon analysis (wt %) of Hastelloy X (heat 4936)

Time (h)	Aging temperature (°C)		
	593	704	871
As received	0.070		
10,000	0.084	0.093	0.083
20,000		0.097	
53,000	0.076	0.086	0.134

Table 3. Carbon analysis (wt %) of Hastelloy X

Time (h)	Aging temperature (°C)		
	593	704	871
As received	0.060[a]		0.080[b]
53,000	0.083		0.161

[a]Transverse weld, heat 4284, base metal.

[b]Heat 4345, filler metal.

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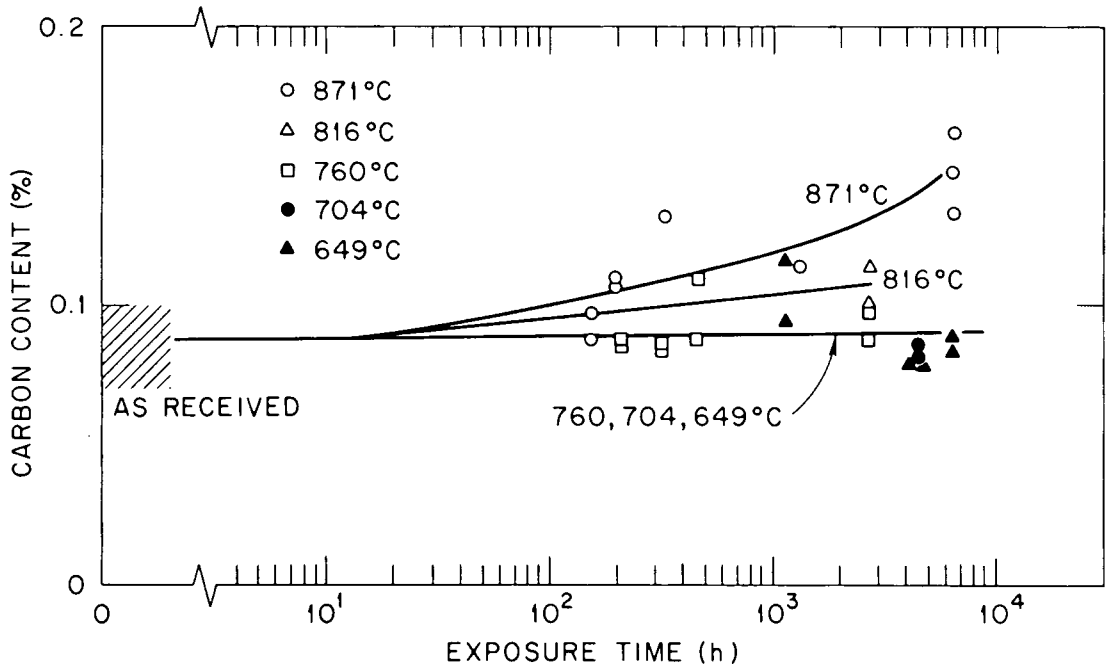


Fig. 7. Carbon content of Hastelloy X exposed to HTGR-He.

the single sample in Fig. 7, which had a carbon content of 0.15% after 7500 h, and the aged sample in Table 2, which had a carbon content of only 0.083% after 10,000 h. Thus the samples aged, particularly at 871°C, will show the full effects of thermal exposure but not the full effects of the carburization that would occur in an undepleted environment.

The carbon results for aged transverse weld samples of Hastelloy X are shown in Table 3. There are only two data points, and those are for samples aged 53,000 h. The analytical samples were taken from the base metal. The results are quite similar to those noted in Table 2 for another heat of base metal; that is, there was little carburization at 593°C and considerable carburization at 871°C.

The carbon concentrations of some of the aged Inconel 617 samples are given in Tables 4 and 5. Data for single creep specimens of Inconel 617 are shown in Fig. 8 for comparison. The data for single specimens show no evidence of carburization at 593°C, and the samples aged at 593°C show no evidence of carburization. Of the single specimens, there is one specimen

Table 4. Carbon analysis (wt %) of Inconel 617 (heat XX1A3US)

Time (h)	Aging temperature (°C)		
	593	704	871
As received	0.070		
10,000	0.065	0.075	0.065
20,000	0.076	0.076	0.080
53,000	0.056	0.065	0.127

Table 5. Carbon analysis (wt %) of Inconel 617

Time (h)	Aging temperature (°C)		
	593	704	871
As received		0.080[a]	0.060[b]
53,000	0.057		0.141

[a]Heat XX09A9UK, filler metal.

[b]Transverse weld, heat XX14A6UK, base metal.

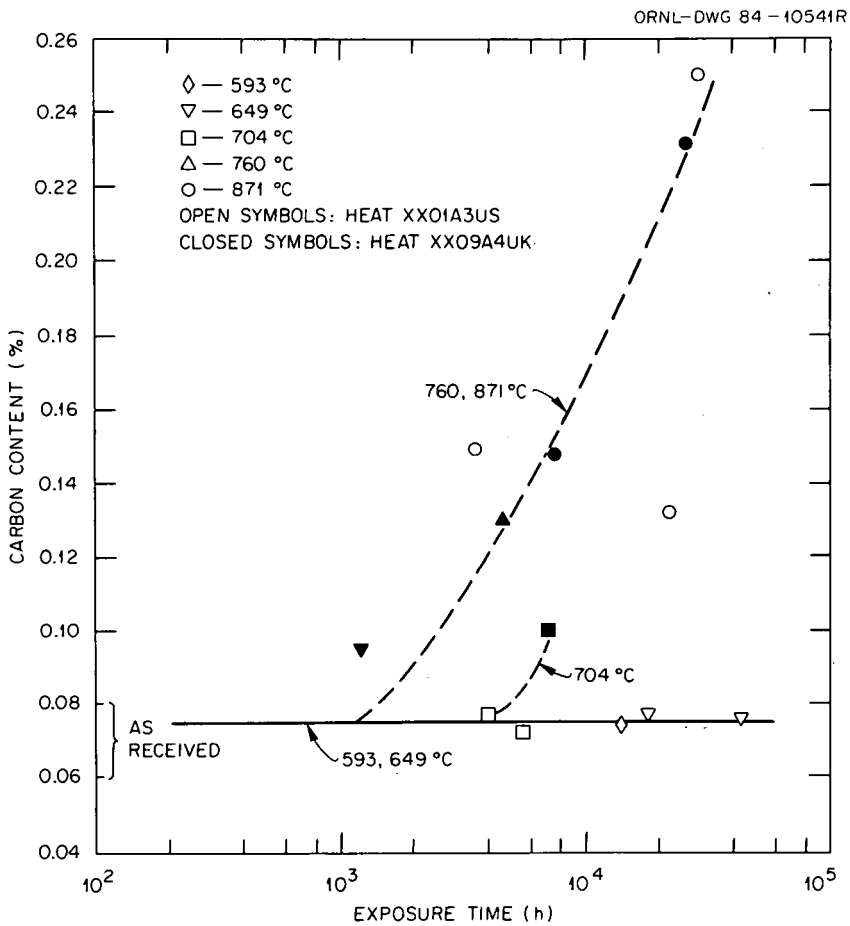


Fig. 8. Post-test carbon concentrations of Inconel 617 creep samples.

at 704°C that appears to be carburized after 7000 h exposure, but the samples aged at 704°C do not show carburization after 53,000 h. At 871°C both the single creep samples (Fig. 8) and the aged samples (Tables 4 and 5) were carburized, but the single samples carburized at a higher rate.

4. RESULTS OF TENSILE TESTS

4.1 HASTELLOY X BASE METAL AND TRANSVERSE WELD SAMPLES

Tensile test results are summarized in Table 6 for Hastelloy X base metal (heat 4936). These test samples were aged in an inert or HTGR-He environment for times up to 53,000 h. The influence of aging on the yield and ultimate tensile strengths at 25°C is shown in Fig. 9. After an aging time of 20,000 h, the strength properties were not altered appreciably whether the aging environment was HTGR-He or inert gas. At an aging temperature of 704°C, the yield and ultimate tensile strengths were increased significantly by aging in inert gas or HTGR-He, but the strength increase was greater for HTGR-He than for inert gas. The samples aged 53,000 h were aged in HTGR-He, so there is no basis for comparing the effects of various aging environments. One interesting effect observable from the samples aged 53,000 h is an overaging tendency at 704 and 871°C, which resulted in the strengths of the samples aged 53,000 h being less than those of the samples aged 20,000 h. At an aging temperature of 593°C, the strengths of the samples aged for 53,000 h were greater than those of the samples aged 20,000 h.

The reductions in area of the samples just discussed are shown in Fig. 10. Aging times up to 20,000 h at 538 or 593°C in HTGR-He or inert gas did not have much influence on the reduction in area. Aging at 704 and 871°C for 20,000 h caused a significant decrease in the reduction in area, with the magnitude of the effect being greater at 704 than at 871°C and greater in HTGR-He than in inert gas. The reduction in area of samples aged 53,000 h showed the same tendency toward overaging as that observed on the basis of yield and tensile strengths. At 871 and 704°C

Table 6. Tensile properties of as-received and aged Hastelloy X base metal (heat 4936)

Test	Aging conditions				0.2% Yield strength		Ultimate tensile strength		Elongation (%)		Reduction in area (%)	Type specimen[b]
	Temperature		Time (h)	Environment[a]					Uniform	Total		
	(°C)	(°F)			(MPa)	(ksi)	(MPa)	(ksi)				
Tested at 22°C (72°F)												
14892			0		359	52.0	762	110.5	43.8	50.5	59.8	A
14893			0		349	50.6	764	110.8	43.9	51.5	59.2	A
14894			0		350	50.8	764	110.8	44.3	51.9	61.4	A
X6-14			0		334	48.4	767	111.3	45.6	49.9	49.9	B
17885	538	1000	2,500	I	333	48.3	764	110.8	49.5	55.5	54.4	A
17886	538	1000	2,500	I	340	49.3	766	111.2	56.8	54.4	50.0	A
17889[c]	704	1300	2,500	I	482	69.9	916	132.9	7.7	7.7	8.7	A
17890[c]	704	1300	2,500	I	476	69.0	939	136.1	9.8	9.8	8.1	A
17893	871	1600	2,500	I	367	53.3	835	121.1	15.8	15.8	14.7	A
17894	871	1600	2,500	I	368	53.4	818	118.7	13.4	13.4	12.3	A
XC7	538	1000	10,000	I	385	55.9	777	112.7	43.1	43.2	39.7	A
XC25	704	1300	10,000	I	357	51.8	854	123.8	18.2	18.2	16.1	A
XC43	871	1600	10,000	I	399	57.9	732	106.2	15.0	15.1	10.5	A
Tested at 24°C (75°F)												
X-100	593	1100	10,000	H	385	55.9	770	111.7	40.2	43.4	49.5	B
X-104	704	1300	10,000	H	579	83.8	1,054	152.9	10.8	10.8	14.0	B
X-108	871	1600	10,000	H	401	58.1	846	122.7	10.0	10.0	9.3	B
XC14	538	1000	20,000	I	355	51.5	787	114.1	44.8	46.8	36.9	B
XC32	704	1300	20,000	I	350	50.7	832	120.7	9.5	9.5	8.1	B
XC50	871	1600	20,000	I	325	47.2	791	114.7	18.0	18.0	14.5	B
X113	593	1100	20,000	H	398	57.7	807	117.1	23.2	23.7	29.7	B
X117	704	1300	20,000	H	540	78.3	1,098	159.3	8.2	8.2	3.5	B
X121	871	1600	20,000	H	325	47.2	779	113.0	10.6	10.6	13.7	B
X115	593	1100	53,000	H	540	78.4	1,048	152.0	11.6	11.6	15.9	B
X119	704	1300	53,000	H	475	68.9	939	136.2	9.6	9.6	10.6	B
X123	871	1600	53,000	H	290	42.0	673	97.6	13.7	13.7	14.7	B

Table 6. (continued)

Test	Aging conditions				0.2% Yield strength		Ultimate tensile strength		Elongation (%)		Reduction in area (%)	Type specimen[b]
	Temperature		Time (h)	Environment[a]								
	(°C)	(°F)			(MPa)	(ksi)	(MPa)	(ksi)	Uniform	Total		
					Tested at 538°C (1000°F)							
14807			0		239	34.6	592	85.8	46.5	49.9	40.9	A
14808			0		221	32.1	596	86.5	43.9	48.8	31.4	A
17887[d]	538	1000	2,500	I	222	32.2	603	87.5	54.0	58.6	48.0	A
17888[c,d]	538	1000	2,500	I	227	32.9	601	87.2	49.3	52.7	35.5	A
XC8[d]	538	1000	10,000	I	236	34.2	600	87.0	48.6	48.7	42.9	A
XC15	538	1000	20,000	I	226	32.8	580	84.1	46.8	47.7	41.7	A
					Tested at 593°C (1100°F)							
X-101	593	1100	10,000	H	176	25.5	521	75.6	62.3	65.2	44.4	B
X-112	593	1100	20,000	H	292	42.3	664	96.3	20.9	21.4	21.2	B
					Tested at 704°C (1300°F)							
14799			0		223	32.4	425	61.6	23.0	48.5	41.7	A
14800			0		215	31.2	452	65.5	29.9	37.3	33.6	A
14801			0		212	30.8	460	66.7	30.8	37.3	35.9	A
17891	704	1300	2,500	I	294	42.6	443	64.3	6.3	58.1	55.0	A
17892	704	1300	2,500	I	284	41.2	429	62.2	5.9	74.2	58.2	A
XC26	704	1300	10,000	I	230	33.3	439	63.7	11.1	36.3	37.5	A
X-105	704	1300	10,000	H	359	52.1	588	85.3	8.6	21.6	29.0	B
XC33	704	1300	20,000	I	249	36.1	420	60.9	10.4	47.3	49.1	A
X-116	704	1300	20,000	H	368	53.4	590	85.6	3.7	25.3	28.1	B

Table 6. (continued)

Test	Aging conditions				0.2% Yield strength		Ultimate tensile strength		Elongation (%)		Reduction in area (%)	Type specimen[b]
	Temperature		Time (h)	Environment[a]								
	(°C)	(°F)			(MPa)	(ksi)	(MPa)	(ksi)	Uniform	Total		
Tested at 871°C (1600°F)												
14815			0		161	23.4	165	24.0	2.1	79.8	88.9	A
14898			0		163	23.6	168	24.3	2.3	84.4	88.2	A
17895	871	1600	2,500	I	146	21.2	161	23.4	2.5	96.4	82.4	A
17896	871	1600	2,500	I	141	20.4	159	23.0	2.5	101.0	86.3	A
XC43	871	1600	10,000	I	151	21.9	162	23.5	5.8	78.5	94.5	A
X-109	871	1600	10,000	H	211	30.6	239	34.7	5.0	59.4	68.9	B
XC51	871	1600	20,000	I	188	27.3	232	33.7	6.4	74.4	70.2	A
X-120	871	1600	20,000	H	191	27.7	249	36.1	1.8	46.0	47.8	B

[a]I, inert; H, HTGR-He.

[b]A, 6.4 mm (0.25 in.) in diameter by 3.18 cm (1.25 in.) long; B, 3.2 mm (0.125 in.) in diameter by 2.54 cm (1.00 in.) long.

[c]Broke at gage mark.

[d]Serration occurred before 0.2% yield.

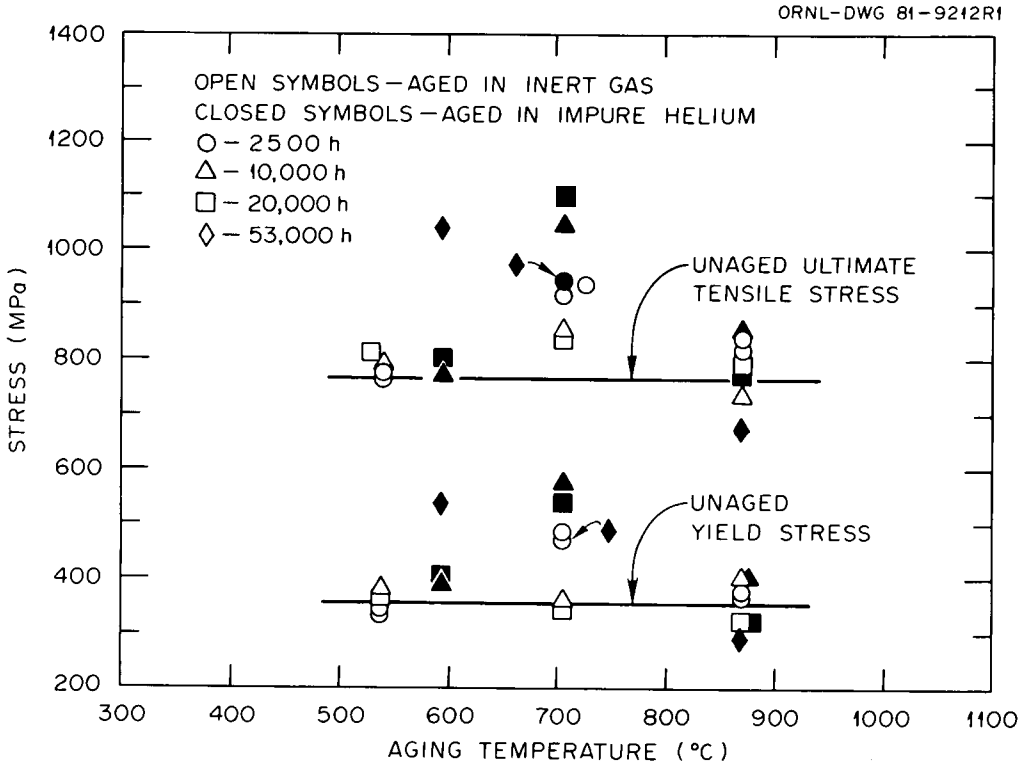


Fig. 9. Influence of aging on the yield and ultimate tensile stress of Hastelloy X base metal (heat 4936) at 25°C.

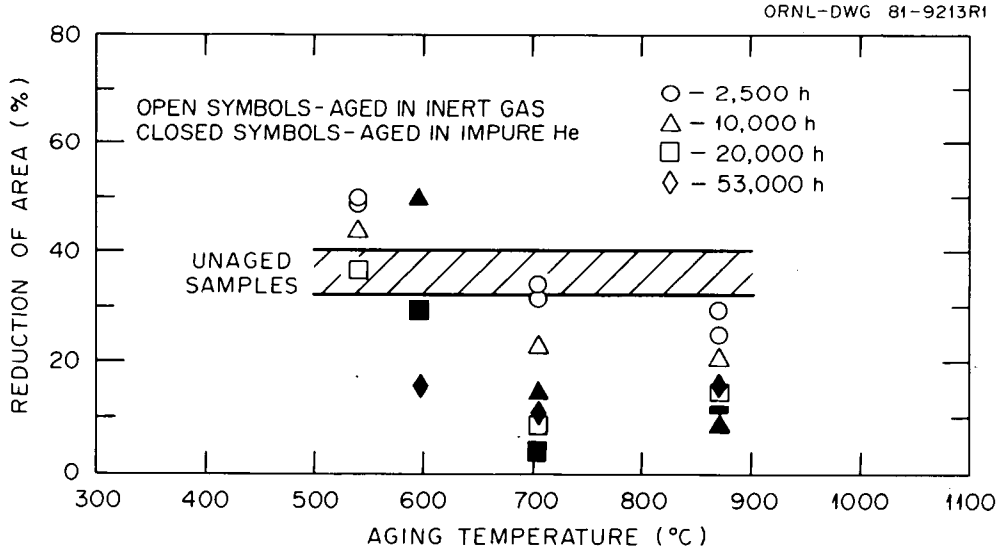


Fig. 10. Influence of aging on the reduction in area of Hastelloy X (heat 4936) tested at 25°C.

the reduction in area was higher after aging for 53,000 h than after aging for 20,000 h. At 593°C the reduction in area continued to decrease with increasing aging time from 10,000 to 53,000 h.

Transverse weld samples of Hastelloy X were also evaluated, and the results are shown in Table 7. The transverse weld samples had slightly higher yield and ultimate tensile strengths than the base metal had; however, failure was usually in the weld metal. The locations of the failures are shown in Table 7, and the reduction in area of the weld was measured even though failure may have occurred in the base metal. The influence of aging on the yield and ultimate tensile strengths at 25°C are shown in Fig. 11. The behavior noted for the transverse welds in Fig. 11 is very similar to that noted for the base metal in Fig. 9. Up to 20,000 h aging, the optimum degree of strengthening occurred from aging in the 650 to 704°C temperature range. Solid lines are shown for the shortest aging time of 2000 h and dashed lines for the longest time of 53,000 h. Most aging treatments at 871°C led to softening. However, aging for 53,000 h in HTGR-He caused large increases in the strength after aging at 593°C and decreases in the strengths due to aging at 704 and 871°C. The fracture strain was reduced by all aging treatments. The strain reductions were greatest after aging at 871°C, but many aging conditions resulted in reduction in area values of 10% or less. A comparison between Fig. 10 for the base metal and Fig. 12 for the transverse weld samples shows that the weld samples failed at slightly lower ductilities than did the base metal.

4.2 INCONEL 617 BASE METAL AND TRANSVERSE WELD SAMPLES

Tensile results for heat XX01A3US are given in Table 8. All of the results shown are for samples aged in HTGR-He. The influence of aging on the yield and ultimate tensile strengths is shown in Fig. 13. Generally the yield strength is increased by aging at 593 and 704°C, but only slightly by aging at 871°C. Aging 53,000 h at 871°C in HTGR-He actually results in a slight reduction in the yield strength. The ultimate tensile strength was increased by aging at 593°C. Aging at 704°C resulted in some strengthening initially, but the material overaged so that its strength

Table 7. Tensile properties of Hastelloy X welded with Hastelloy filler metal by the gas tungsten arc process

Test	Aging temperature		Test temperature		0.2% Yield stress		Ultimate tensile stress		Elongation (%)		Reduction in area (%)	Location of fracture[a]	Weld reduction in area (%)
	(°C)	(°F)	(°C)	(°F)	(MPa)	(ksi)	(MPa)	(ksi)	Uniform	Total			
As welded													
16703			25	77	456	66.2	725	105.2	38.0	40.0	44.1	W	44.1
16704			593	1100	304	44.1	548	79.5	33.0	33.5	52.2	W	44.1
16705			659	1200	308	44.6	535	77.6	36.0	36.8	43.8	W	43.8
16706			760	1400	283	41.0	443	64.2	15.0	30.9	36.4	B	33.0
16707			871	1600	248	35.9	262	38.0	5.0	26.0	79.7	W	79.7
Aged 2000 h in inert environment													
16776	593	1100	25	77	492	71.3	855	124.0		28.8	35.4	B	30.8
16778	649	1200	25	77	720	104.4	1,112	161.3		5.1	13.5	W	13.5
16780	760	1400	25	77	557	80.8	996	144.4		8.6	11.1	W	11.1
16782	871	1600	25	77	376	54.6	784	113.7		12.8	22.8	W	22.8
16777	593	1100	593	1100	334	48.4	628	91.1		25.2	44.7	B	23.3
16779	649	1200	649	1200	524	76.0	695	100.8		9.8	30.5	W	30.5
16781	760	1400	760	1400	351	50.9	437	63.4	9.0	21.3	56.2	W	56.2
16783	871	1600	871	1600	198	28.7	244	35.4	7.0	49.8	75.4	W	75.4
Aged 10,000 h in inert environment													
122-9	593	1100	25	77	511	74.1	820	118.9	17.8	18.5	37.5	B	3.4
122-13	649	1200	25	77	590	85.6	1,113	161.4	6.1	6.1	10.3	W	10.3
122-17	760	1400	25	77	480	69.6	859	124.6	6.5	6.5	9.2	W	9.2
122-21	871	1600	25	77	348	50.4	700	101.6	10.6	10.6	16.5	W	16.5
122-10	593	1100	593	1100	359	52.0	666	96.6	19.6	20.4	39.8	B	5.1
122-14	649	1200	649	1200	516	74.8	704	102.1	6.4	12.7	36.7	W	36.7
122-18	760	1400	760	1400	330	47.8	450	65.3	8.0	23.1	53.6	W	53.6
122-22	871	1600	871	1600	197	28.5	228	33.1	8.0	38.5	66.7	W	66.7

Table 7. (continued)

Test	Aging temperature		Test temperature		0.2% Yield stress		Ultimate tensile stress		Elongation (%)		Reduction in area (%)	Location of fracture[a]	Weld reduction in area (%)
	(°C)	(°F)	(°C)	(°F)	(MPa)	(ksi)	(MPa)	(ksi)	Uniform	Total			
Aged 10,000 h in HTGR-He environment													
X-167	593	1100	25	77	425	61.7	740	107.4	30.2	30.7	35.9	W	35.9
X-171	704	1300	25	77	678	98.3	1,060	153.7	4.1	4.1	12.1	W	12.1
X-175	871	1600	25	77	441	64.0	774	112.3	4.6	4.6	9.9	W	9.9
X-168	593	1100	593	1100	259	37.6	514	74.6	29.6	29.4	33.3	W	33.3
X-172	704	1300	704	1300	408	59.2	566	82.1	4.9	12.1	42.2	W	42.2
X-176	871	1600	871	1600	202	29.3	232	33.6	5.0	25.6	63.3	W	63.3
Aged 20,000 h in HTGR-He environment													
X-180	593	1100	25	77	523	75.8	907	131.5	4.8	5.0	22.6	B	35.9
X-184	704	1300	25	77	589	85.4	996	144.4	1.4	1.4	7.2	W	7.2
X-188	871	1600	25	77	330	47.9	661	95.9	3.1	3.1	10.4	W	10.4
X-179	593	1100	593	1100	434	63.0	736	106.7	4.5	5.1	22.1	B	0.6
X-183	704	1300	704	1300	423	61.3	567	82.3	2.2	10.4	36.3	W	36.3
Aged 53,000 h in HTGR-He environment													
X-182	593	1100	25	77	734	106.5	1,098	159.3	3.9	3.9	19.1	B	3.9
X-186	704	1300	25	77	548	79.4	907	131.5	4.7	4.7	10.6	W	10.6
X-190	871	1600	25	77	296	42.9	544	78.9	4.4	4.4	6.6	W	6.6

[a]W, weld metal; B, base metal.

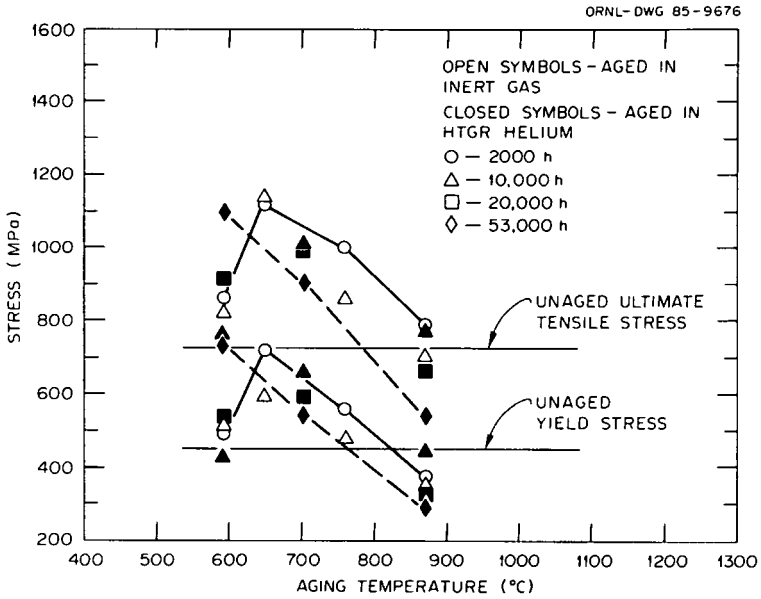


Fig. 11. Tensile properties at 25°C of Hastelloy X transverse weld samples after aging at the indicated temperature. Solid lines denote shortest aging time, and dashed lines indicate longest aging time.

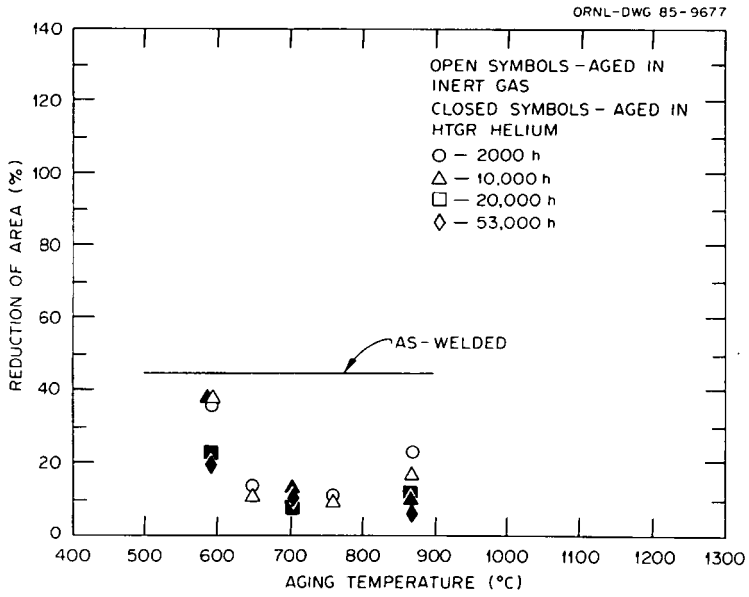


Fig. 12. Reduction in area at 25°C of Hastelloy X transverse weld samples aged at the indicated temperatures.

Table 8. Tensile properties of Inconel 617 (heat XX01A3US) base metal at 25°C (77°F) after aging in HTGR-He

Specimen[a]	Aging condition			0.2% Yield strength		Ultimate tensile strength		Elongation (%)		Reduction in area (%)
	Temperature		Time (h)							
	(°C)	(°F)		(MPa)	(ksi)	(MPa)	(ksi)	Uniform	Total	
16445	As received			303	43.9	747	108.3	54.1	54.3	36.2
16546	As received			301	43.7	757	109.8	59.6	60.9	40.7
16547	As received			305	44.3	754	109.3	57.4	57.9	32.4
17020	As received			319	46.2	785	113.8	61.8	64.2	48.9
I-100	593	1,100	10,000	316	45.9	737	106.9	53.5	59.6	57.1
I-113	593	1,100	20,000	510	74.0	919	133.3	24.4	24.8	29.1
I-115	593	1,100	53,000	527	76.4	946	137.2	24.4	24.4	27.8
I-104	704	1,300	10,000	531	77.0	818	118.6	14.1	14.3	14.6
I-117	704	1,300	20,000	523	75.8	794	115.3	9.6	10.0	15.5
I-119	704	1,300	53,000	473	68.6	732	106.1	9.6	9.6	11.0
I-108	871	1,600	10,000	392	56.8	623	90.3	11.4	11.4	14.2
I-121	871	1,600	20,000	303	44.0	509	73.8	6.0	6.2	6.8
I-123	871	1,600	53,000	288	41.8	386	56.0	2.8	2.8	2.8

[a]Specimens machined before aging. The gage sections of the first three specimens listed were 31.8 mm (1.25 in.) long by 6.4 mm (0.25 in.) in diameter. The gage sections of the other specimens were 25.4 mm (1.00 in.) long by 3.18 mm (0.125 in.) in diameter.

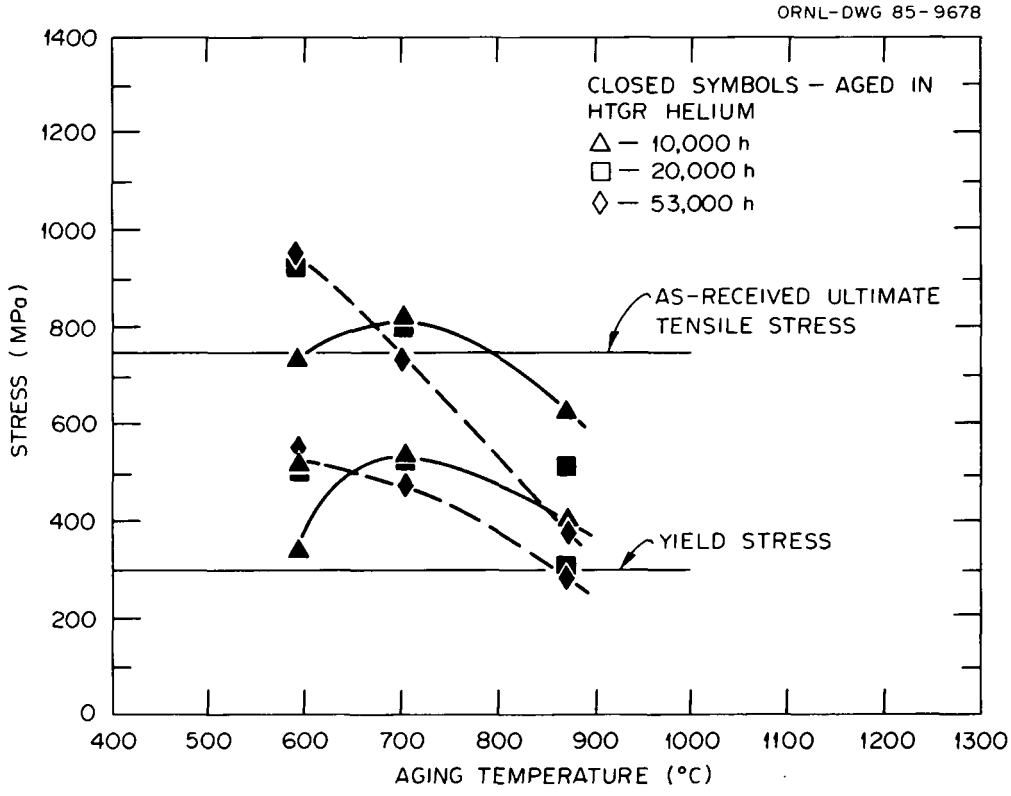


Fig. 13. Influence of aging in HTGR-He on the yield and ultimate tensile strengths of Inconel 617 base metal tested at 25°C. Solid lines denote shortest aging time, and dashed lines indicate longest aging time.

after 53,000 h aging was less than that of the as-received material. Aging only 10,000 h at 871°C resulted in an ultimate tensile strength less than that of the as-received material, and further aging caused further reductions in the ultimate tensile strength. The reduction in area is shown for these same samples in Fig. 14. With the exception of one point, aging caused the reduction in area to decrease. The extent of the decrease increased with longer aging time and with higher aging temperature.

The data from this report are for samples aged in HTGR-He, but the data for samples aged 20,000 h in an inert environment were reported by McCoy and King.[3] Some of the results are combined for the two studies in Fig. 15. The data show that aging in HTGR-He reduces the fracture elongation more than aging in inert gas. The data also show that for aging in HTGR-He there is very little difference in the fracture elongations of

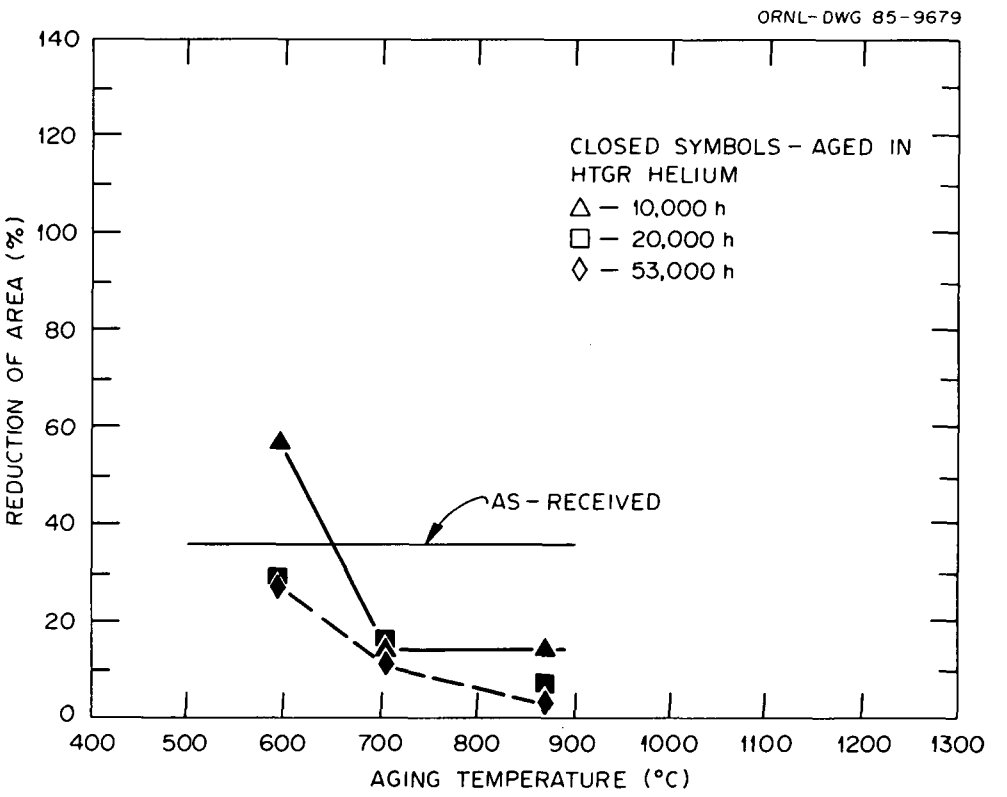


Fig. 14. Influence of aging in HTGR-He on the reduction in area of Inconel 617 base metal tested at 25°C. Solid line denotes shortest aging time, and dashed line indicates longest aging time.

samples aged 20,000 and 53,000 h. Thus the aging effects appear to be near saturation, but the strains are as low as 2.8% after aging at 871°C.

The effects of aging for 10,000, 20,000, and 53,000 h in HTGR-He on the tensile properties at 25°C of transverse welds of Inconel 617 are given in Table 9. The influence of aging on the yield and tensile strengths of these samples is shown in Fig. 16. Test results were not available for an as-received transverse weld sample, so the same values used in Fig. 13 for unaged base metal are used in Fig. 16. Aging at 593°C resulted in significant increases in both the yield and ultimate tensile stresses. Aging at 704°C caused a large increase in the yield strength and a moderate increase in the ultimate tensile strength. Aging at 871°C had a slight effect on the yield strength but caused a large decrease in the ultimate tensile strength. In most cases the property change was at a maximum after 20,000 h aging, and overaging occurred as a result of 53,000 h aging.

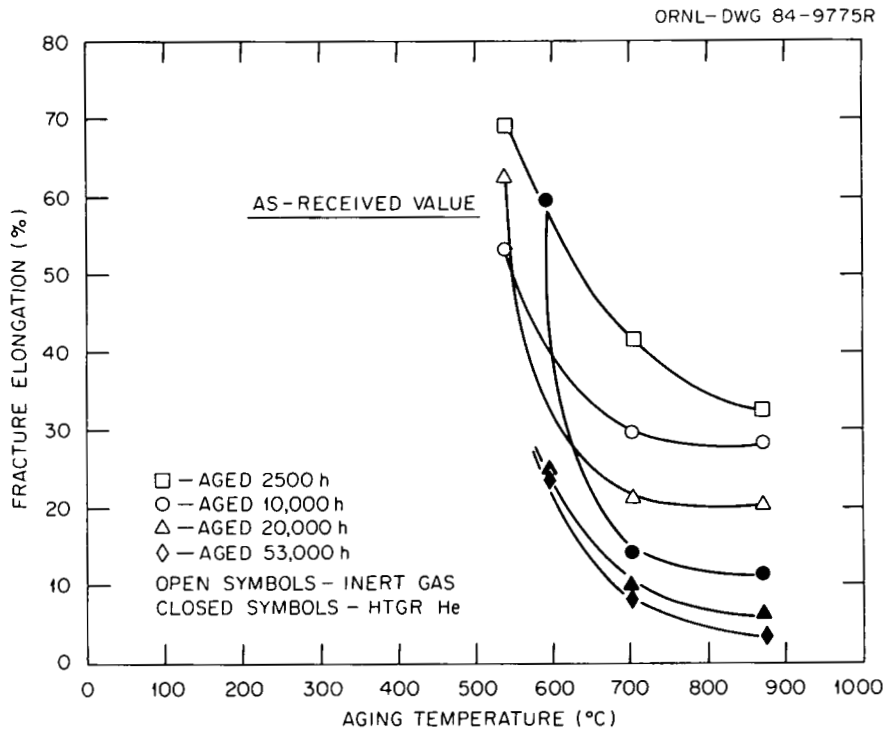


Fig. 15. Influence of aging on the fracture elongation at 25°C of Inconel 617 (heat XX01A3US).

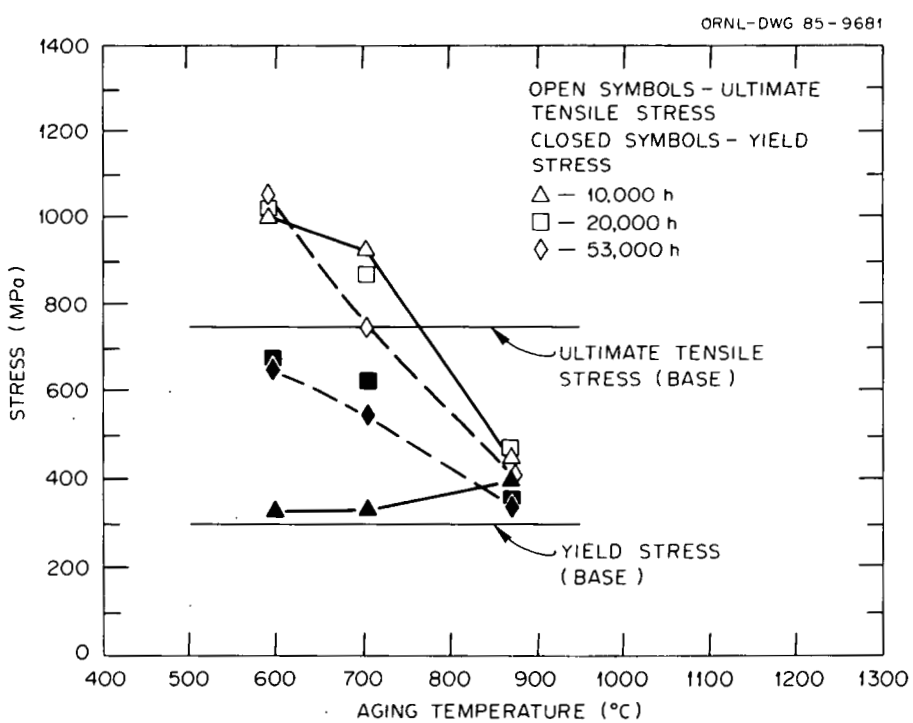


Fig. 16. Yield and ultimate tensile stresses of transverse weld samples of Inconel 617 tested at 25°C. Solid points are the yield stress, and open points are the ultimate tensile stress. Horizontal lines indicate base metal. Solid lines indicate the shortest aging time, and dashed lines are the longest aging time. Samples were aged in HTGR-He.

Table 9. Tensile properties at 25°C (77°F) of Inconel 617 transverse weld specimens after aging in HTGR-He

Specimen[a]	Aging condition			0.2% Yield strength		Ultimate tensile strength		Elongation (%)		Reduction in area (%)	Failure location[b]	Reduction in area of weld (%)
	Temperature		Time (h)									
	(°C)	(°F)		(MPa)	(ksi)	(MPa)	(ksi)	Uniform	Total			
I-148	593	1,100	10,000	330	47.8	1004	145.6	26.0	26.3	28.3	B	
I-161	593	1,100	20,000	676	98.0	1020	148.0	19.5	19.9	10.6	B	
I-173	593	1,100	53,000	650	94.3	1051	152.4	21.0	21.0	28.1	W	
I-152	704	1,300	10,000	331	48.0	927	134.4	14.2	14.4	21.0	B	
I-165	704	1,300	20,000	626	90.3	867	125.8	5.0	5.2	8.3	B	
I-167	704	1,300	53,000	547	79.4	744	107.9	5.3	5.3	6.8	B	2.7
I-156	871	1,600	10,000	394	57.1	450	65.2	0.7	0.7	3.2	B	
I-169	871	1,600	20,000	356	51.6	476	69.0	0.8	0.8	3.3	B	
I-171	871	1,600	53,000	339	49.2	396	57.4	0.6	0.6	0.6	B	0.6

[a]Welds made by the gas-tungsten-arc process utilizing heat XX14A6UK base metal and XX09A9UK filler wire. Specimens machined before aging. Gage sections were 25.4 mm (100 in.) long by 3.18 mm (0.125 in.) in diameter.

[b]W, weld metal; B, base metal.

The fracture strains of the transverse weld samples at 25°C are shown in Fig. 17. The results in Table 9 show that failure usually occurred in the base metal, indicating that the weld metal was stronger. As shown in Fig. 17, aging reduced the reduction in area to quite low values. Generally, the degree of embrittlement increased with aging time and, after 53,000 h at 871°C, the fracture strain at 25°C was 0.6%. The reduction in area of the fracture (base metal) and the weld were also 0.6%. There is not much evidence of the fracture strain saturating with aging time (i.e., the fracture strain is progressively lower with increasing aging time).

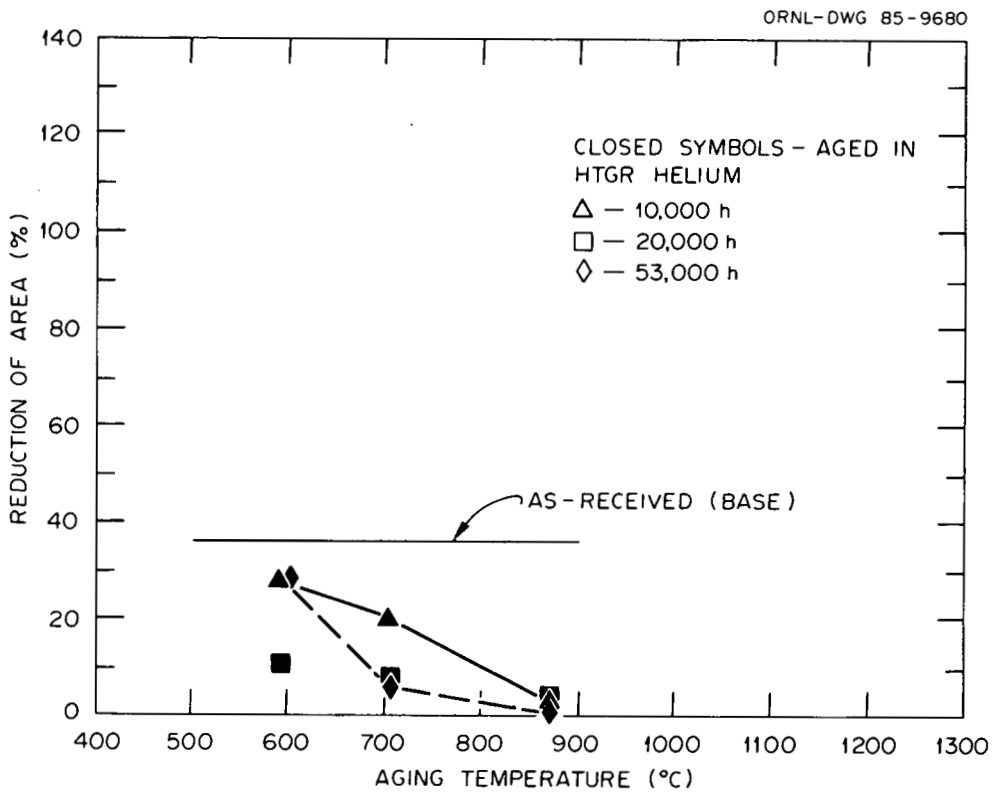


Fig. 17. Reduction in area of transverse weld samples of Inconel 617 at 25°C. Solid line denotes shortest aging time, dashed line longest aging time.

5. SCANNING ELECTRON MICROSCOPY OF FRACTURE SURFACES

Representative tensile samples were selected to view the fracture surfaces in the scanning electron microscope. The same samples also were prepared for viewing by standard metallographic methods, and the results of those examinations are presented in Section 6.

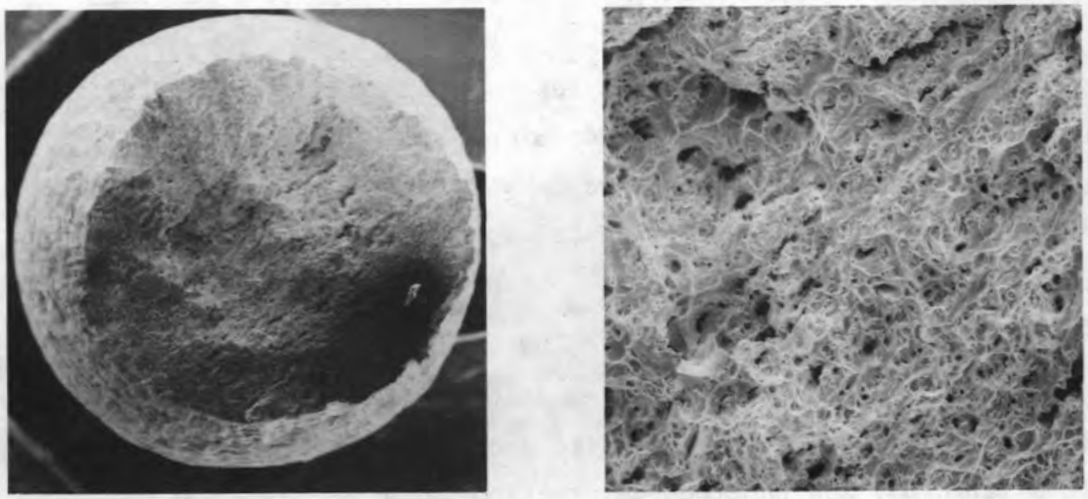
5.1 HASTELLOY X

Typical views of as-received Hastelloy X deformed to fracture are shown in Fig. 18. The fracture is quite ductile, with a large reduction in area (see Table 6), and numerous deformation marks are apparent on the fracture surface. The fracture surface of this material after 53,000 h aging at 593°C and fracturing at 25°C is shown in Fig. 19. The fracture is less ductile and more intergranular than that of the as-received sample shown in Fig. 18. There is some cleavage fracture in the aged sample in Fig. 19. The fracture surface of a sample aged at 704°C is shown in Fig. 20. The fracture is almost totally intergranular, with only a few cleavage areas. The surface when magnified several thousand times shows considerable plastic flow. The fracture of a sample aged 53,000 h at 871°C and deformed at 25°C is shown in Fig. 21. The fracture is entirely intergranular, and the surface detail (when viewed at higher magnification) is localized plastic deformation. The fractured samples in Figs. 18 through 21 had reduction in areas of 60%, 15.9%, 10.6%, and 14.7%, respectively. The amount of plastic flow seems quite consistent with those values.

5.2 HASTELLOY X WELDMENTS

Micrographs of the fracture of a Hastelloy X sample with a transverse weld are shown in Fig. 22. As shown in Table 7, the sample failed in the weld, and the reduction in area was 44%. The entire sample cross section appears deformed, and separation has occurred primarily along dendrites in the weld. The transverse weld sample shown in Fig. 23 failed in the base metal, and the reduction in area was 19.1% (Table 7). The failure is partially intergranular and partially cleavage, very similar to the fracture

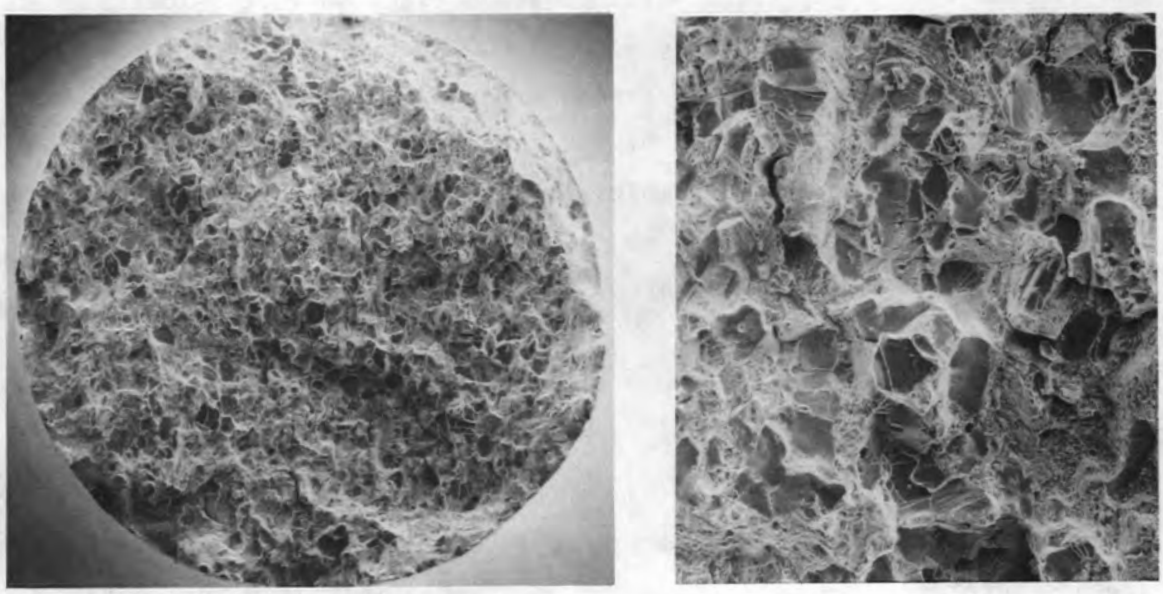
M-1307 M-1308



(*a*) 1 mm (*b*) 100 μm

Fig. 18. Scanning electron micrographs of as-received Hastelloy X base metal (heat 4936, specimen X6-14) strained to failure at 25°C. Failed at 49.9% reduction in area.

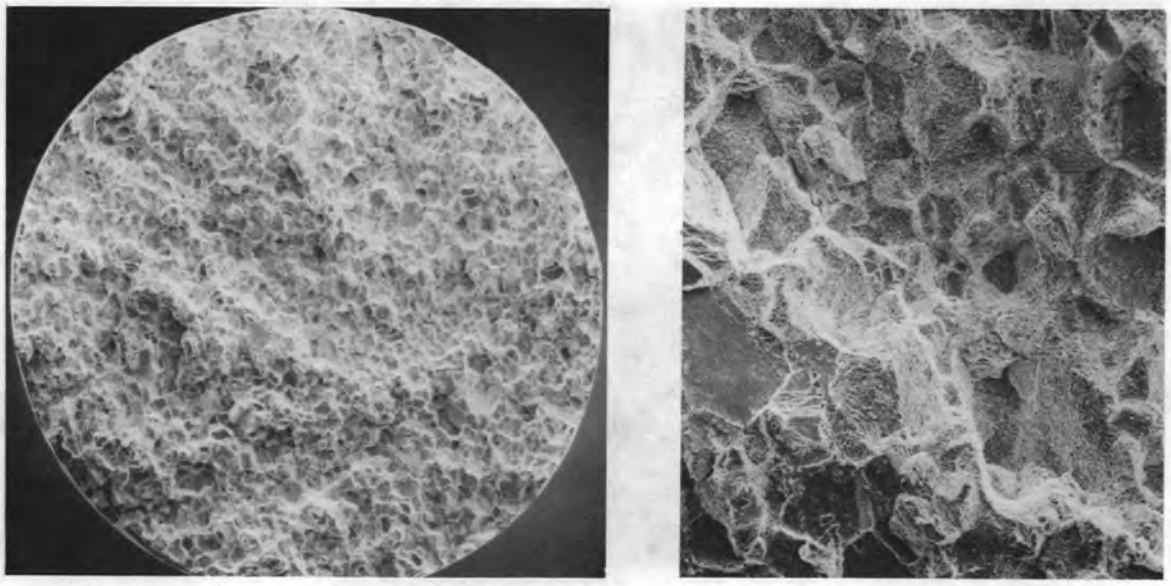
M-1329 M-1330



(*a*) 1 mm (*b*) 100 μm

Fig. 19. Scanning electron micrographs of Hastelloy X base metal (heat 4936, sample X-115) after aging 53,000 h at 593°C in HTGR-He and straining to failure at 25°C. Fractured with 15.9% reduction in area.

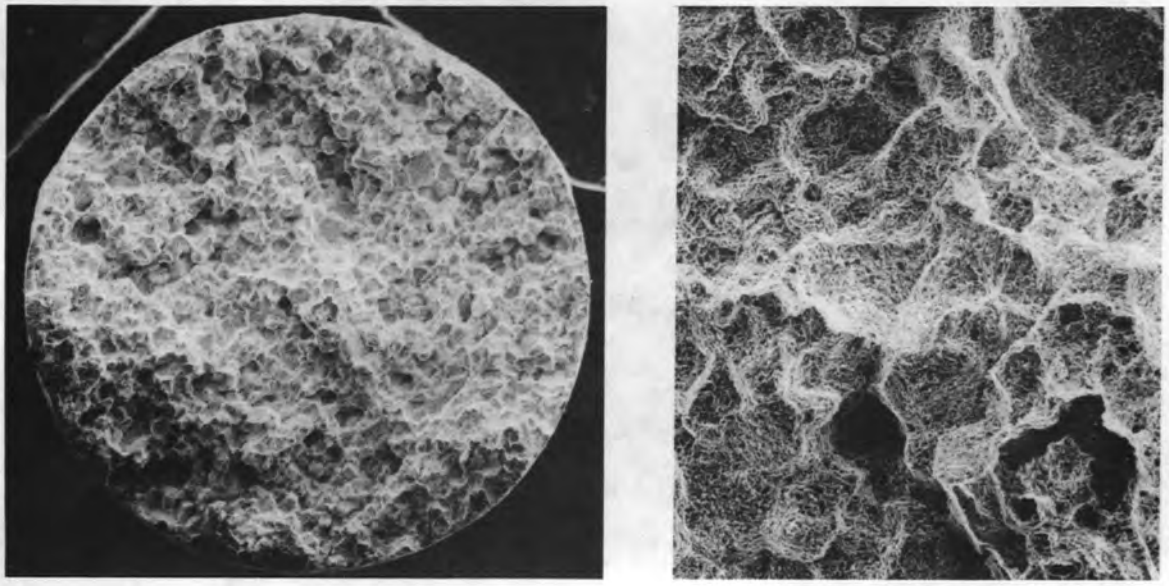
M-1317 M-1318



(*a*) 1 mm (*b*) 100 μm

Fig. 20. Scanning electron micrographs of fracture of Hastelloy X base metal (heat 4936, sample X-119) after aging 53,000 h at 704°C in HTGR-He and straining to failure at 25°C. Fractured with 10.6% reduction in area.

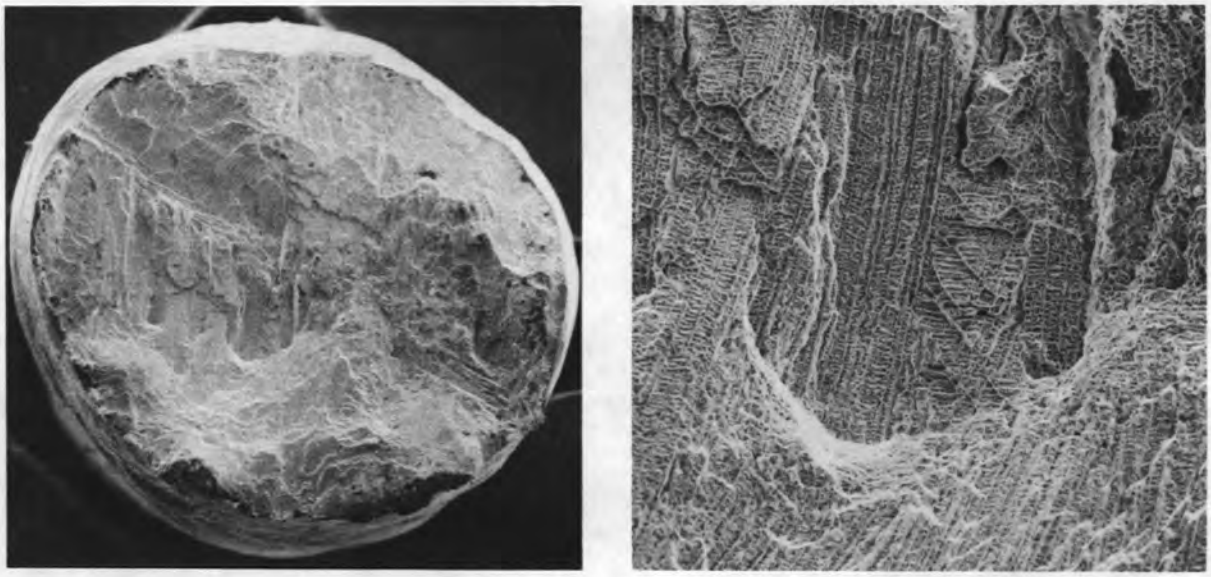
M-1311 M-1312



(*a*) 1 mm (*b*) 100 μm

Fig. 21. Scanning electron micrographs of fracture of Hastelloy X base metal (heat 4936, sample X-123) after aging 53,000 h at 871°C in HTGR-He and straining to failure at 25°C. Fractured with 14.7% reduction in area.

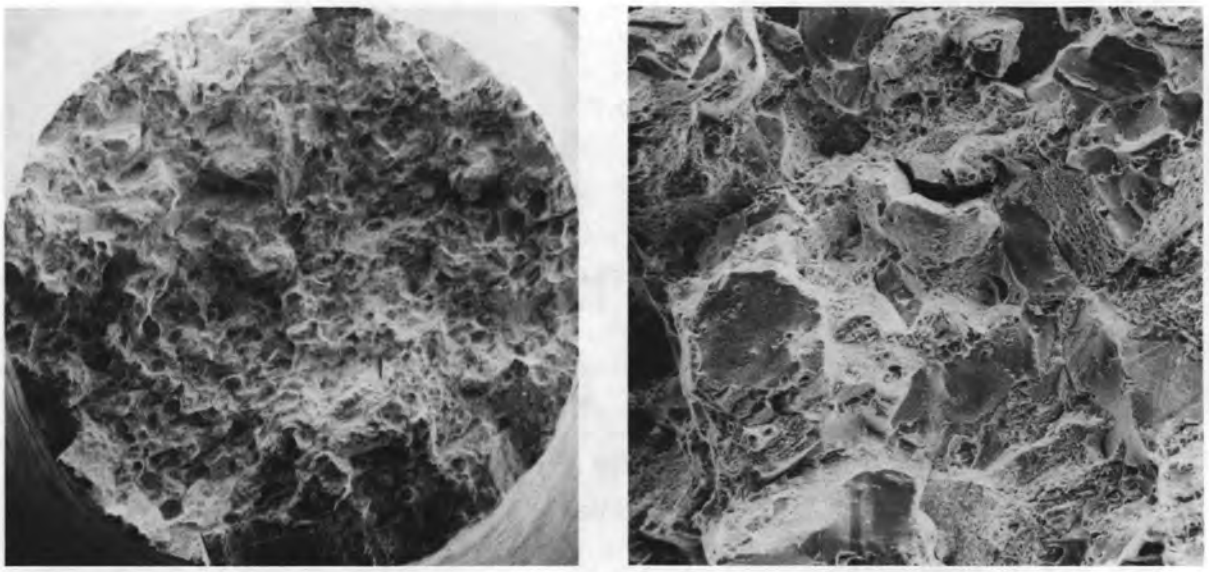
M-1313 M-1314



(*a*) 1 mm (*b*) 100 μm

Fig. 22. Scanning electron micrographs of fracture of transverse Hastelloy X weld (specimen 16703) in the as-received condition and strained to failure at 25°C. Failed with 44.1% reduction in area.

M-1327 M-1328



(*a*) 1 mm (*b*) 100 μm

Fig. 23. Scanning electron micrographs of fracture of transverse Hastelloy X weld (sample X-182) after aging 53,000 h at 593°C in HTGR-He and straining to failure at 25°C. Failed with 19.1% reduction in area.

shown in Fig. 19 in a base metal specimen. The sample aged at 704°C failed in the weld metal, and the reduction in area was 10.6% (Table 7). The photomicrographs of this sample in Fig. 24 show that fracture occurred along the dendrites of the weld with only a small amount of plastic deformation. The sample aged at 871°C failed similarly (Fig. 25), but the reduction in area was less than that of the previous sample aged at 704°C.

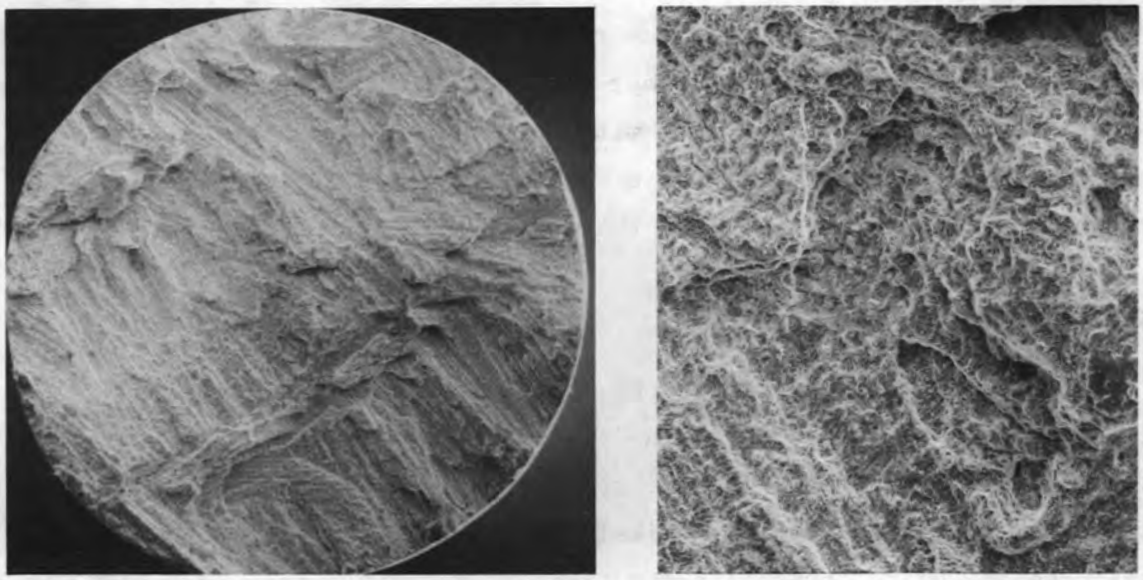
5.3 INCONEL 617 BASE METAL

Scanning electron micrographs of the fracture of as-received Inconel 617 base metal are shown in Fig. 26. This sample failed at a reduction in area of 40.7%. The fracture is primarily intergranular, but the stringers shown in Fig. 3 likely played a major role because their features are quite obvious in the fracture. Some regions indicate rather high localized deformation. Fig. 27 shows the fracture of an Inconel 617 sample aged for 53,000 h at 593°C in HTGR-He. The fracture is largely intergranular, but there are several cleavage areas. The reduction in area of this sample was 27.8%. The fracture surface of a sample aged 53,000 h at 704°C in HTGR-He is shown in Fig. 28. The fracture is predominately intergranular with some cleavage. There is very little evidence of plastic deformation, and there are numerous precipitates in the grain boundaries. The reduction in area of this specimen was 11%. The fracture surface of a sample aged 53,000 h at 871°C in HTGR-He is shown in Fig. 29. This sample had a reduction in area of 2.8%. The fracture is entirely intergranular with numerous precipitates in the boundaries. There is some evidence of plastic deformation.

5.4 INCONEL 617 TRANSVERSE WELD SAMPLES

Transverse weld samples of Inconel 617 were examined after aging 53,000 h in HTGR-He and fracturing at 25°C. The results of these tensile tests are given in detail in Table 9. Note that the fracture usually occurred in the base metal. However, the sample shown in Fig. 30 was aged at 593°C and failed in the weld. The reduction in area was 28%, and there is considerable evidence of plastic flow in the fracture. Failure

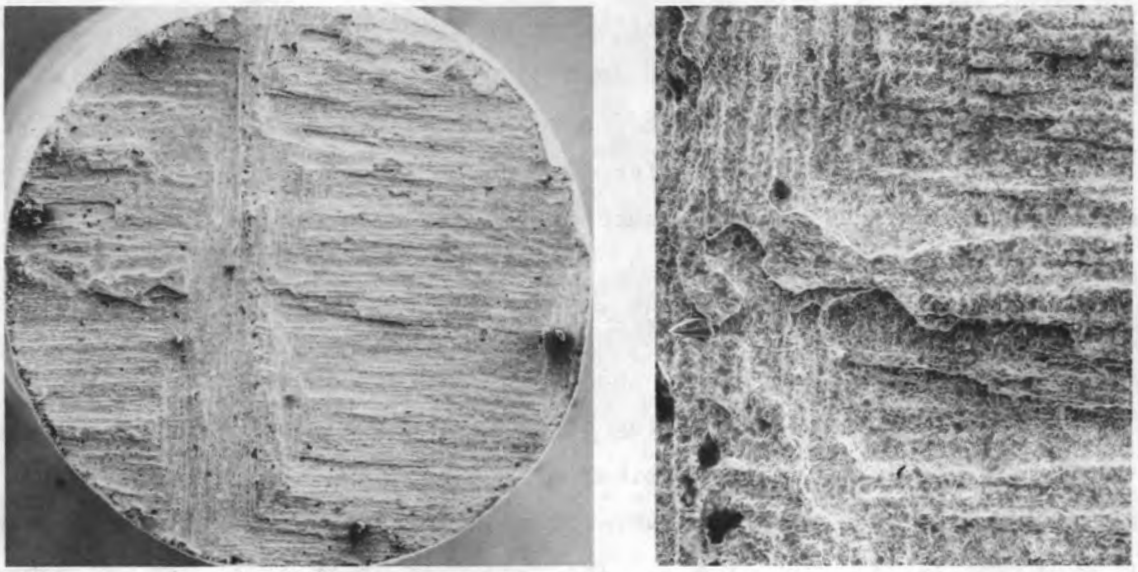
M-1319 M-1320



(*a*) 1 mm (*b*) 100 μm

Fig. 24. Scanning electron micrographs of fracture and transverse Hastelloy X weld (sample X-186) after aging 53,000 h at 704°C in HTGR-He and straining to failure at 25°C. Failed after 10.6% reduction in area.

M-1303 M-1304



(*a*) 1 mm (*b*) 100 μm

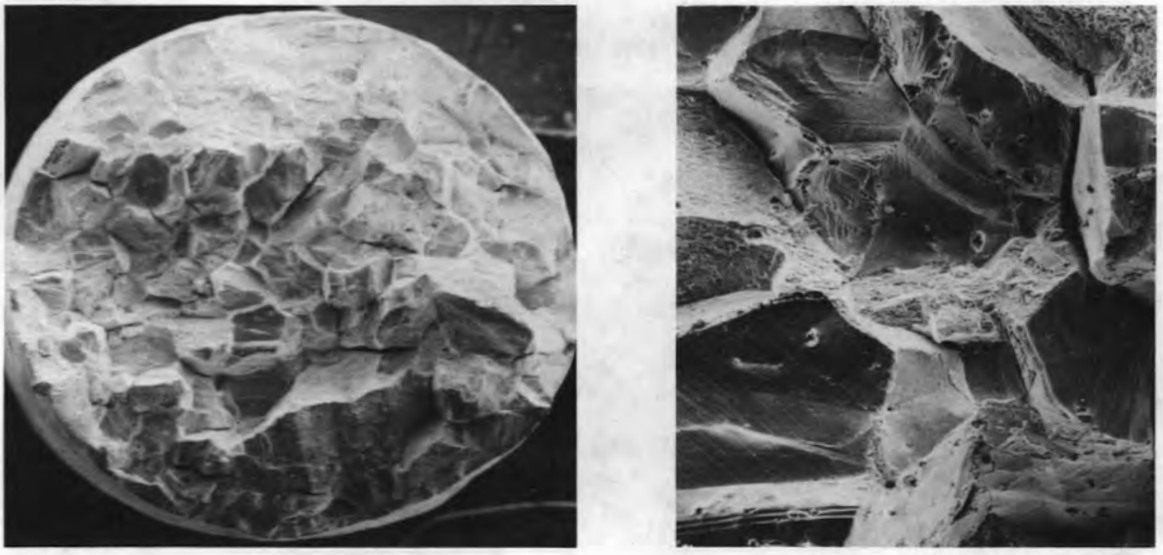
Fig. 25. Scanning electron micrographs of fracture of transverse Hastelloy X weld (sample X-190) after aging 53,000 h at 871°C in HTGR-He and straining to failure at 25°C. Failed after 6.6% reduction in area.

M-1305 M-1306

(*a*) 1 mm (*b*) 100 μm

Fig. 26. Scanning electron micrographs of fracture of Inconel 617 (heat XX01A3US, sample 17020) in the as-received condition and strained to failure at 25°C. Fractured with 48.9% reduction in area.

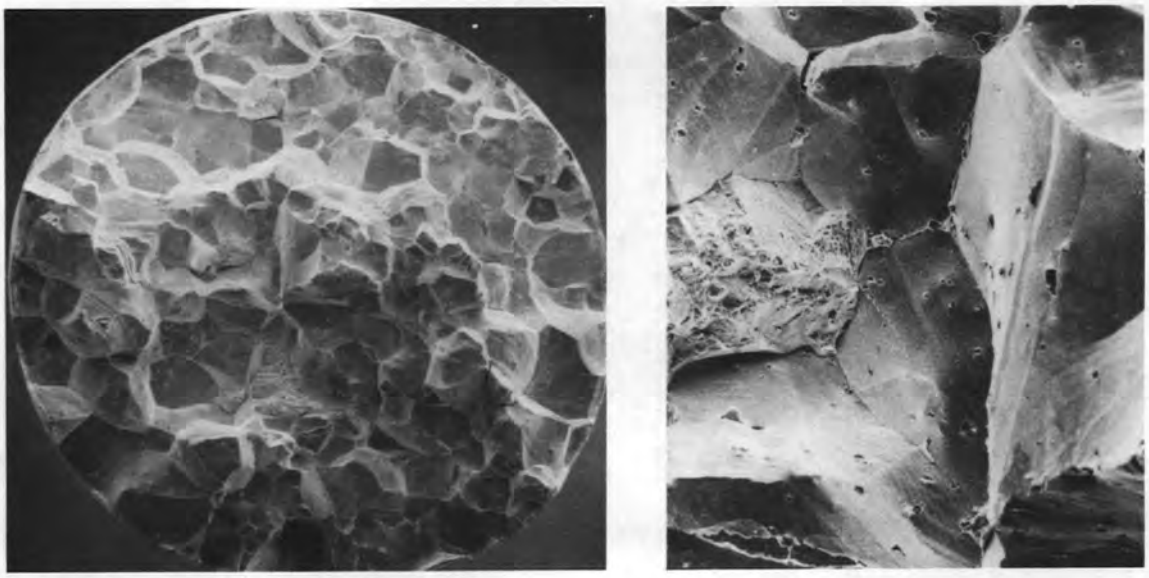
M-1325 M-1326



(*a*) 1 mm (*b*) 100 μm

Fig. 27. Scanning electron micrographs of fracture of Inconel 617 (heat XX01A3US, sample I-115) after aging 53,000 h at 593°C in HTGR-He and straining to failure at 25°C. Failed with 27.8% reduction in area.

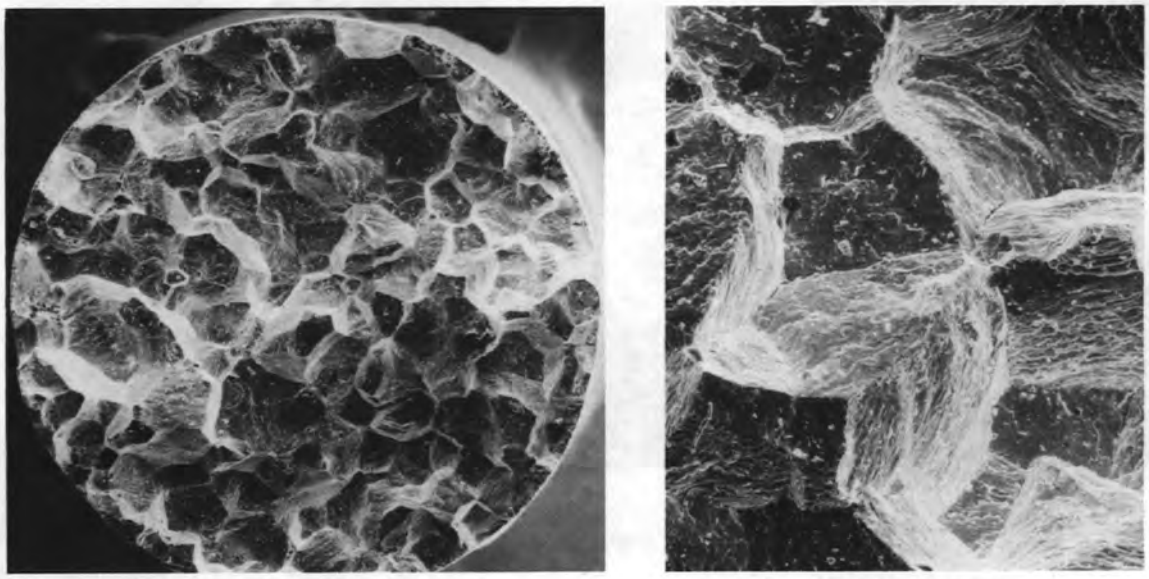
M-1315 M-1316



(*a*) 1 mm (*b*) 100 μm

Fig. 28. Scanning electron micrographs of fracture of Inconel 617 (heat XX01A3US, sample I-119) after aging 53,000 h at 704°C in HTGR-He and straining to failure at 25°C. Failed with 11.0% reduction in area.

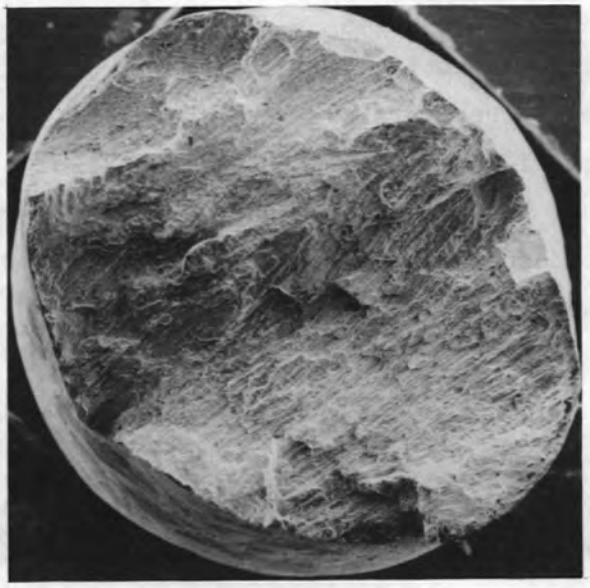
M-1301 M-1302



(*a*) 1 mm (*b*) 100 μm

Fig. 29. Scanning electron micrographs of fracture of Inconel 617 (heat XX01A3US, sample I-171) after aging 53,000 h at 871°C in HTGR-He and straining to failure at 25°C. Failed with 2.8% reduction in area.

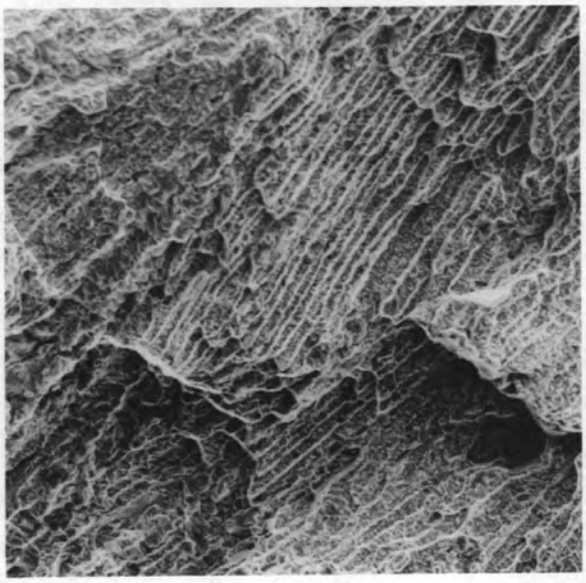
M-1323



(*a*)

1 mm

M-1324



(*b*)

100 μm

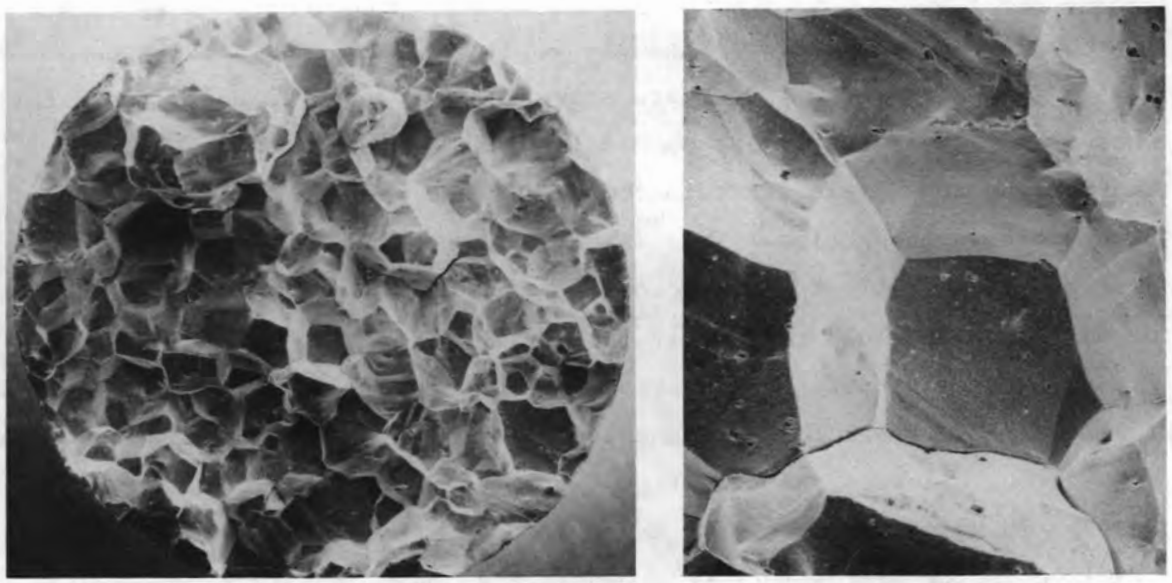
Fig. 30. Scanning electron micrographs of fracture of transverse weld of Inconel 617 (sample I-173) after aging 53,000 h at 593°C in HTGR-He and straining to failure at 25°C. Failed with 28.1% reduction in area.

occurred between the dendrites in the weld metal. The sample in Fig. 31 failed in the base metal, and the reduction in area was 6.8% (Table 9). Note that the failure is entirely intergranular with little evidence of plastic deformation. There are some precipitates in the grain boundaries. The sample aged at 871°C failed with 0.6% reduction in area (Table 9). Note in Fig. 32 that the failure was entirely intergranular with very little evidence of deformation in the grain boundaries. There are many precipitates in the grain boundaries.

6. METALLOGRAPHIC STUDIES

The same samples viewed in the scanning electron microscope (described in Section 5) were sectioned and viewed metallographically. Previous experience had shown that these two materials etch very rapidly in etchants such as aqua regia when they contain such copious amounts of precipitates. This generally results in problems with stains and with false microstructures. Very good results were obtained with 10% oxalic acid used electrolytically with low currents of the order of 0.1 A/cm^2. This etchant was used in the present study. Unfortunately, resources did not permit precipitate extraction and identification studies. Considerable use was made of such studies by Lai[4] for Hastelloy X and by Mankins et al.[5] for Inconel 617. Lai concluded on the basis of 10,000 h of aging at 538 through 871°C that M_6C-type carbides were the major precipitating phase, and small amounts of σ and μ phases were detected at 760 and 871°C, respectively. Mankins et al. concluded that the only active carbide in Inconel 617 was $M_{23}C_6$, and they found small amounts of gamma prime in samples exposed at 649 and 760°C. The samples analyzed were aged at temperatures of 649 to 1093°C for times from 200 to 10,000 h. Analysis of the particular heat of material being studied by PHACOMP predicted that 0.63% gamma prime could form. It is possible that the chemical composition of the three heats involved in the current study could result in the formation of slightly more or less gamma prime.

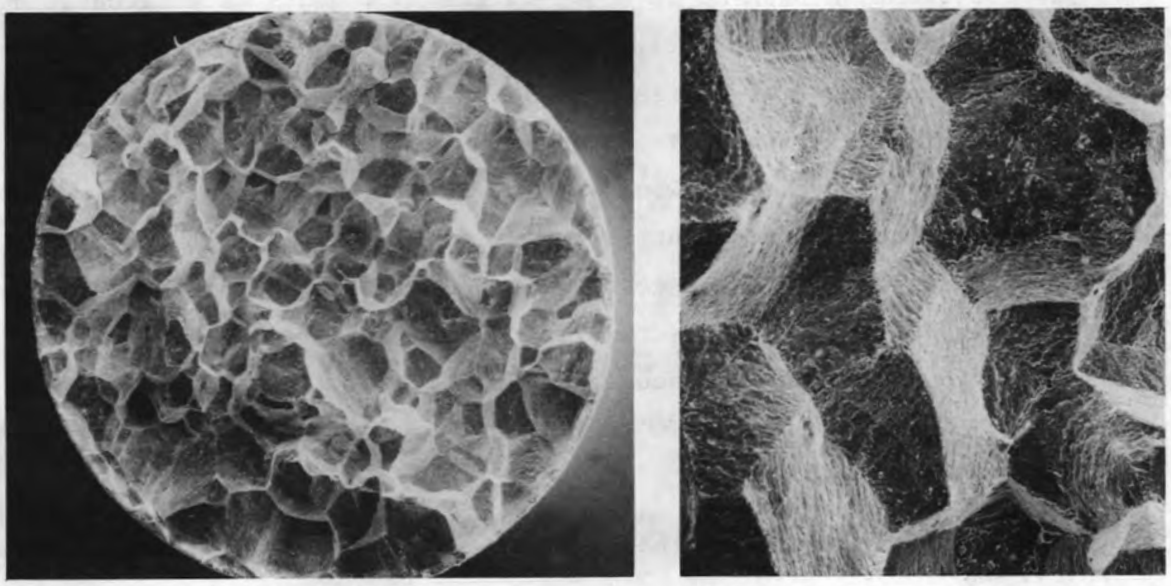
M-1321 M-1322



(*a*) 1 mm (*b*) 100 μm

Fig. 31. Scanning electron micrographs of fracture of transverse weld of Inconel 617 (sample I-167) after aging 53,000 h at 704°C in HTGR-He and straining to failure at 25°C. Failed with 6.8% reduction in area.

M-1309 M-1310



(*a*) 1 mm (*b*) 100 μm

Fig. 32. Scanning electron micrographs of fracture of transverse weld of Inconel 617 (sample I-123) after aging 53,000 h at 871°C in HTGR-He and straining to failure at 25°C. Failed with 0.6% reduction in area.

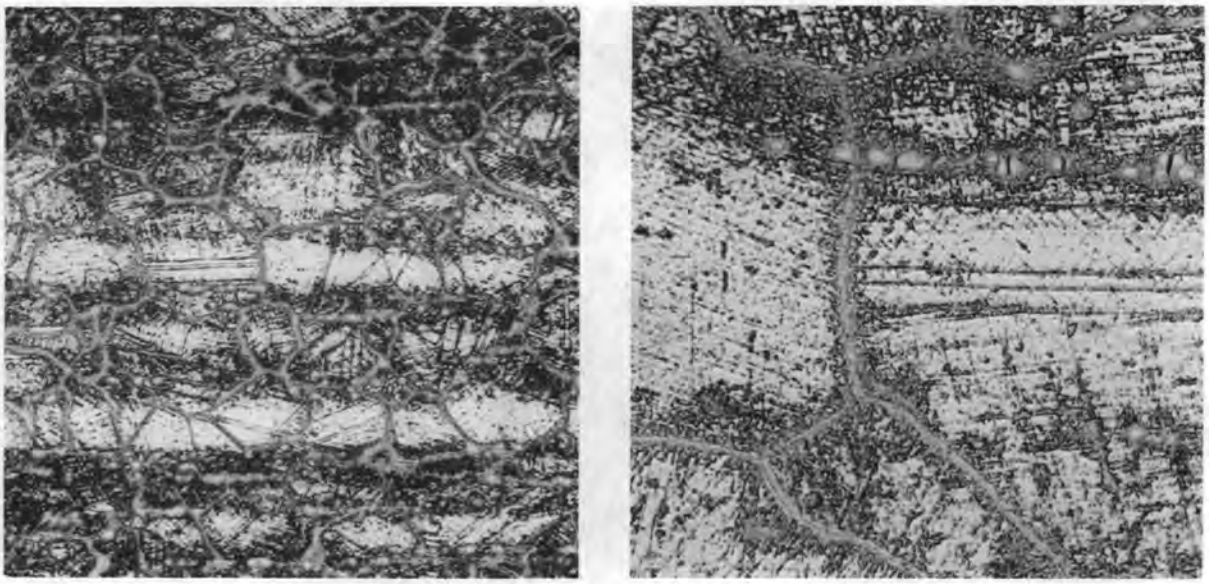
6.1 HASTELLOY X BASE METAL

Typical micrographs of a sample of Hastelloy X base metal aged 53,000 h at 593°C and deformed at 25°C are shown in Fig. 33. The only precipitating phase under these conditions should be M_6C. Note that the carbides are heavy along some grain and twin boundaries and seem to prefer some habit planes within the grains. This inhomogenous precipitate distribution is particularly visible in the photomicrograph magnified 100 times. This type of distribution would reasonably result in the fracture surface shown in Fig. 19 where fracture has occurred by almost every mechanism imaginable. The precipitates are fine and should inhibit yielding and fracture, so the high strength levels shown in Fig. 9 and the reasonable fracture strain in Fig. 10 would be expected. Aging at 704°C for 53,000 h in HTGR-He resulted in the microstructure shown in Fig. 34. At this high level of precipitation it is likely that much of the molybdenum and chromium are tied up as carbides, and the carbides are coarse enough that they likely do not provide much strengthening. Thus, this microstructure is consistent with the predominately intergranular fracture and very localized plastic flow (Fig. 20), with the slightly enhanced strength (Fig. 9), and with the very low fracture strain (Fig. 10). The microstructure after aging at 871°C for 53,000 h in HTGR-He is shown in Fig. 35. This sample contains more carbon than usual (Table 2), and the grain boundaries contain an almost continuous layer of coarse carbides. There are also numerous precipitates within the grains. The fracture shown in Fig. 21 would be expected because the fracture would follow the grain boundary precipitate, but the softer material adjacent to the grain boundaries would deform locally. The carbides are too coarse for much strengthening, so the low strength (Fig. 9) would be expected; the deformation is so localized that the deformation measured macroscopically would be low (Fig. 10).

6.2 HASTELLOY X TRANSVERSE WELD SAMPLES

Typical micrographs near the fracture of a transverse weld Hastelloy X sample aged at 593°C for 53,000 h in HTGR-He are shown in Fig. 36. The

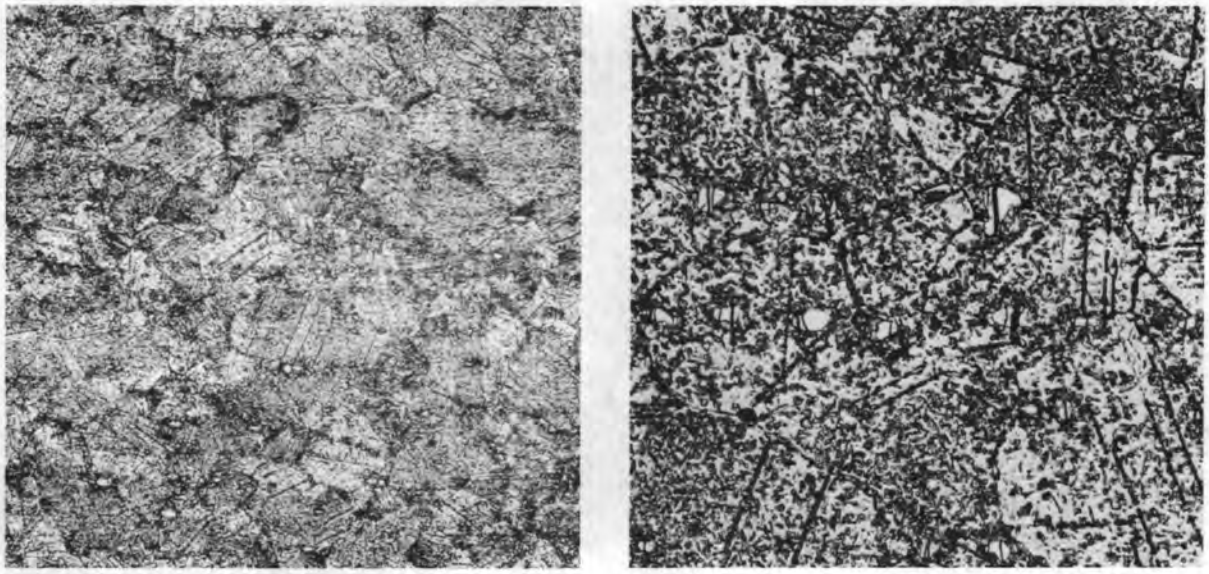
Y-201817 Y-201816



(*a*) 200 μm (*b*) 40 μm

Fig. 33. Hastelloy X base metal (heat 4936, sample X-115) after aging 53,000 h at 593°C in HTGR-He and straining to failure at 25°C. Fractured with 15.9% reduction of area.

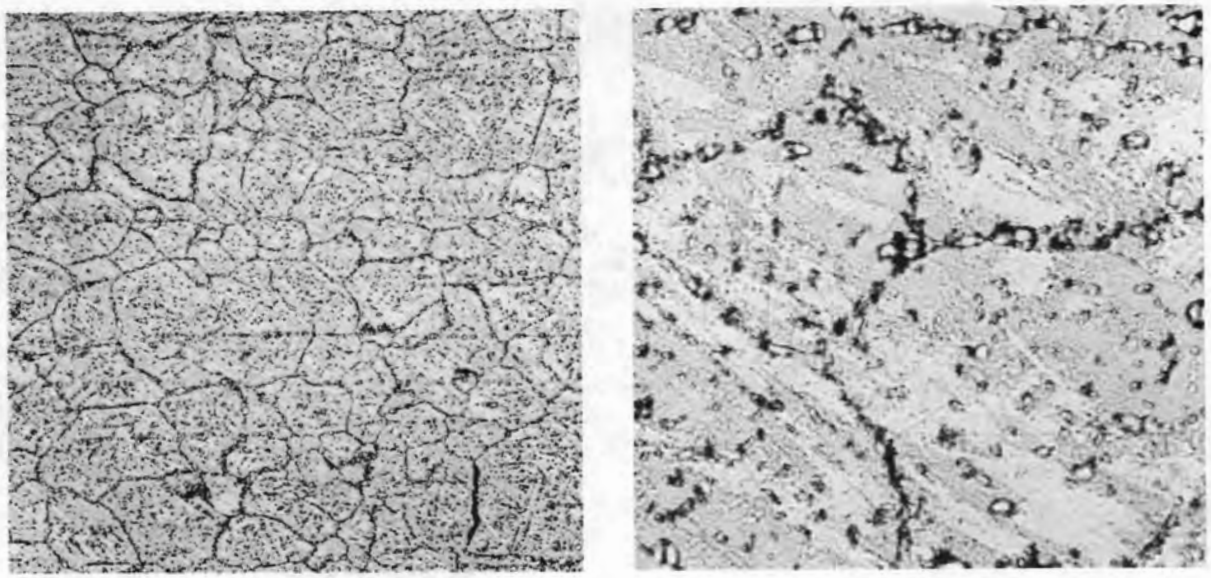
Y-201799 Y-201798



(*a*) 200 μm (*b*) 40 μm

Fig. 34. Hastelloy X base metal (heat 4936, sample X-119) after aging 53,000 h at 704°C in HTGR-He and straining to failure at 25°C. Fractured with 10% reduction in area.

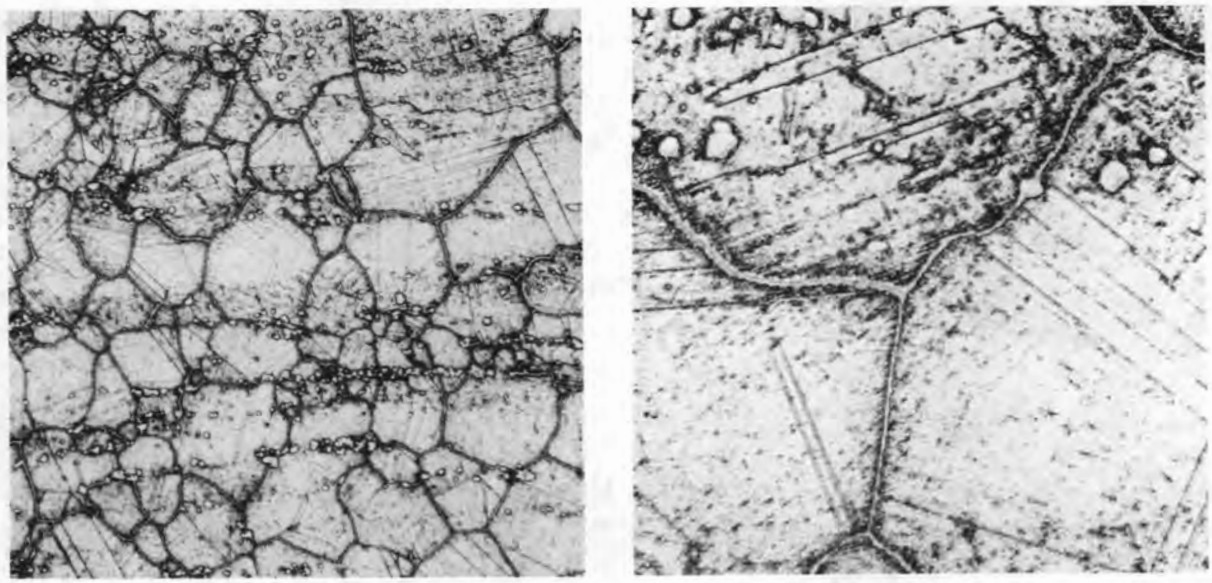
Y-201803 Y-201802



(*a*) 200 μm (*b*) 40 μm

Fig. 35. Hastelloy X base metal (heat 4936, sample X-123) after aging 53,000 h at 871°C in HTGR-He and straining to failure at 25°C. Fractured with 14.7% reduction in area.

Y-201815 Y-201814



(*a*) 200 μm (*b*) 40 μm

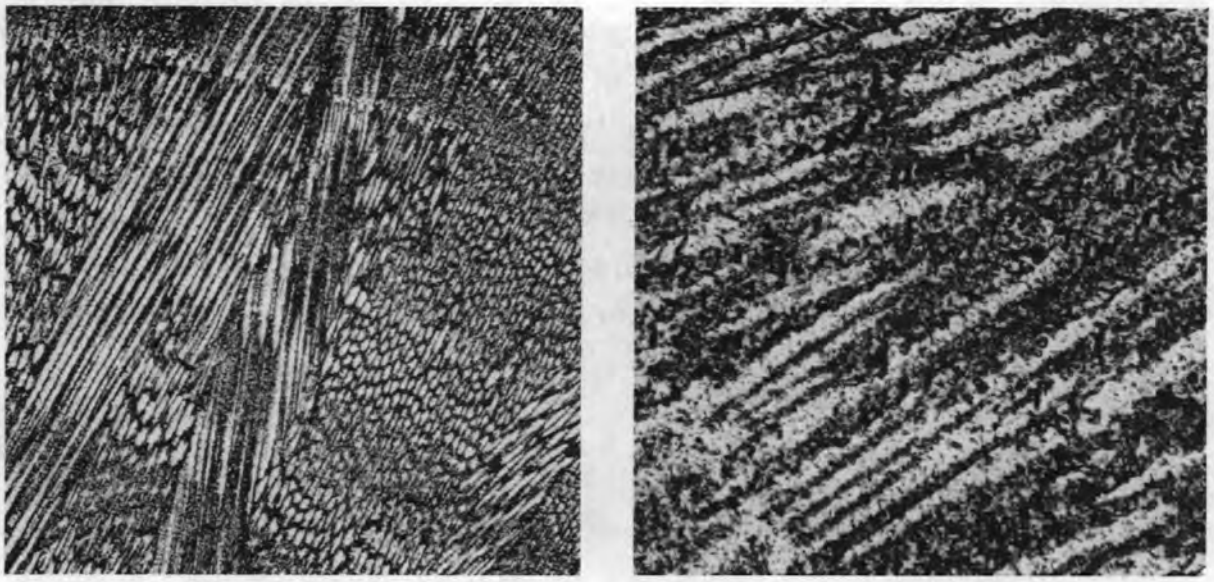
Fig. 36. Photomicrographs near the fracture of a transverse weld sample (X-182) of Hastelloy X after aging 53,000 h at 593°C in HTGR-He and straining to failure at 25°C. Failed with 44.1% reduction in area.

sample failed in the base metal (Table 7), so the microstructure is very similar to that in Fig. 33 where only base metal was involved. The precipitate is concentrated on most grain boundary surfaces, so the mixed-mode fracture shown in Fig. 23 would be expected. The samples aged at 704 and 871°C failed in the weld (Figs. 37 and 38, respectively). At 704°C the weld structure was not altered greatly. During aging at 871°C, the weld structure changed considerably and discrete particles replaced the cast weld metal structure. However, many of the boundaries of the cast structure remain.

6.3 INCONEL 617 BASE METAL

The as-received microstructure of Inconel 617 (XX01A3US) is shown in Fig. 3. Stringers of precipitates are visible in the as-received microstructure, and they seem to play a role in the fracture of the unaged material as was shown in Fig. 26. Fig. 39 shows typical photographs of Inconel 617 (heat XX01A3US) after aging 53,000 h at 593°C in HTGR-He and straining to failure at 25°C. The precipitate is rather fine under these conditions and decorates the grain and twin boundaries. It is reasonable that the fineness of the precipitate results in the strengthening noted in Fig. 13, the good ductility noted in Fig. 14, and the mixed-mode fracture surface shown in Fig. 27. Aging at 704°C resulted in the optical microstructure shown in Fig. 40. There are many fine precipitates in the matrix after this aging treatment, but the grain boundaries have become coated rather continuously and this should lead to poor fracture strain. As shown in Fig. 13, this aging treatment resulted in increased yield strength and no change in ultimate tensile strength. Fig. 14 shows that the reduction in area is low, about 11%, and Fig. 28 shows that the fracture is predominately intergranular and reasonably brittle. Aging at 871°C for 53,000 h in HTGR-He (Fig. 41) resulted in grain boundaries almost continuously coated with coarse precipitates. The fracture is intergranular with surface deformation (Fig. 29), but the macroscopic rupture strain is quite low (0.3%). This aging condition results in an average yield stress and a reduced ultimate tensile strength (Fig. 13).

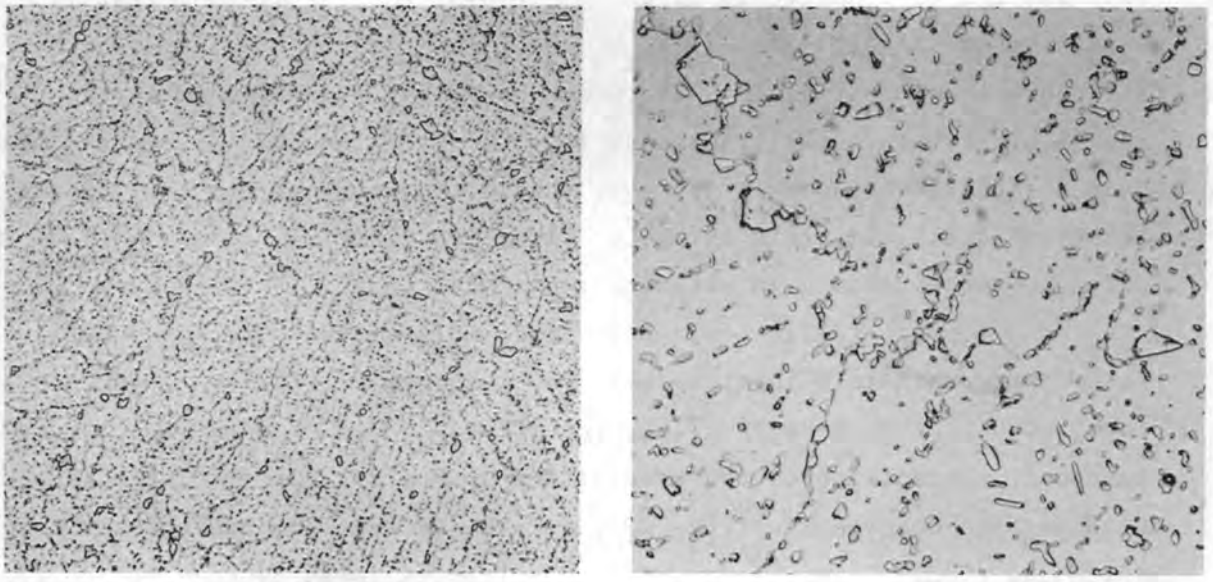
Y-201797 Y-201795



(*a*) 200 μm (*b*) 40 μm

Fig. 37. Photomicrographs near the fracture of a transverse weld sample (X-186) of Hastelloy X after aging 53,000 h at 704°C in HTGR-He and straining to failure at 25°C. Failed with 10.6% reduction in area.

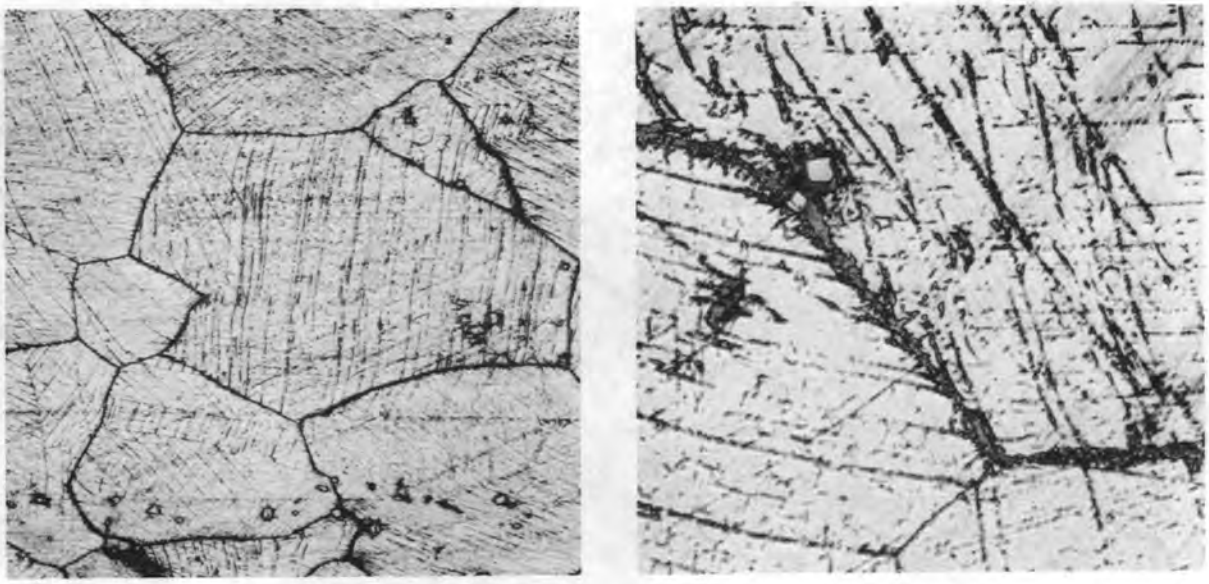
Y-201813 Y-201812



(*a*) 200 μm (*b*) 40 μm

Fig. 38. Photomicrographs near the fracture of a transverse weld sample (X-190) of Hastelloy X after aging 53,000 h at 871°C in HTGR-He and straining to failure at 25°C. Failed with 6.6% reduction in area.

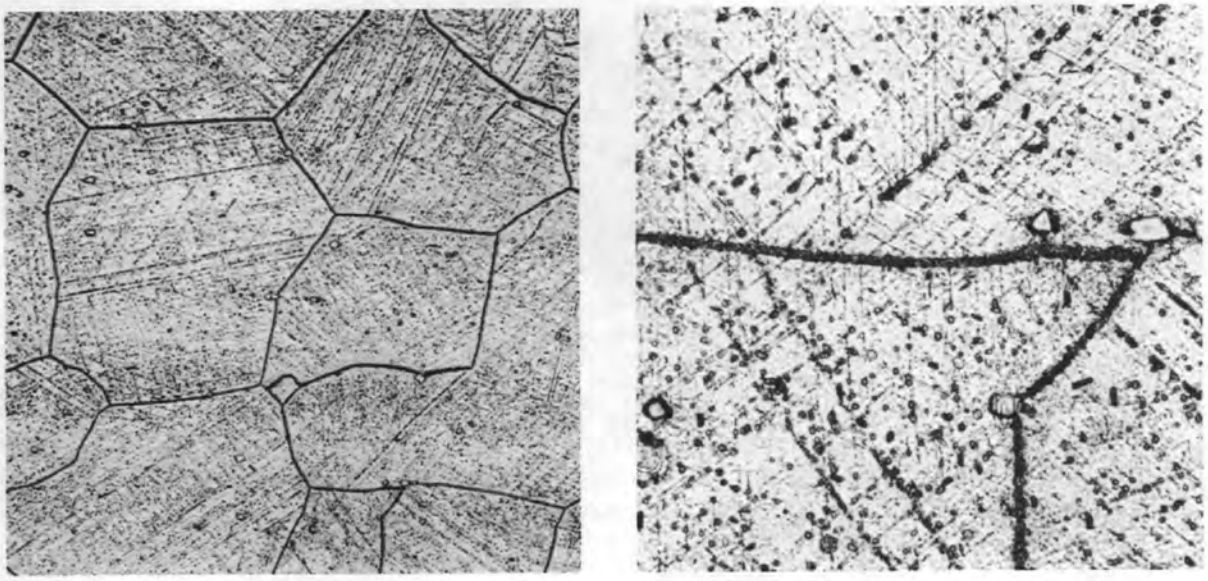
Y-201787 Y-201786



(*a*) 200 μm (*b*) 40 μm

Fig. 39. Inconel 617 base metal (XX01A3US, sample I-115) after aging 53,000 h at 593°C in HTGR-He and straining to failure at 25°C. Fractured with 27.8% reduction in area.

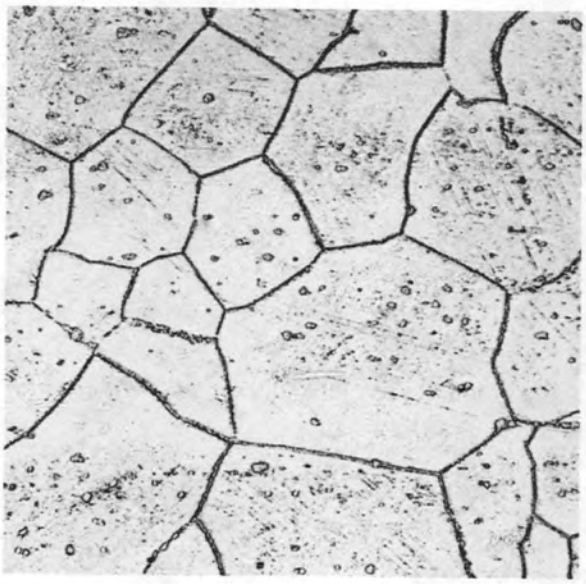
Y-201805 Y-201804



(*a*) 200 μm (*b*) 40 μm

Fig. 40. Inconel 617 base metal (XX01A3US, sample X-119) after aging 53,000 h at 704°C in HTGR-He and straining to failure at 25°C. Failed with 11.0% reduction in area.

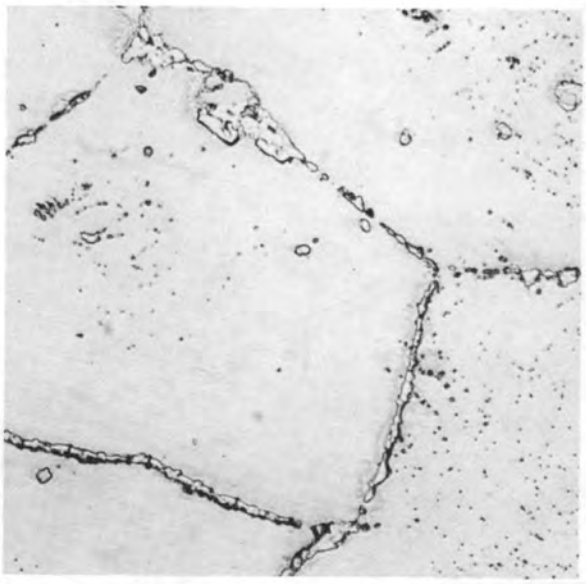
Y-201811



(*a*)

200 μm

Y-201810



(*b*)

40 μm

Fig. 41. Inconel 617 base metal (XX01A3US, sample I-123) after aging 53,000 h at 871°C in HTGR-He and straining to failure at 25°C. Failed with 2.8% reduction in area.

6.4 INCONEL 617 TRANSVERSE WELD SAMPLES

Typical optical micrographs of a transverse weld sample of Inconel 617 after aging 53,000 h at 593°C in HTGR-He are shown in Fig. 42. Failure occurred in the weld metal, and the fracture surface shows considerable deformation (Fig. 30). The yield and ultimate strengths of this sample were increased dramatically by aging at 593°C (Fig. 16), and the fracture ductility was reduced only slightly (Fig. 17). After aging 53,000 h at 704°C, fracture occurred in the base metal, and the microstructure shown in Fig. 43 was obtained. As noted previously for the base metal sample, this material had a higher than average yield strength, an average ultimate tensile strength (Fig. 16), a low reduction in area (Fig. 17), and a fracture that was intergranular and brittle (Fig. 31). Aging at 871°C also resulted in the failure occurring in the base metal. The microstructure with its continuous network of heavy precipitate is shown in Fig. 44. This structure had an average yield strength and a very low ultimate tensile stress (Fig. 16), a reduction of area of only 0.6% (Fig. 17), and a fracture that was intergranular with considerable surface deformation (Fig. 32).

7. CREEP STUDIES

Twelve creep specimens were aged under the same conditions described for the tensile specimens. The creep specimens and their exposure and test conditions are described in Table 10. Four of the tests are completed, five specimens are in test, and three tests have not been started. The completed results will be made available as soon as possible, but in the interim, we will use the limited test results available for tentative conclusions. Three pieces of information besides the partial creep test results are useful:

(1) The microstructures and the tensile results of duplicates of all 12 creep samples are known.

(2) Creep results for samples of Hastelloy X base metal and welds aged 20,000 h were compared with those for as-received materials in Ref. 2.

(3) Creep results for samples of Inconel 617 base metal aged 20,000 h were compared with those for as-received materials in Ref. 3.

Y-201791 Y-201789

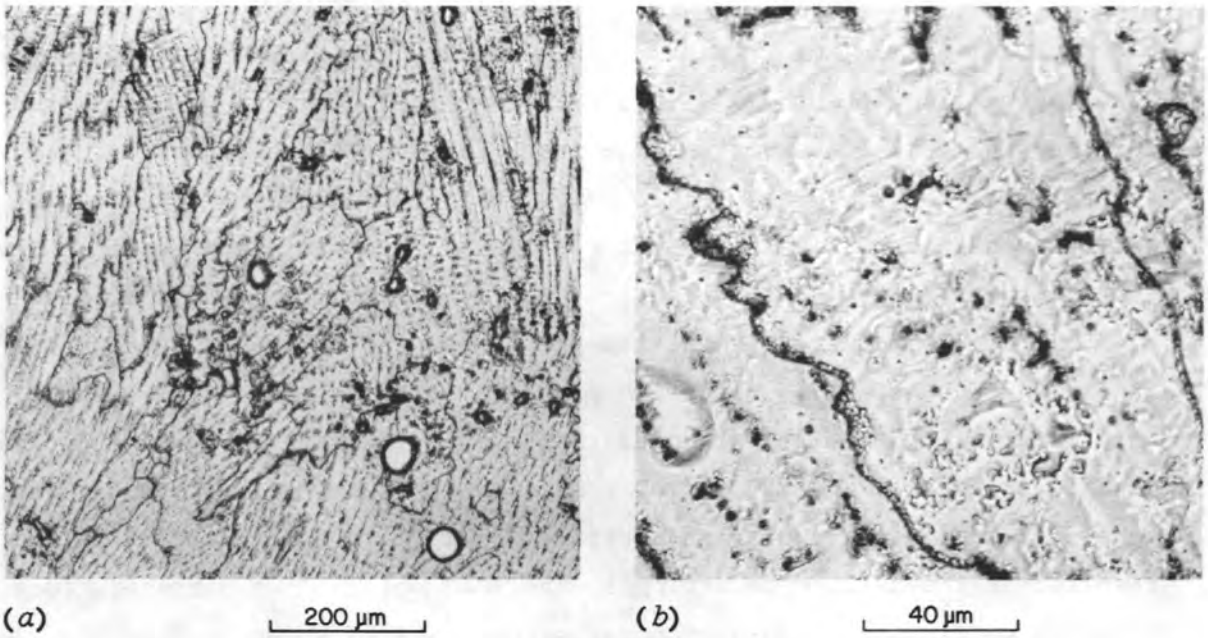


Fig. 42. Photomicrographs near fracture of transverse weld sample (I-173) of Inconel 617 after aging 53,000 h at 593°C in HTGR-He and straining to failure at 25°C. Failed with 28.1% reduction in area.

Y-201793 Y-201792

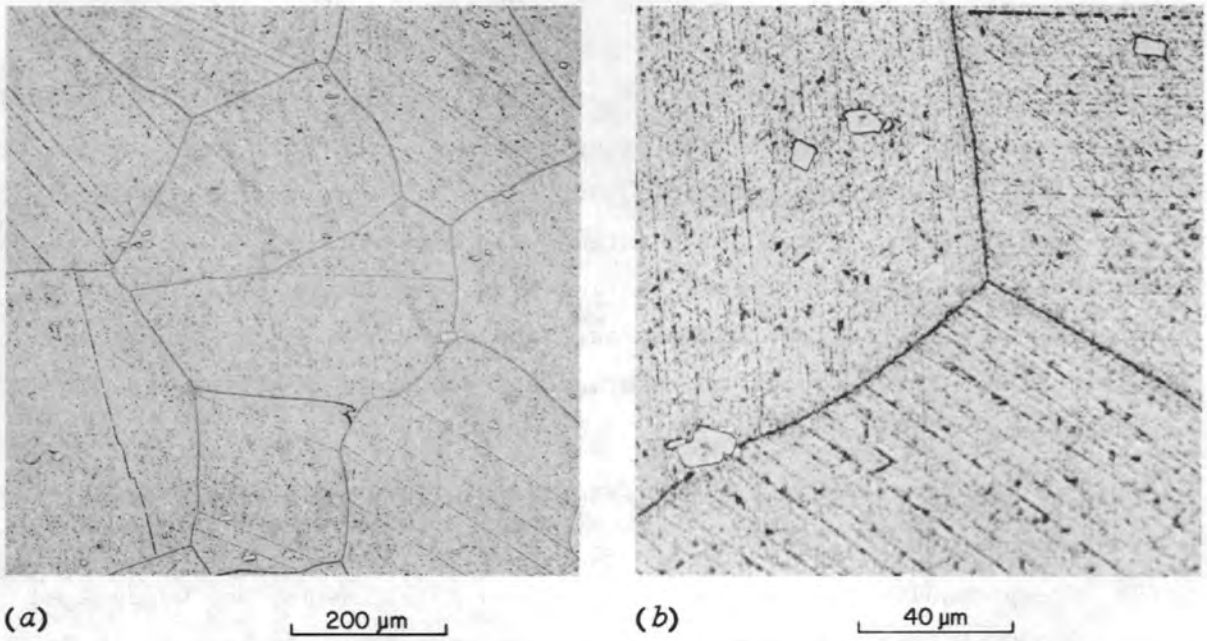
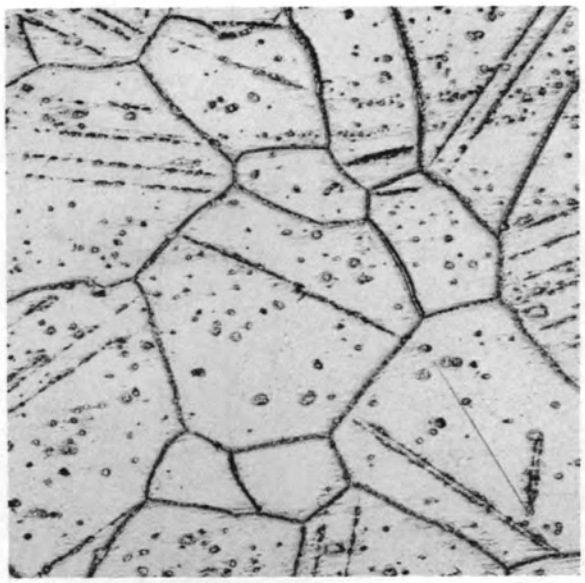


Fig. 43. Base metal near fracture of transverse weld sample I-167 of Inconel 617 after aging 53,000 h at 704°C in HTGR-He and straining to failure at 25°C. Failed with 6.8% reduction in area.

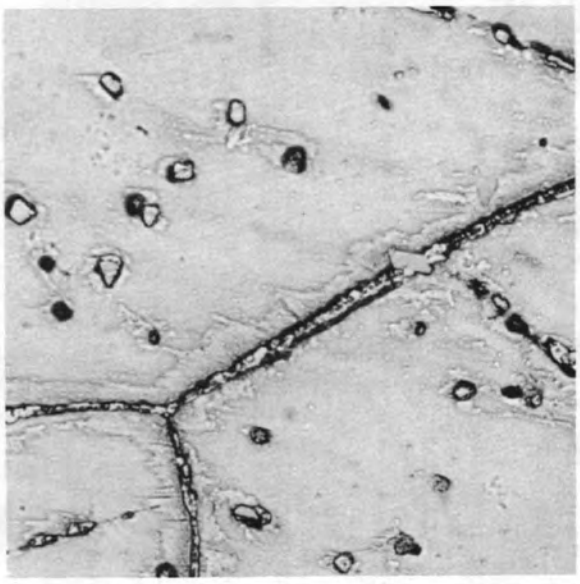
Y-201785



(*a*)

200 μm

Y-201784



(*b*)

40 μm

Fig. 44. Base metal near fracture of transverse weld sample I-171 of Inconel 617 after aging 53,000 h at 871°C in HTGR-He and straining to failure at 25°C. Failed with 0.6% reduction in area.

Table 10. Summary of results of creep tests on samples aged 53,000 h in HTGR-He

Test	Aging temperature (°C)	Test temperature (°C)	Test temperature (°F)	Stress (MPa)	Stress (ksi)	Time to creep (h) 1%	Time to creep (h) 2%	Time to creep (h) 5%	Time to tertiary creep (h)	Rupture life (h)	Steady-state creep rate (h^{-1})	Elongation (%) Loading	Elongation (%) Creep	Reduction in area (%)
Hastelloy X base metal														
24205	593	593	1100	345	50	33	140	460	800	1707	9.1 E-3	0.43	41.6	58.4
24211[a]	704	704	1300	138	20	1400	1950		1600	2350	1.0 E-3		3.0	
24329[a]	871	871	1600	48	7					50				
Hastelloy X transverse weld														
24210	593	593	1100	345	50	90	228	420	250	515	7.5 E-3	0.64	10.6	54.1
24217	704	704	1300	138	20	356	750	1300	860	1524	2.3 E-3	0.13	12.7	54.6
24323[a]	871	871	1600	48	7					570	neg.	0.20	0.15	
Inconel 617 base metal														
24321[a]	593	593	1100	414	50					834	2.4 E-5	*b*	0.21	
24322[a]	704	704	1300	207	30					717	8.7 E-5	*b*	0.15	
24330[a]	871	871	1600	48	7					404				
Inconel 617 transverse weld														
24203[a]	593	593	1100	414	60	*b*	*b*	*b*	*b*	129	6.2 E-4	*b*	2.3	7.3
24337[a]	704	704	1300	207	30					20				
24331[a]	871	871	1600	48	7					405				

[a]Test still in progress.

[b]Data will not be available for these items.

The data available thus far for Hastelloy X are compared with our previous data in Fig. 45. The minimum creep rates for the base metal and transverse weld metal aged 53,000 h at 593°C are in excellent agreement with each other. The data population is rather small at 593°C, but the two samples aged 53,000 h at 593°C and tested at 593°C have creep rates about two orders of magnitude greater than those of other samples tested at this temperature and stress. Thus it appears that the copious amounts of grain boundary precipitate shown in Figs. 33 and 36 have led to reduced creep strength. Similarly the two specimens, base and transverse weld, aged and tested at 704°C have minimum creep rates about two orders of magnitude higher than those of samples tested in the as-received condition or after aging for 10,000 or 20,000 h. It would again appear that the large amount of grain boundary precipitate formed (Figs. 34 and 37) has led to a reduction in the creep strength.

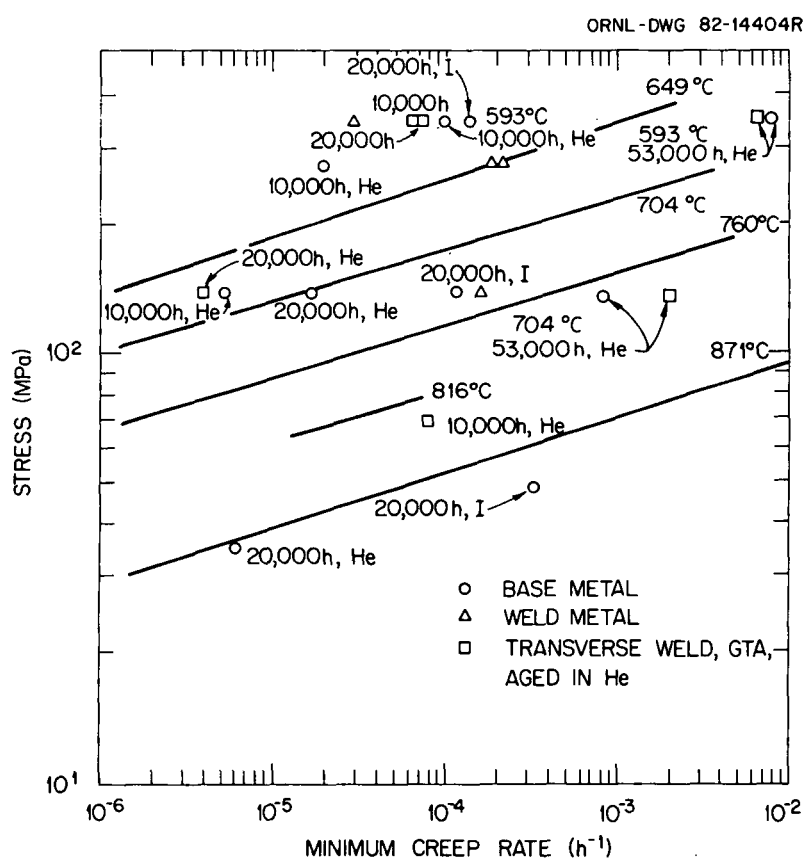


Fig. 45. Comparison of the creep rate of Hastelloy X in a number of conditions. The lines represent as-received Hastelloy X base metal.

The creep results for base and transverse weld samples of Inconel 617 aged 53,000 h are compared in Fig. 46 with the other data that we have on this material. One base metal sample was aged and is in test at 593°C. The agreement with data obtained on unaged material is good. Two samples were aged at 704°C, one base metal and one transverse weld, and both appear to have creep rates from one to one and one-half orders of magnitude greater than that of the unaged material. The massive amounts of precipitate formed in Hastelloy X (Figs. 40 and 44) appear to reduce the creep strength.

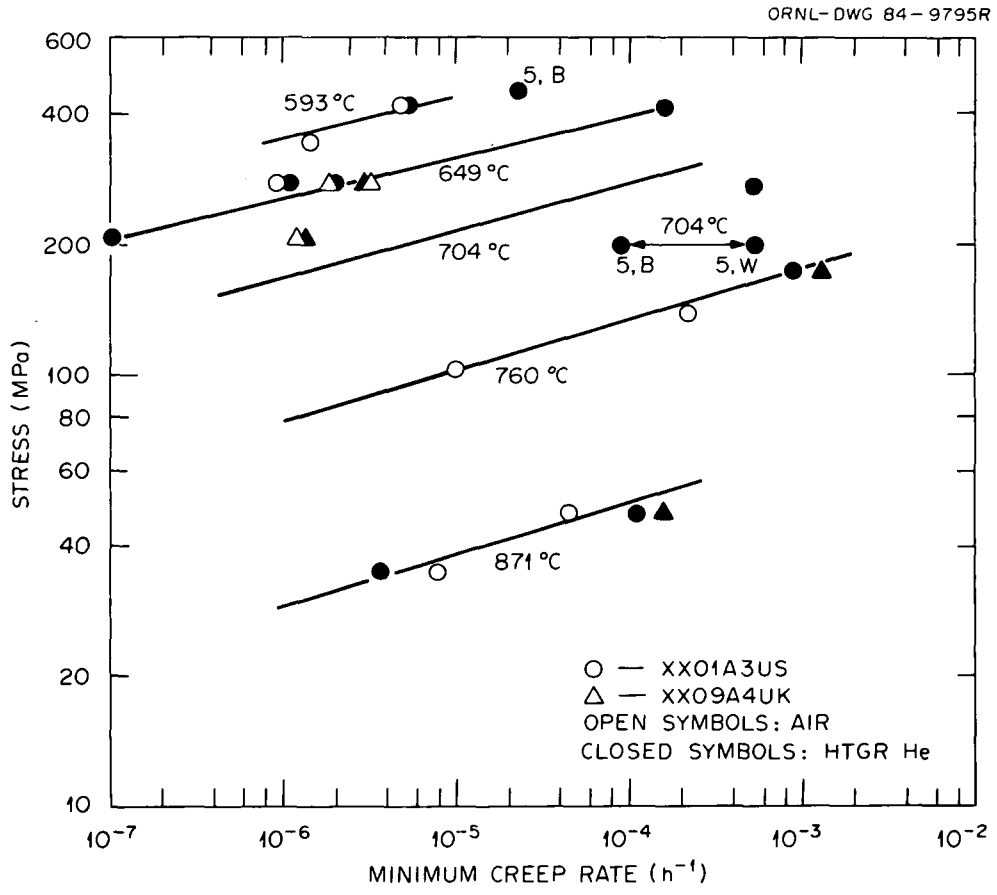


Fig. 46. Minimum creep rate of Inconel 617. Aged samples are denoted by the symbols 5, B, and W denoting 53,000 h aging, base metal, and transverse weld metal, respectively.

8. DISCUSSION

The design of alloys for good long-term high-temperature strength results in rather complex systems that usually have some degree of metallurgical instability at the service temperature. A common change in many alloys involves carbides that may form initially, change morphology, or change stoichometry. Property changes due to carbide formation are usually detectable, but they are seldom very dramatic because most alloys contain only 0.05 to 0.08% carbon. However, in the HTGR environment, there is an external supply of carbon, and the formation of carbide precipitates could continue until all the carbide-forming elements such as Mo, Cr, and Ti were depleted from the matrix. Some of the more complex heat-resistant alloys form intermetallic compounds such as $Ni_3(Al,Ti)$ and a spectrum of others. As noted in Section 6, Lai[4] concluded on the basis of 10,000 h aging data that the principal precipitating phase in Hastelloy X was M_6C-type carbides and that small amounts of two different intermetallics formed during aging at 760 and 871°C. Mankins et al.[5] concluded that Inconel 617 formed $M_{23}C_6$ at all temperatures and small amounts of gamma prime at 649 and 760°C. The aging times in the work of Mankin et al. varied from 200 to 10,000 h. Thus, the literature leads one to conclude that the two alloys being evaluated, Hastelloy X and Inconel 617, are rather simple.

The 24 samples evaluated in this investigation are very important because they were exposed for 53,000 h to a typical HTGR environment at 593, 704, and 871°C. These are realistic operating temperatures, and the exposure time of about 6 years is a significant fraction of the scheduled operating lifetime of an HTGR. The samples were equally divided among Hastelloy X base metal, Hastelloy X transverse weld sample, Inconel 617 base metal, and Inconel 617 transverse weld sample. Also, the samples were equally divided among ambient temperature short-term tensile tests and reasonably long-term (several thousand hours) creep tests conducted at the aging temperature. Both alloys welded well, and the transverse weld sample failed either in the base metal or the weld metal with very little

difference in strength or ductility. The ambient temperature condition is of concern because reactors will be cooled from time to time for maintenance, and so the properties at ambient conditions are important.

The HTGR environment was carburizing to both alloys, but the large amount of reactive surface area in the aging retorts led to some depletion of the carburizing species from the test gas. The greatest amount of carburization occurred at 871°C, the highest temperature examined in this study.

The aging behavior of Hastelloy X was somewhat surprising. Observations were made previously after aging 10,000 and 20,000 h. After aging 10,000 h the strength was a maximum at 704°C, and the fracture strain decreased most markedly due to aging at 871°C. This trend continued after 20,000 h aging. After 53,000 h aging, some overaging effects were noted at 704 and 871°C, and the largest increase in strength and decrease in fracture strain occurred at 593°C. The same trends were noted in the Hastelloy X transverse weld specimens.

The primary difference with the Inconel 617 from the shorter aging times was that large property changes occurred after aging 20,000 h at 593°C, and the changes were even greater after 53,000 h. However, some overaging effects were noted after aging 53,000 h at 704 and 871°C, just as noted with Hastelloy X.

From an absolute standpoint, the fracture strains of all samples aged 53,000 h at 704 and 871°C were 10% or less, and those aged at 593°C fractured at strains of 20 to 30%. The properties after aging at 593°C for 53,000 h are currently quite acceptable, but there is no evidence that the properties will not continue to degrade at this aging temperature.

Several sets of observations (e.g., Fig. 15) show that the fracture strains are less after aging in HTGR-He than in inert gas. This may be due to the carburization in the HTGR-He and the formation of more carbide precipitates. Surface corrosion effects in HTGR-He are another possible explanation.

The microstructures and the fracture surface details agreed very well. There is no doubt that the continuous intergranular precipitate formed at 871°C leads to brittle intergranular fracture. None of the microconstituents were identified.

The creep program is only partially complete, but some trends seem established. Samples of Hastelloy X aged at 593 and 704°C are creeping faster than unaged material. One sample of Inconel 617 aged at 593°C for 53,000 h was creeping at about the same rate as the unaged material. Two Inconel 617 specimens aged and being tested at 704°C are creeping faster than the unaged material. It is postulated that the massive amounts of precipitate formed as a result of 53,000 h aging caused depletion of solid-solution strengthening elements from the matrix with the concomitant reduction in creep strength.

9. CONCLUSIONS

Twenty-four samples of Hastelloy X and Inconel 617 were exposed to HTGR He for 53,000 h. The aging temperatures were 593, 704, and 871°C. The samples were equally divided among base metal and those with a transverse weld. The materials welded quite well, and the weld areas were usually as strong as the base metal. Half of the samples were subjected to short-term tensile tests at 25°C. Aging times of 2500, 10,000, and 20,000 h had produced large changes in properties at 704 and 871°C. The aging duration of 53,000 h caused large increases in strength and decreases in fracture strain for samples aged at 593°C, and some overaging occurred in samples aged at 704 and 871°C. The continued change of properties, particularly fracture strain, with longer times approaching realistic HTGR service lifetimes, is a major concern. The copious quantities of precipitate that form in these two materials during aging reduce the fracture strain when the precipitates form continuous intergranular networks, and they are likely to reduce the creep strength as solid-solution strengthening elements are removed from solution.

10. ACKNOWLEDGMENTS

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