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Mechanical Properties of Inconel 617 and 618

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MECHANICAL PROPERTIES OF INCONEL 617 AND 618

H. E. McCoy and J. F. King

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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MECHANICAL PROPERTIES OF INCONEL 617 AND 618*

H. E. McCoy and J. F. King

ABSTRACT

Inconel 617 and 618 were evaluated for application in high-temperature gas-cooled reactors (HTGRs). Techniques were developed for making sound welds, and tests were performed on base and weld metals. Specimens of both materials were aged to 20,000 h to evaluate thermal stability. Short-term tensile tests on alloy 617 showed that aging severely reduced the strain at fracture at both ambient and elevated temperatures. The impact energy at ambient temperature was severely degraded by aging. Creep tests showed that fracture occurred at 593 through 704°C after only 1 to 2% strain, and higher strains were noted at higher temperatures. There was no detectable difference between the creep behavior in air and that in HTGR helium environments. Inconel alloy 618 had excellent stability during aging. Fracture strains in short-term tensile tests and impact energies in impact tests remained high after aging. The creep properties of alloy 618 were equivalent in air and in HTGR helium. Both alloys were carburized during creep testing in HTGR helium, and the rate of carburization became rather high at 760°C and higher temperatures.

INTRODUCTION

Inconel 617 and 618 were developed for high-temperature applications by the Huntington Alloy Products Division of the International Nickel Company. Inconel 617 has a nominal composition of Ni—22% Cr—12.5% Co—9% Mo—1% Al—0.07% C and is basically a solid-solution alloy. Numerous details on microstructure and phase stability were reported by Markins, Hosier, and Bassford.[1] They noted that the major second phase in the alloy under all aging conditions was $M_{23}C_6$, and a PHACOMP analysis indicated that 0.63% gamma prime could form. The alloy has good strength at elevated temperatures, and Hosoi and Abe noted in creep tests at 1000°C

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that the creep strength in helium was very sensitive to the oxygen concentration over some concentration range.[2] This influence on creep strength was attributed to decarburization.

Inconel 618, which is free of cobalt, was developed for nuclear applications. It has a nominal composition of Ni—22.5% Cr—15% Fe—6% W—0.05% C. There is considerably less experience with this alloy than with Inconel 617.

In high-temperature gas-cooled reactor (HTGR) application, structural alloys will be exposed to a gaseous environment consisting primarily of helium with small amounts of H_2, CH_4, CO, CO_2, N_2, and H_2O. Previous studies of nickel-base alloys have shown that most combinations of those impurities carburize most nickel-base alloys.[3-5] Many of the alloys undergo microstructural modifications when heated for long times, and the infusion of carbon into the alloys during heating can lead to further modifications. Thus, the mechanical properties of structural alloys for HTGRs need to be determined in realistic service environments.

EXPERIMENTAL DETAILS

MATERIALS

Several test materials were involved in this program. Their chemical compositions and product forms are presented in Tables 1 and 2, respectively. Three heats of Inconel 617 base metal and one heat of weld metal were studied. One heat of Inconel 618 base metal and wire was evaluated. All welding was by the gas tungsten arc (GTA) process.

All three heats of Inconel 617 base metal had fairly coarse grains. Typical photomicrographs of heat XX01A3US are shown in Fig. 1. The microstructure is characterized by inhomogeneous stringers in the longitudinal direction. Photomicrographs of heat XX09A4UK are shown in Fig. 2. This heat has a slightly coarser grain size than heat XX01A3US, and the stringers are not nearly as pronounced as in heat XX01A3US. Photomicrographs were not available for as-received heat XX14A6UK; however, the sample shown in Fig. 3 had been aged at 593°C, and grain size would not have been altered. This heat has the largest grain size of the three heats of base metal evaluated but shows no evidence of longitudinal

stringers. The precipitate morphology noted in this sample due to aging at 593°C is quite similar to that noted by Markins, Hosier, and Bassford[1] when this alloy had been aged for 3000 h at 816°C.

Table 1. Alloy compositions of materials studied

Heat	Content (wt %)											
	Ni	Cr	Cu	Co	Mn	Mo	Fe	Al	Si	C	S	W
						Inconel 617						
XX01A3US	57.35	20.30		11.72	0.05	8.58	1.01	0.76	0.16	0.07	0.004	
XX09A4UK	55.11	21.83		12.55	0.02	8.79	0.38	1.15	0.10	0.07	0.001	
XX14A6UK	55.13	21.74		12.32	0.02	8.91	0.53	1.11	0.18	0.06	0.002	
XX09A9UK	54.38	22.14		12.66	0.02	8.79	0.39	1.37	0.17	0.08	0.001	
						Inconel 618						
Y09B9	56.49	22.5	0.04		0.09		14.79	0.01	0.021	0.04	0.005	5.85

Table 2. Product characterization of materials produced by Huntington Alloy Products Division

Heat	Product form	Grain size (μm)	Anneal
	Inconel 617		
XX01A3US	Plate, 13 mm (0.5 in.) thick	130	Solution annealed by vendor
XX09A4UK	Plate, 9.5 mm (0.38 in.) thick	210	Unknown
XX14A6UK	Plate, 13 mm (0.5 in.) thick	270	Solution annealed by vendor
XX09A9UK	Wire, 1.1 mm (0.045 in.) diameter		Unknown
	Inconel 618		
Y09B9	Plate, 13 mm (0.5 in.) thick	32	Unknown
	Wire, 1.6 mm (0.063 in.) diameter		Unknown

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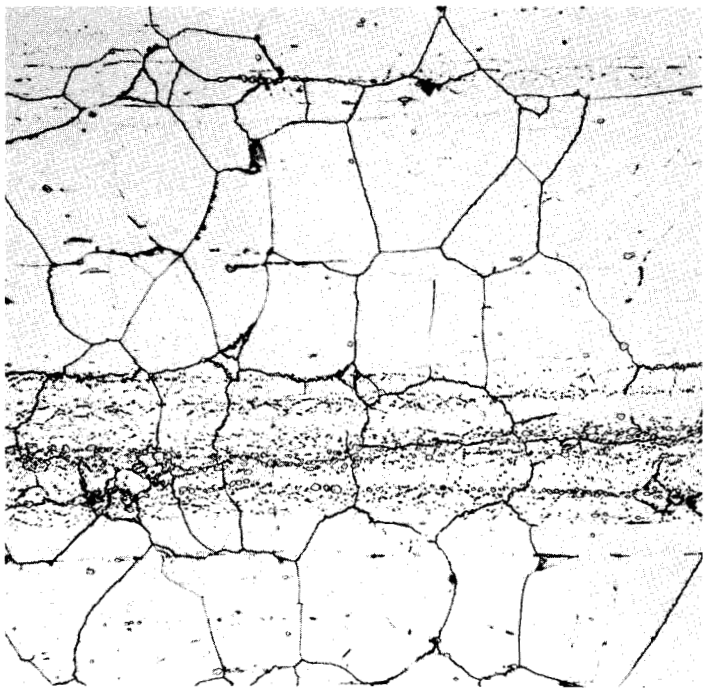

(*a*)

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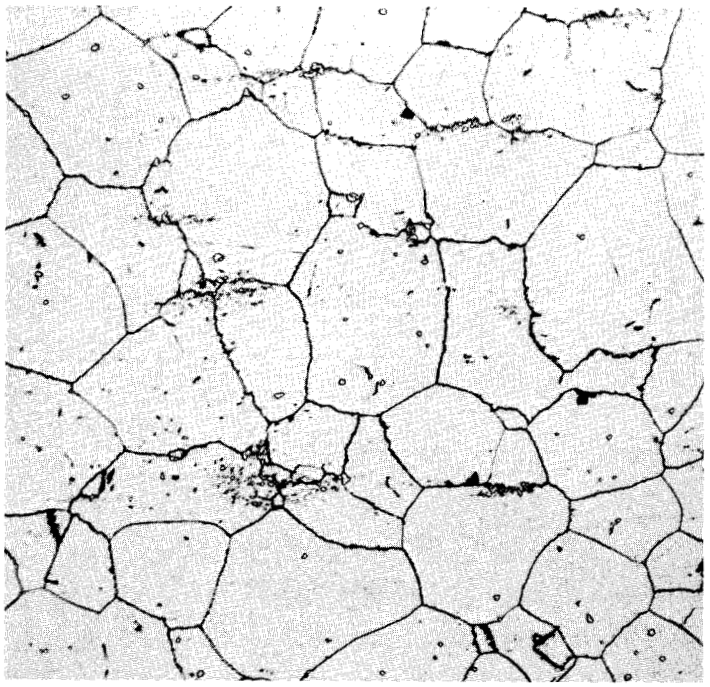

(*b*)

200 μm

Fig. 1. As-received Inconel 617 (heat XX01A3US). (*a*) Longitudinal view. (*b*) Transverse view.

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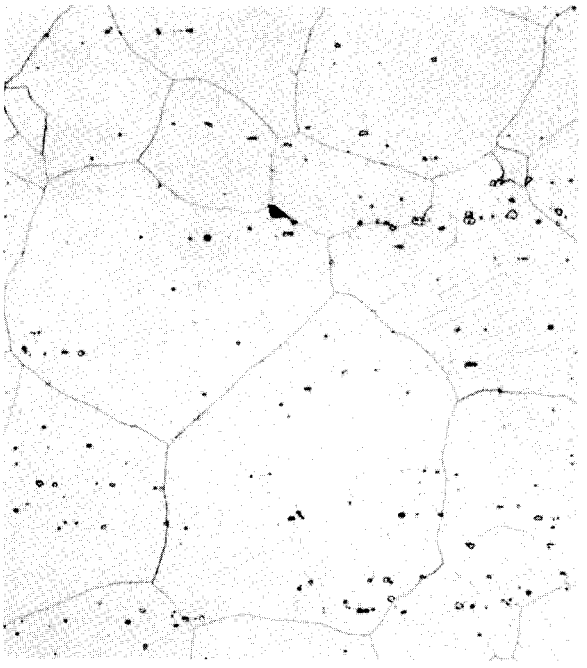

(*a*) 200 μm

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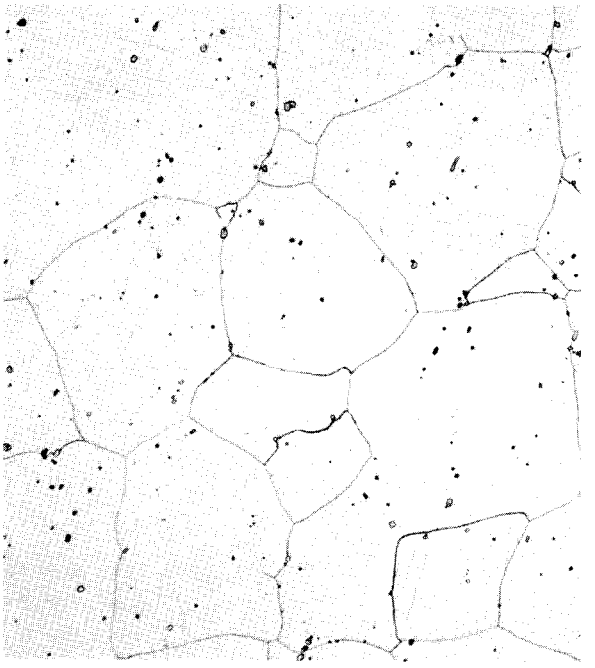

(*b*) 200 μm

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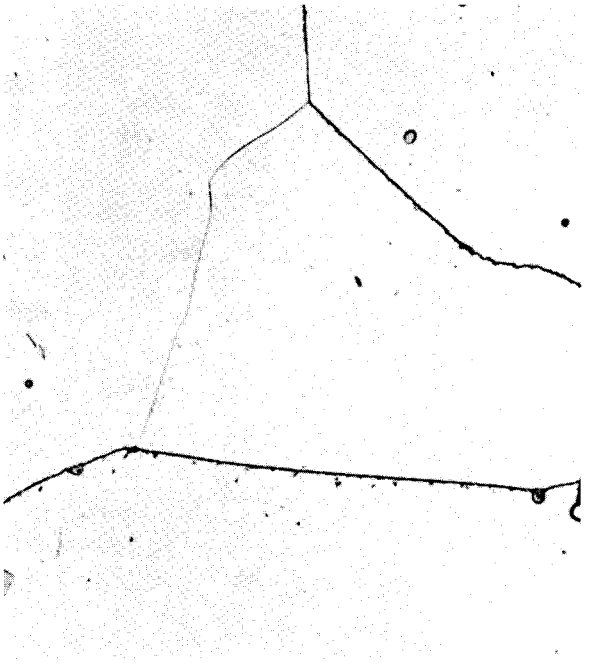


(*c*) 40 μm

Fig. 2. As-received Inconel 617 (heat XX09A4UK). (*a*) Longitudinal view. (*b*) Transverse view. (*c*) Transverse view at higher magnification.

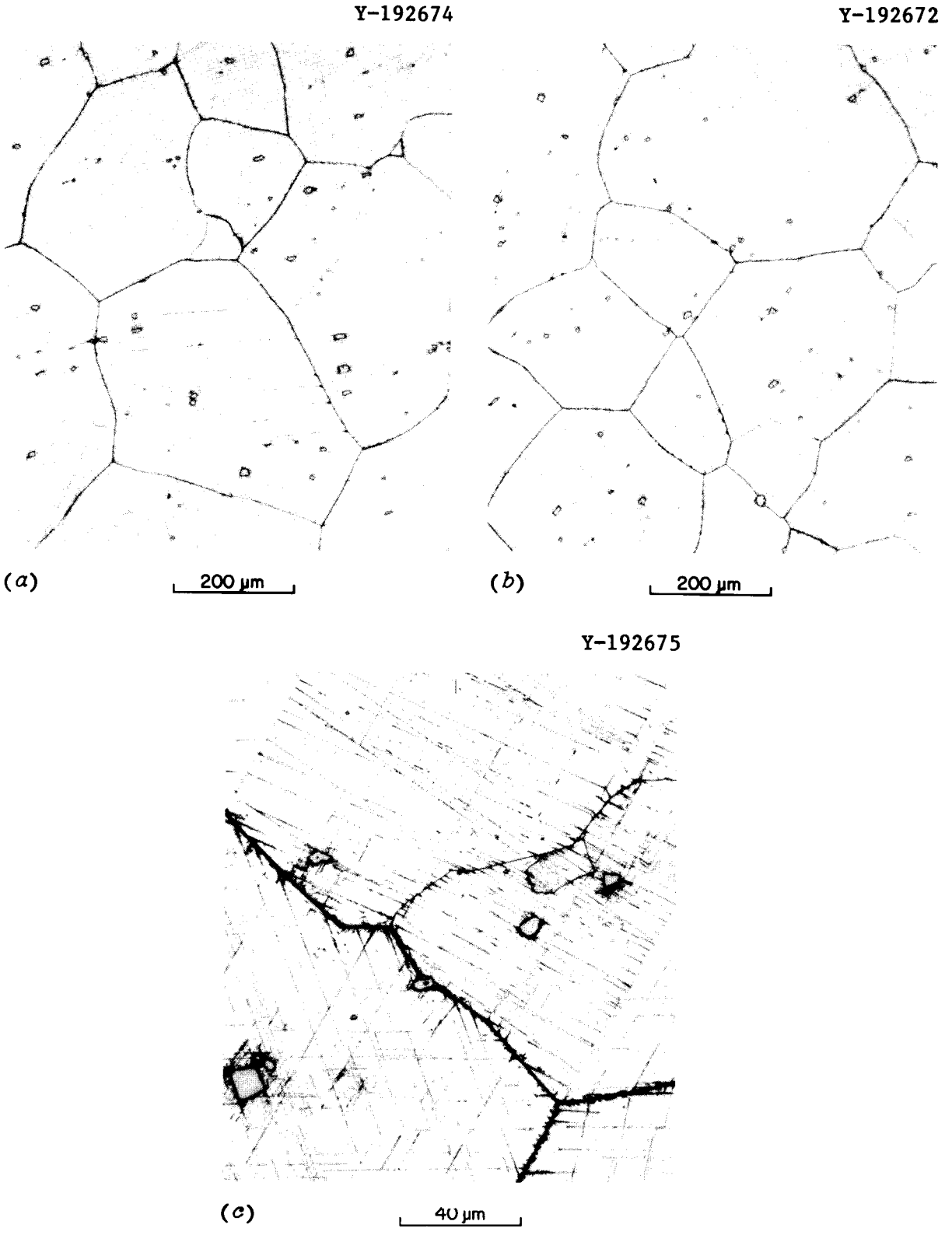


Fig. 3. Inconel 617 (heat XX14A6UK) after aging 10,000 h at 593°C. (*a*) Longitudinal view. (*b*) Transverse view. (*c*) Transverse view at higher magnification.

The microstructure of as-received Inconel 618 is shown in Fig. 4. The annealing treatment is unknown, but the grain size was quite small and there was little evidence of stringers. Some of the material was annealed for 1 h at 1177°C (2150°F) before testing. Typical microstructures of the sample material in this condition are shown in Fig. 5. This anneal increased the grain size considerably.

TEST SPECIMENS

The geometry of the test specimen is shown in Fig. 6. A good feature of the specimen is the absence of threads: the grips align on the 6.35-mm (0.250-in.) diameter and pull against the surface inclined at 60°. The extensometer attaches by setscrews to the two grooves, which have an included angle of 100°.

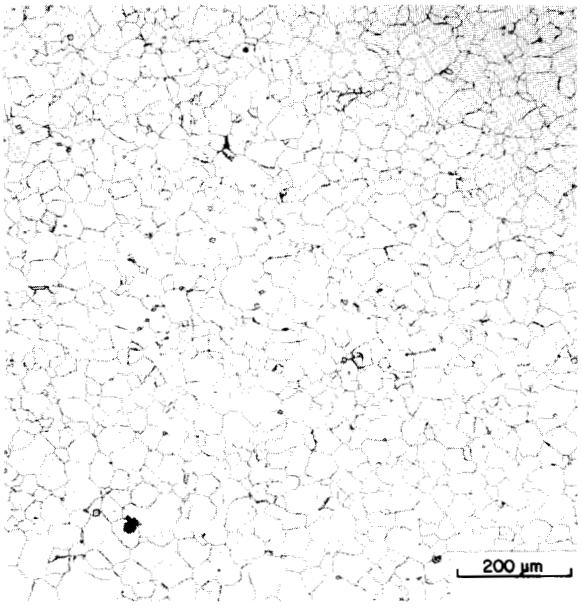
Base-metal specimens were oriented so that the specimen axis was parallel with the primary working direction. Transverse weldment specimens were taken so that the gage section of each specimen spanned the weld and contained the base metal, the weld metal, and the heat-affected zone. All-weld-metal specimens were cut so that they were parallel to the weld, and the gage section consisted totally of weld metal.

TESTING EQUIPMENT

The creep testing chambers used in this study were described in detail previously.[4] 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. Neoprene U-cups are used to seal between the pull rods and the environmental chamber and also to seal the four extensometer rods that exit the bottom of the test chamber.

Some samples aged in inert environments at 538, 704, and 871°C were simply stacked in metal containers, which were later evacuated, filled with an inert gas, and welded shut. Specimens were aged about 20,000 h in HTGR helium in the stainless steel retorts shown in Fig. 7. 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. (The term "HTGR helium" is further abbreviated as

Y-192676



(*a*)

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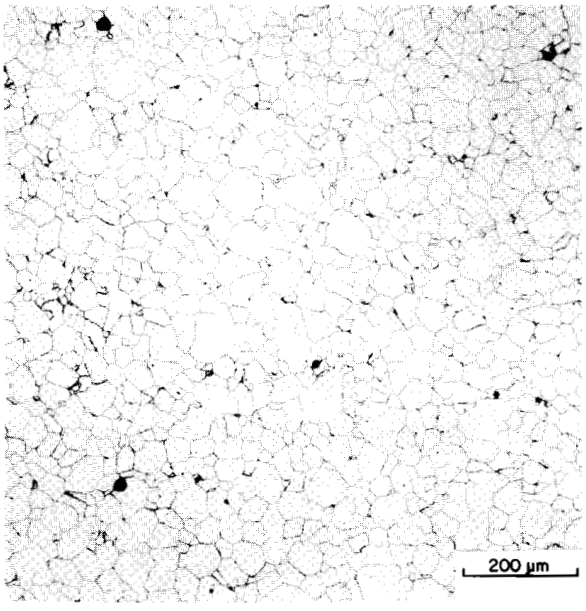

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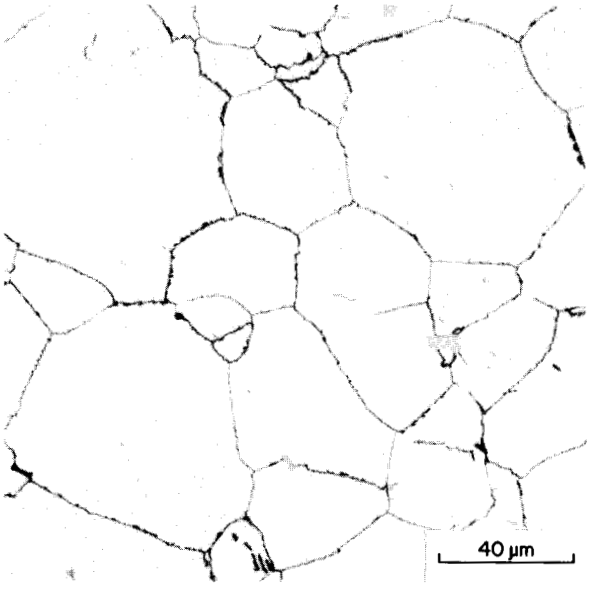

(*c*)

Fig. 4. As-received Inconel 618 (heat Y09B9). (*a*) Longitudinal view. (*b*) Transverse view. (*c*) Transverse view at higher magnification.

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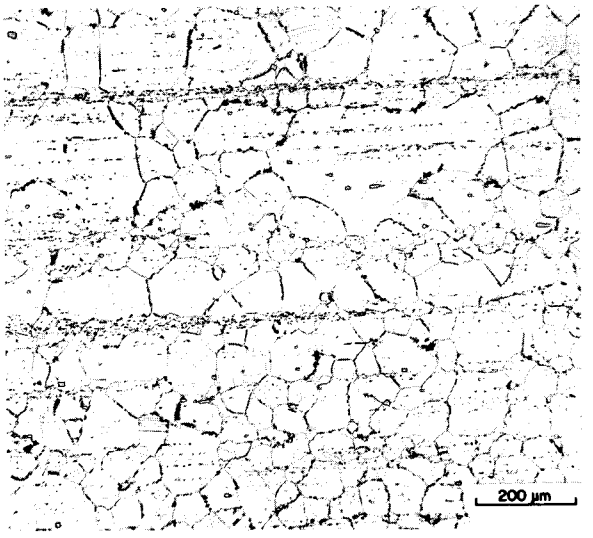

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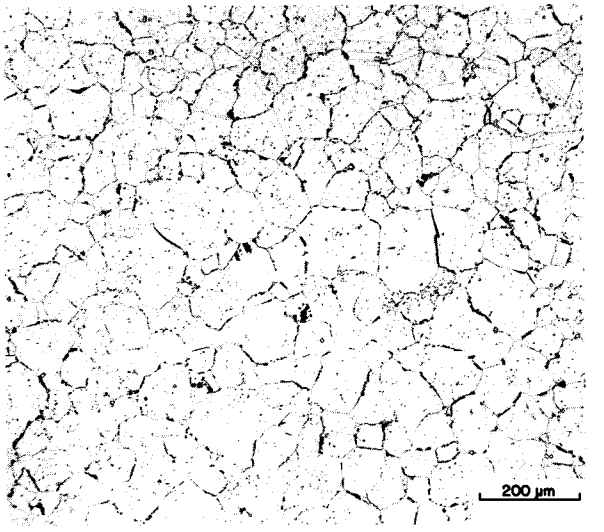

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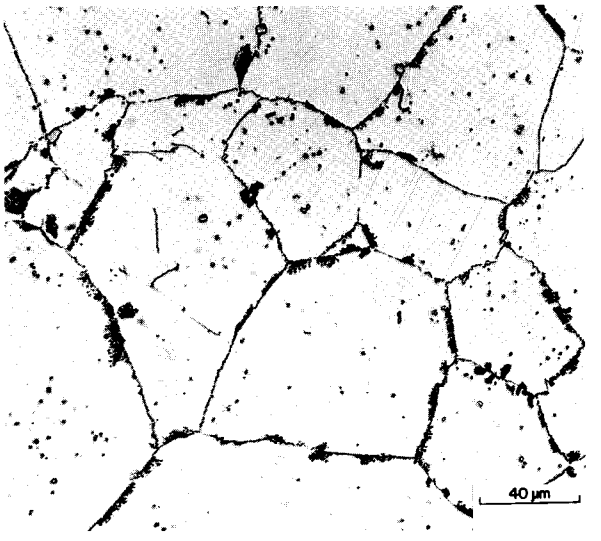

(*c*)

Fig. 5. Inconel 618 (heat Y09B9) after being annealed for 1 h at 1177°C. (*a*) Longitudinal view. (*b*) Transverse view. (*c*) Transverse view at higher magnification.

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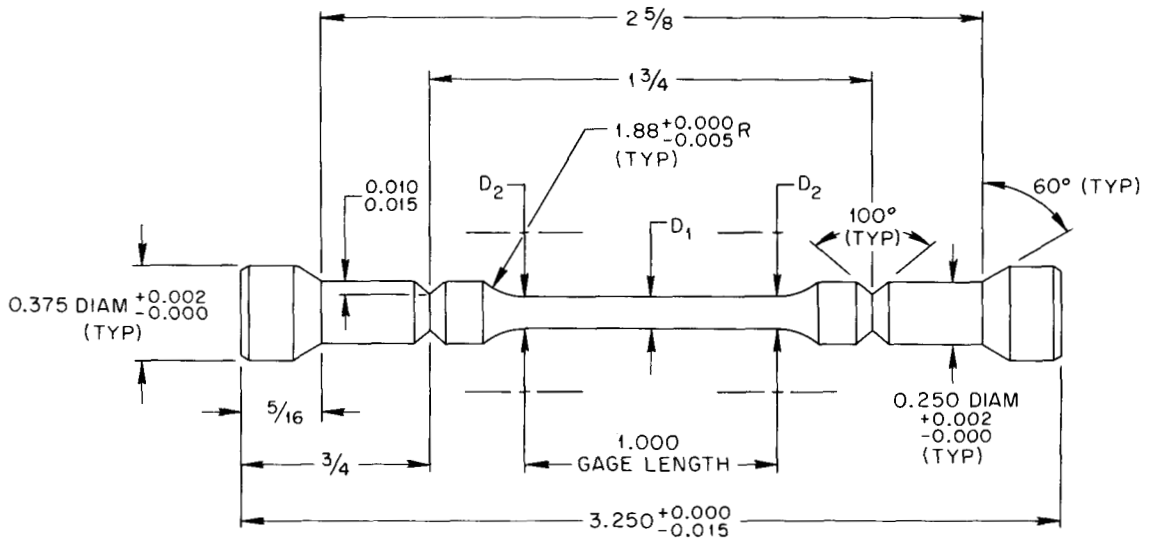


NOTE: ALL DIMENSIONS IN INCHES

$D_1 = 0.125 \pm 0.001$ DIAM

$D_2 =$ FROM 0.0010 TO 0.0015 GREATER THAN D_1

Fig. 6. Details of creep specimens for environmental tests. To convert dimensions to millimeters, multiply by 25.4.

"HTGR-He" in the rest of this report. It refers to a gas mixture, defined in the following subsection, that approximates the composition of the impure helium in an operating HTGR.)

TEST ENVIRONMENT

The test gas (HTGR-He) is premixed by the Matheson Company, Inc. It is received in pressurized cylinders and 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 through a gas chromatograph to monitor moisture 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 flow rate high enough to keep the gas stream from being detectably depleted in reactants. The helium composition in the test chamber is [in pascals (microatmospheres)]

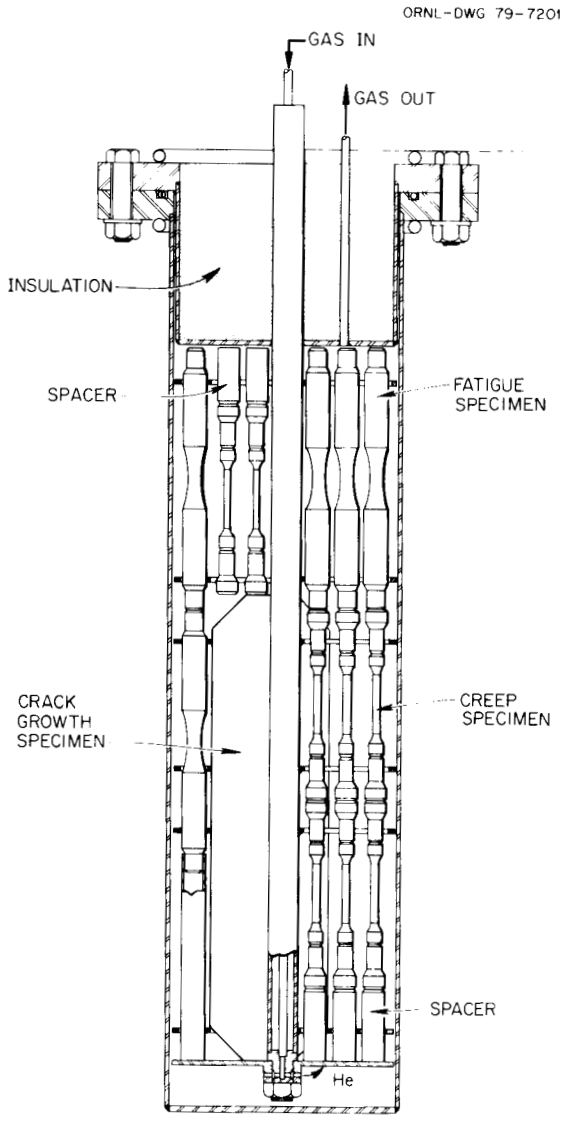


Fig. 7. Environmental exposure chamber.

34.0 (337) H_2, 3.2 (32) CH_4, 1.9 (19) CO, 0.2 (2) H_2O, and <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 against a certified standard prepared by the Matheson Company, Inc., and weekly readings are made routinely on each machine.

The 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.

The standard test startup consisted of assembling the new test specimen in the chamber, evacuating air from the chamber, pressurizing 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 was less than 10 ppm, heating to test temperature, and applying load to start the test. All impurities were initially high, but 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.

RESULTS OF TENSILE TESTS

INCONEL 617 BASE METAL

Tensile properties of the as-received and aged Inconel 617 (heat XX01A3US) are given in Tables 3 and 4. Samples were aged to 20,000 h in an inert environment and in HTGR-He. Tensile properties are given in Table 5 for as-received Inconel 617 (heat XX09A4UK). The properties of the two heats studied vary appreciably in the as-received condition. As shown in Fig. 8, the strength properties of heat XX01A3US decrease smoothly with increasing test temperature, whereas heat XX09A4UK shows some erratic behavior from 600 to 750°C. Heat XX09A4UK is stronger over the entire range of test temperatures, particularly above 600°C. The fracture strain and reduction of area are different, especially above 600°C, where the ductility increases with increasing temperature for heat XX01A3US and decreases with increasing temperature for heat XX09A4UK (Fig. 9).

Table 3. Tensile properties of Inconel 617 (heat XX01A3US) as received and after aging in static air

[Specimens machined after aging. Gage sections were 31.8 mm (1.25 in.) long by 6.4 mm (0.25 in.) in diameter.]

Test	Aging condition			0.2% Yield strength		Ultimate tensile strength		Elongation[a] (%)		Reduction of area (%)
	Temperature		Time (h)							
	(°C)	(°F)		(MPa)	(ksi)	(MPa)	(ksi)	Uniform	Total	
				Tested at 22°C (72°F)						
16545			0	303	43.9	747	108.3	54.1	54.3	36.2
16546			0	301	43.7	757	109.8	59.6	60.9	40.7
16547			0	305	44.3	754	109.3	57.4	57.9	32.4
17897	538	1000	2,500	358	51.9	794	115.2	66.0	68.1	49.3
17898	583	1000	2,500	360	52.2	796	115.5	67.6	69.8	48.6
17901	704	1300	2,500	371	53.8	847	122.9	39.2	39.2	32.0
17902	704	1300	2,500	372	54.7	848	123.1	44.1	44.1	35.0
17905[b]	871	1600	2,500	331	48.0	797	115.6	29.9	29.9	25.0
17906	871	1600	2,500	334	48.5	822	119.2	34.6	34.6	29.3
IC43	538	1000	10,000	392	56.9	805	116.8	53.3	53.4	43.8
IC25	704	1300	10,000	369	53.5	798	115.7	29.6	29.6	23.3
IC7	871	1600	10,000	290	42.0	695	100.8	28.3	28.3	21.0
IC38	538	1000	20,000	379	55.0	822	119.2	62.4	62.7	43.7
IC20	704	1300	20,000	395	57.3	813	117.9	21.1	21.1	17.3
IC02	871	1600	20,000	300	43.5	658	95.4	20.2	20.2	15.9
				Tested at 538°C (1000°F)						
16657			0	216	31.3	610	88.5	64.7	67.3	49.8
16658			0	207	30.0	606	87.9	61.1	64.4	46.1
17899	538	1000	2,500	260	37.7	638	92.5	63.4	68.5	51.7
17900	538	1000	2,500	259	37.6	638	92.5	66.3	69.2	48.9
IC44	538	1000	10,000	222	32.2	669	97.0	36.4	36.5	30.5
IC26	704	1300	10,000	274	39.8	629	91.3	58.9	58.9	46.3
IC39	538	1000	20,000	281	40.8	621	90.0	64.6	66.2	51.4
				Tested at 704°C (1300°F)						
16664			0	199	28.8	466	67.6	29.3	68.0	49.3
16665			0	196	28.4	443	64.2	25.9	69.9	53.1
16666			0	196	28.4	454	65.9	29.9	74.7	52.5
17903	704	1300	2,500	257	37.3	502	72.8	16.2	75.3	57.6
17904[c]	704	1300	2,500	256	37.1	475	69.0	14.4		
IC27	704	1300	10,000	211	30.6	461	66.9	16.9	68.8	56.3
IC21	704	1300	20,000	285	41.3	525	76.2	13.6	53.1	48.5
				Tested at 871°C (1600°F)						
16672			0	181	26.3	194	28.1	3.2	88.0	87.8
16673			0	189	27.4	194	28.1	3.2	91.6	86.0
17907	871	1600	2,500	154	22.3	168	24.3	5.1		
17908	871	1600	2,500	170	24.6	179	25.9	4.3	98.2	91.0
IC8	871	1600	10,000	170	24.7	181	26.2	8.0	90.9	79.2
IC3	871	1600	20,000	180	26.1	255	37.0	8.8	89.9	72.8

[a]Elongation in 31.8 mm (1.25 in.).

[b]Broke at gage mark.

[c]Pull rod broke before specimen broke.

Table 4. Tensile properties of Inconel 617 (heat XX01A3US) after aging in HTGR-He

[Specimens machined before aging. Gage sections were 25.4 mm (1.00 in.) long by 3.18 mm (0.125 in.) in diameter.]

Specimen	Aging condition			0.2% Yield strength		Ultimate tensile strength		Elongation (%)		Reduction of area (%)
	Temperature		Time (h)							
	(°C)	(°F)		(MPa)	(ksi)	(MPa)	(ksi)	Uniform	Total	
Tested at 24°C (75°F)										
I-100	593	1100	10,000	316	45.9	737	106.9	53.5	59.6	57.1
I-113	593	1100	20,000	510	74.0	919	133.3	24.4	24.8	29.1
I-104	704	1300	10,000	531	77.0	818	118.6	14.1	14.3	14.6
I-117	704	1300	20,000	523	75.8	794	115.3	9.6	10.0	15.5
I-108	871	1600	10,000	392	56.8	623	90.3	11.4	11.4	14.2
I-121	871	1600	20,000	303	44.0	509	73.8	6.0	6.2	6.8
Tested at 593°C (1100°F)										
I-101	593	1100	10,000	238	34.5	572	82.9	44.6	48.2	44.4
I-112	593	1100	20,000	378	54.8	687	99.6	24.7	25.6	30.6
Tested at 704°C (1300°F)										
I-105	704	1300	10,000	387	56.1	706	102.4	22.0	25.2	26.6
I-116	704	1300	20,000	378	54.8	698	101.2	17.5	19.3	22.0
Tested at 871°C (1600°F)										
I-109	871	1600	10,000	219	31.7	283	41.0	8.0	59.4	61.2
I-120	871	1600	20,000	171	24.8	280	40.6	5.4	63.4	54.4

Table 5. Tensile properties of as-received Inconel 617 (heat XX09A4UK)

Specimen	Test temperature		0.2% Yield strength		Ultimate tensile strength		Elongation (%)		Reduction of area (%)
	(°C)	(°F)	(MPa)	(ksi)	(MPa)	(ksi)	Uniform	Total	
24	25	77	394	57.1	766	111	53.7	56.5	62.9
200	25	77	363	52.7	765	111	48.7	50.2	53.5
201	593	1100	216	31.4	556	80.7	49.5	52.2	48.6
202	649	1200	235	34.2	549	79.7	46.0	48.3	45.4
203	704	1300	290	42.1	502	72.8	32.0	34.0	39.8
204	760	1400	331	48.1	556	80.7	15.4	19.9	17.4
205	871	1600	268	38.8	284	41.2	13.0	20.4	26.1

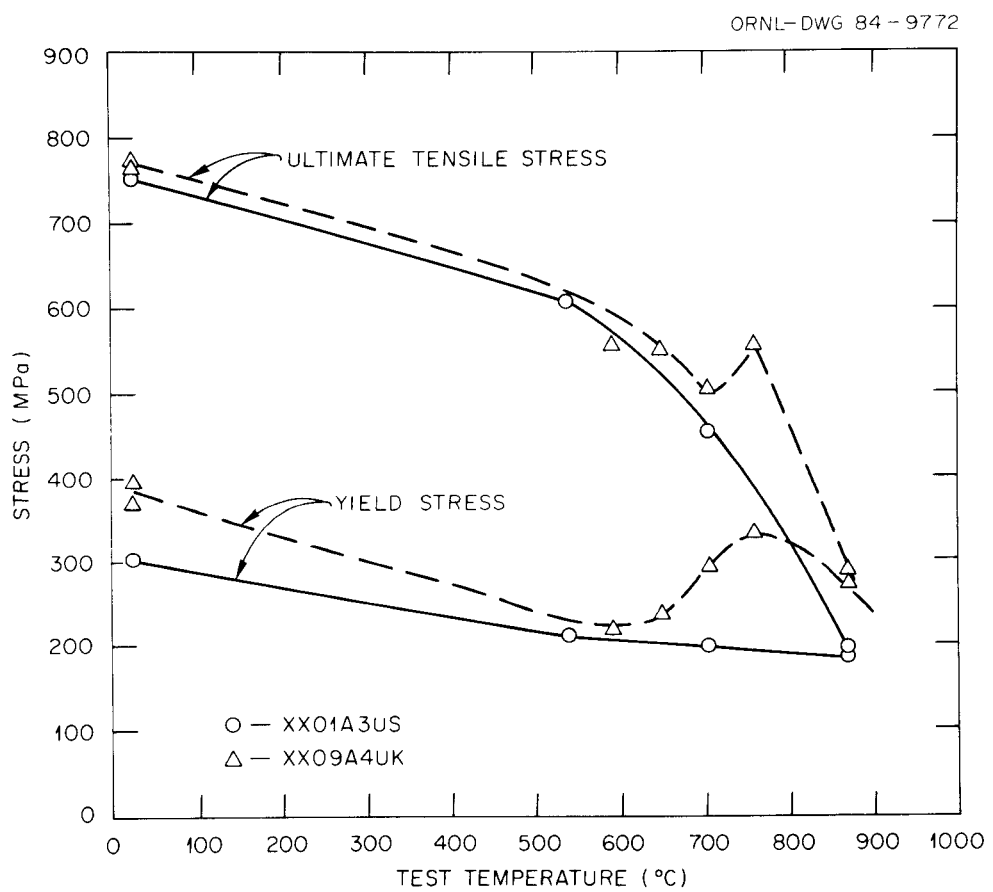


Fig. 8. Tensile properties of Inconel 617 as received.

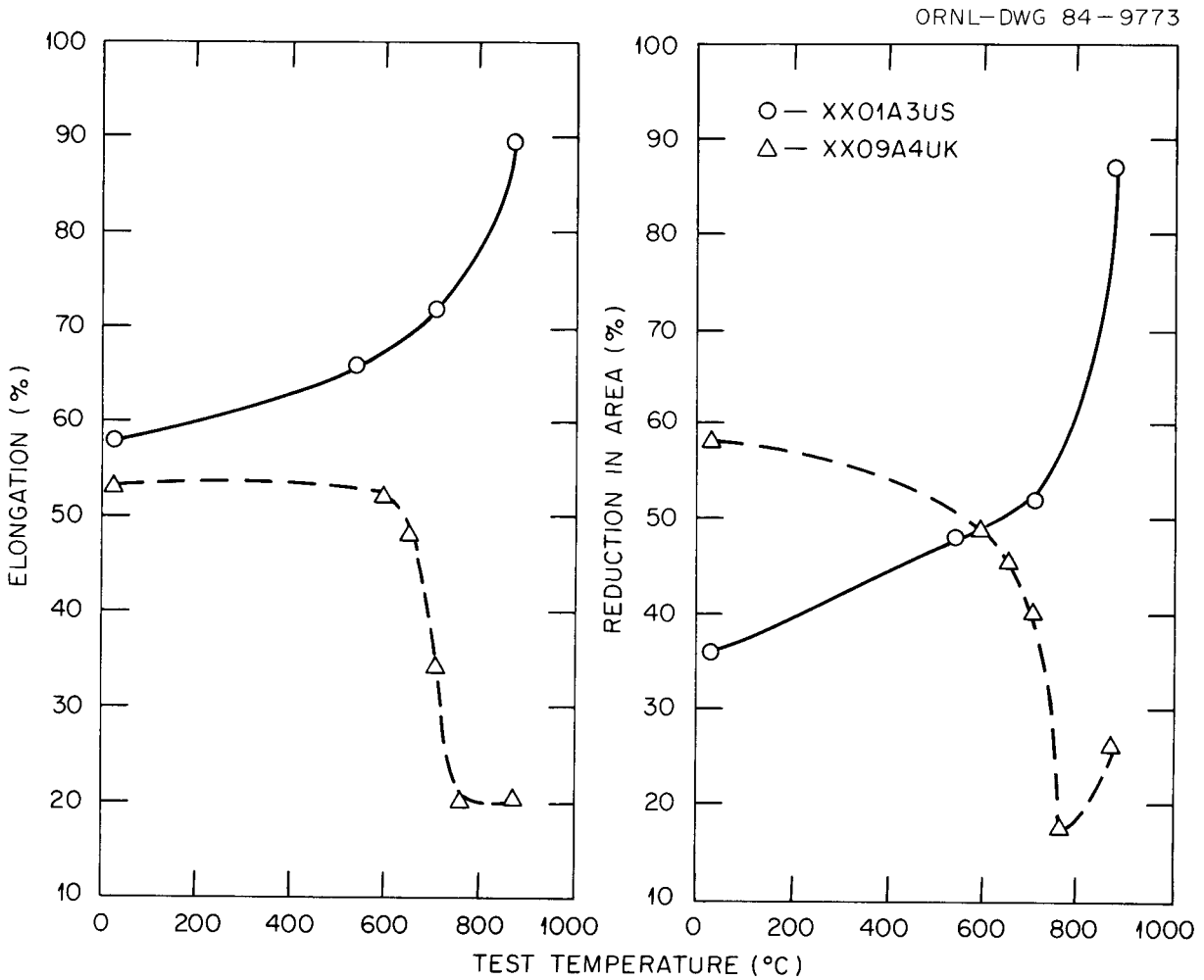


Fig. 9. Ductility of two heats of Inconel 617 tested as received.

Tensile tests were also run on several base metal specimens of heat XX01A3US aged in HTGR-He and in an inert environment (Tables 3 and 4). The specimens were generally tested at 25°C and at the aging temperature. Average carbon concentrations of the samples aged in HTGR-He are given in Table 6. The increases are considerably smaller than those presented later for single creep samples exposed to HTGR-He, so there was some depletion of reactive species from the aging environment. The yield and ultimate tensile strengths at 25°C of the aged samples are shown in Fig. 10. The yield stress is increased about 20% by aging in an inert environment over the range of 538 through 704°C, and the magnitude of strengthening depends only slightly on aging time over the range studied. Aging in HTGR-He at 593 and 704°C increased the yield stress by about 70%.

Table 6. Average carbon concentrations of Inconel 617 (heat XX01A3US) after aging in HTGR-He

Temperature (°C)	Time (h)	Carbon concentration (%)
As received		0.070
538	10,000	0.065
704	10,000	0.075
871	10,000	0.065
583	20,000	0.076
704	20,000	0.076
871	20,000	0.080

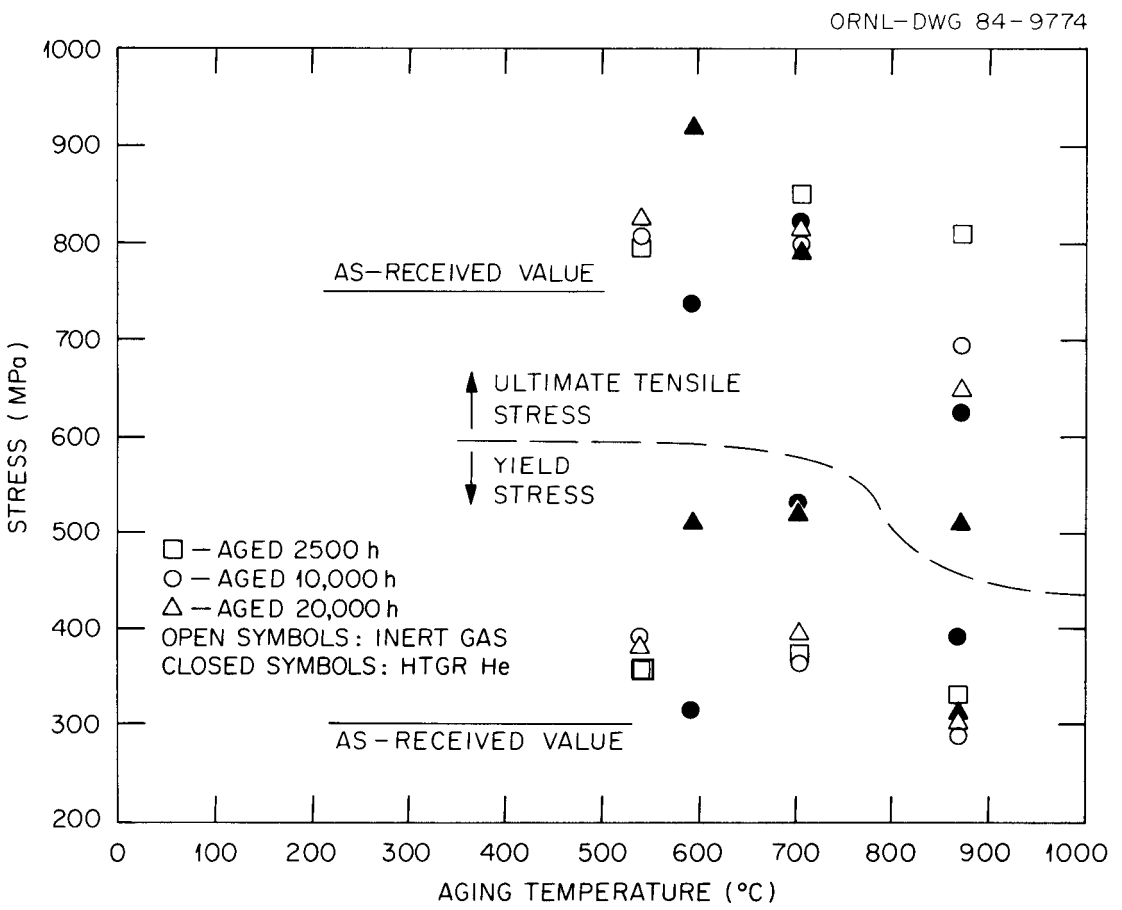


Fig. 10. Influence of aging on the tensile properties of Inconel 617 (heat XX01A3US) measured at 25°C.

At 593°C there seems to be a large effect of aging time, with very little strengthening resulting from 10,000 h aging and a 70% increase from 20,000 h aging. At 704°C there is no detectable effect of aging time. At 871°C aging 10,000 h in HTGR-He resulted in a 30% increase in the yield strength, and all other aging conditions resulted in very small property changes. The ultimate tensile stress is generally modified by aging by about ±10%. In aging at 593°C in HTGR-He, aging 10,000 h had no effect, and aging 20,000 h increased the ultimate tensile strength more than 20%. In aging at 871°C, the ultimate tensile strength decreased for aging times above 2500 h. The strength reductions were greater for samples aged in HTGR-He than for those aged in an inert environment. After 20,000 h aging in HTGR-He, the ultimate tensile strength was reduced more than 30%.

The fracture strains of the aged base metal samples of heat XX01A3US tested at 25°C are shown in Fig. 11. Aging at 538°C had some rather erratic effects, but the elongation values fell in the range of 53 to 69%, all quite high. The general trend observed was that the fracture elongation decreased with increasing aging time and temperature. For an inert environment the lowest fracture strain was 20% for a sample aged 20,000 h at 871°C. Aging in HTGR-He always resulted in a lower fracture strain than the corresponding aging treatment in an inert environment. The lowest value noted following aging in HTGR-He was 6% after 20,000 h at 871°C.

Aged samples were also tested at the aging temperature, and the results have trends similar to those for samples tested at 25°C. The strength values are shown in Fig. 12 and the fracture elongation in Fig. 13. Aging in the range of about 500 to 700°C increased the yield and ultimate tensile strengths, and the increase was greater in the HTGR-He aging environment than in the inert. Aging in either environment at 871°C had very little effect on the yield and ultimate tensile strengths. Aging in inert gas reduced the fracture elongation slightly at 704°C and had no detectable effect at other aging temperatures. Aging in HTGR-He significantly reduced the fracture elongation at all three aging temperatures.

Some material of heat XX01A3US was available from a steam test loop. The material had been exposed for 28,300 h at 482 and 538°C. The exposed form was blocks, and test samples were machined after exposure. Thus, the

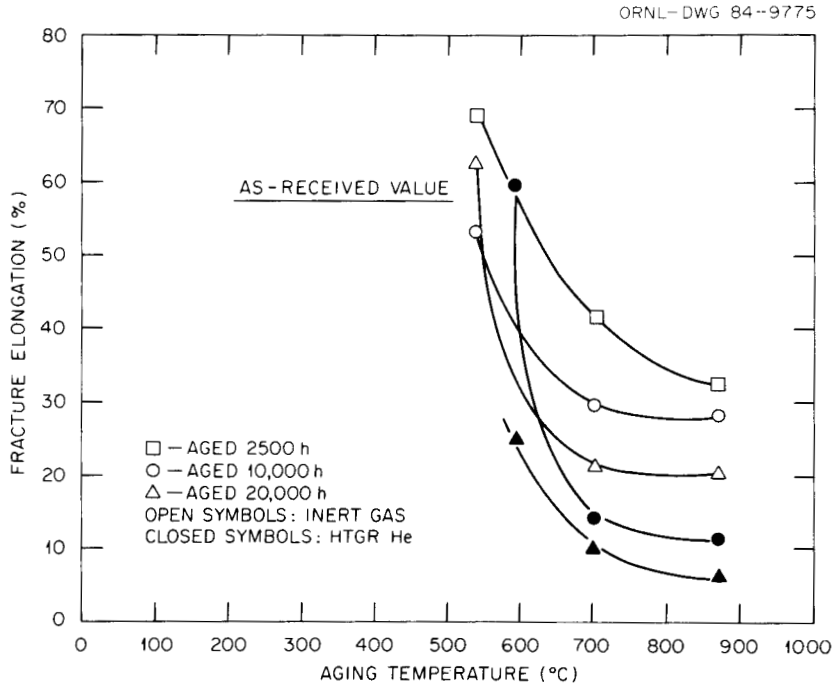


Fig. 11. Influence of aging on the fracture elongation at 25°C of Inconel 617 (heat XX01A3US).

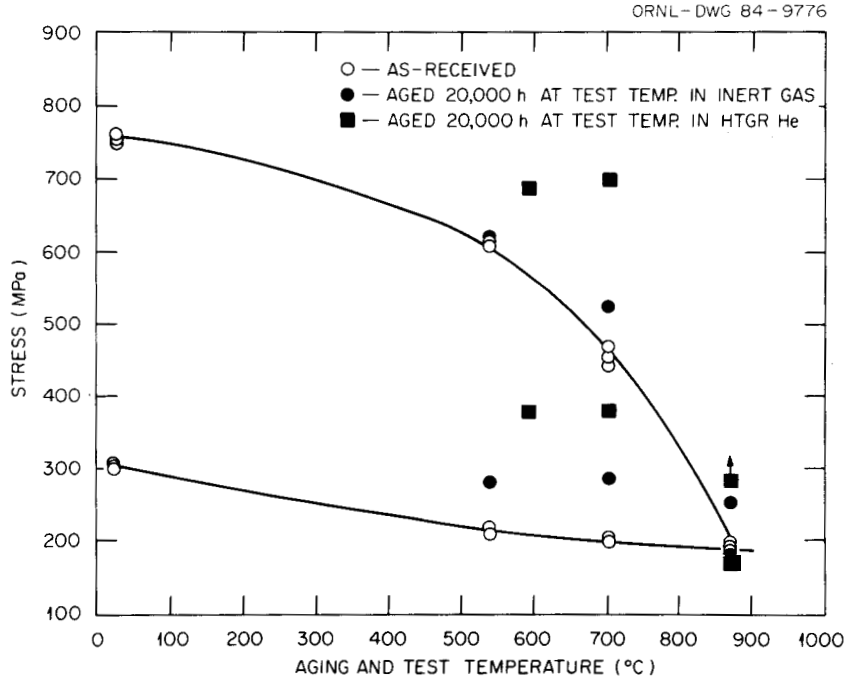


Fig. 12. Influence of aging on the tensile properties of Inconel 617 aged and tested at the same temperature (heat XX01A3US). The lower curve is yield stress, and the upper curve is ultimate tensile stress.

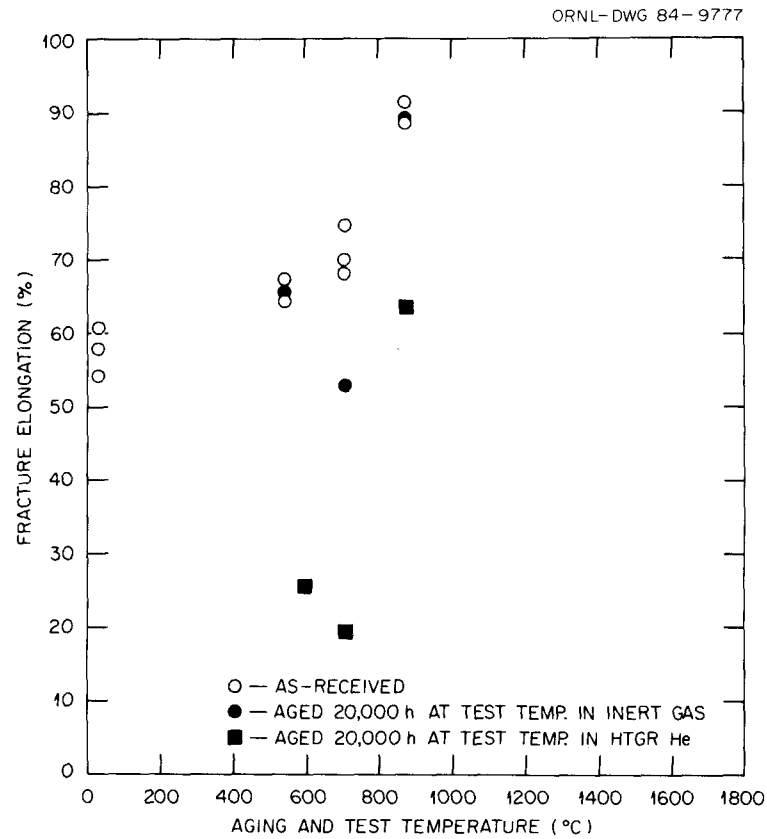


Fig. 13. Influence of aging on the fracture elongation of Inconel 617 (heat XX01A3US) aged and tested at the same temperature.

only important aspect of the exposure was the thermal treatment because any effects of chemical interaction with the test environment were machined away. Test results for some of the samples are given in Table 7. Comparison with the results in Table 3 shows that similar properties resulted from aging over the range of 480 to 600°C in both environments. Aging in this temperature range causes some increase in strength and decrease in ductility.

INCONEL 617 WELDS

Tensile specimens were fabricated so that they consisted entirely of deposited weld metal of heat XX09A9UK. These welds were fabricated by the GTA welding process with heat XX14A6UK base metal. The results of tensile tests on these specimens are given in Table 8. As shown in Fig. 14, the yield strength of the weld metal was almost double that of the base metal

Table 7. Tensile properties of Inconel 617 (heat XX01A3US) base metal at 25°C after aging 28,300 h in steam

(Sample machined from block after exposure, so aging environment was inert)

Aging temperature		0.2% Yield strength		Ultimate tensile strength		Elongation (%)		Reduction of area (%)
(°C)	(°F)	(MPa)	(ksi)	(MPa)	(ksi)	Uniform	Total	
482	900	401	58.2	859	124.6	52.2	52.8	48.9
538	1000	387	56.1	838	121.5	51.8	52.5	46.6

Table 8. Properties of Inconel 617 weld metal as welded

[Base material heat XX14A6UK, filler wire XX09A9UK, specimen gage section 25 mm (1.0 in.) long by 3.2 mm (0.13 in.) in diameter, tested at strain rate of 0.05/min]

Test temperature		0.2% Yield strength		Ultimate tensile strength		Elongation (%)		Reduction of area (%)
(°C)	(°F)	(MPa)	(ksi)	(MPa)	(ksi)	Uniform	Total	
25	77	537	77.9	813	117.9	32.4	35.6	54.6
593	1100	393	57.0	601	87.1	31.3	32.2	45.6
649	1200	367	53.2	567	82.3	35.7	36.8	38.4
760	1400	412	59.8	547	79.4	10.3	25.7	28.8
871	1600	272	39.4	288	41.8	6.0	42.8	71.1

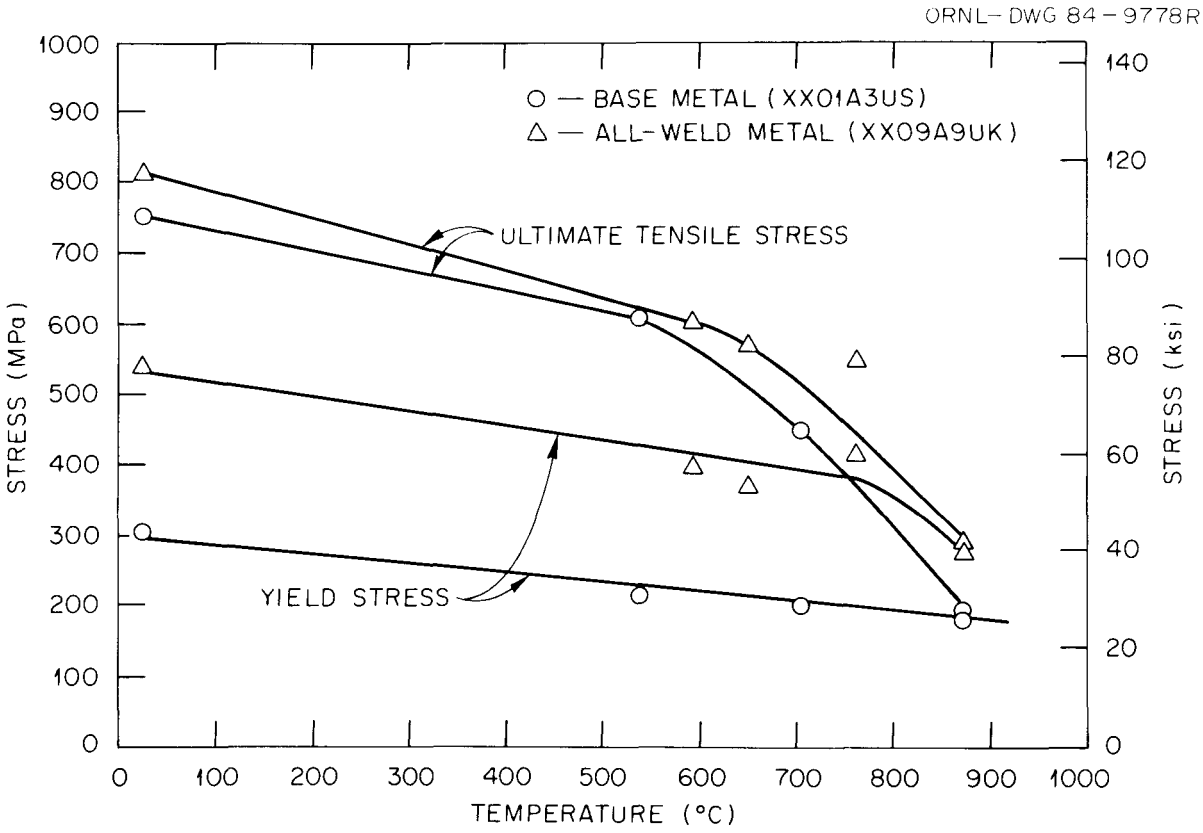


Fig. 14. Tensile properties of Inconel 617 base and weld metal.

from 25 to 871°C. The ultimate tensile strength of the weld metal was only slightly higher than that of the base metal over the same range of test temperature. The elongation and reduction in area are compared for base and weld metal in Fig. 15. The fracture elongation of the weld metal was about half that of the base metal over the entire range of test temperatures. The reduction of area for weld metal was higher at 25°C and lower at elevated temperatures than for base metal. However, the weld metal was very ductile under all test conditions. Note: The base metal properties given here are for unwelded heat XX01A3US, taken from Table 3, and not for the heat XX14A6UK base metal of these welds.

Samples were also made in the transverse orientation so that the specimen gage length crossed the weld metal. These specimens were not tested as welded but were aged 10,000 and 20,000 h in HTGR-He before testing.

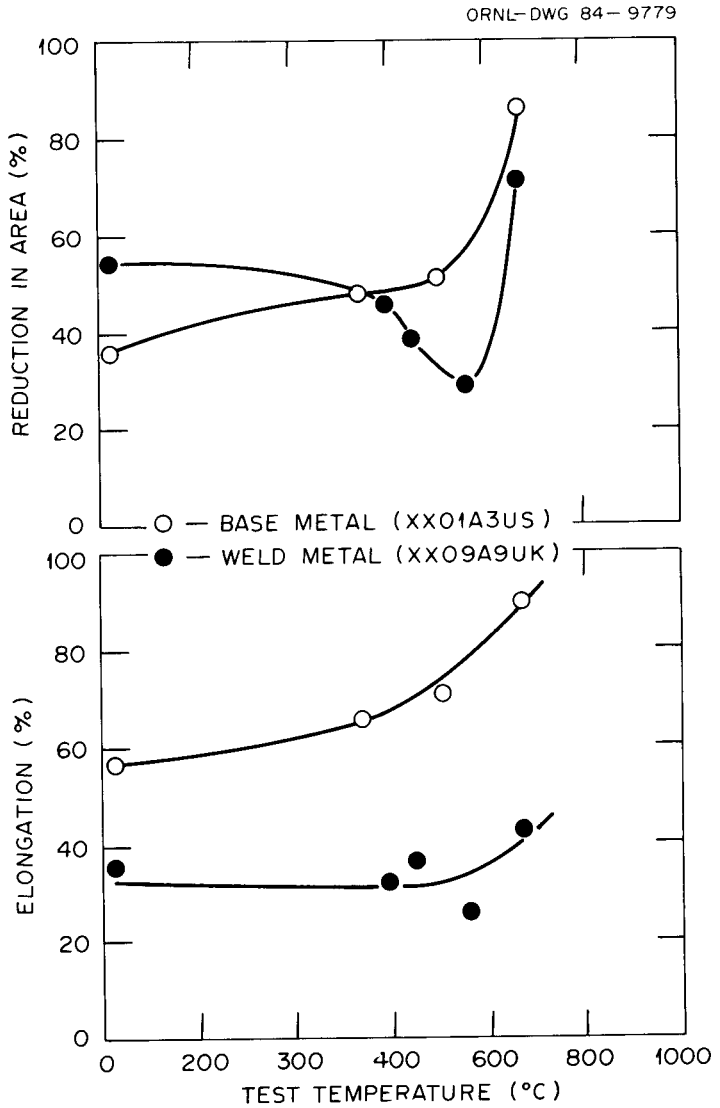


Fig. 15. Tensile ductility of Inconel 617 base and weld metal.

The carbon analyses given in Table 6 are also indicative of the changes in these transverse weld samples. The results of tensile tests on the transverse weld samples are given in Table 9. All the samples tested at 24°C failed in the base metal, as did the samples aged 20,000 h at 593 and 704°C and tested at the same temperatures. The samples aged 10,000 h at 593 and 704°C and tested at those same temperatures and the two samples aged at 871°C failed in the weld.

Table 9. Tensile properties of Inconel 617 transverse weld specimens after aging in HTGR-He

(Welds made by the gas tungsten arc process from heat XX14A6UK base metal and XX09A9UK filler wire)

Specimen	Aging condition			0.2% Yield strength		Ultimate tensile strength		Elongation (%)		Reduction of area (%)	Failure location
	Temperature		Time (h)					Uniform	Total		
	(°C)	(°F)		(MPa)	(ksi)	(MPa)	(ksi)				
Tested at 24°C (75°F)											
I-148	593	1100	10,000	330	47.8	1004	145.6	26.0	26.3	28.3	Base
I-161	593	1100	20,000	676	98.0	1020	148.0	19.5	19.9	10.6	Base
I-152	704	1300	10,000	331	48.0	927	134.4	14.2	14.4	21.0	Base
I-165	704	1300	20,000	626	90.3	867	125.8	5.0	5.2	8.3	Base
I-156	871	1600	10,000	394	57.1	450	65.2	0.7	0.7	3.2	Base
I-169	871	1600	20,000	356	51.6	476	69.0	0.8	0.8	3.3	Base
Tested at 593°C (1100°F)											
I-168	593	1100	10,000	259	37.6	514	74.6	29.0	29.4	33.3	Weld
I-160	593	1100	20,000	623	90.4	800	116.0	15.2	16.2	26.6	Base
Tested at 704°C (1300°F)											
I-153	704	1300	10,000	460	66.7	756	109.7	15.7	19.9	35.0	Weld
I-164	704	1300	20,000	445	64.5	740	107.4	12.2	15.6	8.5	Base
Tested at 871°C (1600°F)											
I-157	871	1600	10,000	223	32.4	270	39.1	6.0	26.8	56.0	Weld
I-168	871	1600	20,000	218	31.6	277	40.2	2.3	23.9	71.8	Weld

The transverse weld samples did change somewhat in strength from aging, but the most important changes involved the fracture elongation. The fracture elongations at 25°C are shown in Fig. 16 for base metal and for transverse weld samples aged in HTGR-He. The fracture elongation of the as-received material was 57%. Aging the base metal at higher temperatures brought about progressive decreases in the fracture elongation. The same trend was noted for transverse weld samples, but the elongations were consistently lower than those for base metal.

For aged transverse specimens tested at the aging temperature, the fracture elongations are shown in Fig. 17. Test results for as-received base metal are shown for comparison. Aging caused large reductions in

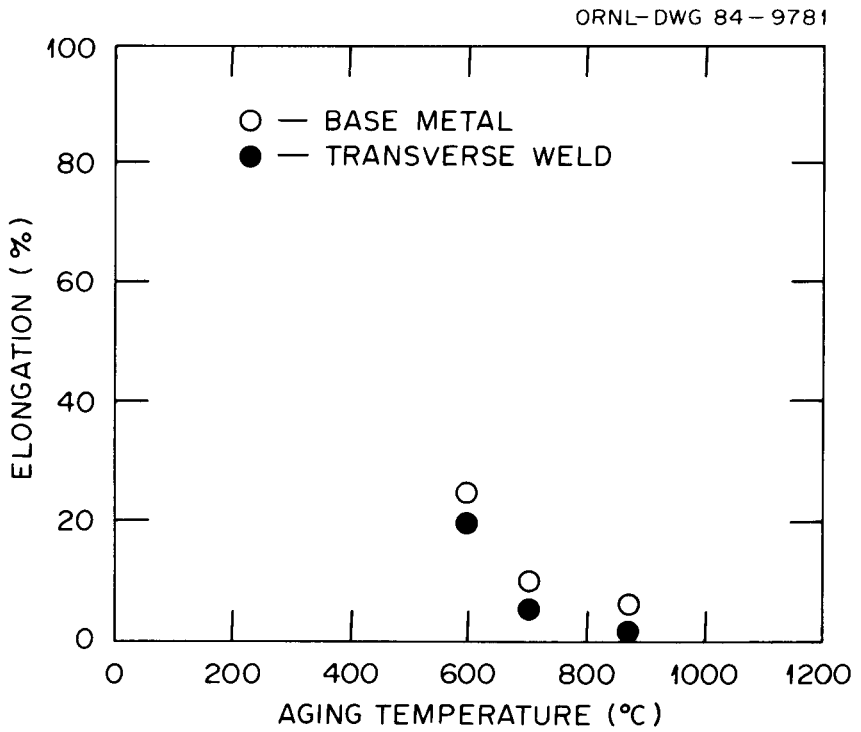


Fig. 16. Elongation at 25°C of Inconel 617 after aging 20,000 h in HTGR-He at the indicated temperature.

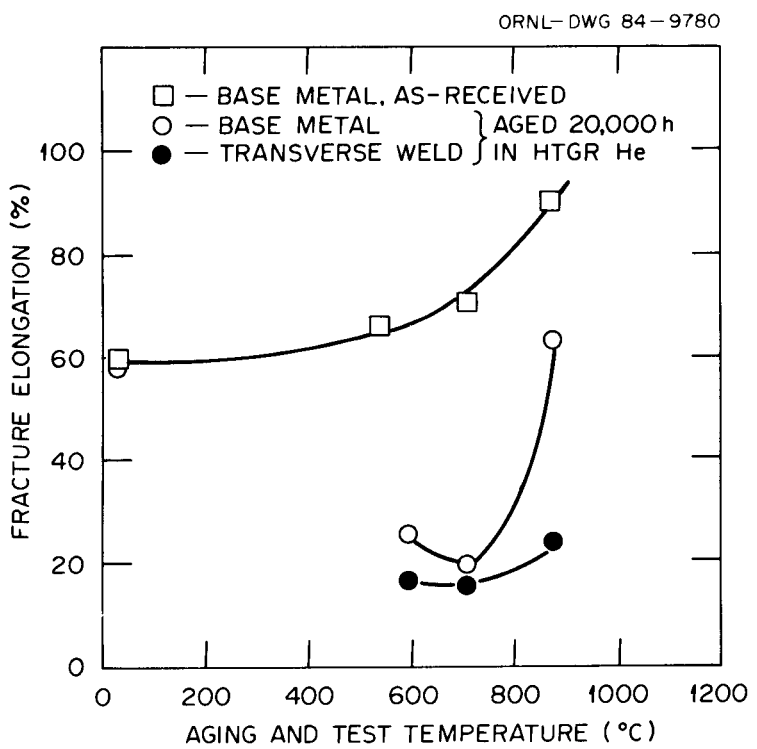


Fig. 17. Fracture elongation of Inconel 617 after several aging treatments.

fracture elongation for the base metal and even lower values for the transverse weld samples. With increasing test temperature, the transverse weld samples failed both in the base metal and in the weld metal. Elongation did not change sharply with change in fracture mode, so the base metal and weld metal likely have similar properties after aging.

INCONEL 618 BASE AND WELD METAL

Welds were made in the Inconel 618 plate with filler metal of the same composition and the GTA welding process. Transverse samples were taken, so the gage length included base and weld metals. Test results for the base and transverse weld samples are given in Table 10. The yield and ultimate tensile stresses of the two types of samples are compared in

Table 10. Tensile properties of as-received Inconel 618 (heat Y09B9)

Specimen		Test temperature		0.2% Yield strength		Ultimate tensile strength		Elongation (%)		Reduction of area (%)	Failure location
Number	Type[a]	(°C)	(°F)	(MPa)	(ksi)	(MPa)	(ksi)	Uniform	Total		
B102	B	25	25	373	54.0	703	102	42.4	45.3	56.4	
W103	TW	25	25	373	54.0	703	102	42.6	47.2	55.7	Weld
B103	B	538	1000	267	38.7	578	83.9	48.9	54.2	55.5	
W104	TW	538	1000	267	38.7	573	83.1	50.1	53.2	53.0	Weld
B104	B	593	1100	220	31.9	550	79.8	49.0	52.1	51.0	
W105	TW	593	1100	203	29.4	549	79.7	45.0	46.8	53.7	Weld
B105	B	649	1200	196	28.4	541	78.4	47.0	48.2	46.0	
W106	TW	649	1200	189	27.5	526	76.2	39.5	40.6	38.3	Weld
B106	B	704	1300	185	26.7	454	65.9	30.5	33.5	31.3	
W107	TW	704	1300	189	27.4	463	67.1	39.5	44.4	34.1	Weld
B107	B	760	1400	184	26.6	363	52.6	13.0	38.5	43.0	
W108	TW	760	1400	171	24.8	361	52.3	17.0	40.6	39.7	Base
B108	B	871	1600	175	25.4	197	28.6	8.0	59.8	69.3	
W109	W	871	1600	172	25.0	205	29.8	55.0	62.1	70.0	Weld

[a]B = base metal; TW = transverse weld sample including base metal, heat-affected zone, and weld metal.

Fig. 18. The strengths of both sample types are about equivalent, but the transverse weld samples tend to fail at a slightly lower stress. This tendency is supported by the test results in Table 10, where only one sample failed in the base metal. The two ductility measures — elongation and reduction of area — are shown in Fig. 19, which also shows that the properties of base metal and transverse weld samples are similar.

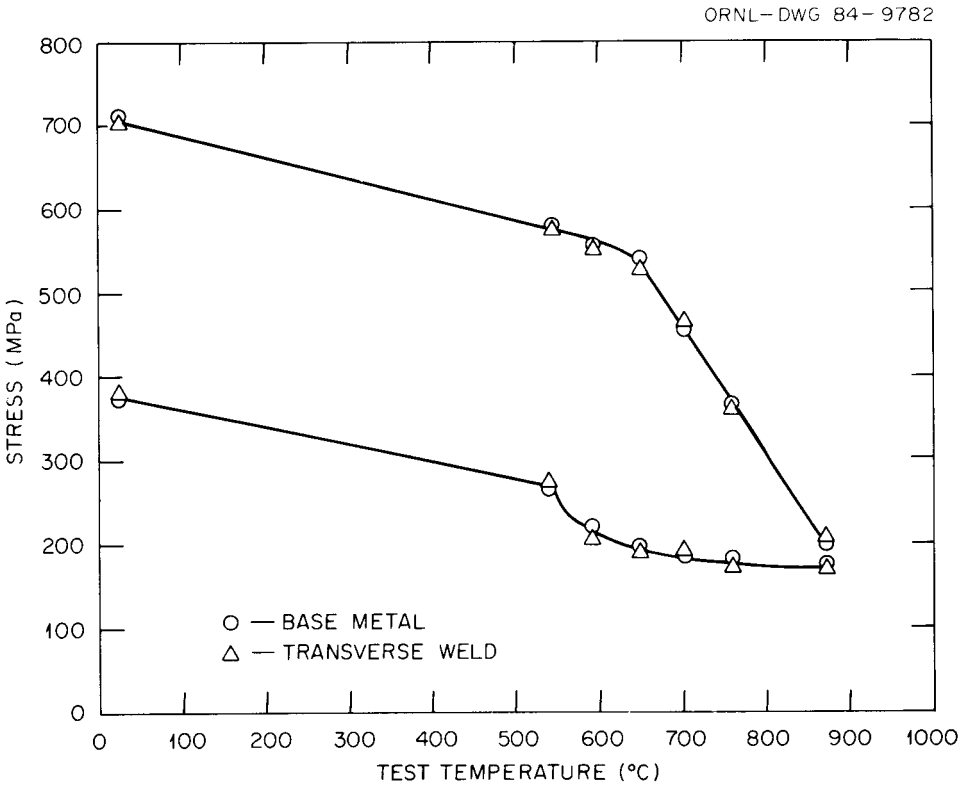


Fig. 18. Tensile properties of Inconel 618. The upper curve is the ultimate tensile stress, and the lower curve is the yield stress.

A few samples of Inconel 618 base metal and transverse welds were aged 12,500 h in HTGR-He. The carbon analyses of these samples are given in Table 11. Some of the aged samples were tested at 25°C, and the results are given in Table 12. Note that all the samples failed in the weld metal. The yield stress shows some sizable variations, but the results are not sufficient to verify the trends. The yield and ultimate stresses are generally not more than 10% less than the strength of the

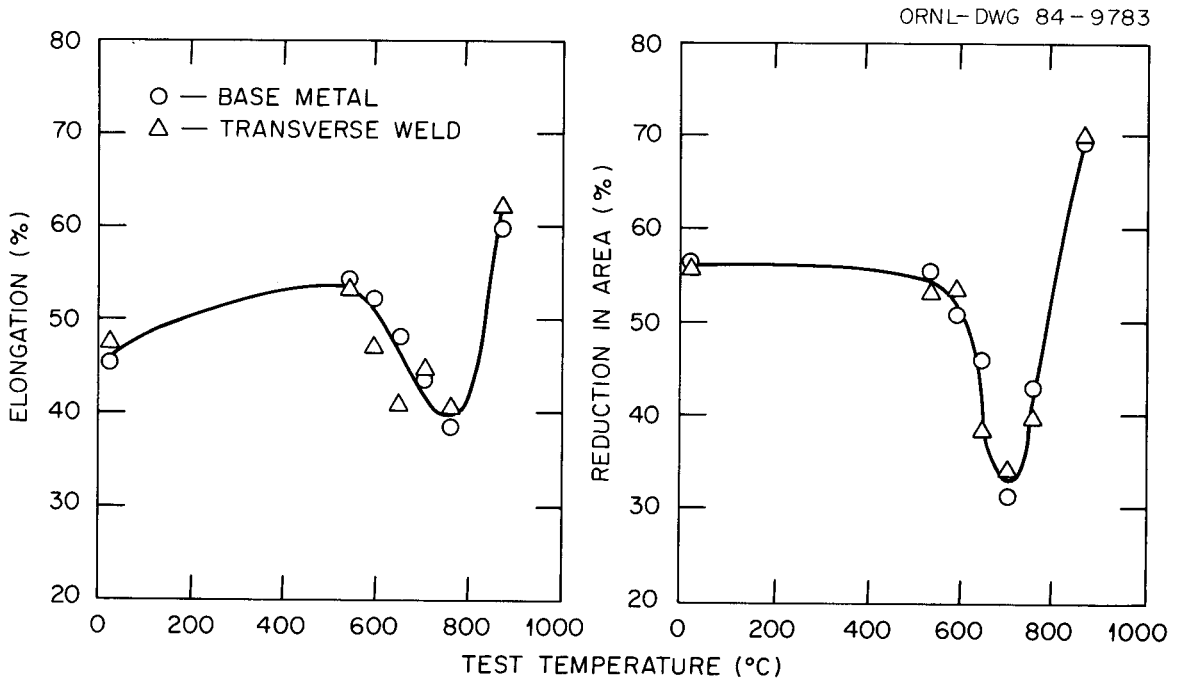


Fig. 19. Tensile ductility of Inconel 618.

Table 11. Average carbon concentration in Inconel alloy 618 after aging 12,500 h in HTGR-He at indicated temperature

Temperature (°C)	Carbon content (%)
As received	0.047
593	0.061
704	0.063
871	0.102

Table 12. Tensile properties of Inconel 618 (heat Y09B9) at 24°C (75°F) after aging for 12,500 h in HTGR-He

Specimen		Aging temperature		0.2% Yield strength		Ultimate tensile strength		Elongation (%)		Reduction of area (%)
Number	Type[a]	(°C)	(°F)	(MPa)	(ksi)	(MPa)	(ksi)	Uniform	Total	
52	B	593	1100	303	43.9	716	103.9	35.4	39.6	54.0
54	B	704	1300	298	43.2	717	104.0	36.2	39.7	51.2
56	B	871	1600	492	71.4	689	100.0	41.1	41.4	36.3
1	TW	593	1100	516	74.8	749	108.7	17.4	18.7	37.6
2	TW	704	1300	503	73.0	740	107.3	39.5	40.5	40.5
4	TW	871	1600	323	46.8	632	91.6	22.7	23.2	14.4

[a]B = base metal; TW = transverse weld sample including base metal, heat-affected zone, and weld metal. All transverse weld specimens failed in the weld metal.

as-received material, but there are some large isolated property changes, particularly at 871°C. Fracture elongations are shown in Fig. 20 for aged base and transverse weld samples tested at 24°C. The base metal shows a small decrease in fracture elongation. The transverse weld samples show very little decrease due to aging at 704°C, but they show decreases of about 50% when aged at 593 and 871°C.

NOTCHED TENSILE AND IMPACT PROPERTIES AT AMBIENT TEMPERATURE

Notched tensile specimens of Inconel 617 base metal were prepared. The specimens were machined after aging for 10,000 h in air, so the aging environment was inert for all practical purposes. The gage section of the specimens was 31.8 mm (1.25 in.) long and 6.4 mm (0.25 in.) in diameter, notched to a diameter of 3.2 mm (0.13 in.) with an included angle of 60° and a root radius of 0.13 mm (0.005 in.). The specimens were tested at 25°C. Ratios of notched to unnotched values of ultimate tensile strengths for the aging temperatures of 538, 704, and 871°C were, respectively, 1.15, 1.03, and 0.94. Thus, the material was notch strengthened by aging at

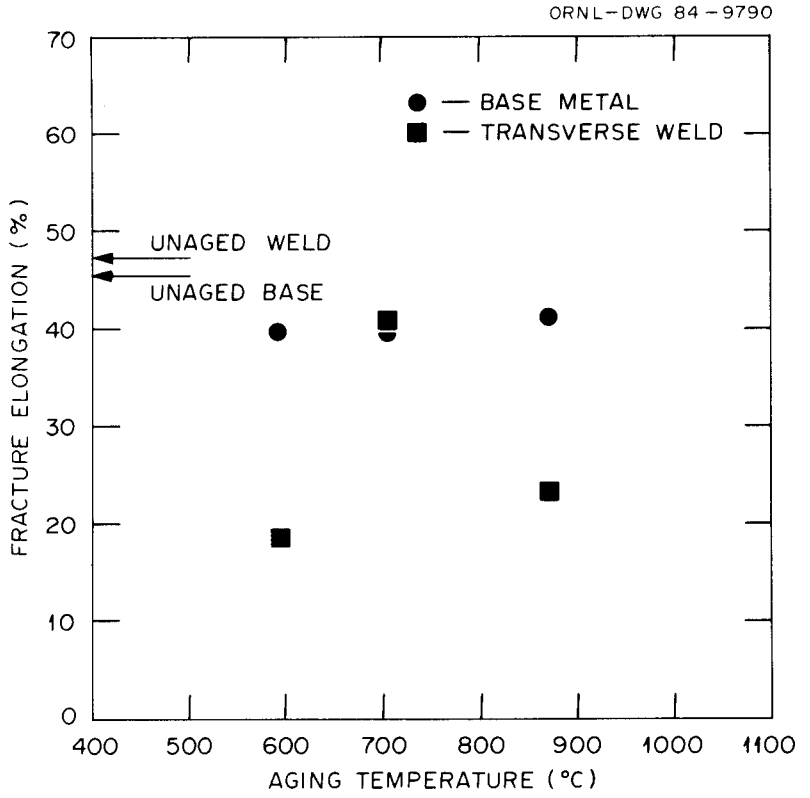


Fig. 20. Fracture elongation at 24°C of Inconel 618 after aging 12,500 h in HTGR-He.

538°C, weakened by aging at 871°C, and strengthened very slightly by aging at 704°C. These tests were carried out at a very slow strain rate compared with the strain rate in an impact test.

Impact tests were run on both heats of Inconel 617 base metal and Inconel 618 base and weld metal following aging at a variety of conditions, and the test results are given in Table 13. The impact test results for samples aged at 871°C are shown in Fig. 21. The shapes of the curves between 0 and 100 h are conjecture because tests were not run in that time interval. Impact energies of both heats of Inconel 617 base metal fell to low values rather quickly. The two heats have different properties after short aging times, but low values of 25 J or less will likely be obtained from both heats following 10,000 h aging at 871°C. Inconel 618 base metal initially has an impact energy less than that of

Table 13. Charpy-V impact properties at 25°C for Inconel 617 and 618 after various treatments

Heat	Aging temperature (°C)	Impact energy [J (ft-lb)] after each aging time						
		0	100 h	1000 h	10,000 h	11,786 h	20,000 h	28,300 h
				Inconel 617				
XX01A3US	492	175 (129)						190 (140)[a]
XX01A3US	538	175 (129)			81 (60)		103 (76)	138 (102)[a]
XX01A3US	704	175 (129)			31 (23)		20 (15)	
XX01A3US	871	175 (129)	91 (67)[b]					
			98 (72)	53 (39)	26 (19)		20 (15)	
XX09A4UK	871	298 (220)	61 (45)[b]	26 (19)				
				Inconel 618				
Y09B9								
Base metal	871	94 (69)	88 (65)	94 (69)		100 (74)	104 (77)	
Weld metal	871	151 (111)	134 (99)	115 (85)		109 (80)	91 (67)	

[a]Exposed to steam environment before specimens were machined.

[b]Aged in HTGR-He; all other samples machined after aging in inert gas or air.

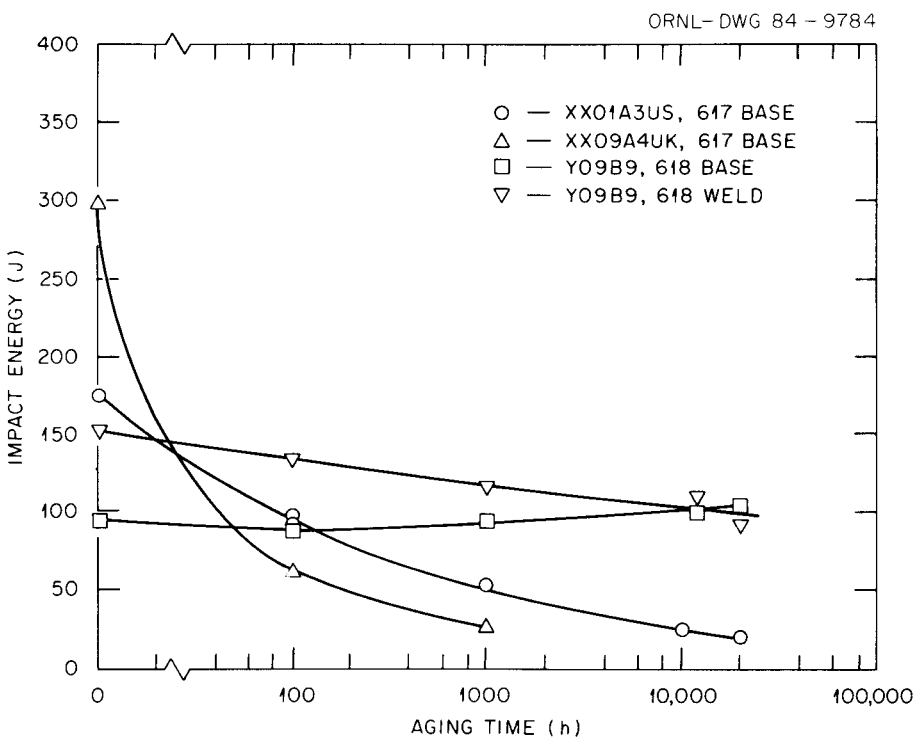


Fig. 21. Impact energy at 25°C of Inconel 617 and 618 after aging at 871°C.

Inconel 618 weld metal, but with aging the value for the base metal remained constant within 10% and the value for the weld metal decreased slightly to reach a common value of about 100 J after 10,000 h. Thus, Inconel 618 has very good impact properties after aging while Inconel 617 has low or poor impact properties after aging a short time at 871°C.

Impact results for Inconel 617 after aging at several temperatures are shown in Fig. 22. The rate of embrittlement is very slow at 492°C and very fast at 871 and 704°C. The rate is slow enough at 538°C that the impact energy would likely not drop below 50 J in a 30-year (263,000-h) plant life. Thus, it is quite likely that the service temperature of this material must not exceed about 600°C to keep the impact energy above 50 J.

Impact samples of heat XX01A3US that had been aged 20,000 h at various temperatures before testing were examined in the electron microscope.

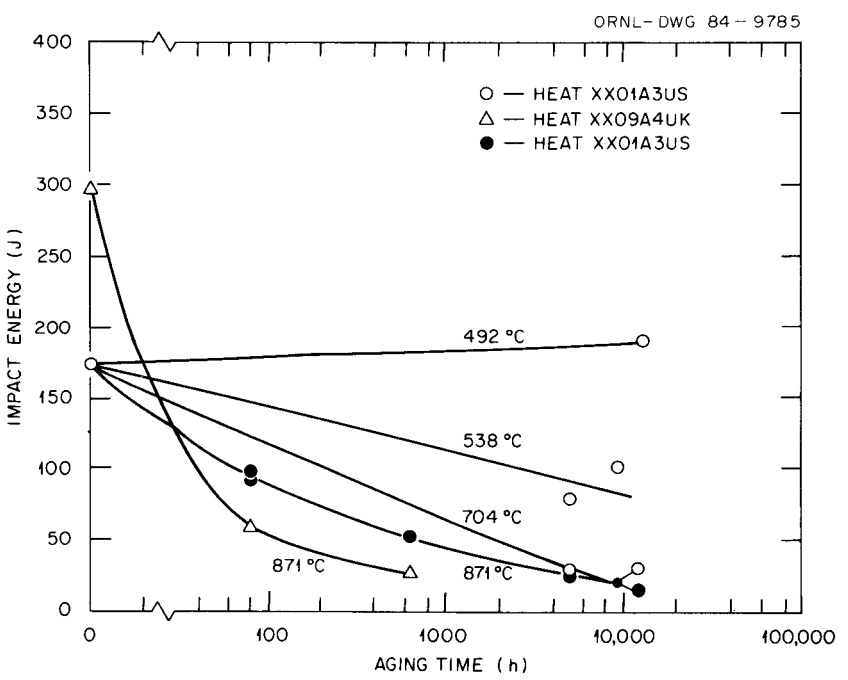


Fig. 22. Impact energy of Inconel 617 at 25°C after aging at various temperatures.

Typical scanning electron micrographs of the fracture surface of a sample tested in the as-received condition are shown in Fig. 23. The fracture is mostly transgranular ductile shear with some intergranular fracture segments. Similar micrographs for a sample aged 20,000 h at 538°C are shown in Fig. 24. The fracture mode is primarily transgranular shear, and the impact energy is still quite high. The fracture surfaces of samples aged at 704 and 871°C are shown in Figs. 25 and 26, respectively. The fracture path in both of these samples is intergranular, but there is considerable evidence of localized shear on the fracture surface. The impact energies of the samples aged at 704 and 871°C was quite low (20 J).

RESULTS OF CREEP TESTS

INCONEL 617

Creep tests were run on two heats of Inconel 617 base metal, and the results are summarized in Table 14. Selected properties are shown graphically: the time to 1% strain in Fig. 27 and the time to rupture in Fig. 28 as functions of stress. Many variables and several test temperatures are involved, so it is very difficult to draw straight lines through the data in Figs. 27 and 28. The Larson-Miller parameter was used as a means of consolidating the data obtained at several temperatures into a single line. The time to 1% strain is shown in Fig. 29 and the time to rupture in Fig. 30 as functions of stress. These data do not show any consistent effect of environment on the creep strength.

The minimum creep rates for the two base metal heats are shown in Fig. 31. The line at 704°C was constructed mostly on the basis of parallelism to lines at 649 and 760°C because the extremes of the line at 704°C are defined by two different heats. Environment has no apparent effect on minimum creep rate.

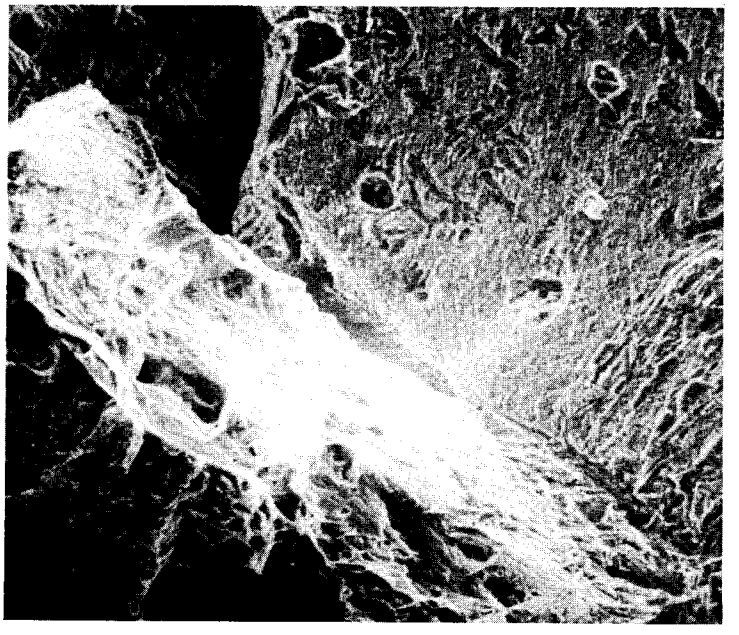
Fracture strains of the various base metal tests are shown in Fig. 32. The lines are meant only to illustrate trends, and there are many deviations of data above and below them. Also, the lines were drawn

M-10827

(*a*)

250 μm

M-10826

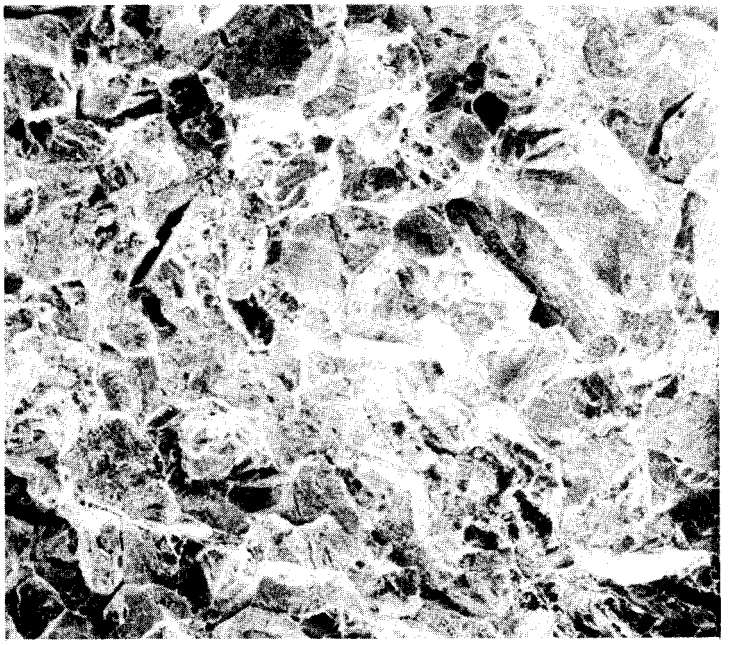


(*b*)

50 μm

Fig. 23. Scanning electron micrographs of the fracture surface of a sample of Inconel 617 (heat XX01A3US) impact tested as received. Impact energy was 175 J.

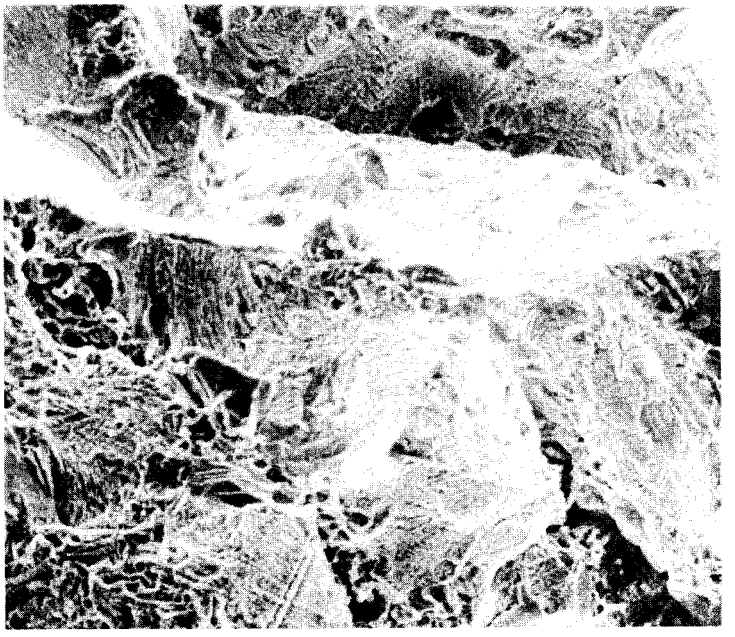
M-10830



(*a*)

250 μm

M-10829

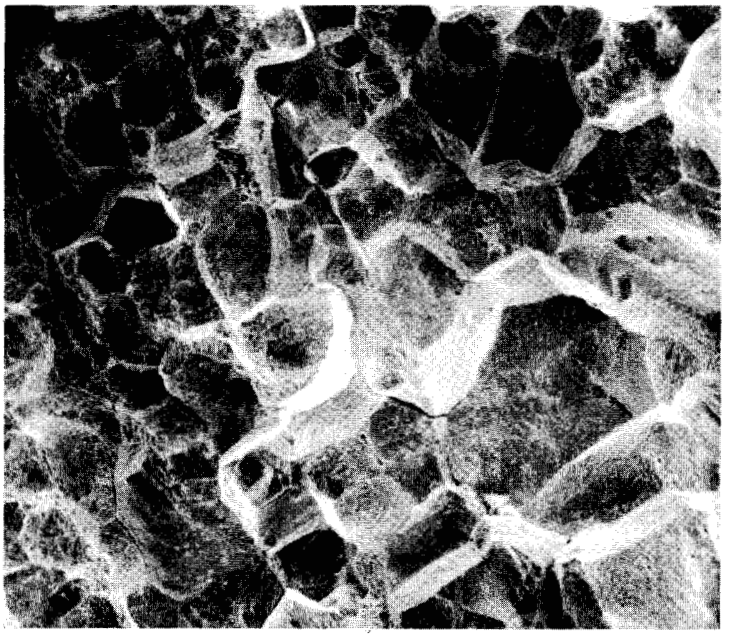


(*b*)

50 μm

Fig. 24. Scanning electron micrographs of the fracture surface of Inconel 617 (heat XX01A3US) aged 20,000 h at 538°C and impact tested at 25°C. Impact energy was 103 J.

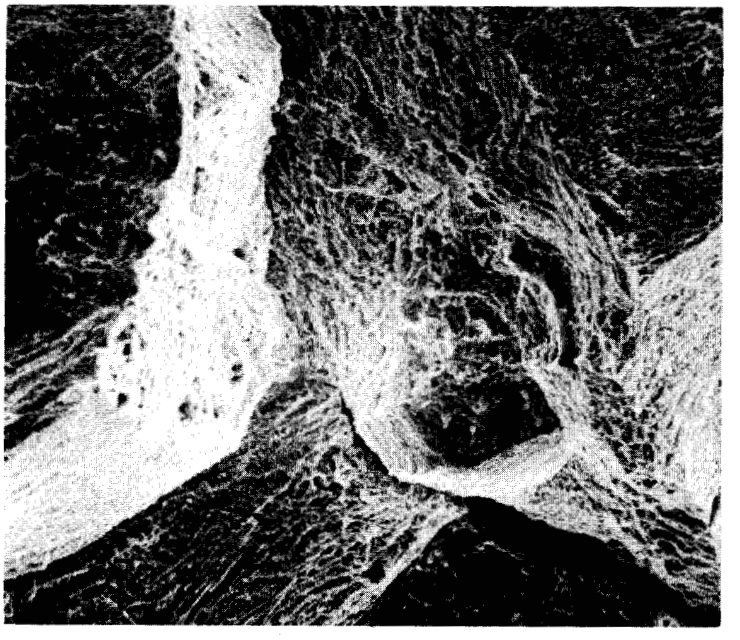
M-10843



(*a*)

250 μm

M-10842

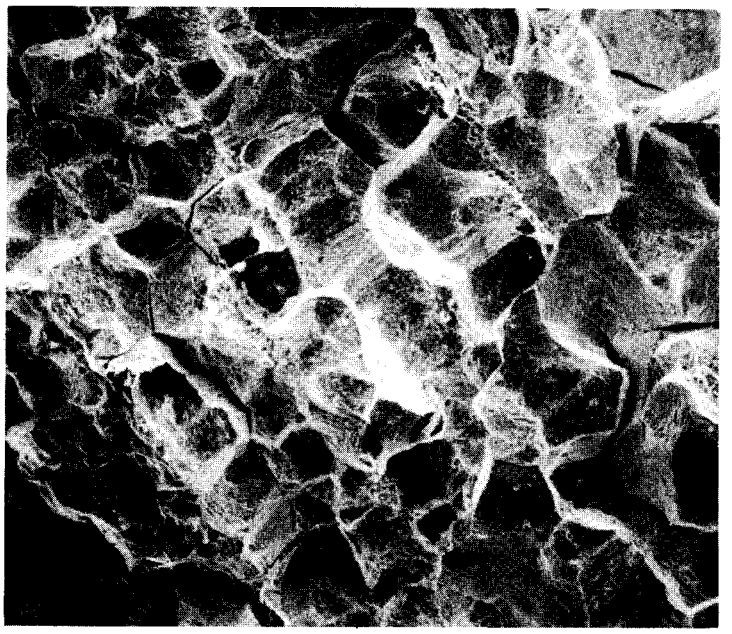


(*b*)

50 μm

Fig. 25. Scanning electron micrographs of the fracture surface of Inconel 617 (heat XX01A3US) aged 20,000 h at 704°C and impact tested at 25°C. Impact energy was 20 J.

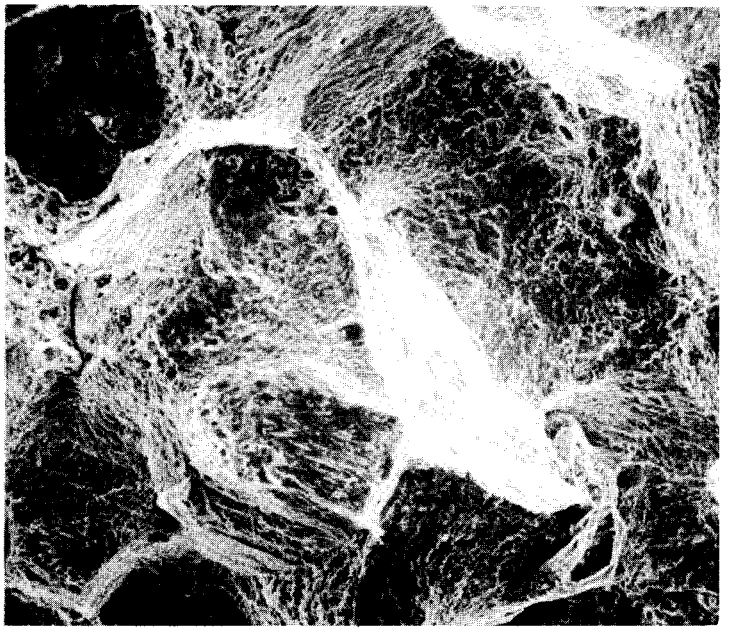
M-10840



(*a*)

250 μm

M-10839



(*b*)

50 μm

Fig. 26. Scanning electron micrographs of the fracture surface of Inconel 617 (heat XX01A3US) aged 20,000 h at 871°C and impact tested at 25°C. Impact energy was 20 J.

Table 14. Results of creep tests on Inconel 617 base metal

Test	Heat[a]	Stress (MPa)	Stress (ksi)	Environment	Condition[b]	Time to indicated creep strain (h) 1%	2%	5%	Time to tertiary creep (h)	Rupture or test (h)	Steady-state creep rate (h^{-1})	Elongation (%) Loading	Elongation (%) Creep	Reduction of area (%)
					Tested at 593°C (1100°F)									
22188	A	414	60	Air	As received	20		3,350	3,445	3,570	4.8 E-6	16.0	6.9	18.2
20531	A	414	60	He	As received	380	1,700		4,534	4,534	4.9 E-6	18.0	3.7	21.3
20521	A	414	60	He	593/10,000/He	500				548	1.0 E-5		2.2	22.0
22645	A	414	60	He	593/20,000/He	2,000			2,600	2,676	2.4 E-6	1.3	3.9	12.0
22200	A	345	50	He	593/10,000/He					5,629	5.5 E-7	4.4	5.5	14.4
23088	B	345	50	Air	As received					2,107	1.4 E-6	9.3	3.4	15.2
					Tested at 649°C (1200°F)									
19106	A	414	60	He	As received	0.5	5	145	168	186	1.6 E-4	15.3	7.5	16.1
21589	A	345	50	Air	482/28,300/Steam	280	1,100		1,700	1,813	1.3 E-5	8.4	5.6	13.5
21610	A	345	50	Air	538/28,300/Steam	189	837			1,063	2.4 E-5	6.8	6.5	13.5
19182	A	276	40	He	As received	5,000	8,600	15,100	8,000	16,718	2.0 E-6	0.03	9.3	14.9
18483	A	276	40	He	As received	7,600				15,850	1.1 E-6	2.5	6.1	8.3
18440	A	276	40	Air	As received	8,850	16,480	25,510	16,000	25,566	9.3 E-7	1.5	6.4	15.7
19182	A	207	30	He	As received					26,515[c]	9.9 E-8	0.3	0.3	
23085	B	276	40	Air	As received					1,121	1.9 E-6	1.5	1.4	4.5
23339	B	276	40	Air	As received					667	3.1 E-6	4.6	1.9	10.0
22751	B	276	40	He	As received					1,208	3.0 E-6	2.3	0.8	10.6
					Tested at 704°C (1300°F)									
22742	A	276	40	He	As received	7	13	47	212	329	5.1 E-4	3.5	33.0	45.2
21579	A	276	40	He	704/10,000/He					584			9.1	17.8
20526	A	207	30	He	704/10,000/He	3,450	5,050		3,800	5,424	2.0 E-6		3.7	2.4
22180	A	207	30	He	704/20,000/Inert	925	2,285	4,529	2,500	5,585	7.1 E-7	0.2	10.9	14.3
23342	B	207	30	Air	As received					2,621	1.2 E-6	0.4	1.2	1.6
22753	B	207	30	He	As received	6,200			6,700	7,034	1.2 E-6	0.4	2.5	3.6
22760	B	172	25	He	As received	9,600	16,400			18,526[d]	6.5 E-6	0.5	2.9	
					Tested at 760°C (1400°F)									
17563	A	172	25	He	As received	9	20	57	105	280	8.9 E-4	0.2	53.6	74.1
23091	A	138	20	Air	As received	10	28	123	780	1,305	2.2 E-4		48.5	56.9
21578	A	138	20	He	538/28,300/Steam	209	440	870	550	1,762	3.9 E-5	0.06	24.5	46.2
20032	A	103	15	Air	As received	125	750	3,500	17,000	20,702	1.0 E-5		28.0	31.9
22754	B	172	25	He	As received	513	740	935	517	948	1.3 E-3		5.4	9.4
22755	B	138	20	He	As received					4,585		0.3	9.4	21.8
					Tested at 871°C (1600°F)									
21611	A	69	10	Air	As received	5	20			576			45.1	44.7
20492	A	48	7	Air	As received	240	500	1,200	2,400	4,800	4.1 E-5	0.09	26.2	28.3
20545	A	48	7	He	As received	2,500	3,500		3,000	3,562	1.1 E-4		11.8	18.4
20544	A	48	7	He	871/20,000/Inert	40	92	345	2,200	2,282	5.7 E-5		21.3	28.7
22185	A	48	7	He	871/20,000/He	40	100	320	1,880	1,933	5.7 E-5		18.8	24.0
19393	A	35	5	He	As received	3,200	5,848	13,875	26,800	29,544	3.7 E-6	0.002	16.5	22.3
19761	A	35	5	Air	As received	3,100	4,780	8,670	23,200	34,231	7.9 E-6	0.07	30.3	23.3
23827	B	48	7	He	As received	2,900	3,700	5,670	2,300	7,543	1.6 E-4	0.02	11.4	11.6
21603	B	21	3	Air	As received					26,117[e]	1.9 E-7	0.01	1.3	
21602	B	21	3	He	As received					25,699[e]	2.6 E-7		0.27	

[a]A = heat XX01A3US; B = heat XX09A4UK.

[b]Aging temperature (°C)/aging time (h)/aging environment.

[c]Test reloaded at a higher stress.

[d]Test in progress.

[e]Discontinued before failure.

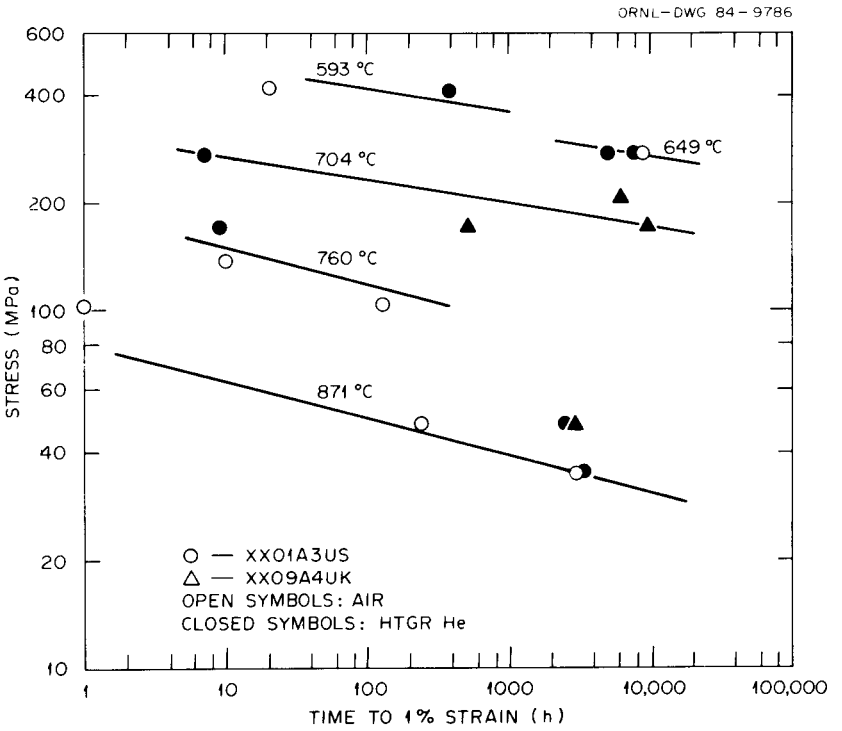


Fig. 27. Creep strength of Inconel 617.

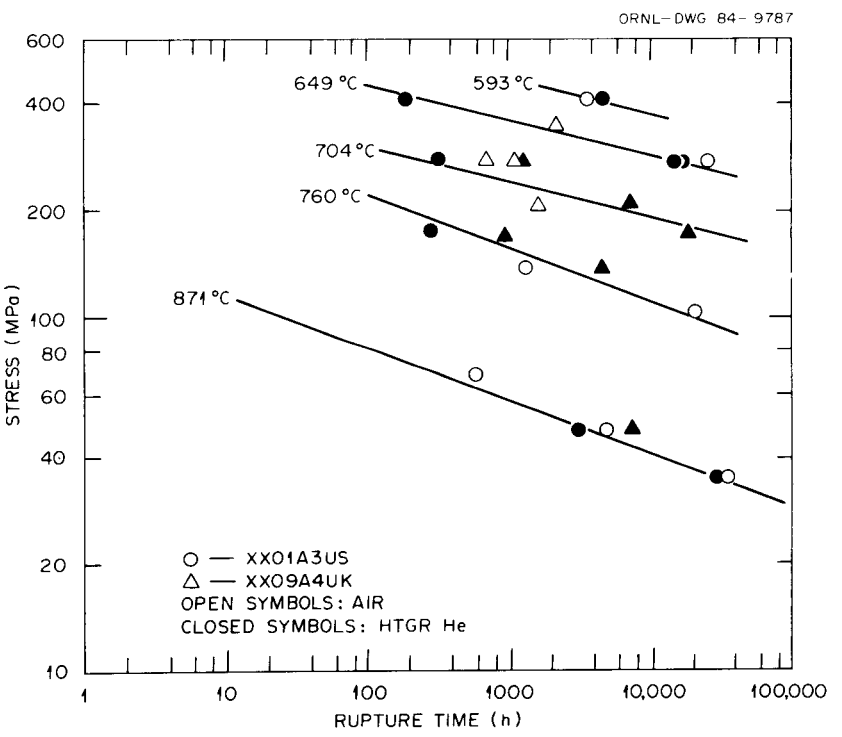


Fig. 28. Stress-rupture properties of Inconel 617.

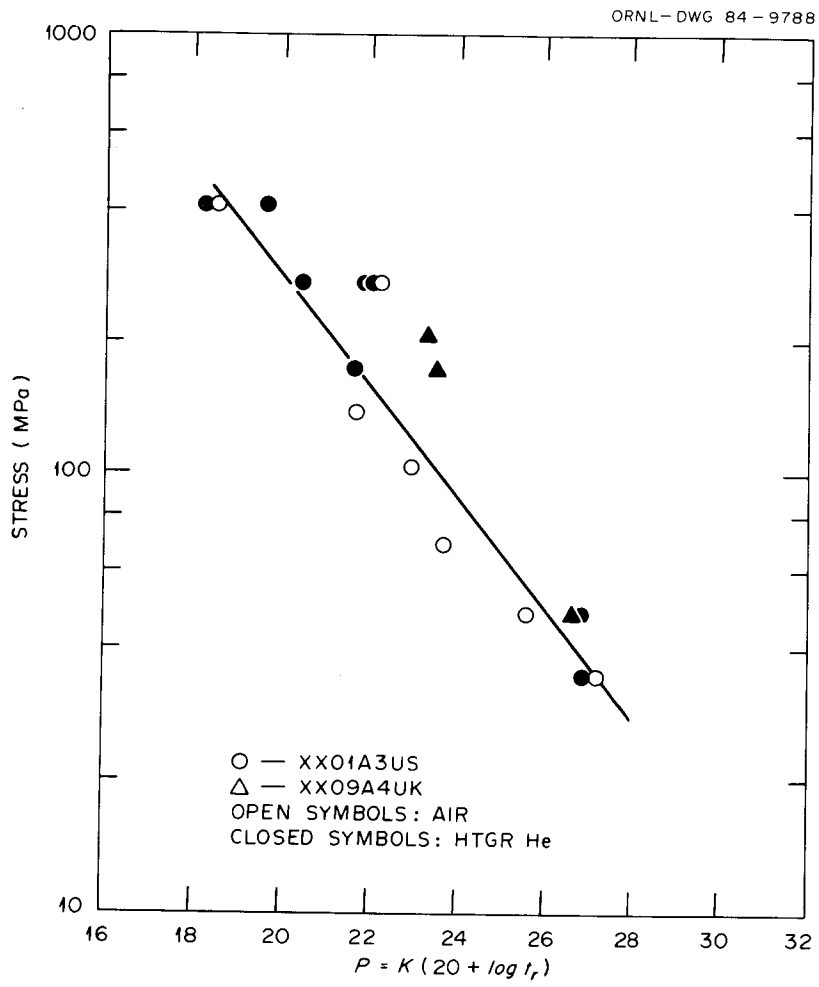


Fig. 29. Larson-Miller plot for 1% strain in Inconel 617.

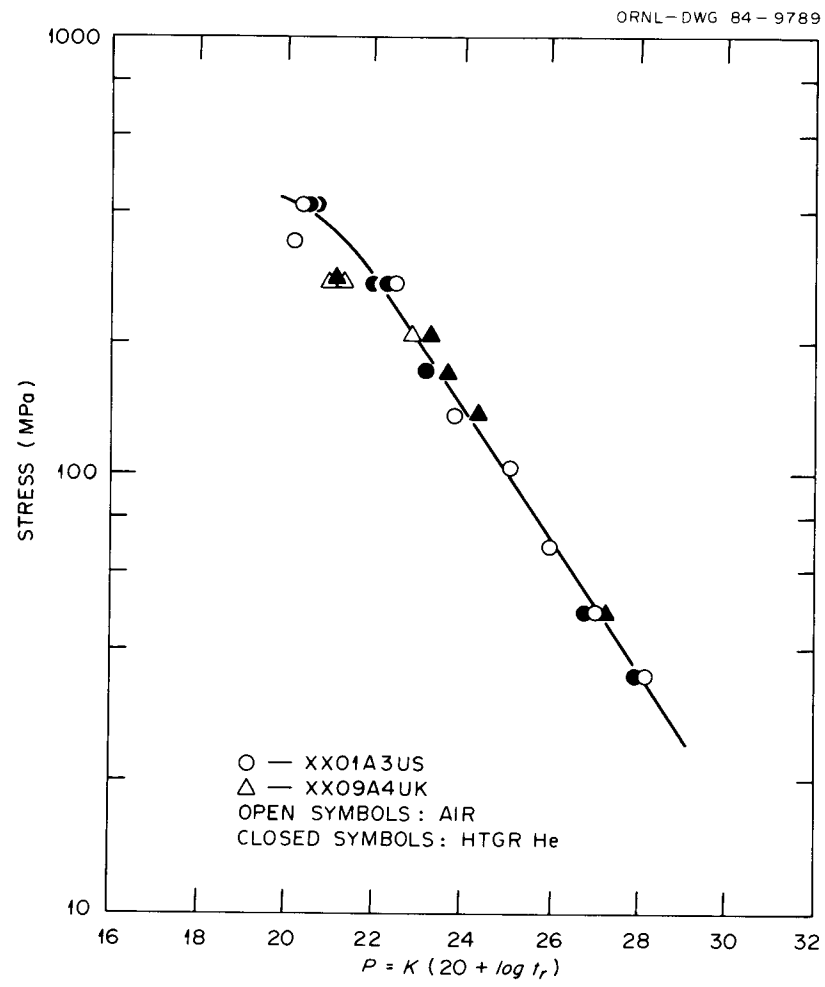


Fig. 30. Larson-Miller plot for rupture in Inconel 617.

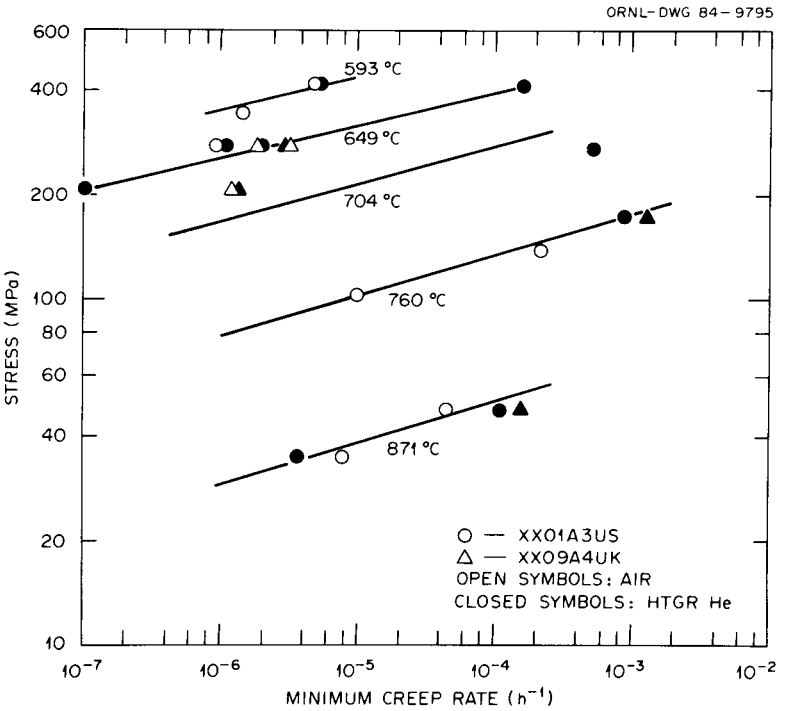


Fig. 31. Minimum creep rate of Inconel 617.

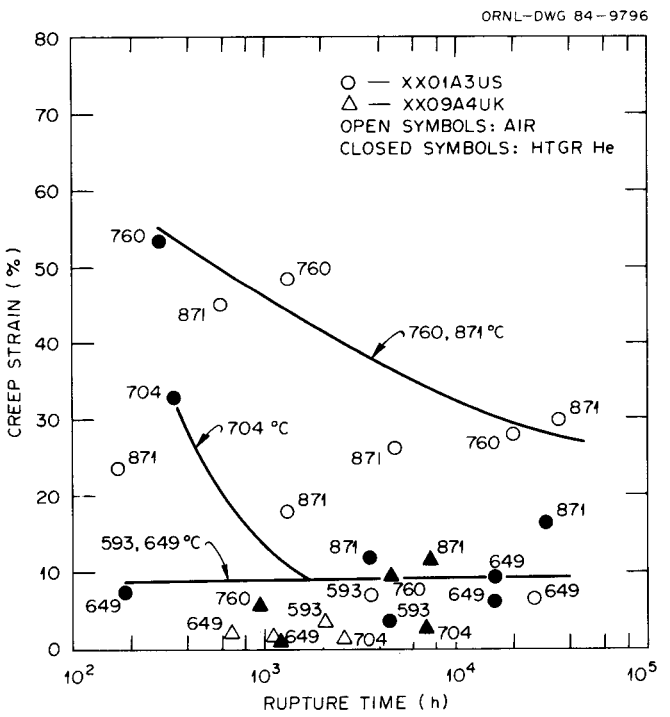


Fig. 32. Creep fracture strains of Inconel 617 at various temperatures.

to represent heat XX01A3US primarily. Heat XX09A4UK clearly has lower fracture strains. At test temperatures below 760°C, many tests failed with less than 10% creep strain and many with only 1 to 2%. At 760 and 871°C, the fracture strains were much higher. There is no obvious effect of test environment on the fracture strain.

Figure 33 shows that the time to tertiary creep is about 0.75 times the rupture life. This correlation seems to hold reasonably well for both heats, both environments, and all test temperatures.

Some results were also included in Table 14 for base metal aged up to 20,000 h in an inert environment and in HTGR-He. The rupture lives and minimum creep rates of the aged specimens (Figs. 34 and 35) deviated very little from lines representing the as-received material taken from Figs. 28 and 31.

Some samples of all weld metal and some with transverse welds were tested. The results of these tests are summarized in Tables 15 and 16, and some of the test data are included in Figs. 34 and 35. Samples of all weld metal were slightly stronger than the base metal in that they had a longer rupture life and a lower minimum creep rate. However, the three transverse weld samples failed in the weld metal, indicating that the weld metal was slightly weaker. These comparisons are clouded by the fact that the base-metal tests involved two heats and the transverse weld tests involved a third heat of base metal. The fracture strain data in Tables 15 and 16 show that the fracture strains are about the same as those noted for the as-received base metal (Table 14) (i.e., values of 1 to 2% were obtained at 704°C and below, and higher values were obtained at 760 and 871°C).

Tests on aged and welded Inconel 617 are few at this time, so conclusions must be valued accordingly. It appears, however, that the changes in creep properties due to welding or aging up to 20,000 h are very small.

INCONEL 618

The results of creep tests on Inconel 618 are given in Table 17. This material was tested as received (Fig. 4) and after annealing for 1 h

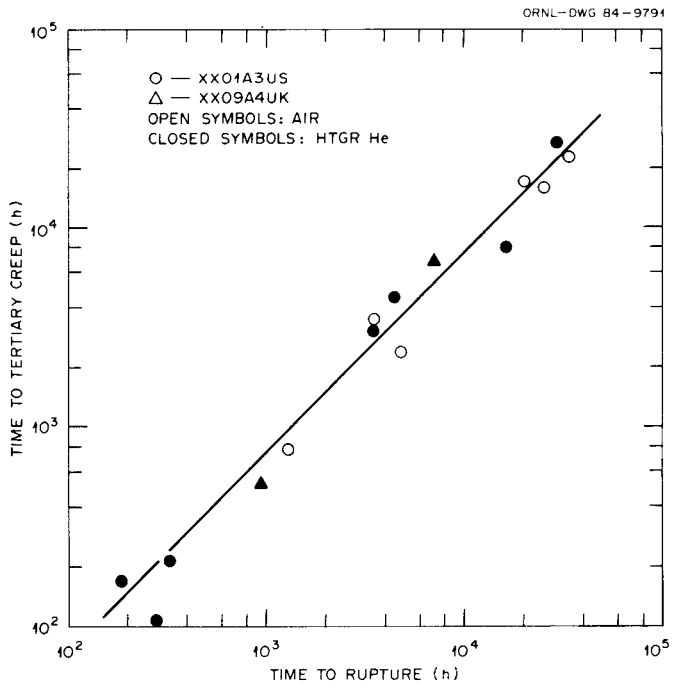


Fig. 33. Correlation of times to rupture and tertiary creep for Inconel 617.

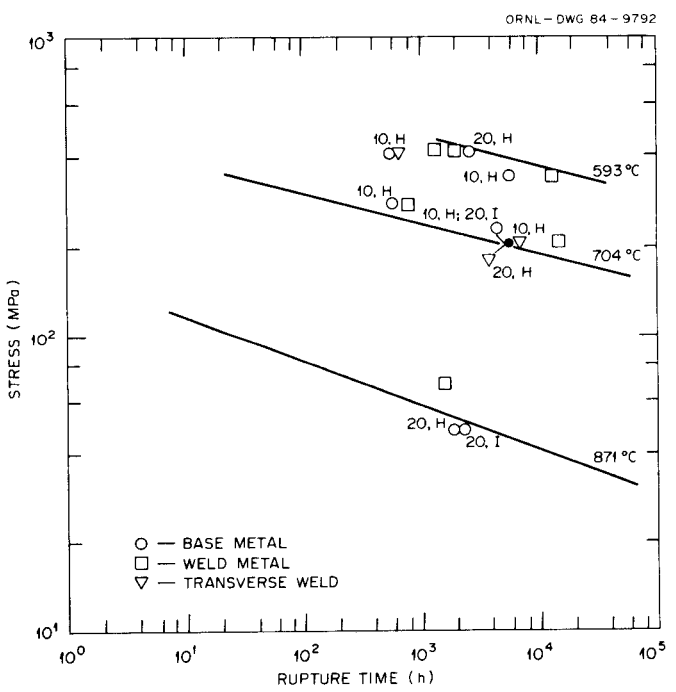


Fig. 34. Influence of welding and aging on stress-rupture properties of Inconel 617 (heat XX01A3US). Lines represent as-received base metal. Points are for weld samples and for aged samples, with numbers indicating aging time in 1000 h and "H" or "I" denoting HTGR-He or inert aging environments, respectively.

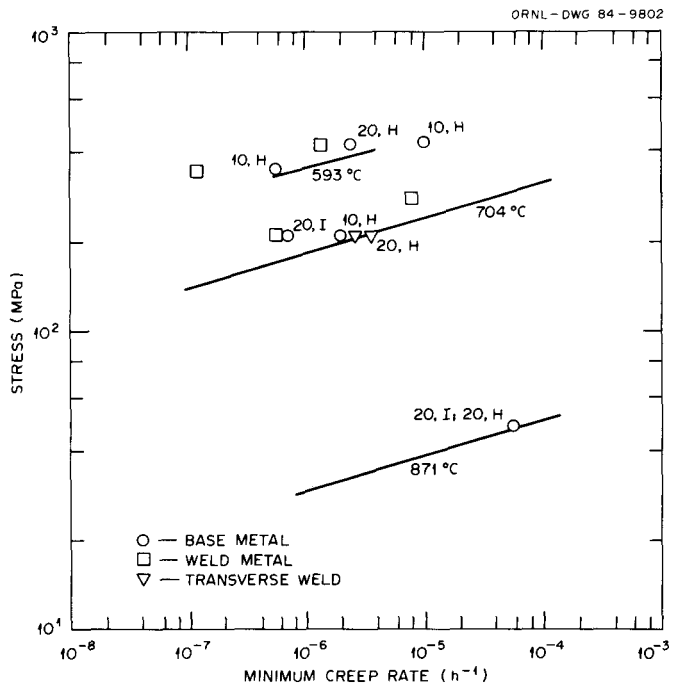


Fig. 35. Influence of welding and aging on minimum creep rate of Inconel 617 (heat XX01A3US). Lines represent as-received base metal. Points are for weld samples and for aged samples, with numbers indicating aging time in 1000 h and "H" or "I" denoting HTGR-He or inert aging environments, respectively.

Table 15. Results of creep tests on Inconel 617 weld metal in helium environment

(Weld metal heat XX09A9UK)

Test	Stress (MPa)	Stress (ksi)	Time to indicated creep strain (h) 1%	2%	5%	Time to tertiary creep (h)	Rupture or test (h)	Steady-state creep rate (h^{-1})	Elongation (%) Loading	Elongation (%) Creep	Reduction of area (%)
Tested at 593°C (1100°F)											
20515	345	50					13,726[a]	1.2 E-7	0.6	0.4	
	414	60					1,296		0.09	0.1	
23071	414	60					2,034	1.3 E-6	0.72	2.0	5.1
Tested at 649°C (1200°F)											
21399	345	50	525				526	8.1 E-6	0.1	2.6	3.3
20513	276	40					3,467	5.4 E-7	0.1	2.3	1.4
20553	242	35	15,400				18,182	4.7 E-7	0.2	1.8	2.9
Tested at 704°C (1300°F)											
22741	276	40	775			785	790	8.0 E-6	0.5	4.4	5.9
22763	207	30	4,250	10,500	14,900	6,000	14,978	5.5 E-7	0.04	7.7	14.1
Tested at 760°C (1400°F)											
20735	138	20	2,900	3,750	4,750	2,600	5,190	1.8 E-6		17.1	43.5
Tested at 871°C (1600°F)											
20533	69	10	1,467	1,570			1,623		0.07	10.7	21.0

[a]Discontinued before failure.

Table 16. Results of creep tests on Inconel 617 transverse welds in helium environment

(Base plate heat XX14A6UK, weld metal XX09A9UK)

Test	Condition[a]	Time to indicated creep strain (h)			Time to tertiary creep (h)	Rupture or test (h)	Steady-state creep rate (h^{-1})	Elongation (%)		Reduction of area (%)
		1%	2%	5%				Loading	Creep	
					Tested at 593°C (1100°F) and 414 MPa (60 ksi)					
20522	593/10,000/He					645			1.7	5.2
					Tested at 704°C (1300°F) and 207 MPa (30 ksi)					
20528	704/10,000/He	4,830	6,380		5,100	6,932	2.6 E-6	0.3	3.7	17.1
21612	704/20,000/He	1,100	3,250	5,470	3,025	5,496	3.7 E-6	0.0	6.5	17.6

[a]Aging temperature (°C)/aging time (h)/aging environment.

at 1177°C (Fig. 5). The alloy was also welded by the GTA process with filler metal of the same heat as the base metal. Transverse weld samples were tested in the as-welded and annealed conditions.

The data obtained at 760°C are used to illustrate several points. The time to 1% strain is shown in Fig. 36 as a function of stress level, and the time to rupture is shown in Fig. 37. The data show considerable scatter, but the fine-grained as-received material is weaker in most instances than the coarser grained annealed material.

The data at various temperatures were correlated by using the Larson-Miller parameter, and the results are shown in Figs. 38 and 39 for the times to 1% strain and rupture, respectively. These results show some tendency for the as-received material to be weaker in short-term tests, but the data for welds and base metal tested in air and in HTGR-He fall within a narrow scatter band. The only two transverse weld specimens (Table 17) that have failed thus far failed in the base metal, indicating that the weld metal seems to be stronger in creep. The test at 760°C on an all-weld-metal sample (T.N. 23081) indicates that the weld metal is creeping slower than the base metal.

Fracture strains of the samples are shown in Fig. 40. Most of the samples had fracture strains above 20%, but the two annealed transverse weld samples tested at 871°C failed after only 6% strain. Test temperature and environment do not appear to affect the fracture strain.

Table 17. Results of creep tests on Inconel 618

Test	Specimen type[a]	Anneal[b]	Stress (MPa)	Stress (ksi)	Environment	Time to indicated creep strain (h) 1%	2%	5%	Time to tertiary creep (h)	Rupture or test (h)	Steady-state creep rate (h^{-1})	Elongation (%) Loading	Creep	Reduction of area (%)
Test temperature 593°C (1100°F)														
23070	B	No[c]	345	50	He	3.5	10	31		66.7	1.4 E-3	6.7	19.4	22.3
Test temperature 649°C (1200°F)														
22021	B	No	172	25	He	50	90	260	830	1,584	1.6 E-4	0.4	47.5	46.0
Test temperature 704°C (1300°F)														
23614	B	No[d]	69	10	He	730	5,350			5,859[e]	2.1 E-6		1.9	
23346	B	No	69	10	He	2,430	5,560			10,990[e]	2.6 E-6		4.0	
23347	TW	No	69	10	He					9,638[e]	4.0 E-7		0.3	
23616	TW	No[d]	69	10	He					2,631[e]		0.2	0.1	
Test temperature 760°C (1400°F)														
20637	B	Yes	131	19	He	2	7	12	45	85	3.1 E-3	0.1	43.2	45.6
20388	B	No	103	15	He	5	8	15	5	34	2.0 E-3	0.0	36.6	41.3
20679	B	Yes	103	15	He	9	20	60	290	392	5.6 E-4	0.2	29.9	31.1
20494	B	Yes	90	13	He	20	50	180	840	1,263	1.5 E-4	0.1	29.2	37.4
20398	B	No	69	10	He	18	30	51	20	117	3.8 E-4	0.1	43.7	36.9
22199	B	Yes	69	10	He	340	1,340	4,800	8,300	14,436	8.6 E-6	0.1	22.3	28.3
23083	TW	Yes	69	10	Air					11,158[f]		0.1	0.5	1.4
23079	B	No	41	6	Air	4,400	9,500			15,776[e]	2.0 E-6		3.2	
23080	TW	No	41	6	Air	3,150	6,700			15,718[e]	2.3 E-6	0.1	4.1	
23081	W	No	41	6	Air					13,677[e]			0.5	
Test temperature 871°C (1600°F)														
20493	B	Yes	62	9	He	3	7	20	65	93	2.2 E-3	0.2	29.9	29.4
22189	B	Yes	35	5	Air	95	245	950	1,900	3,671	5.1 E-6		22.1	19.8
22196	B	Yes	35	5	He	715	1,275	2,530	1,250	4,269	1.4 E-5		17.2	17.9
23087	TW	No	35	5	Air	450	2,600	4,650	3,250	5,110	4.6 E-6	0.1	6.5	10.5[g]
23615	TW	No	35	5	He	2,600				3,348	2.2 E-6	0.01	5.8	4.9[g]

[a]B = Base metal; TW = transverse weld; W = weld metal.

[b]Samples annealed 1 h at 1177°C (Yes); tested in the as-received condition (No).

[c]Aged 12,500 h at 593°C in helium.

[d]Aged 12,500 h at 704°C in helium.

[e]Test in progress.

[f]Discontinued before failure.

[g]Failure located in base metal.

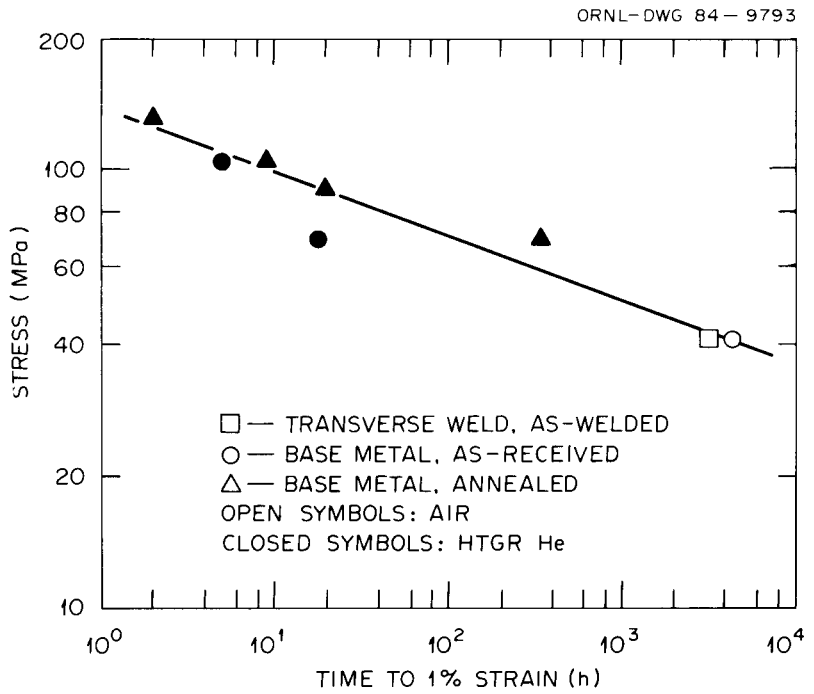


Fig. 36. Creep strain behavior of Inconel 618 at 760°C.

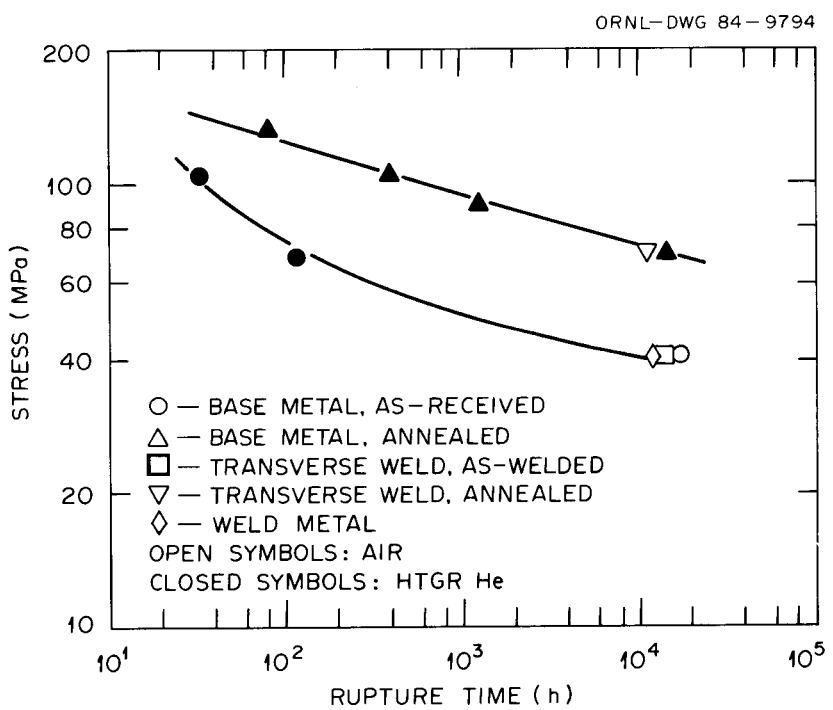


Fig. 37. Stress-rupture properties of Inconel 618 at 760°C.

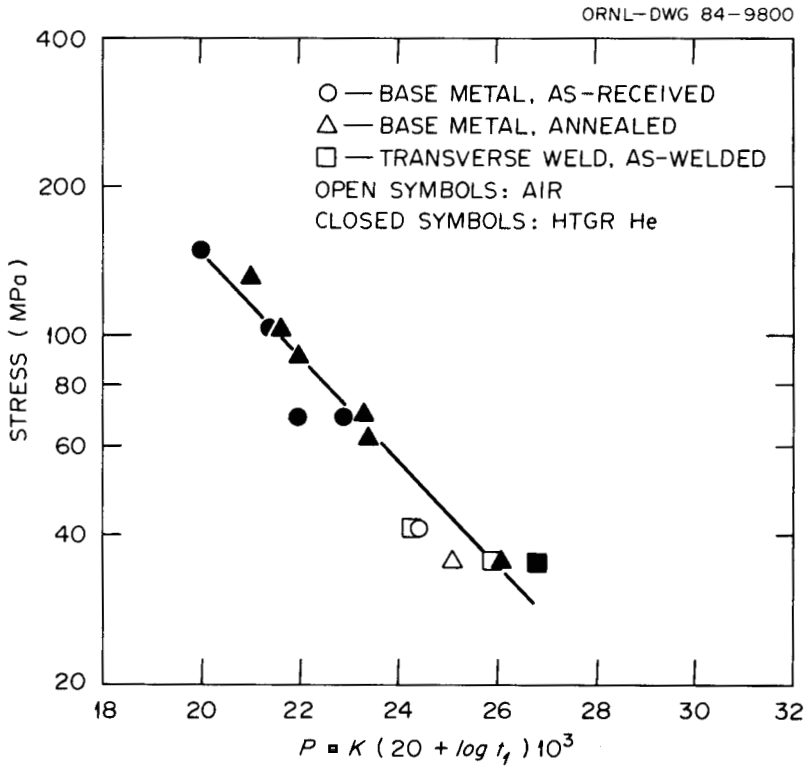


Fig. 38. Larson-miller correlation for Inconel 618 to 1% strain.

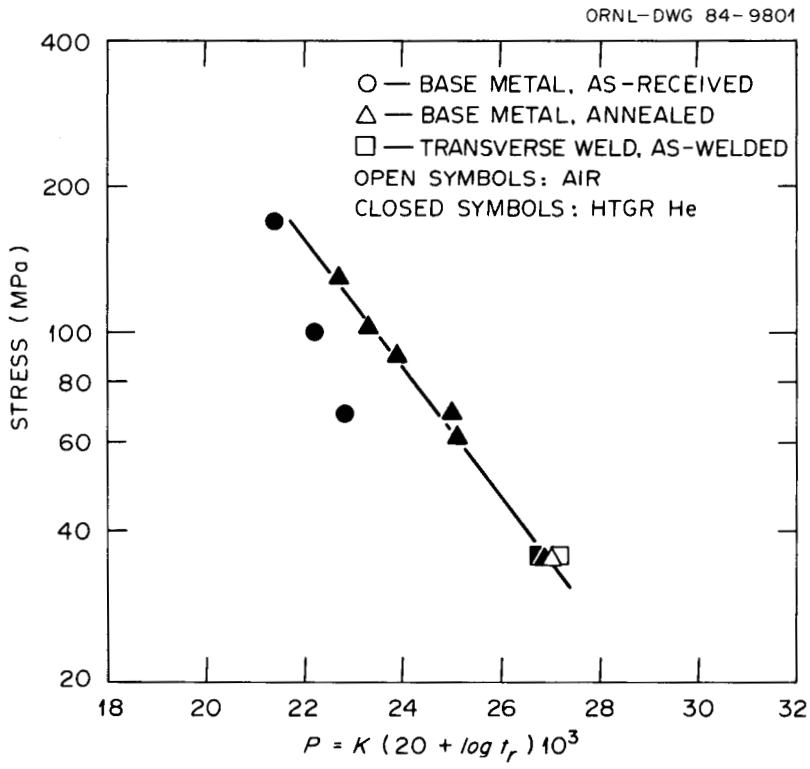


Fig. 39. Larson-Miller correlation for Inconel 618 to rupture.

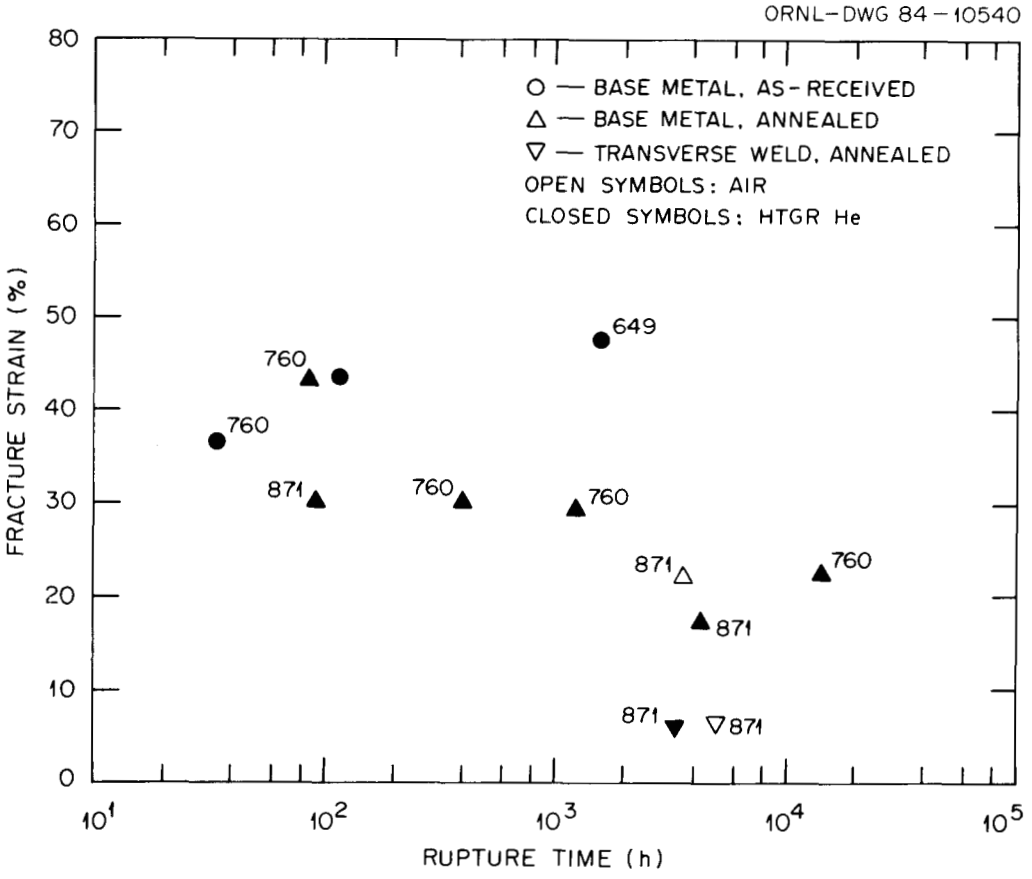


Fig. 40. Creep fracture strains for Inconel 618 tested at the indicated temperatures.

TENSILE TESTS OF DISCONTINUED CREEP SAMPLES

One sample each of Inconel 617 and 618 was exposed to a creep load for a long period of time and then subjected to a short-term tensile test at ambient temperature. The results of these tests are described in Table 18. The first test involved a sample of Inconel 617 that had been creep tested at 871°C for 26,117 h and had strained 0.3%. The sample was heavily carburized and was found to contain 0.233% C. The results for this test can be compared with those for samples aged 20,000 h in static air (Table 3) and in HTGR-He (Table 4). The yield stress of the creep sample is higher, and the ultimate tensile stress and elongation lower. The carbon content of the creep sample is much higher than that of the

Table 18. Tensile properties at 25°C of Inconel 617 and 618 after exposures in creep tests

Material	Pretest condition	0.2% Yield stress		Ultimate tensile stress		Elongation (%)		Reduction of area (%)
		(MPa)	(ksi)	(MPa)	(ksi)	Uniform	Total	
Inconel 617, XX09A4UK	Test 21602, 871°C, 21 MPa, 26,117 h, 0.233% C	326	47.3	446	64.7	4.0	4.0	0.3
Inconel 618, transverse weld, Y09B9	Test 23083, 760°C, 69 MPa, 11,158 h, 0.062% C	565	53.0	582	84.4	19.7	19.7	22.3

sample aged in HTGR-He, and the main effects of the higher carburization on tensile properties appear to be to decrease the ultimate tensile stress and the fracture elongation.

The second creep sample subjected to a short-term tensile test was an Inconel 618 sample creep tested for 11,158 h at 760°C. During the test the sample deformed 0.5%, and the carbon content increased from 0.047 to 0.062%. Tensile properties of this alloy are shown in Table 10 for the as-received condition and in Table 12 for material aged 12,500 h in HTGR-He. The samples in Table 12 were aged at 704 and 871°C, whereas the creep sample was tested at 760°C; therefore, some interpolation is necessary. The yield stress of the creep sample is as expected, the ultimate tensile stress is about 20% lower, and the fracture elongation is about 50% lower. The carbon content of the creep sample is about the same as that for the samples aged in HTGR-He, so it is not obvious whether the larger property changes are associated with creep damage or with exposure to the environment.

EXAMINATION OF FRACTURES OF CREEP SAMPLES

The fracture surfaces of several specimens were examined in the scanning electron microscope (Table 19). Figure 41 shows the fracture of an Inconel 617 specimen tested at 593°C. The fracture was largely intergranular cleavage with interspersed ductile transgranular fracture areas. The particles on the fracture surface were very high in titanium.

Table 19. Summary of samples examined in scanning electron microscope

(Tested in HTGR-He)

Test	Material	Specimen type	Temperature (°C)	Time (h)	Creep strain (%)
20521[a]	617	Base metal	593	548	2.2
20522[a]	617	Transverse weld	593	645	1.7
20513	617	Weld metal	649	3467	2.3
20735	617	Weld metal	760	5190	17.1
20545	617	Base metal	871	3562	11.8
23021	618	Base metal	649	1584	47.5

[a]Aged 10,000 h at 593°C in HTGR-He.

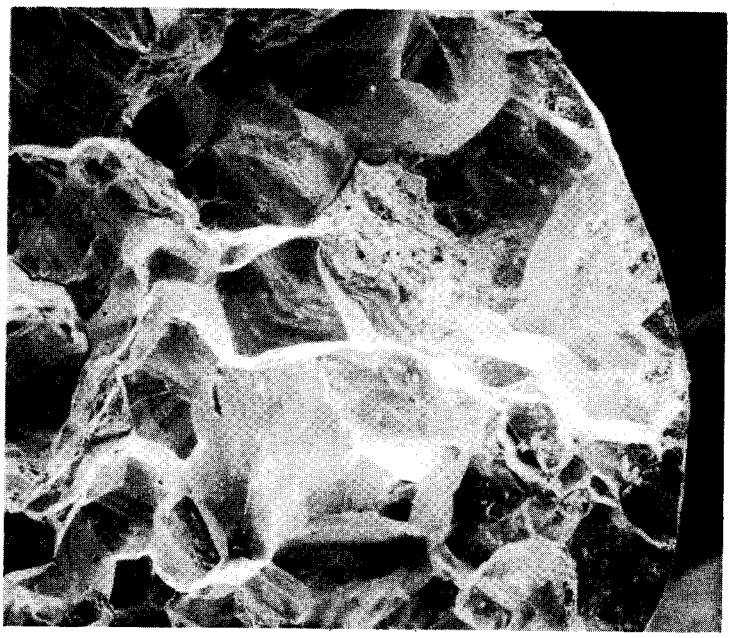
The fracture of a transverse weld sample of Inconel 617 is shown in Fig. 42. The fracture occurred near the fusion line and included portions of base and weld metal. The base metal failed almost entirely by intergranular cleavage, and the weld metal seemed to fail along dendrites.

The fracture of a sample of weld metal of Inconel 617 tested at 649°C is shown in Fig. 43. Fracture occurred along the dendrites with a fracture strain of only 2.3%. Figure 44 shows a fracture from a similar sample of weld metal tested at 760°C. The fracture path was influenced by the dendrites in the weld, but there is more evidence of plastic flow than noted at 649°C (Fig. 43). The fracture strain of the specimen tested at 760°C was 17.1%.

The fracture of an Inconel 617 sample tested at 871°C (Fig. 45) is predominately intergranular, but there is considerable evidence of plastic deformation and intergranular cracking. The fracture strain of this sample was 11.8%.

A sample of Inconel 618 tested at 649°C was examined, and typical micrographs are shown in Fig. 46. The fracture was largely transgranular, the deformation mode was ductile shear, and the fracture strain was 47.5%. This type of high-ductility failure was not noted at any of the test conditions examined for alloy 617.

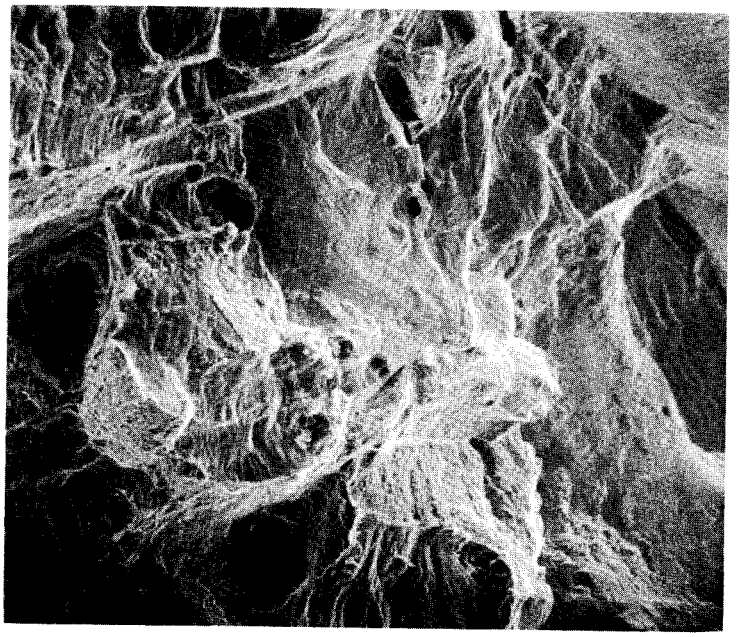
M-10818



(*a*)

250 μm

M-10820



(*b*)

50 μm

Fig. 41. Scanning electron micrographs of specimen 20521 of Inconel 617 tested at 593°C with a 2.2% strain.

M-10791

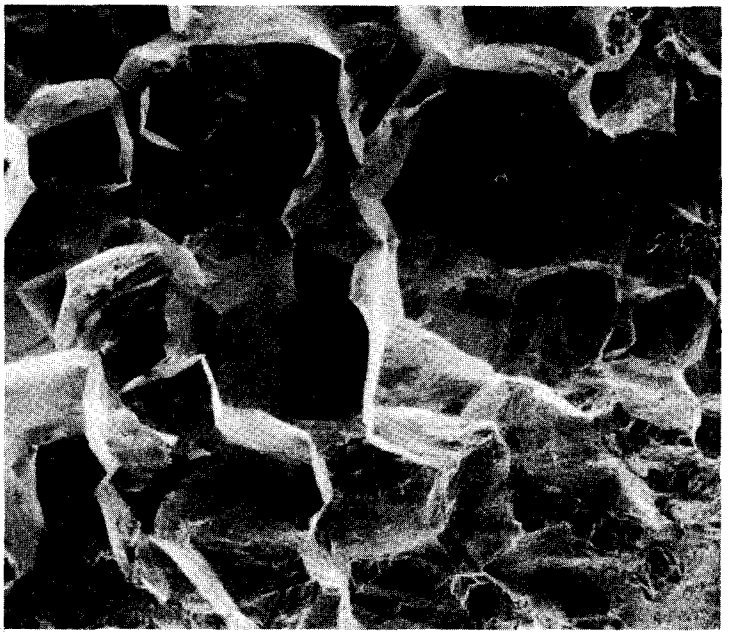
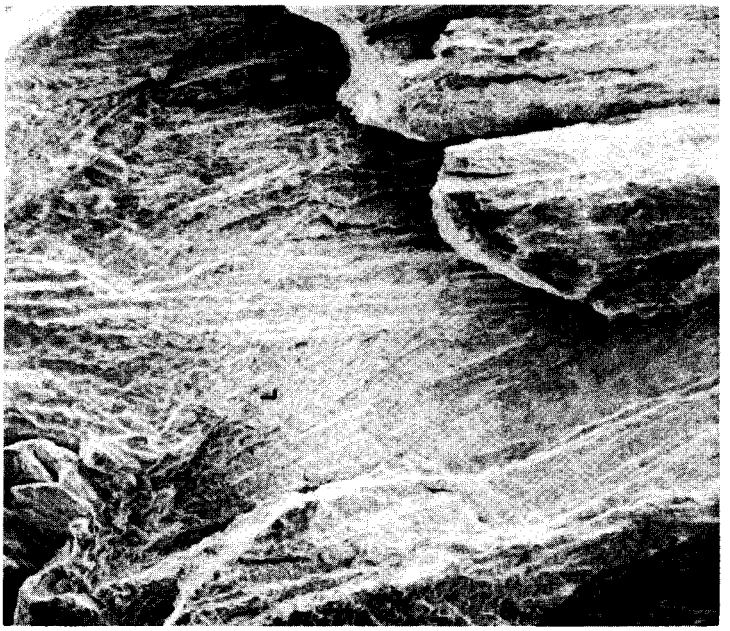


Fig. 42. Scanning electron micrograph of specimen 20522 of Inconel 617 transverse weld sample tested at 593°C with a strain of 1.7%.

Selected area diffraction was used to determine whether variations in grain boundary chemical composition accounted for the low fracture strain of alloy 617 at temperatures of 593 and 649°C. The fracture surface of samples tested at 593°C (brittle) and those tested at 871°C (ductile) contained all the chemical constituents of the alloy, including small amounts of aluminum and titanium; however, there were no detectable differences in the compositions of the fracture surfaces of ductile and brittle fractures. The composition of the fracture surface of a sample of alloy 618 resembled very closely that of the bulk metal, including small amounts of titanium and aluminum.

M-10814



(*a*)

250 μm

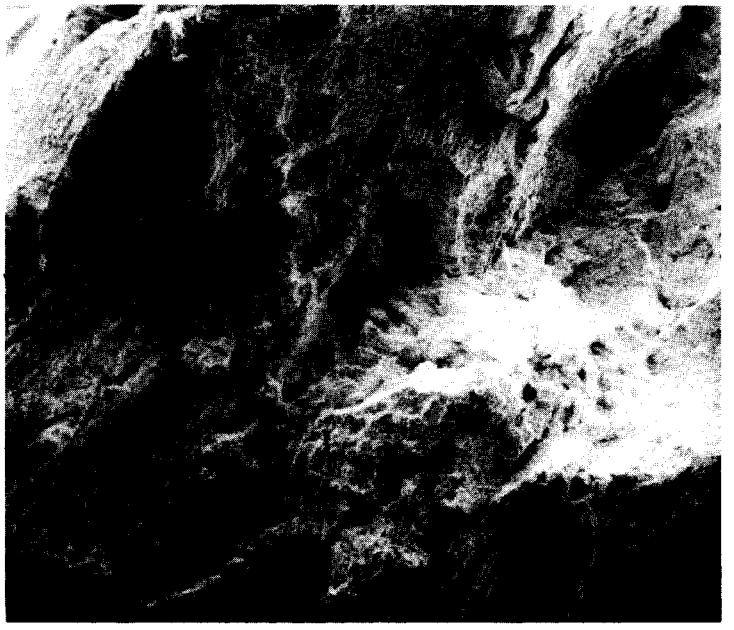
M-10815

(*b*)

50 μm

Fig. 43. Scanning electron micrographs of specimen 20513 of Inconel 617 weld metal tested at 649°C with a strain of 2.3%.

M-10809



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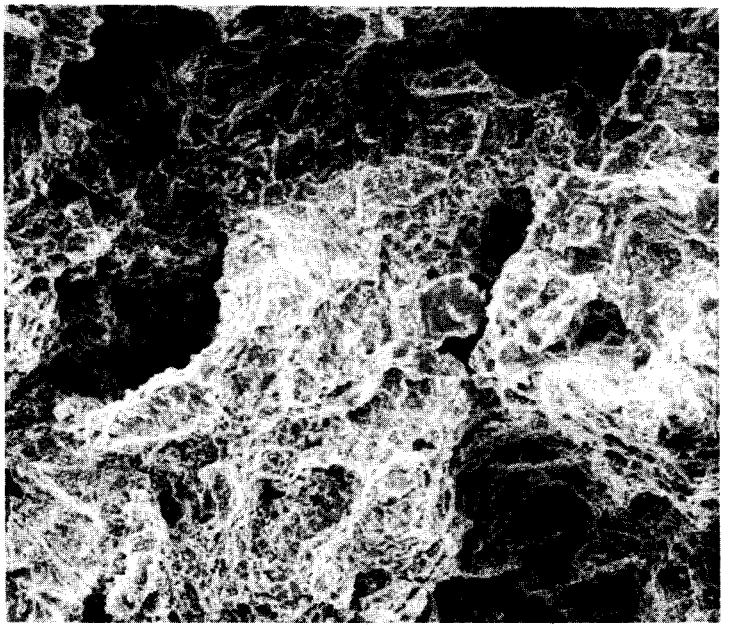

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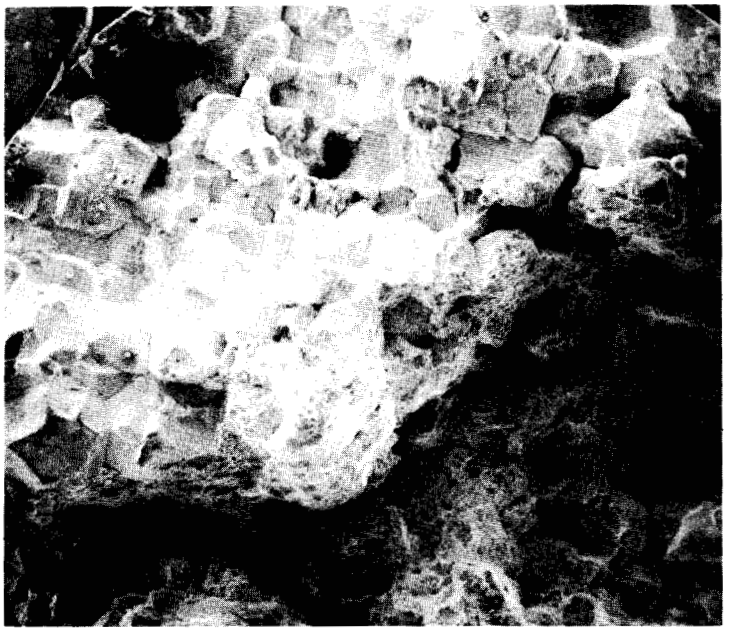
Fig. 44. Scanning electron micrographs of sample 20735 of Inconel 617 weld metal tested at 760°C with a strain of 17.1%.

M-19479



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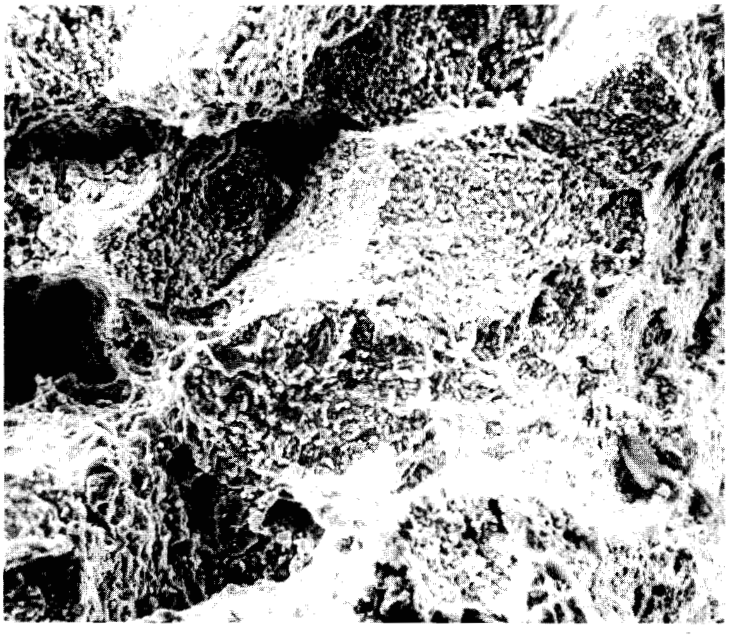

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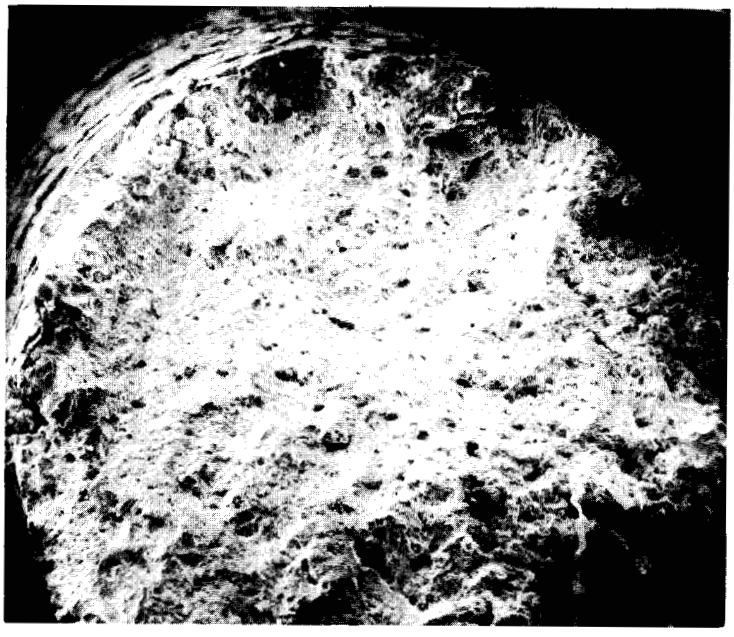
Fig. 45. Scanning electron micrographs of specimen 20545 of Inconel 617 tested at 871°C with a strain of 11.8%.

M-19482



(*a*)

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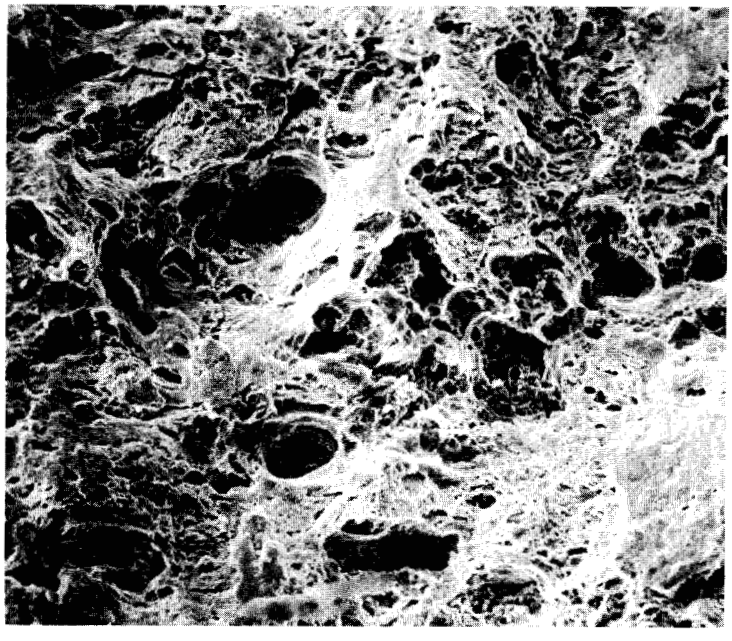
M-19480



(*b*)

50 μm

Fig. 46. Scanning electron micrographs of specimen 23021 of Inconel 617 tested at 649°C with a strain of 47.5%.

POSTTEST CARBON ANALYSES

The carbon content of most samples tested in HTGR-He was determined. Two small disks were cut from the gage section, and each was totally consumed in the analysis. Thus, the carbon concentrations determined were averages of the cross section of the 3.2-mm-diam (1/8-in.) samples. The values for Inconel 617 are shown in Fig. 47. The various heats of material are not distinguished in this figure. The initial carbon concentrations of the alloy 617 lots studied ranged from 0.06 to 0.08%. There are only a few data points at each temperature; consequently, the lines drawn in Fig. 47 are not very precise. At 593 and 649°C, no carburization could be detected. At 704°C one sample exposed for 7000 h had appreciable carburization, and one exposed for 4600 h at 760°C was carburized. At 871°C carburization was noted in a sample exposed for only 3500 h. The point

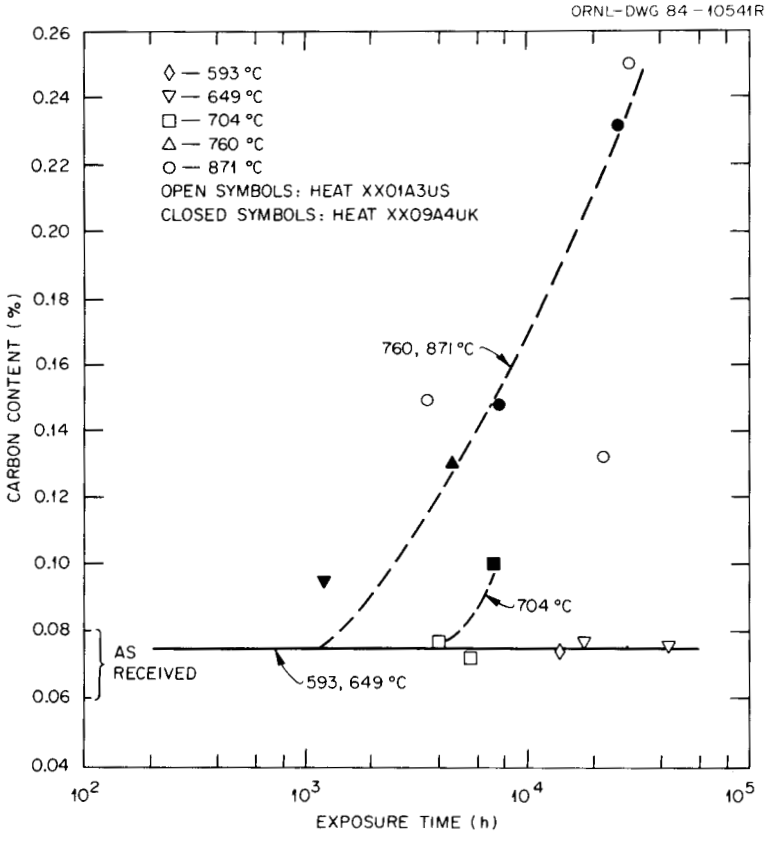


Fig. 47. Posttest carbon concentrations of Inconel 617 creep samples.

corresponding to 22,000 h exposure at 871°C and 0.132% carbon was for a sample aged for 10,000 h in a retort with other samples and tested for 12,000 h in a retort where it was the only sample. The amount of carburization that occurred during exposure with several samples was less than that noted where a single sample was involved. The CO and CH_4 most likely were depleted in the aging retort at 871°C limiting the amount of carburization.

The data for Inconel 618 are even fewer than for alloy 617. The available data (Fig. 48) indicate that measurable carburization occurred at 871 and 760°C, but there is a hint of decarburization at 649°C. On the basis of the sparse data, alloys 617 and 618 do not appear appreciably different in their carburization tendencies.

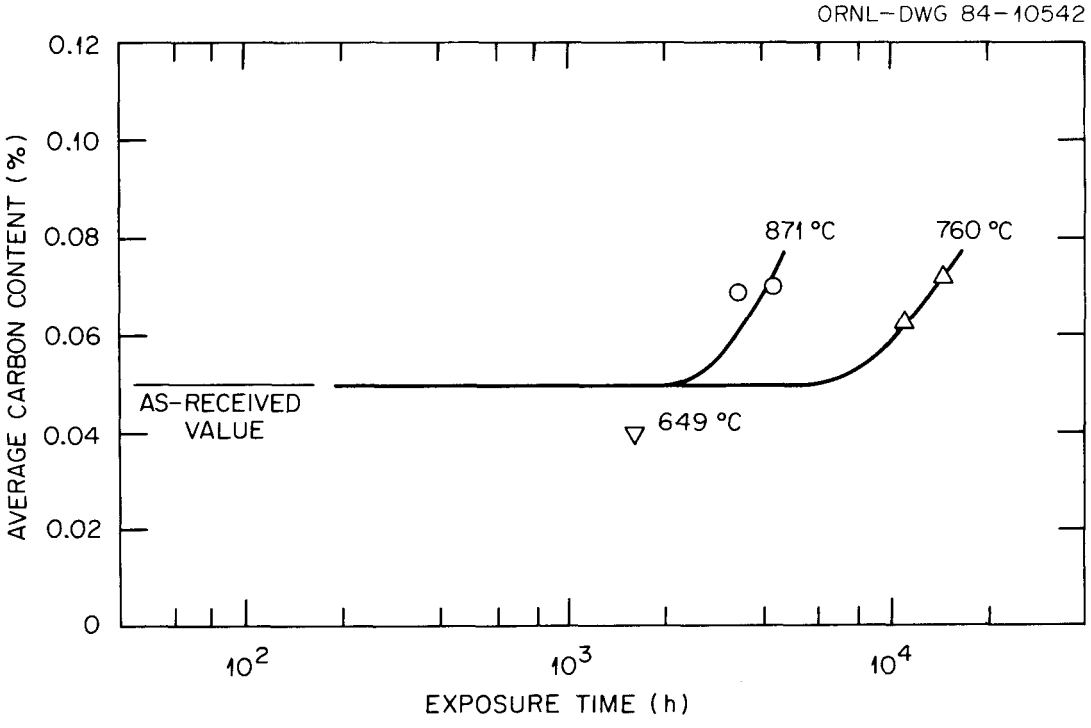


Fig. 48. Posttest carbon concentrations of Inconel 618 creep samples.

DISCUSSION OF RESULTS

One of the first points of concern is how the tensile and creep properties measured in this study for Inconel 617 compare with those in literature from the manufacturer. International Nickel Company, Inc. (INCO).[6] The yield and tensile strengths of the solution-annealed material for our two heats and those in INCO literature compare very well. The fracture elongation trend described in INCO literature is one of increasing elongation with increasing temperature. The results for our heat XX01A3US follow this trend, but the trend for our heat XX09A4UK is one of decreasing elongation with increasing temperature above 600°C.

Because Inconel 618 is not as popular as alloy 617, our best source of information about it was a personal communication[7] from Huntington Alloy Products Division of INCO. There is some question about the condition of our material, but its very fine grain size indicates that it was either as hot formed or annealed at 1000 to 1100°C. Nevertheless, our test results and those in the communication[7] agree very well. The ultimate tensile strength of our material consistently fell below the Huntington Alloy results by about 10%, but that agreement is good.

Creep-rupture data were available for Inconel 617 and 618 in the two sources just listed. Those data are compared with ours in Fig. 49 in the form of a Larson-Miller correlation. The agreement was excellent for alloy 617, and there is only a minor variation in the shapes of the two curves for alloy 618. On the basis of the small number of data points used by the manufacturer and ORNL in determining these curves, the agreement is quite acceptable. Thus, the data obtained in this study compare well with the data obtained by the supplier.

The tensile properties of Inconel 617 were very good in the solution-annealed condition (i.e., the yield and ultimate strengths were high, and the elongation and reduction of area were high). Exposing this material to temperatures over the range of 538 to 871°C brought about some very significant changes in the fracture behavior. The yield and ultimate tensile strengths went through small variations, but those changes were secondary to the decreases noted in fracture elongation. The fracture

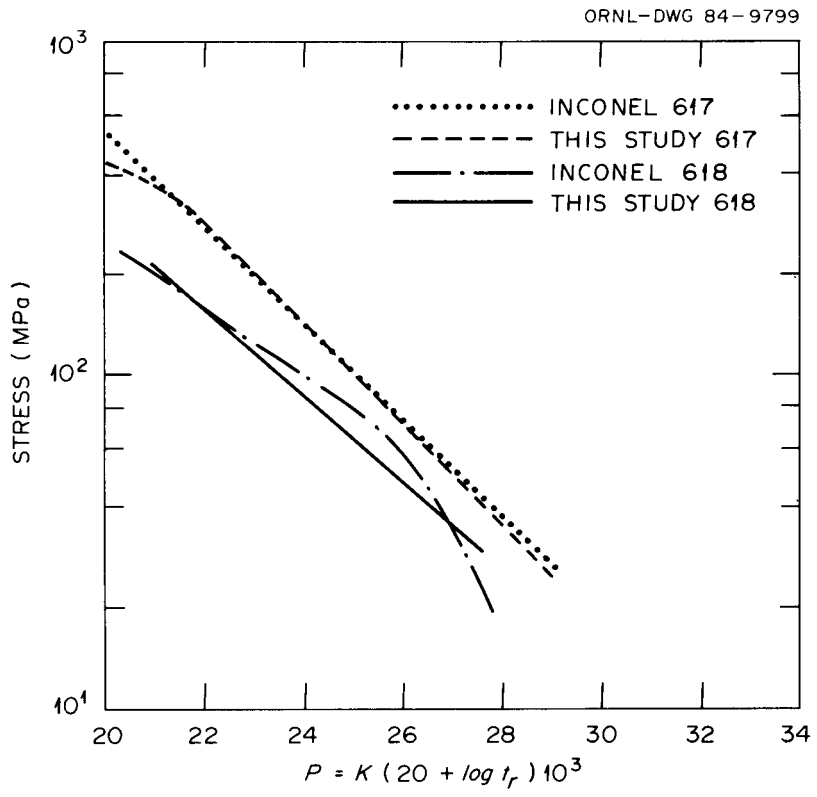


Fig. 49. Comparative stress-rupture properties for Inconel 617 and 618. *Source of Inconel 617 data*: Circular 20M 9-72 T-46, manufacturer's literature of Huntington Alloy Products Division, The International Nickel Company, Inc., Huntington, W. Va., 1972. *Source of Inconel 618 data*: D. W. Rahoi, Huntington Alloy Products Division, the International Nickel Company, Inc., Huntington, W. Va., personal communication to J. F. King, Oak Ridge National Laboratory, Mar. 7, 1979.

strain was lower in specimens tested at ambient temperature than in those tested at elevated temperatures and lower for specimens aged in HTGR-He than for specimens aged in inert gas. That trend was also noted for specimens aged and impact tested. The impact energy decreased with increasing aging time, and the kinetics of the reduction increased with increasing temperature.

Inconel 617 could be welded quite soundly with alloy 617 filler wire by the GTA process. Tensile specimens made totally of weld metal were tested and found to have much higher yield strengths, slightly higher ultimate tensile strengths, and lower (50%) fracture elongations than those of base metal. In transverse weld specimens, fracture initially occurred in the weaker base metal, but increasing aging time and higher

temperatures shifted the fracture location to the weld metal. Aging reduced the fracture strain of the weld metal qualitatively (as noted for the base metal), but the absolute values of fracture strain were much lower for the weld metal than for the base metal.

Creep tests were run on two heats of Inconel 617. Some tests have reached times of 25,000 to 35,000 h, but so many variables are involved that trends can be noted with confidence but conclusions must be tentative. At test temperatures of 593, 649, and 704°C, fracture occurred after creep strains of only 1 to 2%. At test temperatures of 760 and 871°C, high creep strains occurred. This was also true for aged samples and for weld metal. Scanning electron microscopy revealed that the low-strain fractures were largely intergranular (interdendritic for weld metal) cleavage fractures and that the higher strain fractures were intergranular (or interdendritic) shear fractures. There were no detectable general variations in the chemical composition of the fracture surfaces, but some individual particles quite high in titanium were noted on the brittle fracture surfaces. Markins, Hosier, and Bassford[1] noted small amounts of gamma prime in Inconel 617 exposed at 649 and 760°C, but they performed a PHACOMP analysis to predict that only 0.63% gamma prime could form in this alloy. Thus, gamma prime may or may not be associated with the low creep strain of this alloy in the intermediate temperature range.

Creep samples were analyzed for carbon following testing. Samples exposed at 593 and 649°C were not detectably carburized. Samples tested at 704, 760, and 871°C were carburized to an extent that increased with temperature. Although the carbon content increased, none of the creep properties was detectably affected by whether the test environment was HTGR-He or air.

Tensile tests were run on Inconel 618 both as received and welded. Transverse weld samples generally failed in the weld metal, but the strength and fracture strain of the weld metal were only slightly less than those of the base metal. Both base-metal and transverse weld samples were aged 12,500 h in HTGR-He, and only small property changes were noted. Impact samples notched in the base and weld metal were aged 20,000 h at

871°C in an inert environment, and the impact energy remained almost unchanged. Thus, the thermal stability of this alloy is very good. Creep tests revealed some sensitivity of strength to grain size in short-term tests; however, that factor seems unimportant in long-term tests. Testing began with the material in a rather fine-grained condition, and a grain-coarsening anneal of 1 h at 1177°C was later incorporated. The creep strength of this alloy may be improved by a pretest anneal. The creep test results thus far do not show any effects of test environment.

Creep samples were measurably carburized at test temperatures of 760 and 871°C, and slight decarburization may have occurred at 649°C. The scant data available fail to indicate any basic differences in carburization resistance between Inconel 617 and 618.

SUMMARY AND CONCLUSIONS

Tensile and creep properties were measured for Inconel 617 and 618 base metal and welds. Techniques were developed for welding the materials by the GTA process, and sound test welds were made. Samples of both materials were aged up to 20,000 h, and subsequent tests were run to determine the effects of aging on tensile, impact, and creep properties. The influence of HTGR-He containing small amounts of H_2, H_2O, CO, and CH_4 on the properties was of concern in all phases of this study.

Inconel 617 had good tensile and creep strength. Transverse weld samples were as strong as base metal initially, but the weld became slightly weaker than the base metal with increasing aging time and increasing test temperature. A major problem with this alloy is the decrease in fracture strain due to aging. The decrease was greater for tests at ambient temperature, but the fracture strain was also lower when the tests were conducted at the aging temperature. The ductility was reduced more for welds than for base metal, more in aging environments of HTGR-He than in inert environments, and more for increasing aging time and temperature. This reduction in short-term fracture strain also manifests itself in a large reduction in the impact energy. Creep tests did not reveal any effects of environment. Rather low creep strains of 1 to 2%

occurred at 593, 649, and 704°C before fracture, and much higher strains were observed at 760 and 871°C. This creep ductility problem was equivalent for base and weld metals.

Inconel 618 was characterized by good metallurgical stability, with very minimal effects of aging on tensile and impact properties. The weld metal was about as strong as the base metal. Long test times and high test temperatures favored failure in the weld metal of transverse specimens. The test environment had no detectable influence on the creep properties.

Both alloys were carburized quite rapidly at 760 and 871°C in HTGR-He. The carburization rates at 593 and 649°C were very slow, and only small changes in carbon content were measured.

It is informative to compare some of the mechanical properties of interest for Inconel 617 and 618 and Hastelloy X. The data from this study and data from a previous study[5] for Hastelloy X were used to develop Table 20. The three alloys are very similar on the basis of short-term

Table 20. Comparison of alloy properties

	Hastelloy X	Inconel 617	Inconel 618
Yield stress at 25°C, MPa	350	303	373
Ultimate tensile stress at 25°C, MPa	764	752	703
Yield stress at 871°C, MPa	162	185	175
Ultimate tensile stress at 871°C, MPa	167	194	197
Impact energy at 25°C following aging 20,000 h at 871°C, J	11	20	104
Stress, MPa, for 1% creep in 100,000 h at			
649°C	84	195	70
704°C	60	125	51
760°C	42	76	36
871°C	20	29	18

yield and ultimate tensile strengths. The main "short-term parameter" affected is the impact energy following aging for 10,000 h at 871°C. Hastelloy X and Inconel 617 have the lowest impact energies, and alloy 618 has by far the highest. For a creep criterion 1% strain in 100,000 h was chosen. The allowable stresses for Inconel 617 are the highest, and Hastelloy X and Inconel 618 are weaker but similar to each other. There is a definite trend for the difference in allowable stress to become less as the temperature is increased. Thus, Inconel 617 is the strongest, but it has severe ductility limitations. Inconel 618 is the most stable and does not appear to have ductility problems, but it is not as strong in creep as alloy 617. Hastelloy X has even worse ductility problems than does Inconel 617, but its creep strength is only slightly better than that of Inconel 618. Thus, there are several trade-offs in selecting from these alloys.

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