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ORNL-TM-3321

THERMAL STABILITY OF TITANIUM-MODIFIED HASTELLOY N AT 650 AND 760° C

C. E. Sessions and E. E. Stansbury

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C. E. Sessions and E. E. Stansbury[1]

ABSTRACT

We have investigated the influence of small titanium additions on the thermal stability of Ni–12% Mo–7% Cr–0.07% C. The mechanical properties at 650°C (tensile tests at 0.002/min strain rate and creep tests at 40,000 psi stress) were measured for four heats of this alloy with titanium contents from 0.15 to 1.2%. Solution annealing temperatures were 1177 or 1260°C, and subsequent precipitation heat treatments were conducted at 650 and 760°C. Titanium increases the stability of a complex MC-type carbide. At low titanium levels the MC carbide is stable at 650°C but is unstable at 760°C, where an M_2C-type carbide is precipitated, resulting in inferior properties. For the higher titanium concentrations the MC carbide is stable on aging at 760°C and results in excellent properties after a solution anneal at 1177°C. However, high-titanium alloys are significantly less ductile if they are solution annealed at 1260°C and aged at either 650 or 760°C. The heat with the lowest carbon content (0.04% C) was most resistant to property changes on aging up to 10,000 hr at both 650 and 760°C.

INTRODUCTION

We are concerned with modifying the composition of Hastelloy N for better performance as a material for a molten-salt thermal breeder reactor with an expected lifetime of 30 years or more. The existing Hastelloy N developed primarily at Oak Ridge over the past 10 to 12 years functioned well in an 8-MW(t) reactor built five years ago and operated over the past three years.[2] However, this reactor operated at

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[1]Consultant from the University of Tennessee, Knoxville, Tennessee.

[2]H. E. McCoy, Jr., An Evaluation of the Molten-Salt Reactor Experiment Hastelloy N Surveillance Specimens — Fourth Group, ORNL-TM-3063 (March 1971).

650°C, and when it was constructed very little was known about the problem of high-temperature irradiation damage to this alloy. Since that time we have observed[3] damage by thermal neutrons on many materials, including nickel-base alloys such as Hastelloy N, in which a larger deterioration in properties is observed than in most iron-base alloys. Thus, an alloy with greater resistance to radiation damage is needed.

We found that small additions of reactive elements, notably titanium, can significantly enhance the postirradiation creep-rupture life and ductility.[4] Thus, we have proceeded with the development of a modified Hastelloy N containing titanium. This report represents the phase of the program concerned with the effect of thermal and mechanical treatment on the creep and tensile behavior of these experimental alloys.

EXPERIMENTAL PROCEDURE

The material investigated in this program was produced by a commercial vendor. These four 100-lb heats were vacuum induction melted with initial ingot breakdown at 1177°C and final fabrication to 0.5-in. plate at 870°C. The titanium content was the primary variable, being 0.15, 0.27, 0.45, and 1.20% as shown in Table 1. Although a nominal carbon level of 0.07% was specified, the carbon content of Heat 466-548 was lower than the others. The various thermal and mechanical treatments applied are given in Table 2. The variables include solution annealing conditions, cold work, aging time, and aging temperature.

The dimensions of the mechanical property test specimen are shown in Fig. 1. This specimen was used so that the results could be compared directly to postirradiation mechanical property tests for which this

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[3]H. E. McCoy, "Variation of the Mechanical Properties of Irradiated Hastelloy N with Strain Rate," J. Nucl. Mater. 31, 67–85 (1969).

[4]H. E. McCoy, Jr., Influence of Titanium, Zirconium, and Hafnium Additions on the Resistance of Modified Hastelloy N to Irradiation Damage at High Temperature — Phase I, ORNL-TM-3064 (January 1971).

Table 1. Chemical Analyses of Alloys in Aging Program

Element	Chemical Analyses, wt %			
	466-535[a]	466-541	466-548	467-548
Mo	12.8	13.2	12.4	12.0
Cr	7.2	6.8	7.7	7.1
Ti	0.15	0.27	0.45	1.20
C	0.073	0.07	0.04	0.08
Fe	0.03	0.03	0.03	0.04
Si	0.07	0.05	0.05	0.03
Mn	0.12	0.10	0.14	0.12
Mg	0.034	0.01	0.023	
W	0.007	0.07	0.007	0.004
Zr	0.0002	0.0007	< 0.0001	0.002
V	< 0.01	< 0.01	< 0.01	0.001
Co	0.02	0.04	0.03	0.15
Cu	0.002	0.005	0.0005	0.01
Nb	< 0.0001	0.0001	0.0003	0.0005
Al	< 0.03	< 0.03	< 0.03	0.05
S	0.003	< 0.002	0.003	< 0.002
P				0.0004
B	0.0002	0.0002	0.00007	0.0007
O	0.0012	0.0008	0.001	0.0003
N	0.007	0.0005	0.0009	0.0002
H	0.0002	0.0002	0.0002	0.0003
Ni	Balance	Balance	Balance	Balance

[a]Heat numbers of 100-lb heats obtained from the Stellite Division of the Cabot Corporation.

Table 2. Variables Considered in This Study

Variable	Levels or Treatments
Titanium content, %	0.15, 0.27, 0.45, and 1.2
Preage treatment	Solution anneal 1 hr at 1177°C Solution anneal 1 hr at 1260°C Anneal 1 hr at 1177°C, then prestrain 10% at room temperature
Aging temperature, °C	650 and 760
Aging time, hr	1500, 3000, and 10,000
Testing after aging	
Tensile	650°C, 0.002/min strain rate
Creep	650°C, 40,000 psi stress

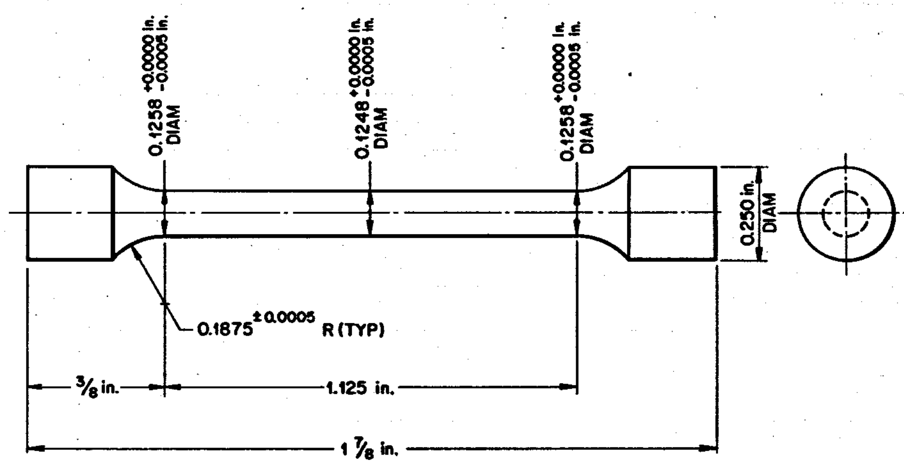

Fig. 1. Mechanical Property Test Specimen.

specimen was designed. Specimens were solution annealed in argon and aged in stainless steel capsules that had been evacuated and backfilled with argon. Conventional tensile measurements were carried out using an Instron tensile machined equipped with a specimen furnace. Two Chromel-P vs Alumel thermocouples were used with a proportioning controller to keep the test specimen at 650 ± 3°C.

The microstructures developed during aging were studied by optical and electron transmission microscopy, extraction replication, and scanning microscopy. In addition, precipitated phases present in certain alloys were identified by x-ray diffraction using the Debye-Scherrer technique on precipitates extracted electrolytically with a methanol–10% HCl solution. Details of the microstructure examination have been reported,[5] and the alloying effects of titanium have been summarized.[6] We shall concern ourselves here with the detailed mechanical property response to heat treatment and a limited discussion of microstructures revealed in optical metallography.

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[5]C. E. Sessions, Influence of Titanium on the High-Temperature Deformation and Fracture of Some Nickel Based Alloys, ORNL-4561 (July 1970).

[6]C. E. Sessions, E. E. Stansbury, R. E. Gehlbach, and H. E. McCoy, Jr., "Influence of Titanium on the Strengthening of a Ni-Mo-Cr Alloy," pp. 626–630 in Second International Conference on the Strength of Metals and Alloys, Conf. Proc., Vol. II, The American Society for Metals, Metals Park, Ohio, 1970.

Because of the large number of variables included in this study and the attempt to measure the complicated and very subtle differences introduced by titanium additions, these experiments were statistically designed as a full factorial replication[7] including the variables in Table 2. The purpose of this experimental design was to provide better evaluation and separation of the influence of the specific variables on the tensile property response. Results of these statistical analyses have been discussed previously.[8]

RESULTS AND DISCUSSION

Typical results of the influence of preage thermal-mechanical treatment, aging time, and temperature on the tensile and creep properties at 650°C will be presented first. The phase identification will then be presented and discussed in light of the mechanical property changes. Strength and ductility values for all specimens tested are given in the Appendix.

Effect of Preaging on the Mechanical Properties

Figure 2 shows the yield strength (0.2% offset) and total elongation values for the three preage treatments investigated. Solution annealing 1 hr at 1260°C lowered the strength and increased the ductility at each titanium level as compared to a 1-hr solution anneal at 1177°C. Prestraining 10% at room temperature after a 1-hr 1177°C solution anneal doubled the high-temperature yield strength and reduced the ductility by one-third. The yield strength increased with titanium content for these unaged specimens, but the tensile ductility was not appreciably affected by the titanium content for these pretest treatments.

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[7]E. M. Bartee, Engineering Experimental Design Fundamentals, Prentice-Hall, Englewood Cliffs, N.J., 1968.

[8]C. S. Lever and C. E. Sessions, Fuels and Materials Development Quart. Progr. Rept. Sept. 30, 1969, ORNL-4480, pp. 275–279.

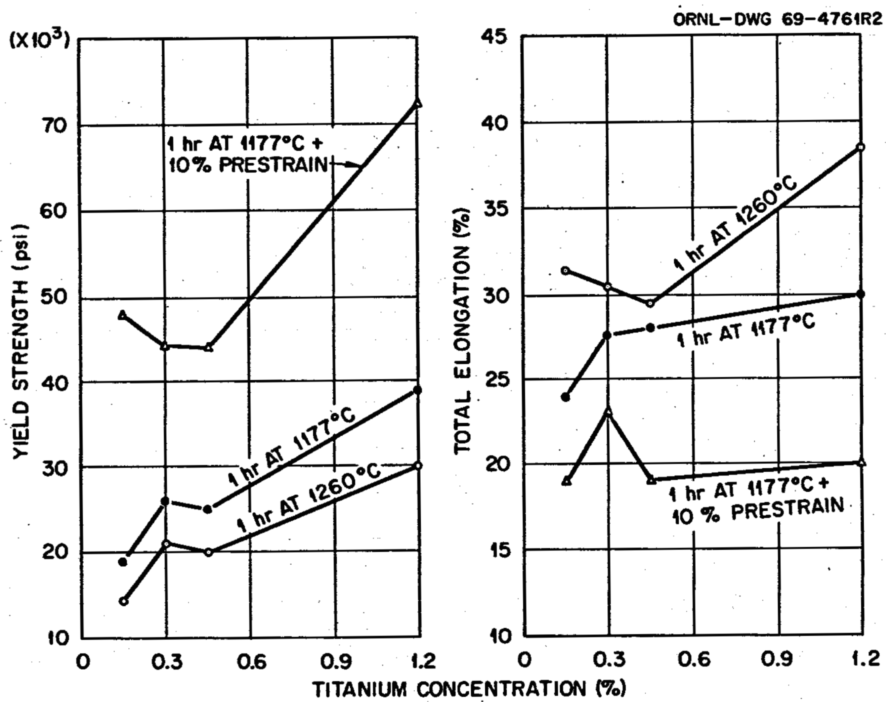

Fig. 2. Effect of Titanium Content and Preaging Treatments on the Strength and Ductility of Modified Hastelloy N at 650°C.

Effect of Aging on Ductility

Changes in tensile elongation with aging time are shown in Figs. 3 through 5 for both aging temperatures. Aging at either 650 or 760°C after a solution anneal at 1177°C enhanced the ductility at 650°C for the 1.2% Ti heat and decreased it for the 0.15% Ti heat. After a 1-hr solution anneal at 1260°C, the 1.2% Ti heat lost ductility on aging at either aging temperature to the same extent as did the lowest titanium level, 0.15%. Thus, the larger grain size or greater amount of solute in solid solution after the 1260°C treatment made the high-titanium heat susceptible to a detrimental aging reaction to which it was immune after the 1177°C anneal.

The effect of prestraining on the aging behavior is shown in Fig. 5. Samples were solution annealed 1 hr at 1177°C, prestrained 10% at room temperature, aged at either 650 or 760°C, and then tested at 650°C. The variation in tensile ductility with aging time and titanium was similar to that shown in Fig. 3 for samples not prestrained; however, the overall ductility was lower for the prestrained material.

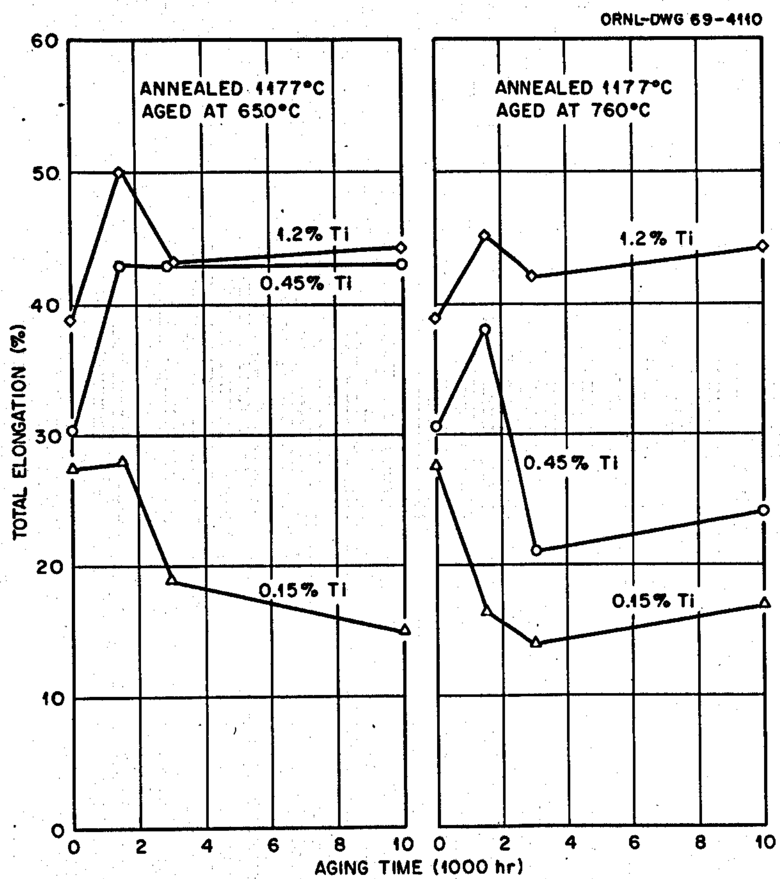

Fig. 3. Effect of Titanium Content and Aging Time on the Tensile Ductility at 650°C after a 1-hr Solution Anneal at 1177°C.

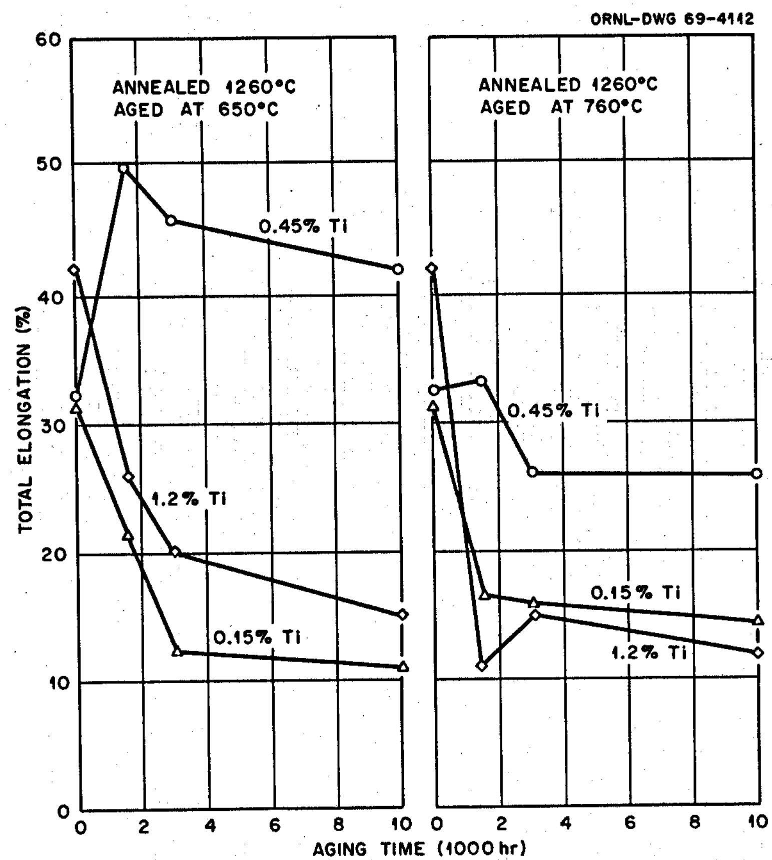

Fig. 4. Effect of Titanium Content and Aging Time on the Tensile Ductility at 650°C after a 1-hr Solution Anneal at 1260°C.

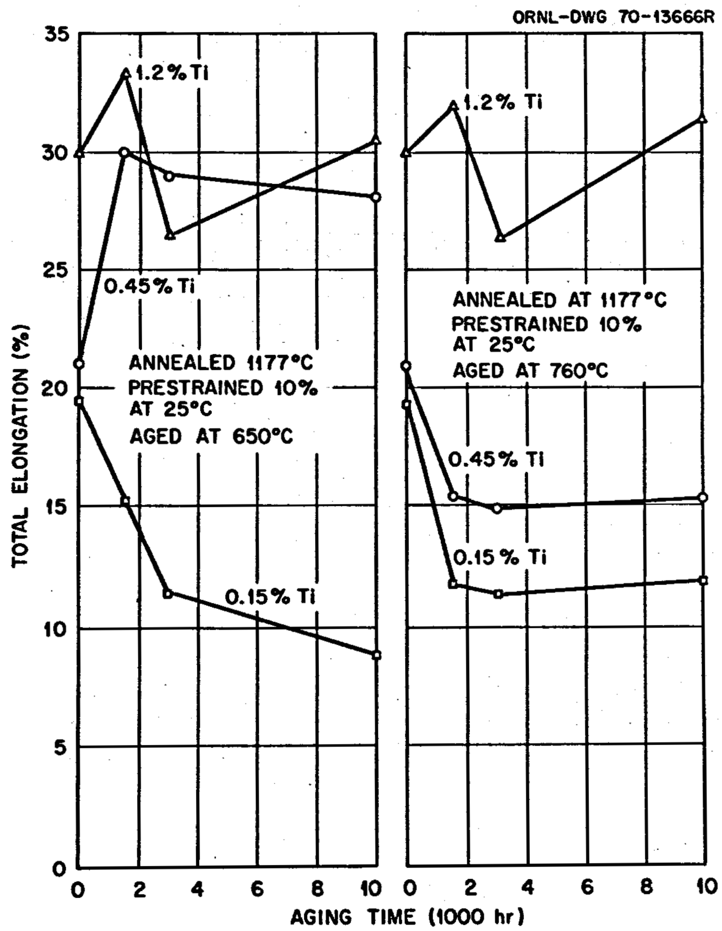

Fig. 5. Effect of Titanium Content and Aging Time on the Tensile Ductility at 650°C for Samples Solution Annealed 1 hr at 1177°C and Prestrained 10% at Room Temperature.

For aging at 650°C the intermediate (0.45%) and high (1.2%) titanium heats showed an increase in ductility with aging time. The low (0.15%) titanium heat exhibited a rapid loss in hot ductility with aging time at 650°C. Aging at 760°C affected the ductility of the low- and high-titanium heats the same as aging at 650°C, whereas the intermediate (0.45%) titanium heat showed a ductility deterioration with time at 760°C in contrast to the ductility enhancement found after aging at 650°C.

Effect of Aging on the Strength

The changes in the yield strength on aging are shown for only the 1-hr solution anneal at 1260°C in Fig. 6. For aging at both 650 and

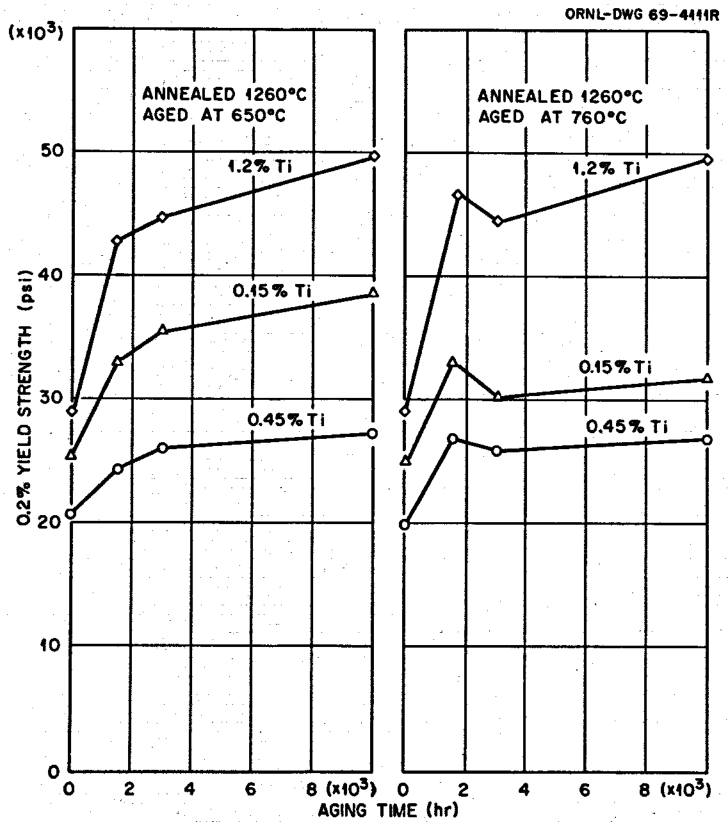

Fig. 6. Yield Strength Changes as a Function of Aging Time and Titanium Content for a 1-hr Solution Anneal at 1260°C.

760°C the largest strength increase occurred for the 1.2% Ti heat. The intermediate (0.45%) titanium heat exhibited the smallest change in yield strength for aging at either temperature. The lower yield strength values in this 0.45% Ti heat were most likely due to the lower carbon concentration for this heat (0.04% as compared to 0.07%).

The changes in yield strength for samples solution annealed at 1177°C before aging (not plotted) were comparable in most cases to those shown in Fig. 6. For the 1.2% Ti heat solution annealed at 1177°C and aged at 650°C, the yield strength at 650°C peaked at 46,000 psi after 1500 hr aging and exhibited classical overaging by decreasing to 37,500 psi after 10,000 hr. The increase in strength on aging the 1.2% Ti heat at 760°C after the 1177°C solution anneal was appreciably lower than for the 1260°C anneal. The yield strength peaked at 41,000 psi after 1500 hr and decreased to 37,000 psi after 10,000 hr.

The strength of the prestrained alloys did not increase on aging. The 1.2% Ti heat actually showed a yield strength decrease with aging time at both 650 and 760°C, and the intermediate- and low-titanium heats did not show any large change in yield strength after aging.

Stress-Strain Curves

The stress-strain curves for the two heats that showed the lowest and highest strengths are shown in Figs. 7 and 8. The engineering stress and strain are plotted up to the ultimate stress for the 0.45 and 1.2% Ti heats in Figs. 7 and 8, respectively. For the 0.45% Ti heat

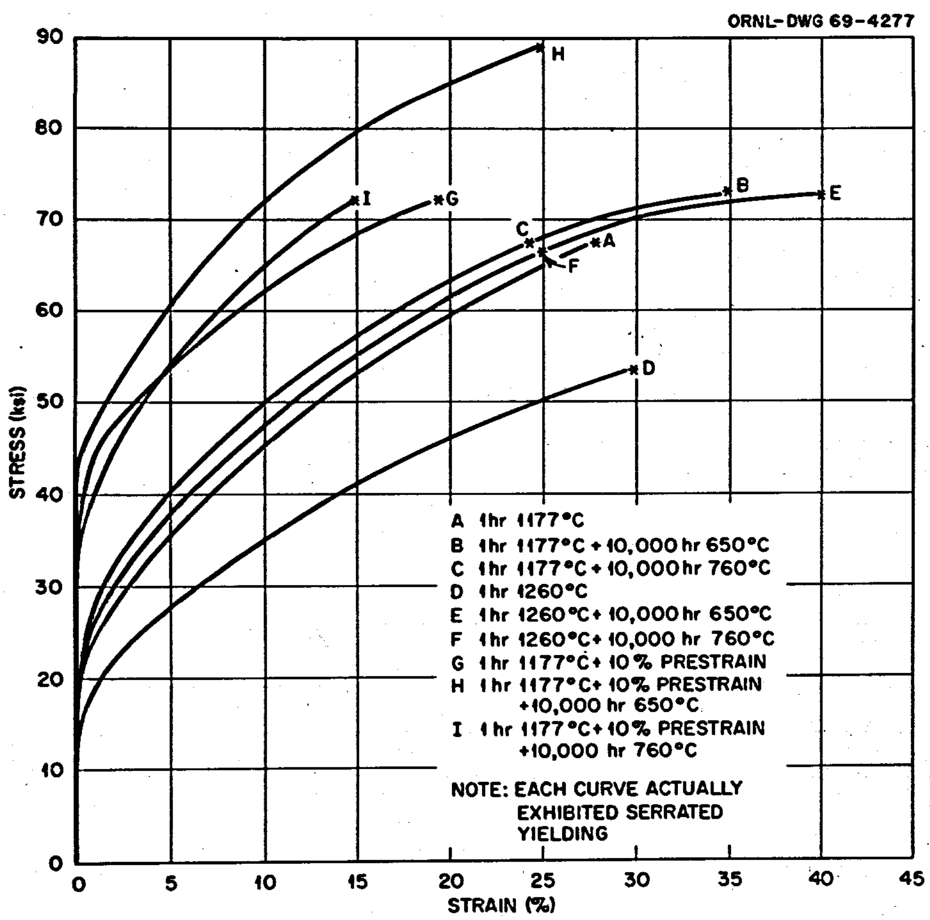

Fig. 7. Stress-Strain Curves for the 0.45% Ti Heat at 650°C for Various Heat Treatments.

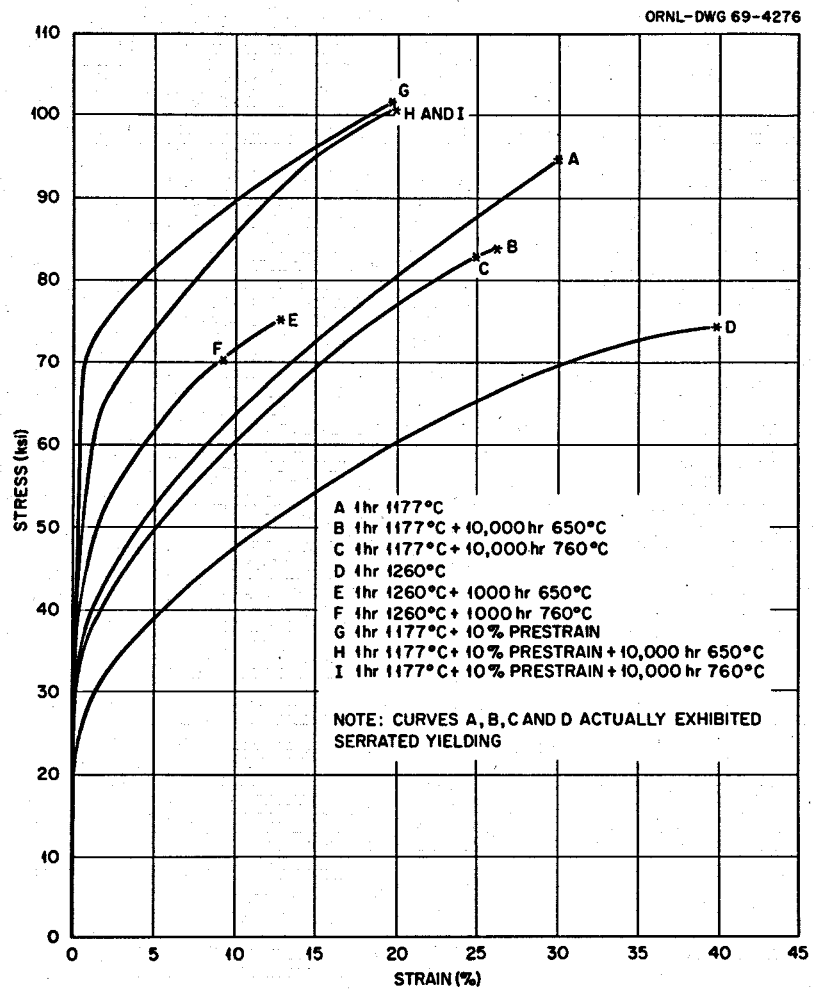

Fig. 8. Stress-Strain Curves for the 1.2% Ti Heat at 650°C for Various Heat Treatments.

ultimate tensile stresses from 50,000 to 90,000 psi and uniform elongations from approximately 15 to 40% were achieved by the thermal-mechanical treatments investigated. The 1.2% Ti heat (Fig. 8) exhibited a range of ultimate tensile strengths from 75,000 to 105,000 psi and uniform elongations from 8 to 40%.

Comparison of Measures of Ductility

Depending on the application, our criteria of usable ductility often vary. Thus, Fig. 9 compares three measures of the tensile ductility at 650°C for a 1-hr anneal at 1177°C and aging at 650°C. This particular treatment produced a high total elongation for the 0.45 and

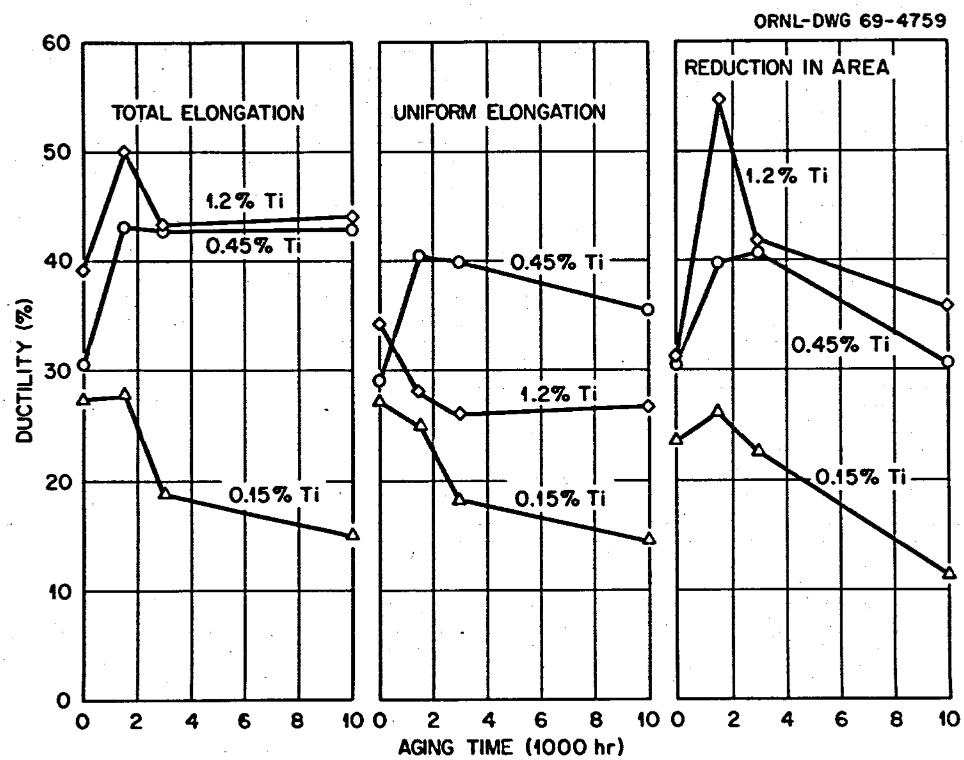

Fig. 9. Comparison of Measures of Tensile Ductility for a 1-hr Solution Anneal at 1177°C and Aging at 650°C.

1.2% Ti levels, as previously discussed, but the 1.2% Ti heat actually deteriorated in uniform elongation from 35 to approximately 26% with aging time. Therefore, to account for the high total elongation observed, the nonuniform elongation must have increased to compensate for the observed decrease in uniform strain. That is to say, for this heat, which exhibited the maximum ductility after this particular treatment, we found a change in the stress-strain relationship induced by precipitation. The change corresponded to an increase in yield strength, an increase in reduction in area, and an increase in total elongation; however, it corresponded to a decrease in ultimate tensile strength and uniform elongation.

Room-Temperature Tensile Properties

The effect of aging treatment on the room-temperature tensile properties is given in Table 3. The yield strengths and ductilities

Table 3. Room-Temperature Tensile Properties at 0.002/min Strain Rate for Titanium-Modified Hastelloy N After Various Heat Treatments

Alloy	Heat Treatment: Solution Anneal (°C)	Heat Treatment: Aging Time (hr)	Heat Treatment: Aging Temperature (°C)	Strength, psi: Yield	Strength, psi: Ultimate Tensile	Elongation, %: Uniform	Elongation, %: Total	Reduction in Area (%)	Specimen
				$\times 10^3$	$\times 10^3$				
466-535	1177	0		42.2	119.5	61.0	64.2	66.4	5292
	1177	1500	650	61.5	135.9	40.5	40.8	29.5	5997
	1177	1500	760	53.3	139.1	43.5	46.0	47.8	5998
466-541	1177	0		38.2	112.7	69.6	73.3	57.1	6093
	1177	1500	650	54.3	125.6	52.5	53.7	41.9	6099
	1177	1500	760	42.4	110.5	50.1	52.3	40.6	6100
466-548	1177	0		37.4	113.3	68.8	72.2	62.6	6270
	1177	1500	650	47.6	124.1	57.4	59.0	41.8	6277
	1177	1500	760	48.1	129.6	53.6	55.4	50.9	6278
467-548	1260	0		35.7	104.9	67.0	68.2	18.2	6344
	1177	0		47.6	134.9	62.5	63.8	46.2	11416
	1177	1500	650	61.3	145.6	50.5	52.0	41.2	6321
	1177	1500	760	50.1	121.3	47.8	49.0	43.6	6322
	1260	1500	760	59.6	122.6	41.6	41.7	36.7	6989

are similar for the various heats as solution annealed at 1177°C. Aging 1500 hr at 650°C increases yield strength more than at 760°C. However, the ductility decrease on aging was approximately the same for the two aging temperatures. These trends in strength and ductility were found for heats 466-535, 466-541, and 466-548. For heat 467-548 (1.2% Ti) we measured the effect of using a higher solution anneal temperature (1260°C) on the aging response. Aging 1500 hr at 760°C increased the room-temperature yield strength from 36,000 to 60,000 psi and decreased the creep elongation from 68 to 42% strain. This change in the room-temperature tensile behavior for the 1260°C solution anneal was larger than that found after annealing at 1177°C. However, the microstructural differences found for this high-titanium (1.2% Ti, 467-548) heat are significant and have been discussed previously.[9] The point is that, regardless of the differences in microstructure, the room-temperature properties are similar.

Creep Properties After 3000 hr of Aging

The effect of titanium content on the creep properties of solution-annealed and aged alloys was measured for 36 specimens. Each test was conducted at 650°C and under 40,000 psi stress so that the rupture lives and creep rates indicate the effects of aging. Creep ductility in these tests reflected the influence of strain rate, which varied with the strength of the alloy under this constant-load creep test. The ductilities, rupture lives, and creep rates are given in Figs. 10 and 11 for some of the treatments used.

Effect of Preage Treatments

For each alloy, except the 1.2% Ti heat, the lowest rupture life and highest elongation corresponded to samples solution annealed at the higher temperature (i.e., 1 hr at 1260°C). In most instances the lowest secondary creep rate, maximum rupture life, and minimum ductility corresponded to the samples prestrained 10% at room temperature.

[9] C. E. Sessions, Influence of Titanium on the High-Temperature Deformation and Fracture of Some Nickel Based Alloys, ORNL-4561 (July 1970).

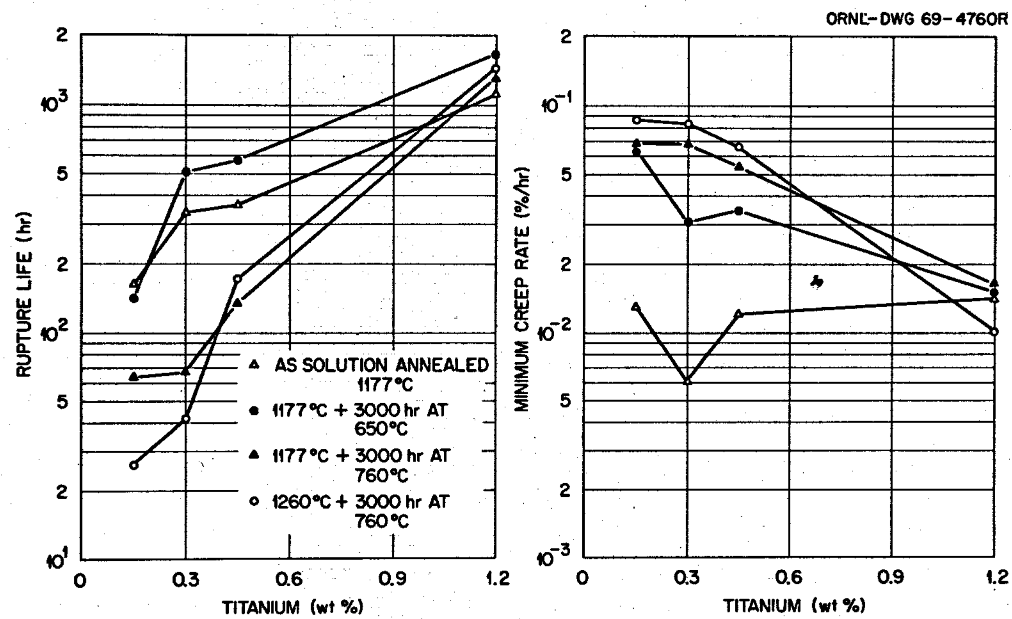

Fig. 10. Variation in Rupture Life and Creep Rate with Titanium Content and Heat Treatment at 650°C and 40,000 psi Stress.

Effect of Aging Temperature

In most cases (10 of 12 tests) the rupture life after aging 3000 hr at 760°C was reduced by from 20 to 90% as compared with the unaged samples. Correspondingly, the creep rates were significantly increased by a factor of 3 to 10, depending on the alloy and heat treatment. The creep ductility of the samples aged 3000 hr at 760°C ranged from 6 to 37%, with the higher values corresponding to the higher titanium level (1.2%).

Less property deterioration was observed for aging 3000 hr at 650°C. In fact, several heats, notably the 0.45% Ti heat, exhibited significantly enhanced rupture lives and fracture strains even though the creep rate was increased appreciably by the aging treatment. In general, the prestrained samples and those with the larger grain size (i.e., solution annealed at 1260°C) showed the largest reduction in rupture life after aging at 760°C.

Effect of Titanium

Generally, the creep-rupture lives and ductilities increased with increasing titanium content for both the solution-annealed and the aged samples. The creep rates decreased with increasing titanium content.

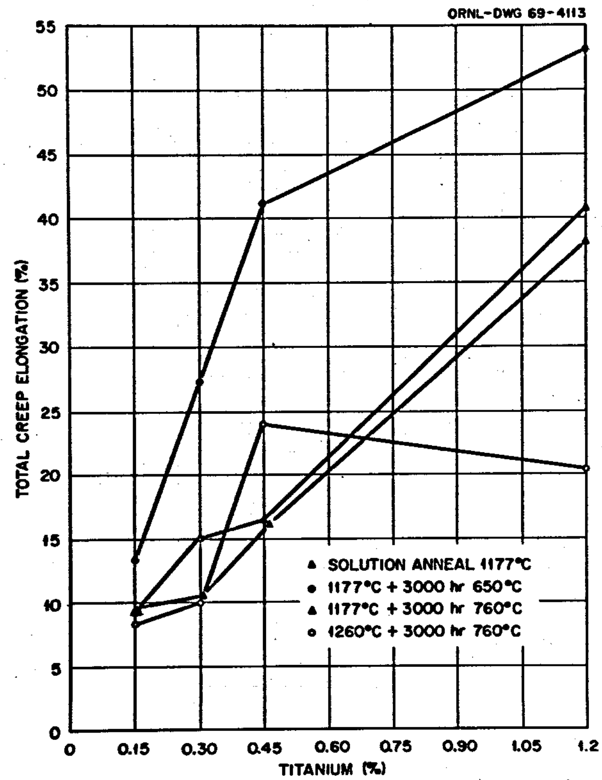

Fig. 11. Effect of Titanium Content and Heat Treatment on the 40,000-psi Creep Elongation at 650°C.

The titanium content influenced the magnitude of the aging effect. For most treatments, the alloy with the highest titanium content exhibited the smallest property deterioration after aging. However, solution annealing for 1 hr at 1260°C led to creep property changes on aging that were quite significant for all levels of titanium.

Comparison of Tensile and Creep Results

Table 4 lists the heat treatments for each heat of material that corresponded to the maximum and minimum property value found in creep and tensile testing of the samples aged 3000 hr. Since three treatments were included in this study [i.e., solution anneals of (1) 1 hr at 1177°C (2) 1260°C, and (3) 1 hr at 1177°C plus 10% prestrain at room temperature] the treatment that is not listed under either minimum or maximum in Table 4 would result in properties intermediate between the minimum and maximum values.

Table 4. Thermal-Mechanical Treatments That Produced the Maximum and Minimum Creep and Tensile Properties at 650°C for Various Heats of Titanium-Modified Hastelloy N

Condition	Property	Heat Treatment[a] to Give Extreme Property for Each Titanium Content in Percent			
		0.15	0.27	0.45	1.2
	Tensile Results[b]				
Unaged	Maximum yield strength	cw	cw	cw	cw
	Minimum yield strength	1260	1260	1260	1260
	Maximum ultimate strength	cw	cw	cw	cw
	Minimum ultimate strength	1260	1260	1260	1260
	Maximum total elongation	1260	1260	1260	1260
	Minimum total elongation	cw	cw	cw	cw
Aged	Maximum yield strength	cw/650	cw/650	cw/650	cw/650
	Minimum yield strength	1260/760	1177/760	1260/760	1177/760
	Maximum ultimate strength	1177/760	cw/650	cw/650	cw/650
	Minimum ultimate strength	1260/650	1260/650	1260/760	1260/760
	Maximum total elongation	1177/760	1177/650	1177/650	1177/650, 760
	Minimum total elongation	cw/650	1260/650	cw/760	1260/760
	Creep Results[c]				
Unaged	Maximum rupture life	cw	1177	cw	1260
	Minimum rupture life	1260	1260	1260	1177
	Maximum secondary creep rate	1177	1260	1177	1177
	Minimum secondary creep rate	cw	cw	cw	1260
	Maximum creep elongation	1260	1260	1260	1177
	Minimum creep elongation	cw	cw	cw	1260
Aged	Maximum rupture life	1260/650	cw/650	cw/650	cw/650
	Minimum rupture life	1260/760	1260/760	cw/760	1177/760
	Maximum secondary creep rate	1260/760	1260/760	1260/650	1177/760
	Minimum secondary creep rate	1177/760	cw/650	cw/650	cw/650
	Maximum creep elongation	cw/650	1177/650	1177/650	1177/650
	Minimum creep elongation	cw/760	cw/760	cw/760	1260/760

[a]The preage heat treatments are 1177, 1 hr at 1177°C; 1260, 1 hr at 1260°C; cw, 1 hr at 1177°C, then cold worked. For aged specimens this is followed by the aging temperature in °C. Aging times were 3000 hr.

[b]Tensile properties measured at 650°C and 0.002/min strain rate.

[c]Creep properties measured at 650°C and 40,000-psi stress.

Results of Unaged Samples

In each case the maximum yield strength and maximum ultimate tensile strength at 650°C were found for the 10% prestrained samples. The minimum yield and ultimate strengths were found for the 1260°C treatment. The minimum and maximum tensile ductilities corresponded to the strongest and weakest alloys in tensile tests, with one exception, the 0.45% Ti heat.

In creep the maximum resistance to rupture (maximum rupture life) corresponded to the prestrained sample in only two of the four heats, but prestrained samples crept at a lower rate in three of the four cases. Notably, the 1.2% Ti heat exhibited the minimum and maximum creep resistance (as measured by the creep rate) for the 1177 and 1260°C anneals, respectively. This indicates that creep of this higher titanium heat was affected more by solution annealing treatment than by 10% cold working. The minimum and maximum creep ductilities of the other alloys corresponded to the prestrained and to the 1260°C annealed material, respectively.

Results for Aged Samples

The prestrained material aged 3000 hr at 650°C exhibited the maximum yield strength for each heat. The minimum yield strength corresponded to either 1177 or 1260°C samples aged 3000 hr at 760°C, depending on the alloy. The maximum ultimate strength was for prestrained samples aged 3000 hr at 650°C with one exception, the 0.15% Ti heat. The lowest ultimate strength (UTS) was found for the 1260°C treatment; at lower titanium level an aging treatment of 3000 hr at 650°C caused the lowest UTS, and at the higher titanium levels aging 3000 hr at 760°C caused the lowest UTS. The rupture life was a maximum for prestrained samples aged at 650°C, and the treatment that produced the minimum rupture life differed for each alloy. An interesting point is that the 0.15% Ti heat gave both the maximum and the minimum rupture life for the larger grain size 1260°C treatment, depending on whether it was aged at 650 or 760°C.

Phases Identified in Aged Alloys

Table 5 lists the phases identified by Debye-Scherrer x-ray diffraction analysis of electrolytically extracted precipitate particles after 3000 hr of aging at the indicated temperatures. In the 0.15% Ti heat a combination of MC and M_2C type phases was present, but at 760°C only the M_2C carbide was found. At the intermediate titanium level (0.45%) only MC was present at 650°C, while at 760°C a weak trace of MC and strong lines for M_2C were found. At the highest titanium level investigated (1.20%) the MC structure was present after aging at either temperature. Thus, we conclude that the higher titanium content has stabilized the MC-type carbide at higher aging temperatures.

Table 5. Phases Identified in Ni-Mo-Cr-Ti Alloys by Debye-Scherrer Analysis

Heat	Heat Treatment	Titanium Content (%)	Phases Present
466-535	1 hr 1177°C + 3000 hr at 650°C	0.15	MC[a] + M_2C
	1 hr 1177°C + 3000 hr at 760°C	0.15	M_2C
466-548	As annealed 1177°C		b
	As annealed 1260°C		b
	1 hr 1177°C + 3000 hr at 650°C	0.45	MC
	1 hr 1177°C + 3000 hr at 760°C	0.45	MC[a] + M_2C
467-548	As annealed 1177°C	1.20	MC
	As annealed 1260°C	1.20	MC
	1 hr 1177°C + 3000 hr at 650°C	1.20	MC
	1 hr 1177°C + 3000 hr at 760°C	1.20	MC
	1 hr 1260°C + 3000 hr at 650°C	1.20	MC
	1 hr 1260°C + 3000 hr at 760°C	1.20	MC

[a]Minor amounts present.

[b]Too little precipitate to extract electrolytically.

The carbides found in these alloys were complex compounds involving chromium, molybdenum, and titanium, and in several cases we found two different lattice parameters that generally fit the hexagonal M_2C-type structure with the same c/a ratio of 1.63. In the case of the MC structure the lattice constant did not correspond to that of TiC (4.33 Å) but rather was significantly lower, which indicates that the smaller molybdenum or chromium atoms were substituting for titanium in the face-centered cubic MC structure.

Metallography

Initial Microstructures

The effect of solution annealing at 1177 or 1260°C on the grain size for three of the four alloys is shown in Fig. 12. Annealing 1 hr at 1177°C did not produce a single-phase alloy for most of these heats. In the alloys containing 0.15 and 1.2% Ti after an 1177°C anneal, large stringers of second-phase particles were evident. The 0.27 and 0.45% Ti heats were virtually single phase after the 1177°C heat treatment. After 1 hr at 1260°C the grain size of each heat was 2 to 4 times that found after the 1177°C anneal. Also, fewer precipitate particles and stringers were present after 1 hr at 1260°C.

Microstructures Developed During Aging

Typical structures obtained by interference contrast microscopy are shown in Fig. 13. This figure compares the precipitate distributions for samples solution annealed at 1177°C and aged 10,000 hr at either 650 or 760°C. These photographs were taken near the fracture sites of tensile samples tested at 650°C after aging at the indicated temperatures. After aging at 650°C the precipitate was generally finer, and the concentration of precipitate near grain boundaries appeared to be fairly heavy, particularly for the low-titanium alloy. After aging 10,000 hr at 760°C the precipitate was rather coarse, and the concentration of precipitate near the grain boundary was lower than that found at 650°C. The precipitate within the grain boundary was also coarse, as shown by the micrographs of the 0.27 and 0.45% Ti heats in Fig. 13. Little precipitation occurred in the 1.2% Ti heat, since the stringers evident in Fig. 13 were actually present to some extent even before aging; this was shown for the as-solution-annealed condition in Fig. 12.

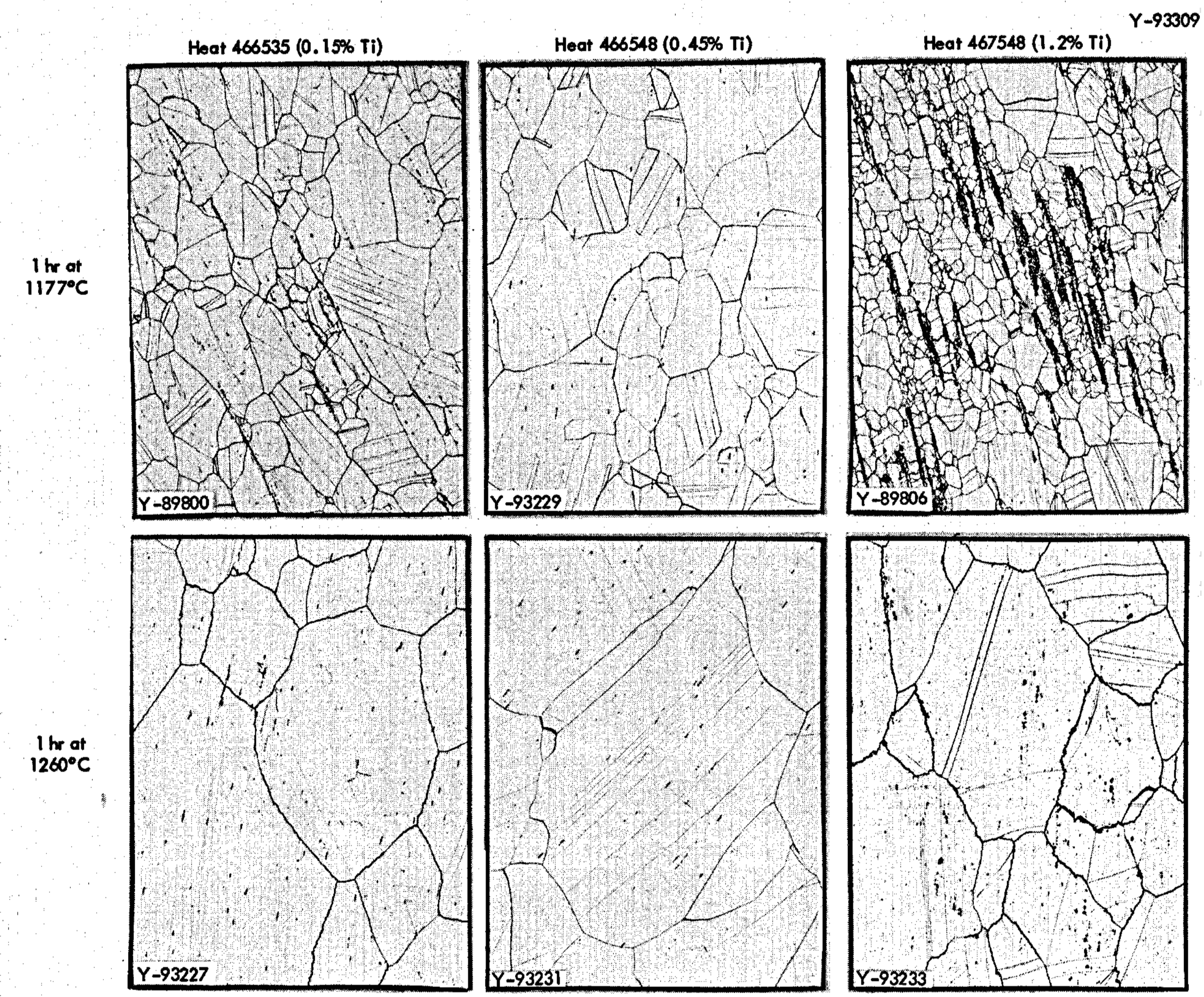

Fig. 12. Microstructures of Titanium-Modified Alloys in the Solution-Annealed Condition. 100X. Reduced 38%.

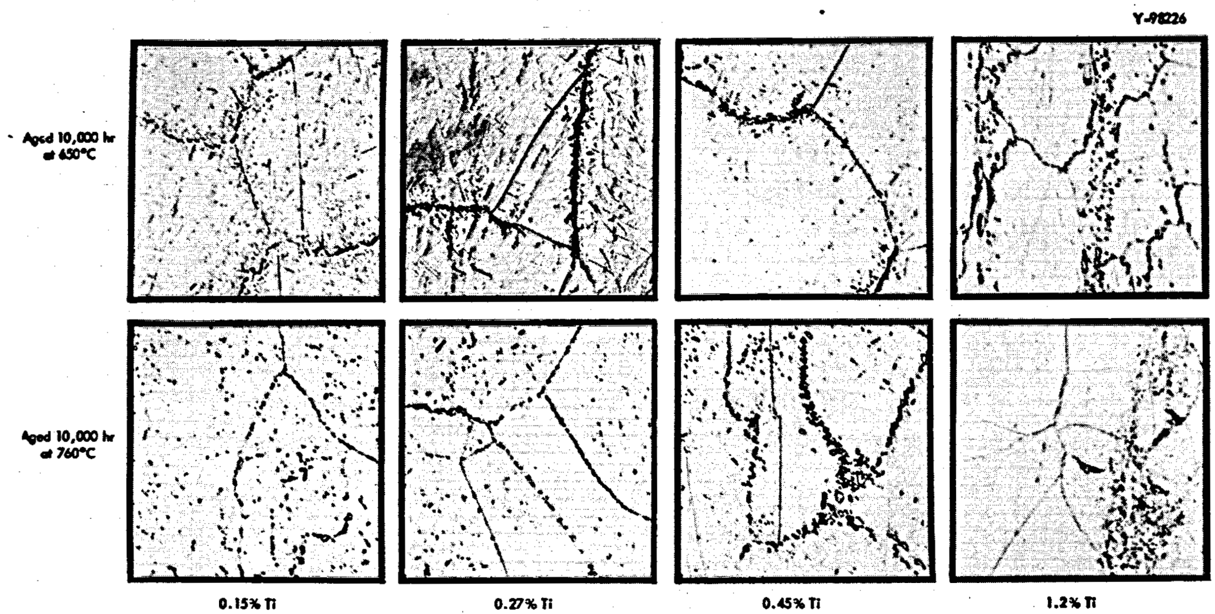

Fig. 13. Microstructures Developed in Ni–12% Mo–7% Cr–0.07% C Alloys Modified with Titanium, Solution Annealed 1 hr at 1177°C, and Aged 10,000 hr. 1000×. Reduced 65%.

Microstructures developed during aging 10,000 hr after a 1-hr solution anneal at 1260°C are shown in Fig. 14. These structures are typical also of those developed in 1500 and 3000 hr of aging and indicate a large variation in optical microstructure with titanium content. Alloys with lower titanium contents exhibited relatively coarse precipitate distributions. The 1.2% Ti alloy showed bands of precipitate stringers after aging at 650°C. Apparently, the stringers present initially along the hot-working direction of the rod stock were not completely eliminated even though the material was solution annealed at 1260°C. During subsequent aging these residual second-phase particles coarsened. Thus, the longitudinal cross section of these aged alloys exhibited a banded structure, which is similar to that found in the cold-worked rod before solution annealing. In the 0.45 and 1.2% Ti heats during aging at 650°C, however, some additional precipitation occurred on both subgrain and grain boundaries.

The microstructures developed at 760°C after a 1260°C solution anneal, shown in Fig. 14, were actually quite similar to those found after the 1177°C anneal and 760°C age (Fig. 13). For the 1.2% Ti heat, Y-98225

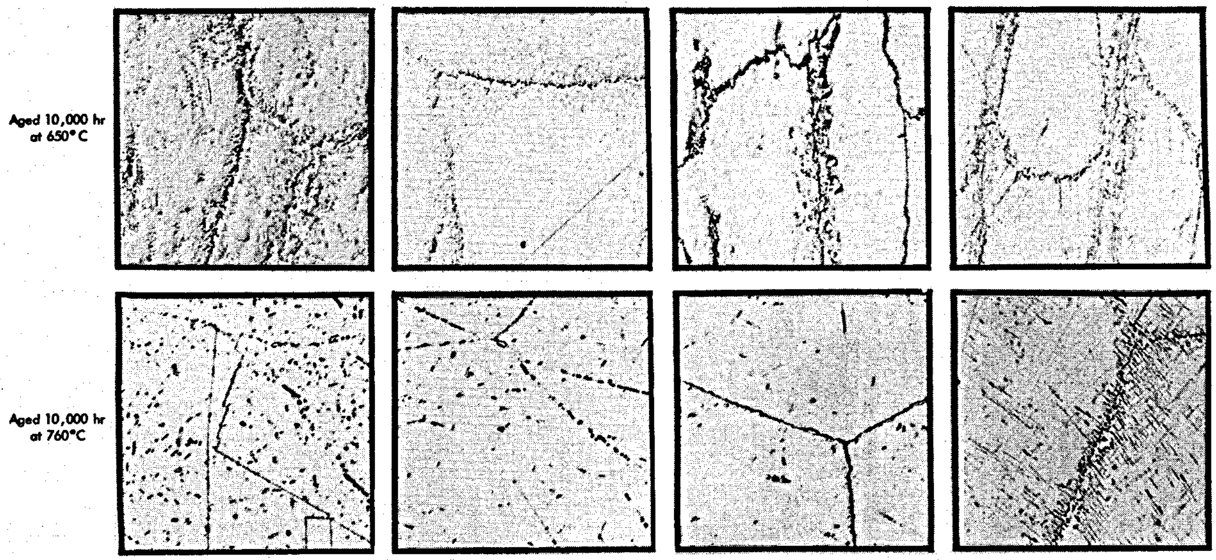

Fig. 14. Microstructures Developed in Ni–12% Mo–7% Cr–0.07% C Alloys Modified with Titanium, Solution Annealed 1 hr at 1260°C, and Aged 10,000 hr. 1000×. Reduced 65%.

however, the structure developed after a 1260°C anneal was much different from that developed in samples annealed at 1177°C. The precipitation appeared to be a Widmanstätten distribution; however, the concentration of precipitate platelets shown in Fig. 14 varied within a given grain. Generally, a higher concentration of "platelets" was found near grain boundaries and near clusters of primary precipitate particles. This "platelet"-type precipitate, which forms primarily after the 1260°C solution anneal and 760°C age, has been shown[10] by transmission electron microscopy to result from precipitation on stacking faults.

Structures Developed in Prestrained Material

The microstructures developed during aging 10,000 hr at 650 and 760°C after annealing 1 hr at 1177°C and prestraining 10% at room temperature are shown in Fig. 15 for the 0.15 and 1.2% Ti heats. Prior prestraining greatly reduced the coarseness of the precipitate in both

[10]C. E. Sessions, Influence of Titanium on the High-Temperature Deformation and Fracture of Some Nickel Based Alloys, ORNL-4561 (July 1970).

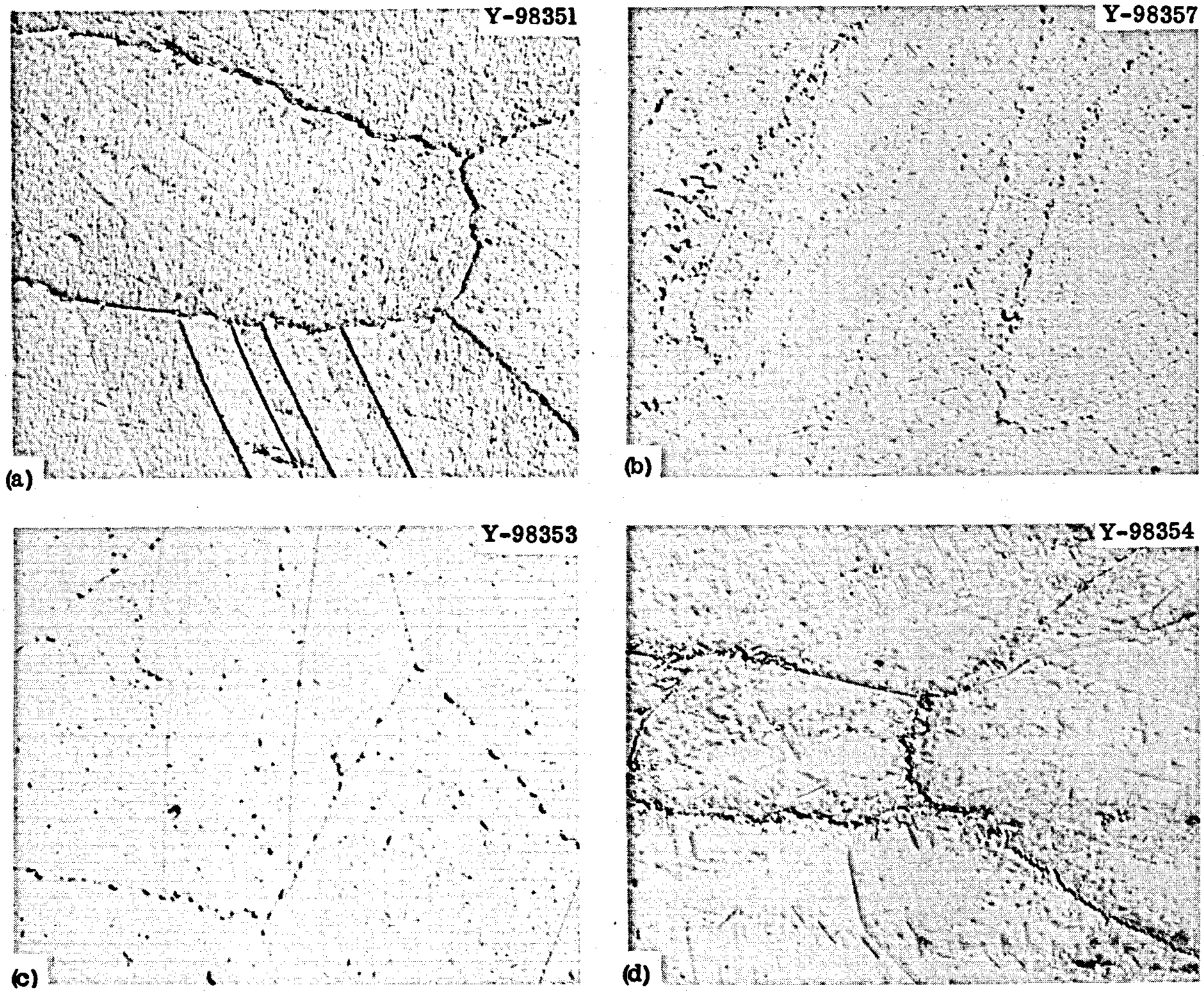

Fig. 15. Microstructures Developed in Ni–12% Mo–7% Cr–0.07% C Alloys Modified with Titanium, Solution Annealed 1 hr at 1177°C, Prestrained 10% at Room Temperature, and Then Aged 10,000 hr. 1000×. (a) 0.15% Ti, aged at 650°C, (b) 1.2% Ti, aged at 650°C, (c) 0.15% Ti, aged at 760°C, and (d) 1.2% Ti, aged at 760°C. Reduced 18%.

alloys at 650°C. In samples aged at 760°C, prestraining 10% produced very little effect on the structure of the 0.15% Ti heat, but the 1.2% Ti heat showed much more precipitate in the prestrained samples (Fig. 15) than in samples aged without prestraining (Fig. 13). In the prestrained 1.2% Ti alloy, both a fine random matrix precipitate and a crystallographic platelet-type precipitate are shown in Fig. 15. Evidently prestraining has enhanced the precipitation of this platelet-type precipitate (the platelets are actually a precipitate on stacking faults,[11] as mentioned previously). This apparent enhancement in formation of stacking fault precipitate produced by straining this 1.2% Ti alloy before aging is consistent with the effect of prior straining on formation of stacking fault precipitates in niobium-doped stainless steels as reported by Silcock and Tunstall[11] and Naybour.[12] However, some differences in the effect of strain on the tendency to precipitate on stacking faults in steels and in this alloy are discussed in another paper.[13]

Fracture of Samples Tested in Creep

The fracture appearance of creep samples tested after solution annealing at 1177°C and aging 3000 hr at 760°C is compared in Fig. 16. The rupture lives (64, 135, and 1335 hr) and the total elongation values (9.5, 16.7, and 36.8%) of samples with these fractures increased progressively with increasing titanium content. The fracture appearance confirms the measured ductility trends, since the fracture mode exhibits a transition with increasing titanium content from intergranular fracture for the 0.15% Ti heat [Fig. 16(a)] to transgranular for the 1.2% Ti heat [Fig. 16(c)]. The intermediate titanium content (0.45% Ti) also failed in a predominantly transgranular fracture mode; however, the tips of the fractured grains [Fig. 16(b)] show regions that appear to have recrystallized during the creep test. Nevertheless, the creep elongation of this 0.45% Ti heat after the 760°C heat treatment was only 16.7%, which indicates that the apparent recrystallization that occurred during creep testing at 650°C did not significantly enhance the ductility. That is, other heat treatments of this 0.45% Ti heat yielded post-age creep elongation values appreciably greater than 16.7% even though they showed no evidence of recrystallization.

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[11] J. M. Silcock and W. J. Tunstall, Phil. Mag. 10, 360–389 (1964).

[12] R. D. Naybour, Acta Met. 13, 1197–1207 (1965).

[13] C. E. Sessions and R. E. Gehlbach, "Effects of Heat Treatment and Straining on Formation of Stacking Fault Precipitates in Hastelloy N," (in preparation).

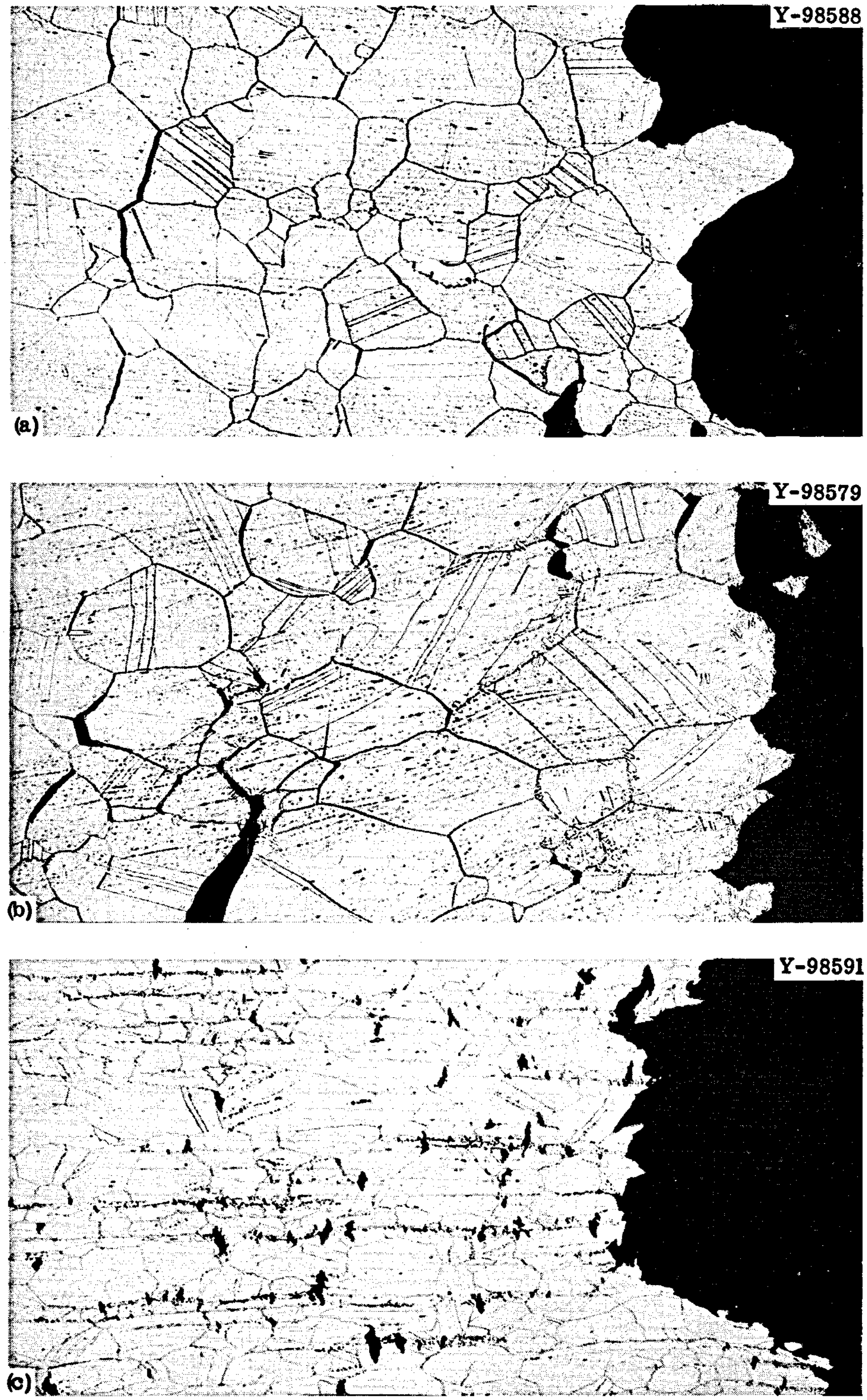

Fig. 16. Creep Fractures at 650°C and 40,000 psi for Heats of Titanium-Modified Hastelloy N After a 1 hr at 1177°C Solution Anneal and 3000 hr Age at 760°C. 100X. (a) 0.15% Ti, (b) 0.45% Ti, and (c) 1.2% Ti. Reduced 18%.

In Fig. 16 it is also apparent that the grain size is smallest for the 1.2% Ti heat. Although the fracture was primarily transgranular in that heat, many intergranular cracks have opened up along the gage length in contrast to the alloys with lower titanium concentrations. Apparently, the grain boundary structure in the higher titanium heat inhibits the propagation of intergranular cracks even though the nucleation of cracks in this alloy is apparently quite easy.

Another comparison of fractures for creep tests after various aging treatments is given in Fig. 17 for three heats with different titanium contents. The total creep elongations for the samples shown in Fig. 17(a), (b), and (c) were 8.3, 42.1, and 24.5%, respectively. Thus, the alloy with the intermediate titanium content [0.45% Ti, Fig. 17(b)] was the most ductile of the three alloys, in contrast to the results in Fig. 16. The grain size and the amount of precipitation evident at this magnification is greater in Fig. 17(a) and (c) than in Fig. 17(b). As in Fig. 16, the incidence of intergranular cracks is higher in the most ductile alloy, Fig. 17(b), than in the other materials. The fracture mode is a mixture of transgranular and intergranular fracture in Fig. 17(a) and (c) and is transgranular in Fig. 17(b).

An influence of aging time at 760°C on the post-age fracture appearance for the 0.45% Ti heat is found by comparing Fig. 17(b) (1500 hr at 760°C) with Fig. 16(b) (3000 hr at 760°C). These two samples from the same heat were both solution annealed at 1177°C, aged, and then creep tested at 40,000 psi and 650°C. The creep properties (594 hr rupture life and 42.1% elongation) for the sample aged 1500 hr at 760°C [Fig. 17(b)] were much better than the properties (135 hr rupture life and 16.7% elongation) obtained after aging 3000 hr at 760°C. Similarly, the fracture appearance is more transgranular and the amount of grain deformation is much greater for the samples aged only 1500 hr [Fig. 17(b)]. Although a significant difference was found in creep properties after aging 1500 and 3000 hr at 760°C, the optical microscopy did not reveal any large differences in structure. This effect of aging time between 1500 and 3000 hr at 760°C is rather surprising since normally aging follows a logarithmic rate, which would give small changes for a factor of 2 difference in aging time, but it should change significantly

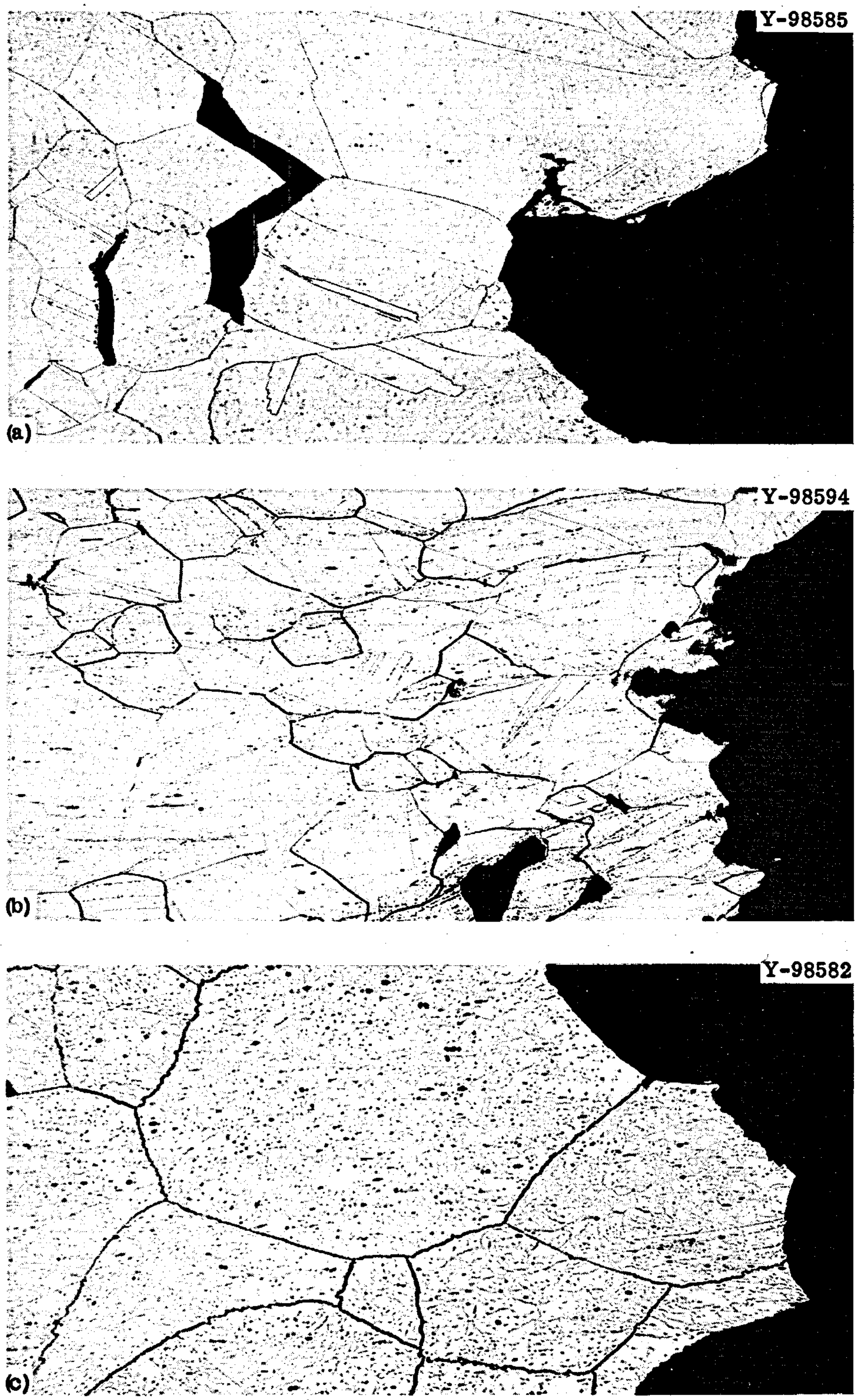

Fig. 17. Creep Fractures for Tests at 650°C and 40,000 psi in Titanium-Modified Hastelloy N for Various Heat Treatments. 100×. (a) 0.15% Ti solution annealed 1 hr at 1260°C and aged 3000 hr at 760°C, (b) 0.45% Ti solution annealed 1 hr at 1177°C and aged 1500 hr at 760°C, and (c) 1.2% Ti solution annealed 1 hr at 1260°C and aged 3000 hr at 650°C. Reduced 18%.

for a tenfold difference in aging time. Nevertheless, this observed large difference in creep behavior between 1500 and 3000 hr of aging was confirmed by the tensile ductility for this 0.45% Ti heat, as shown in Fig. 3, p. 7. The 650°C tensile ductility decreased from about 38 to 21% between 1500 and 3000 hr of aging at 760°C. Thus in the 0.45% Ti heat, evidently either the type of precipitate present or its amount and distribution differed significantly between 1500 and 3000 hr of exposure at a temperature of 760°C (phases were not identified in the samples aged 1500 hr).

SUMMARY AND CONCLUSIONS

We have investigated the effects of titanium content, solution annealing temperature, aging temperature, and aging time on the mechanical properties of modified Hastelloy N at 650°C. In the solution-annealed condition the 0.002/min yield strength increased and the ductility was approximately constant with increasing titanium content from 0.15 to 1.2%. Aging and prestraining both generally increased the yield strength, but the changes in ductility during aging depended on (1) alloy content, (2) annealing temperature, and (3) aging temperature.

Phase identification of electrolytically extraced precipitates indicated two types of carbides present. An M_2C-type was favored in heats with low titanium contents or alloys aged at the higher temperature, 760°C. An MC-type carbide was favored at higher titanium contents or for alloys with a given titanium concentration aged at the lower temperature, 650°C. Thus, the titanium content determined the carbide type, whether M_2C or MC. The carbide type, its stability, and its morphology subsequently influenced the deformation and fracture in creep and tensile tests. Correlations of microstructures and high-temperature mechanical properties as influenced by the carbon and titanium contents have previously been investigated.[14]

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[14]C. E. Sessions and E. E. Stansbury, "Correlation of Structures and High-Temperature Properties in a Ti-Modified Ni-Mo-Cr Alloy," submitted to Metallurgical Transactions.

Increasing the titanium content resulted in improved post-age properties. Apparently, the M_2C carbide precipitate was more detrimental to ductility than was the MC carbide precipitate. However, certain distributions of MC carbides were also detrimental to the ductility. This was indicated by the 1.2% Ti alloy, which lost ductility and increased its strength on aging after the 1260°C solution anneal. Since only MC carbides were identified in this alloy, the strengthening and embrittlement are attributed to the precipitation on stacking faults. Correlations of microstructures and mechanical properties for stacking fault precipitates have also been made for stainless steels,[15,16] Inconel,[17] and Hastelloy N.[18,19]

Several additional points should be made from these results. The alloy with the lowest carbon content (i.e., heat 466-548) generally had the lowest strength and highest ductility. This low carbon alloy, however, did show a pronounced difference in aging response at 650 and 760°C, which we attribute to the fact that different carbides were precipitated at the two aging temperatures (i.e., MC at 650°C and M_2C at 760°C).

Overaging could be expected in most of these alloys before 3000 hr at 760°C, as indicated from the peak in yield strength at 1500 hr in Fig. 6, p. 9.

The comparison of creep and tensile properties indicated that the higher solution anneal, which increases the grain size and amount of

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[15]J. M. Silcock and W. J. Tunstall, Phil. Mag. 10, 360–389 (1964).

[16]R. D. Naybour, Acta Met. 13, 1197–1207 (1965).

[17]P. S. Kotval, Trans. Met. Soc. AIME 242, 1651–1656 (1968).

[18]C. E. Sessions, Influence of Titanium on the High-Temperature Deformation and Fracture of Some Nickel Based Alloys, ORNL-4561 (July 1970).

[19]C. E. Sessions, E. E. Stansbury, R. E. Gehlbach, and H. E. McCoy, Jr., "Influence of Titanium on the Strengthening of a Ni-Mo-Cr Alloy," pp. 626–630 in Second International Conference on the Strength of Metals and Alloys, Conf. Proc., Vol. II, The American Society for Metals, Metals Park, Ohio, 1970.

solute in solution, actually produces the lowest yield strength. The weakest alloys in creep (as measured by the highest secondary creep rate) were found for the 1177°C anneal, which produces a smaller grain size relative to the structure produced at 1260°C. This dependence of low yield strength on large grain size and of low creep resistance on small grain size is typical of the effects of grain size on the strength of alloys at elevated test temperatures. At such temperatures, intergranular fracture rather than transgranular fracture is favored.

Of the treatments investigated, cold working 10% at room temperature after a 1-hr solution anneal at 1177°C had the greatest effect on raising the strength and lowering the ductility, both before and after aging. An exception to this was found for the 1.2% Ti heat, which exhibited the lowest post-age tensile and creep ductility after a 1260°C solution anneal and 760°C age. Samples that showed this large ductility loss had a high concentration of the platelet-type stacking fault precipitate.

The heat treatments investigated produced variations in the high-temperature ductility between 5 and 50%, yet the significant point is that most of these treatments change the room-temperature properties only slightly. The irony, however, is that these subtle microstructural modifications caused by titanium additions have affected the creep behavior of this material even more than we found from this systematic look at tensile behavior. However, testing in creep does not give a constant rate of deformation, and the strain rate sensitivity of the ductility would have made it more difficult to separate effects of aging time and temperature than it has been for this tensile evaluation. However, the creep properties follow similar trends with precipitate type and distribution and are, in fact, most important in reactor applications of this alloy.

APPENDIX

Table 6. Heat Treatment Designations

Designation	Solution-Anneal Temperature[a] (°C)	Prestrain[b]	Age Treatment Time (hr)	Age Treatment Temperature (°C)
109	1177	yes	0	
121	1177	no	0	
123	1260	no	0	
130	1177	no	1,500	650
131	1177	no	1,500	760
132	1177	no	3,000	650
133	1177	no	3,000	760
134	1177	no	10,000	650
135	1177	no	10,000	760
136	1260	no	1,500	650
137	1260	no	1,500	760
138	1260	no	3,000	650
139	1260	no	3,000	760
140	1260	no	10,000	650
141	1260	no	10,000	760
142	1177	yes	1,500	650
143	1177	yes	1,500	760
144	1177	yes	3,000	650
145	1177	yes	3,000	760
146	1177	yes	10,000	650
147	1177	yes	10,000	760

[a]For 1 hr.

[b]Strained 10% in tension at room temperature after the 1-hr anneal at 1177°C.

Table 7. Tensile Data for Heat 466-535 (0.15% Ti) for Various Heat Treatments

Specimen	Heat Treatment	Test Temperature (°C)	Strength, psi Yield	Strength, psi Ultimate Tensile	Elongation, % Uniform	Elongation, % Total	Reduction in Area (%)
			$\times 10^3$	$\times 10^3$			
5989	121	650	29.3	71.4	24.0	24.6	27.28
5991	121	650	28.6	76.1	30.4	30.9	21.18
6019	123	650	26.1	66.7	29.0	30.0	22.61
6017	123	650	24.5	64.5	31.8	33.0	32.09
5993	130	650	38.7	77.5	24.9	27.7	26.61
6040	130	650	40.7	81.9	27.0	32.5	23.40
5994	131	650	35.2	73.9	15.9	16.4	17.59
6041	131	650	35.2	76.2	16.6	17.4	13.54
5999	132	650	39.3	73.9	18.5	18.8	22.93
6000	133	650	32.7	66.2	13.4	13.8	16.71
6001	134	650	36.2	62.6	14.8	15.1	11.49
6002	135	650	32.3	73.4	16.6	16.9	13.29
6021	136	650	33.1	66.1	21.0	21.6	24.20
6022	137	650	33.0	66.7	13.8	16.5	17.43
6027	138	650	35.2	57.5	12.1	12.6	17.73
6028	139	650	30.5	63.3	14.3	16.0	17.86
6029	140	650	38.4	62.2	10.2	10.7	10.54
6030	141	650	31.9	65.3	13.7	14.4	17.36
5997	130	25	61.5	135.9	40.5	40.8	29.54
5998	131	25	53.3	139.1	43.5	46.0	47.81
6003	109	650	47.8	78.4	18.7	19.3	21.38
6005	109	650	49.3	79.4	19.1	19.6	14.72
6007	142	650	57.9	90.2	15.2	15.6	11.67
6008	143	650	49.5	80.9	11.8	12.0	9.67
6013	144	650	54.1	80.7	10.8	11.5	13.71
6014	145	650	44.5	74.7	11.0	11.4	17.31
6015	146	650	48.9	66.2	7.4	7.7	3.98
6016	147	650	39.5	68.1	11.8	12.2	7.14

Table 8. Tensile Data for Heat 466-541 (0.27% Ti) for Various Heat Treatments

Specimen	Heat Treatment	Test Temperature (°C)	Strength, psi Yield	Strength, psi Ultimate Tensile	Elongation, % Uniform	Elongation, % Total	Reduction in Area (%)
			$\times 10^3$	$\times 10^3$			
6090	121	650	26.5	68.8	29.2	30.6	28.41
6092	121	650	27.5	76.2	33.7	34.6	28.30
6121	123	650	21.5	54.4	31.8	33.4	31.22
6119	123	650	21.4	53.9	29.7	30.5	41.14
6128	123	650	21.8	53.6	29.8	31.3	27.48
6094	130	650	35.1	77.3	35.1	43.7	38.68
6095	131	650	33.3	68.8	17.8	18.3	23.17
6101	132	650	34.6	74.3	32.9	39.6	35.31
6102	133	650	31.4	63.4	15.4	16.0	20.73
6103	134	650	33.8	69.8	28.6	34.6	27.36
6104	135	650	28.3	61.0	19.0	19.5	16.01
6123	136	650	26.6	46.8	16.4	17.4	26.43
6124	137	650	31.3	59.2	15.5	16.7	15.92
6129	138	650	29.9	46.1	10.4	11.4	25.41
6120	139	650	29.7	57.6	13.6	14.9	16.03
6131	140	650	36.0	48.5	7.4	8.9	10.39
6132	141	650	31.0	58.1	13.1	14.7	13.57
6099	130	25	54.3	125.6	52.5	53.7	41.88
6100	131	25	42.4	110.5	50.1	52.3	40.64
6107	109	650	45.4	77.5	22.4	23.2	17.77
6105	109	650	44.4	76.4	22.9	23.7	27.82
6109	142	650	48.1	81.3	21.6	24.4	20.33
6110	143	650	45.1	71.5	11.1	11.6	13.07
6115	144	650	49.3	81.6	18.0	19.7	19.39
6116	145	650	42.9	68.4	9.8	10.4	11.66
6117	146	650	52.2	81.6	13.7	14.5	12.60
6118	147	650	44.4	73.1	11.8	12.2	11.42

Table 9. Tensile Data for Heat 466-548 (0.45% Ti) for Various Heat Treatments

Specimen	Heat Treatment	Test Temperature (°C)	Strength, psi Yield	Strength, psi Ultimate Tensile	Elongation, % Uniform	Elongation, % Total	Reduction in Area (%)
			$\times 10^3$	$\times 10^3$			
6268	121	650	25.2	66.6	28.2	29.2	35.43
6270	121	650	24.2	66.9	30.4	31.9	26.33
6299	123	650	20.3	54.2	29.8	32.2	27.74
6297	123	650	20.6	55.3	30.2	32.9	34.73
7400	130	650	29.7	76.6	40.4	43.3	40.10
7402	131	650	30.1	79.2	35.3	37.7	28.94
6279	132	650	29.9	74.3	39.6	42.9	41.38
6280	133	650	30.1	67.3	20.1	20.8	27.04
6281	134	650	29.4	71.9	35.4	43.3	31.45
6282	135	650	28.9	70.3	23.6	24.3	20.41
6301	136	650	24.3	71.6	47.2	49.8	46.04
6302	137	650	27.2	86.5	32.3	33.5	27.64
6307	138	650	26.1	68.6	44.4	46.2	52.56
6308	139	650	25.8	61.8	24.9	26.2	23.42
6309	140	650	29.6	72.4	39.3	41.7	40.83
6310	141	650	27.0	63.9	24.9	26.2	26.33
6277	130	25	47.6	124.1	57.4	59.0	41.79
6278	131	25	48.1	129.6	53.6	55.4	50.95
6285	109	650	41.7	70.1	20.7	21.7	18.04
6283	109	650	43.9	71.9	19.3	20.4	24.16
6287	142	650	45.1	84.3	27.2	30.0	22.49
6288	143	650	41.3	69.7	14.8	15.5	13.90
6293	144	650	44.3	82.0	26.8	28.8	31.47
6284	145	650	40.8	68.8	14.5	14.9	16.75
6295	146	650	47.7	88.2	24.7	28.2	20.49
6296	147	650	42.1	71.9	15.0	15.6	12.60

Table 10. Tensile Data for Heat 467-548 (1.2% Ti) for Various Heat Treatments

Specimen	Heat Treatment	Test Temperature (°C)	Strength, psi Yield	Strength, psi Ultimate Tensile	Elongation, % Uniform	Elongation, % Total	Reduction in Area (%)
			$\times 10^3$	$\times 10^3$			
6313	121	650	39.4	96.3	29.8	35.3	31.58
6315	121	650	33.8	96.7	39.3	42.9	31.22
6341	123	650	29.7	74.6	39.4	40.8	39.95
6343	123	650	28.6	82.4	43.2	43.9	28.53
6317	130	650	46.5	92.1	28.3	50.2	45.00
6355	130	650	42.9	93.6	25.7	45.4	38.51
6357	131	650	41.1	94.5	26.5	44.7	37.44
6318	131	650	41.8	92.8	25.9	45.0	39.99
6323	132	650	43.1	87.9	26.1	43.3	42.18
6324	133	650	40.8	87.1	24.5	42.1	41.05
6325	134	650	38.2	84.3	26.7	44.1	35.74
6326	135	650	36.6	83.1	25.5	44.1	36.05
6345	136	650	42.7	83.3	24.4	25.7	24.07
6346	137	650	46.3	71.5	9.6	11.0	10.60
6351	138	650	44.8	79.0	17.3	20.1	27.41
6352	139	650	44.6	91.7	13.3	15.4	18.51
6353	140	650	49.6	81.4	14.0	15.2	18.29
6354	141	650	49.8	76.4	10.8	11.7	9.52
6321	130	25	61.3	145.6	50.5	52.0	41.21
6322	131	25	50.1	121.3	47.8	49.0	43.57
6327	109	650	73.1	102.2	20.0	27.7	30.35
6329	109	650	64.5	105.5	26.3	32.2	25.14
6331	142	650	64.5	104.5	21.6	33.2	30.56
6332	143	650	62.0	101.9	20.7	32.2	27.30
6337	144	650	61.4	100.3	18.3	26.4	25.45
6338	145	650	56.1	94.1	20.6	26.5	25.79
6339	146	650	62.9	103.3	20.0	30.4	24.07
6340	147	650	61.6	101.9	18.7	31.5	28.11
6989	137	25	59.6	122.8	41.6	41.7	36.73
11416	121	25	47.6	134.9	62.5	63.8	46.25

Table 11. Creep-Rupture Properties at 40,000 psi and 650°C for Commercial Heats Used to Determine the Thermal Stability

Heat	Heat Treatment	Test	Rupture Life (hr)	Secondary Creep Rate (%/hr)	Total Elongation (%)	Reduction in Area (%)
466-535	121	7130	166.5	0.0133	9.50	10.60
	123	7132	50.9	0.0119	13.60	24.39
	131	7277	136.5	0.0489	11.30	17.36
	132	7368	137.8	0.0630	13.60	14.27
	133	7369	63.6	0.0683	9.50	13.19
	138	7336	217.5	0.0346	13.70	15.22
	139	7340	26.2	0.0868	8.30	9.51
	109	7131	193.8	0.0048	3.60	6.41
	144	7464	135.4	0.0179	5.70	11.24
	145	7465	103.2	0.0420	6.70	10.79
466-541	121	7133	341.6	0.0060	15.20	17.75
	123	7135	189.0	0.0080	21.50	26.61
	131	7276	217.0	0.0538	18.90	19.48
	132	7370	507.6	0.0315	27.60	26.93
	133	7371	67.8	0.0692	10.50	18.42
	138	7337	50.0	0.0716	9.70	12.96
	139	7341	41.2	0.0844	9.90	17.62
	109	7134	308.9	0.0043	5.90	13.90
	144	7470	515.4	0.0159	14.00	20.73
	145	7471	43.1	0.0730	6.36	11.96
	121	7136	358.0	0.0117	16.70	22.31
466-548	123	7138	125.1	0.0110	19.30	28.19
	130	7024	569.6	0.0203	27.20	30.09
	131	7275	594.0	0.0251	42.10	34.19
	132	7372	574.6	0.0391	41.30	41.30
	133	7373	135.5	0.0535	16.70	21.94
	138	7338	345.5	0.0968	29.10	37.27
	139	7342	175.2	0.0660	24.10	37.54
	109	7137	620.6	0.0065	11.70	17.77
	144	7468	855.8	0.0166	25.60	22.62
	145	7469	130.7	0.0255	6.50	15.97
	130	7024	669.6	0.0203	27.20	30.09
467-548	121	7434	1207.6	0.0144	36.60	33.44
	123	7141	2472.9	0.0038	21.60	17.62
	131	7303	1006.2	0.0195	32.40	47.74
	132	7374	1582.7	0.0159	54.40	47.74
	133	7375	1335.2	0.0162	36.80	35.17
	138	7339	2095.5	0.0072	24.50	26.93
	139	7343	1454.2	0.0099	20.40	28.64
	109	7140	2149.2	0.0049	30.50	28.94
	144	7466	2638.1	0.0053	33.30	33.60
	145	7467	1853.1	0.0089	32.10	30.33

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