EFFECT OF HEAT TREATMENT ON MECHANICAL

PROPERTIES AND MICROSTRUCTURE OF ALLOY 901

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Summary

Two types of heat treatments are commonly used for alloy 901, one to optimize resistance to creep and rupture and the other, to improve tensile and fatigue properties. Solution and aging-treatment parameters (temperature, time, quench) for both types of heat treatment were varied within the ranges permitted by common specifications to define heat treatments resulting in improved combinations of room-temperature strength and ductility along with good stress-rupture properties. Room-temperature tensile and 1200°F stress-rupture tests were performed and microstructures were evaluated using light and electron microscopy. Structure/property relationships were defined. The results showed that relatively small variations in heat-treatment parameters resulted in significant differences in mechanical properties and microstructure. Room-temperature tensile properties were optimized without degrading 1200°F stress-rupture properties. Mechanical properties were dependent on grain size and the characteristics of the grain boundary carbide and matrix hardening (γ') precipitates.

Superalloys 1988
Edited by S. Reichman, D.N. Duhl,
G. Maurer, S. Antolovich and C. Lund
The Metallurgical Society, 1988

Introduction

Pyromet® alloy 901 is a nickel-iron-base superalloy used for aerospace and land-based turbine components requiring high tensile and stress-rupture strengths along with good hot-corrosion resistance at 1000-1200°F. Two different types of heat treatments are commonly used for the alloy, one to optimize resistance to creep and rupture at elevated temperatures (Type I) and the other, to improve tensile and fatigue properties (Type II). Commercial specifications for alloy 901 generally include solution and aging temperature ranges of 50-75°F and also allow some adjustment of treatment time and cooling rate from the solution temperature. Although the effects of heat treatment on alloy 901 have been summarized (1), a comprehensive study has not been published. The purpose of this study was to determine the effects of heat-treatment parameters, within the limits of commercial specifications, on mechanical properties and microstructure of alloy 901 with the goal of defining heat treatments resulting in improved combinations of room-temperature strength and ductility along with good stress-rupture properties at 1200°F.

Experimental Procedure

Material from a vacuum-induction melted/vacuum-arc remelted production heat of Pyromet alloy 901 was used for all experiments. The chemical composition of this material is listed below:

Element:	C	Cr	Ni	Mo	Ti	Al	B	Fe
Weight %:	0.03	11.5	42.1	5.3	3.0	0.3	0.012	Balance

Longitudinal sections of 6-inch diameter bar (2-inch square x 6-inch long) were heated to 2000°F and press forged to provide 3/4-inch square bars for testing.

Room-Temperature Tensile. Collective ranges for the two solution + double-age heat treatments commonly specified for alloy 901 are listed below:

Type I - 1950-2025°F (±25°F)/2h/AC or faster + 1400-1475°F (±15°F)/ 2-4h/AC or faster + 1300-1375°F (±15°F)/24h/AC
Type II - 1800-1850°F (±25°F)/1-2h/AC or faster + 1300-1350°F (±15°F)/ 6h min./AC or faster + 1175-1225°F (±15°F)/12h min./AC

Three-factor, two-level, full-factorial experimental designs were used to determine the effects of heat treatment on room-temperature tensile properties. For Type I heat treatments, the factors and levels chosen for initial evaluations were: solution temperature (1950,2000°F), mid-age temperature (1400,1450°F) and final-age temperature (1300,1350°F). For Type II heat treatments, the factors and levels were: solution temperature (1800,1850°F), mid-age temperature (1300,1350°F), and mid-age time (6, 18 hours). The final-age treatment was held constant at 1200°F/12h/AC for Type II treatments. Heat treatments representing the mid-points of each range were used to determine whether any non-linear relationships existed (curvature effects). Thus, a total of nine heat treatments of each type were evaluated initially.

Fully heat-treated 3/4-inch square blanks were machined to 0.252-inch gage-diameter threaded specimens for room-temperature tensile tests. Duplicate specimens were tested for each treatment. Sampling and testing procedures were designed to randomize any variations unrelated to heat treatment. Yield strength and elongation results were statistically analyzed (2) to identify factor and curvature effects that were significant with a 95%

® - registered trademark of Carpenter Technology Corporation level of confidence. Following the initial experiments, additional tensile tests were performed to confirm optimum treatments, to define non-linear relationships, and to determine effects of cooling rate from the solution temperature (water quenching vs. air cooling).

Stress-Rupture. Testing was concentrated on heat treatments resulting in the best combinations of room-temperature strength and ductility although limited testing was done to define heat-treatment effects. Fully heat-treated blanks were machined to 0.178-inch gage-diameter combination smooth-notched (K_t=3.8) stress-rupture specimens and tested at 1200°F with a constant load of 90 ksi.

Light and Electron Microscopy. Longitudinal sections of heat-treated tensile specimens were metallographically prepared and examined using light microscopy to determine grain size and relative amounts of grain boundary precipitation. The fracture surfaces of several room-temperature tensile specimens were examined using scanning electron microscopy (SEM) to determine fracture modes. Thin foils, structural replicas (chromium-shadowed parlodion), and carbon extraction replicas were examined using transmission electron microscopy (TEM) to identify and characterize matrix and grain boundary phases. Phases were identified using convergent-beam electron diffraction analysis and energy dispersive spectroscopy (EDS). Precipitates were also identified using X-ray diffraction analysis of extracted residues. Larger matrix phases were analyzed in situ using the electron microprobe.

Gamma-prime (γ') size distributions were determined using thin foil electron micrographs representing several heat treatments. About 500 particle size measurements and about 50 measurements of interparticle spacing were made for each heat treatment. Because of the complexities involved in measuring fine γ' particles, the size distributions reported in this study should be used to compare heat treatments rather than to obtain absolute size data.

Results and Discussion

Type I Heat Treatments

Room-Temperature Tensile Properties. Room-temperature tensile and stress-rupture test results are listed in Table I. Ratings of room-temperature yield strength and elongation were calculated to indicate the degree to which the results exceeded minimum requirements of common specifications. One point was given for each increment of the pooled standard deviation (s) by which the result exceeded the minimum requirement plus one standard deviation e.g. 120 ksi + 3s = 2 points; 120 ksi + 1s = 0 points. The products of the yield strength and elongation ratings listed in Table I indicate the combination of strength and ductility offered by a particular heat treatment. It should be noted that the properties and ratings listed in Table I also depend on composition, forging procedure, and bar size but only heat-treatment effects will be discussed in this paper.

The minimum significant factor effects (95% confidence level) were 2 ksi for yield strength and 1% for elongation. Heat-treatment effects less than these values would not be separable from variation due to forging, sampling, specimen preparation and testing. Mid-age and final-age temperatures, and the interaction between these factors, had the most significant effects on yield strength. Solution and mid-age temperatures, and the interaction between the mid-age and final-age temperatures, had the most significant effects on tensile elongation. For Type I heat treatments, the lower solution, mid-age, and final-age temperatures resulted in the best combinations of room-temperature strength and ductility as evidenced by the ratings in Table I.

Table I. Mechanical Properties - Type I Heat Treatments

Heat Treatment(°F)* Solution +	Mid Age +	Final Age	Room-Temperature Tensile (avg.) 0.2% Y.S. (ksi)	U.T.S (ksi)	El. (%)	R.A. (%)	YS-El. Rating	Stress Rupture (1200°F/90 ksi) Life (h)	El. (%)	R.A. (%)
1950/WQ	+ 1400	+ 1300	136	184	17	26	39	-	-	-
"	+ 1425	+ 1300	138	185	17	22	42	58	7.5	14
								294	10.5	18
								174	10.5	17
" /AC	+ 1425	+ 1300	135	186	16	25	24	-	-	-
" /WQ	+ 1450	+ 1300	140	186	16	19	32	-	-	-
"	+ 1400	+ 1350	135	183	16	20	30	-	-	-
"	+ 1450	+ 1350	133	184	16	21	21	-	-	-
1975/WQ	+ 1400	+ 1300	135	184	18	21	38	304	10.0	14
								314	9.0	14
" /AC	+ 1400	+ 1300	128	181	19	24	21	138	6.0	11
								201	6.5	9
" /WQ	+ 1425	+ 1300	138	183	14	19	6	289	11.5	15
								262	10.5	15
" /AC	+ 1425	+ 1300	131	183	17	25	22	286	9.0	11
								234	9.5	13
" /WQ	+ 1425	+ 1325	138	181	13	18	4	-	-	-
2000/WQ	+ 1400	+ 1300	134	175	15	18	19	-	-	-
"	+ 1425	+ 1300	-	-	-	-	-	271	10.0	17
								239	6.5	15
"	+ 1450	+ 1300	138	176	11	14	0	-	-	-
"	+ 1400	+ 1350	134	174	12	17	0	-	-	-
"	+ 1450	+ 1350	133	174	11	13	0	-	-	-
Minimum Requirements			120	165	12	15		23	5	-

* Solution treated 2h/WQ or AC + mid-aged 2h/AC + final-aged 24h/AC.

Results of the microstructural evaluation of heat-treated tensile specimens using light and electron microscopy are listed in Table II. X-ray diffraction and electron microscopy studies revealed that phases present in in the matrix were Ti and Mo-rich MC carbides (<1-25μm), Mo-rich M_3B_2 and the γ' [Ni_3(Ti,Al)] hardening phase (2-40 nm). Smaller amounts of Mo-rich M_6C carbide (<2μm) were also identified. Grain boundary precipitates were primarily Ti and Mo-rich MC carbides (50-250 nm). Similar phases have been found by others (1,3,4). The same types of phases were observed for all heat treatments although in varying amounts.

Figure 1 illustrates the detrimental effect of increasing the solution temperature from 1950°F to 2000°F on tensile ductility. It appears that some curvature exists in this relationship, depending on the aging treatment. Five of eight specimens solution treated at 2000°F before aging failed the minimum elongation requirement of 12%. The reduction in ductility is attributed to coarser grain size and more extensive precipitation of

Table II. Microstructural Observations - Type I Heat Treatments

Heat Treatment(°F)[1] Solution +	Mid Age +	Final Age	Tensile Fracture Mode[2]	ASTM Grain Size	Grain Boundary Precipitation Rel. Amt.[3]	Continuity[4]	Gamma Prime Distribution Avg. Diam. (nm)	Min. Diam. (nm)	Max. Diam. (nm)	Avg. Spacing (nm)
1950/WQ	+ 1400	+ 1300	50%DT/50%I	4/5	S	D	10	2	26	21
"	1425	+ 1300	---	4/5	M	S	14	3	35	19
"	1450	+ 1300	25%DT/75%I	3/5	M/L	C	15	3	30	16
"	1450	+ 1350	---	4/5	L	C	19	5	39	18
1975/WQ	+ 1400	+ 1300	---	3/5	S/M	D/S	12	3	30	19
" /AC	"	"	---	-	M/L	-	-	-	-	-
" /WQ	+ 1425	+ 1300	Mostly I	3/4	M	S	11	3	27	17
" /AC	"	"	Mostly I	3/5	M/L	S	10	2	22	18
" /WQ	+ 1425	+ 1325	Mostly I	3/4	M/L	C	15	4	29	15
2000/WQ	+ 1400	+ 1300	---	2/4	M	-	-	-	-	-
"	1450	+ 1300	Mostly I	2/4	L	C	16	6	29	15

1 - Solution treated 2h/WQ or AC + mid-aged 2h/AC + final-aged 24h/AC
2 - Fracture mode: DT = ductile transgranular; I = intergranular
3 - Relative amount: S = small; M = moderate; L = large
4 - Continuity: D = discontinuous; S = semi-continuous; C = continuous

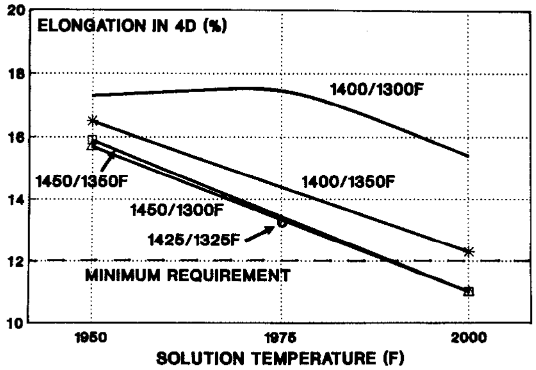


Figure 1. Effect of solution temperature on room-temperature tensile ductility - Type I treatments (mid-age/final-age temperatures are shown)

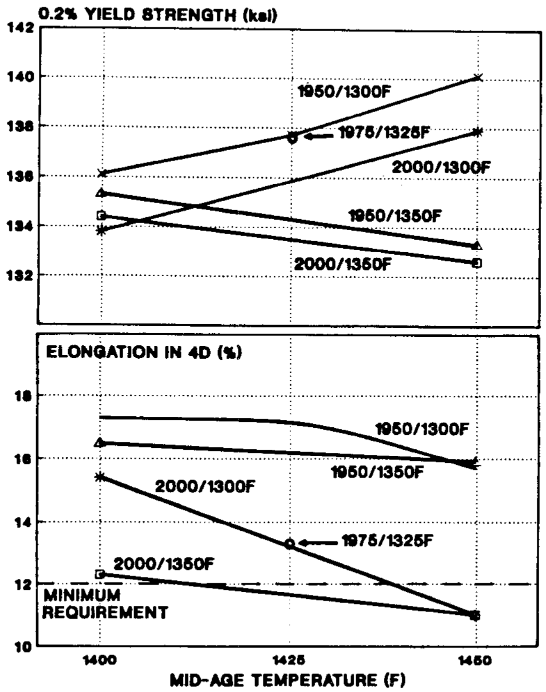


Figure 2. Effect of aging temperatures on room-temperature tensile properties - Type I treatments (sol'n/final-age temps. are shown).

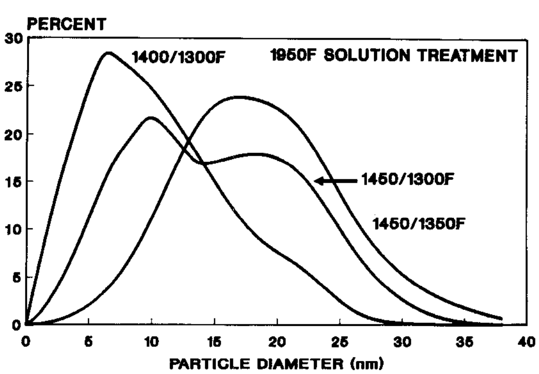
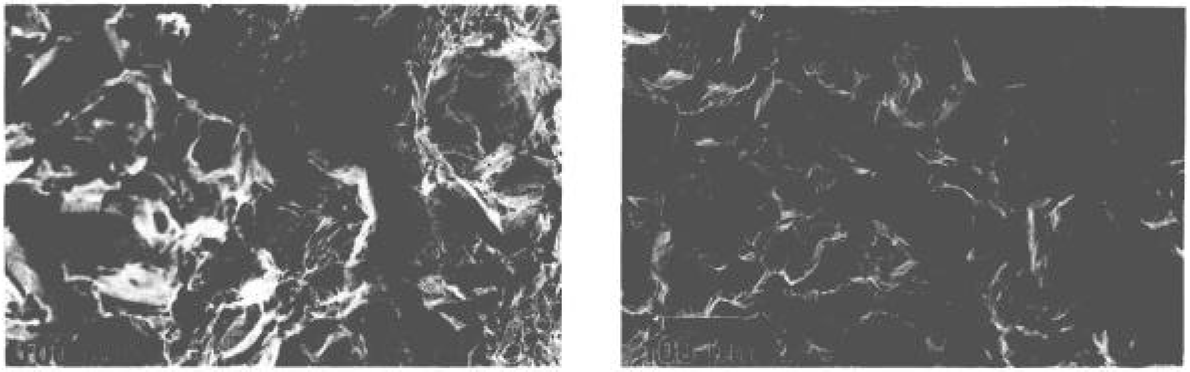


Figure 3. Effect of aging treatment on gamma-prime size distribution-Type I treatments.

semi-continuous or continuous grain boundary carbides (MC) which resulted in a predominantly intergranular tensile fracture mode. Higher solution temperatures were most detrimental when used in conjunction with higher mid-age temperatures (1425-1450°F) because grain boundary precipitation was further increased. Slower cooling from the solution temperature (air vs. water quench) decreased yield strength. Air cooling resulted in a slightly finer distribution of γ' compared to water quenching. Hammond and Ansell have reported that vacancies frozen in position during quenching from the solution temperature provide nucleation sites for γ' particles (5). It is postulated that water-quenching stresses may enhance the diffusion of solutes thereby accelerating the aging process.

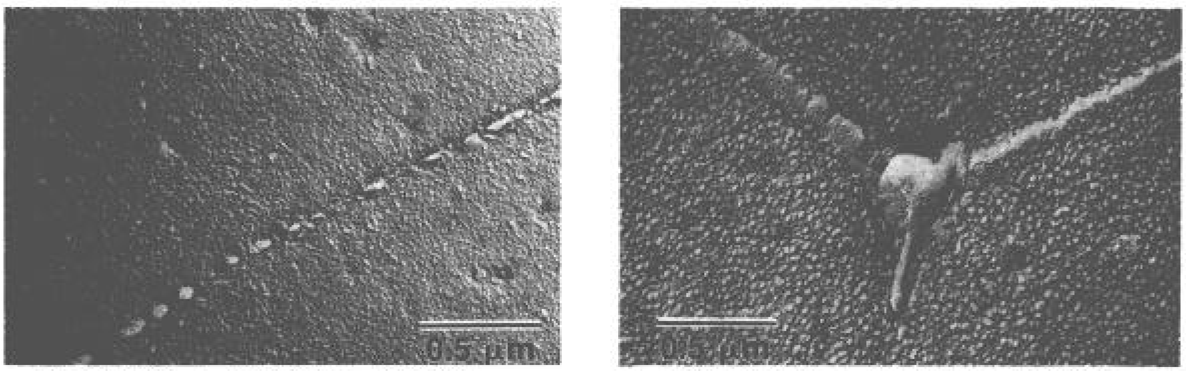
Figure 2 illustrates the significant effect of mid-age temperature on tensile properties and the strong interaction between mid-age and final-age temperatures. Yield strength increased and elongation decreased with increasing mid-age temperature (1400-1450°F) when a 1300°F final age was used while yield strength and elongation both decreased slightly with increasing mid-age temperature when a 1350°F final age was used. Figure 3 contains size distribution plots showing the effects of mid-age and final-age temperatures on γ' particle size. Increasing mid-age temperature from 1400°F to 1450°F (1300°F final age) resulted in a coarser and more bimodal distribution of γ'. The higher strength of material aged at 1450°F + 1300°F appears to be a result of this duplex γ' size and reduced interparticle spacing. The lower ductility of samples solution treated at 1975-2000°F and mid-aged at the higher temperatures was associated with mostly intergranular tensile fracture modes (See Figure 4). Typical replica electron micrographs are shown in Figure 5. The relative amount, size and continuity of grain boundary precipitates increased as mid-age temperature increased leading to more intergranular fracture.



1950°F/WQ + 1400°F + 1300°F
50% intergranular - 17.6% El

1975°F/WQ + 1425°F + 1300°F
mostly intergranular - 13.9% El

Figure 4. SEM micrographs of fracture surfaces of room-temperature tensile specimens with different ductility values.



1950°F/WQ + 1400°F + 1300°F

1950°F/WQ + 1450°F + 1300°F

Figure 5. Replica electron micrographs showing the effect of mid-age temperature on gamma-prime and grain boundary carbide structure.

Results in Table I and Figure 2 indicate that increasing the final-age temperature above 1300°F provided no beneficial effect on room-temperature tensile properties, regardless of the solution/mid-age combination. The yield strength of material mid-aged at 1450°F was reduced significantly when final-age temperature was increased from 1300°F to 1350°F. Based on hardness measurements and γ' size distributions shown in Figure 3, it is concluded that the lower strength is a result of γ' coarsening (overaging).

Stress-Rupture Properties. All Type I heat treatments evaluated resulted in stress-rupture properties exceeding the requirements of common specifications (Table I). Ratings similar to those reported for tensile properties were not useful for stress-rupture data because of considerable variation between duplicate tests. The 1975°F/WQ + 1400°F + 1300°F treatment resulted in the best combination of room-temperature tensile and stress-rupture properties. Using the above treatment temperatures, air cooling from 1975°F resulted in significantly lower stress-rupture life and ductility than did water quenching. Air-cooled specimens contained more extensive precipitation of semi-continuous to continuous grain boundary carbides which resulted in larger amounts of intergranular fracture. Specimens with higher rupture lives generally displayed a mixed mode of fracture. Increasing solution temperature from 1950°F to 1975-2000°F appeared to have a beneficial effect on stress-rupture life.

Table III. Mechanical Properties - Type II Heat Treatments

Heat Treatment*		Room-Temperature Tensile (avg.)					Stress Rupture (1200°F/90 ksi)		
Solution (°F)	+ Mid-Age (°F) (h)	0.2% Y.S. (ksi)	U.T.S (ksi)	El. (%)	R.A. (%)	YS-El. Rating	Life (h)	El. (%)	R.A. (%)
1800/WQ	+ 1300/6	115	184	29	41	0	-	-	-
"	+ 1350/6	134	193	24	43	5/17	102	10.5	23
							42	8.5	27
"	+ 1300/18	135	194	22	43	2/11	-	-	-
"	+ 1350/18	148	198	20	35	0/6	-	-	-
1825/WQ	+ 1350/6	134	193	24	41	6/18	161	8.5	18
							72	6.5	14
"	+ 1325/12	135	191	23	40	5/16	112	9.5	16
							55	6.5	14
" /AC	+ 1325/12	127	190	24	38	0/8	87	Notch Break	
							20	Notch Break	
1850/WQ	+ 1300/6	107	176	34	45	0	-	-	-
"	+ 1350/6	129	188	26	42	3/15	186	9.5	21
							116	6.5	19
" /AC	+ 1350/6	122	190	22	40	0	-	-	-
" /WQ	+ 1325/12	133	190	23	39	2/12	186	10.5	19
							166	11.0	17
"	+ 1350/12	144	195	21	35	0/10	-	-	-
"	+ 1300/18	130	190	26	40	4/16	-	-	-
"	+ 1350/18	148	196	21	34	0/16	-	-	-
Minimum Requirements		120/ 125	170/ 175	18/20	25		23	5	

* Solution treated 1.5h/WQ or AC + mid-aged as shown/AC + final-aged 1200°F/12h/AC.

Type II Heat Treatments

Room-Temperature Tensile Properties. Room-temperature tensile and stress-rupture test results are listed in Table III. Yield strength/elongation ratings were calculated as described for Type I treatments except that typical minimum requirements of 120 and 125 ksi for yield strength and 18 and 20% for elongation were used for calculation. The minimum significant factor effects (95% confidence level) were 2.2 ksi for yield strength and 1.6% for elongation. Mid-age time and temperature had the largest effects on room-temperature tensile properties (Figure 6) although

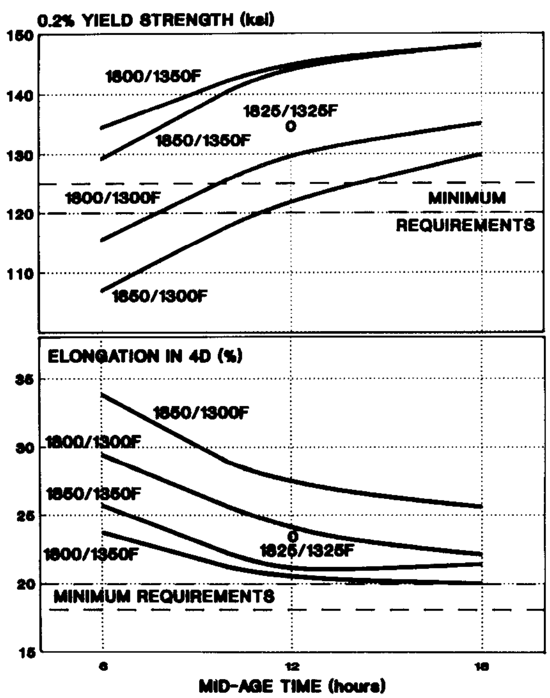


Figure 6. Effect of mid-age time on room-temperature tensile properties - Type II treatments (sol'n/mid-age temperatures are shown).

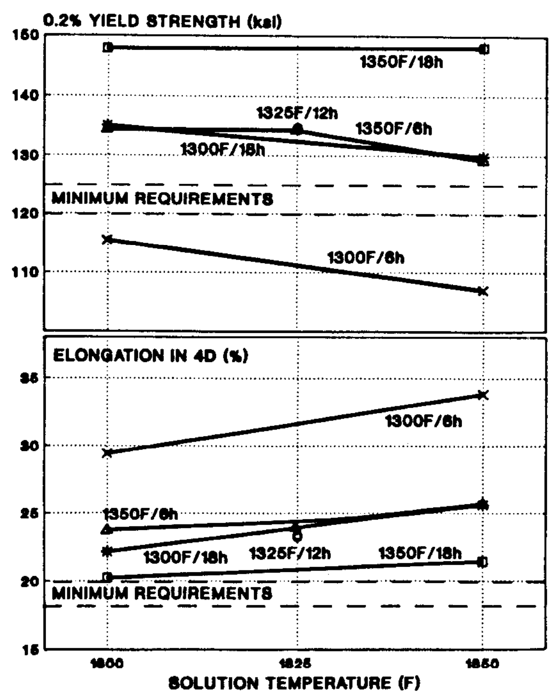


Figure 7. Effect of solution temperature on room-temperature tensile properties - Type II treatments (mid-age treatment is shown).

the solution temperature effect (Figure 7), and the interaction effect between mid-age time and temperature, were also significant with a 95% confidence level. The relationships between mid-age time and yield strength/elongation were not linear. For Type II treatments, a mid-age treatment at 1350°F for the minimum aging time of 6 hours resulted in the best combination of room-temperature strength and ductility.

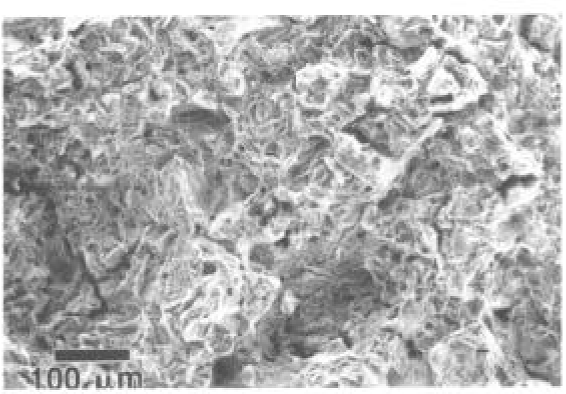
Table IV. Microstructural Observations - Type II Heat Treatments

Heat Treatment[1] Solution (°F)	+	Mid-Age (°F)(h)	Tensile Fracture Mode[2]	ASTM Grain Size	Grain Boundary Precipitation Rel. Amt.[3]	Contin-uity[4]	Gamma Prime Distribution Avg. Diam. (nm)	Min. Diam. (nm)	Max. Diam. (nm)	Avg. Spacing (nm)
1800/WQ	+	1300/6	DT	7/8	S	D	9	4	17	21
"	+	1350/6	DT	7/8	M/L	S	12	4	22	17
"	+	1300/18	DT	8	L	D/S	10	4	17	18
1825/WQ	+	1350/6	-	6/7	S/M	-	-	-	-	-
"	+	1325/12	Mostly DT	6/7	L	C	11	2	26	17
1825/AC	+	1325/12	-	6/7	VL	S/C	10	2	22	15
1850/WQ	+	1350/6	DT	5/7	S/M	D	12	4	22	14
1850/AC	+	1350/6	75%DT/25%I	5/7	L	-	-	-	-	-
1850/WQ	+	1325/12	-	5/7	S/M	-	-	-	-	-

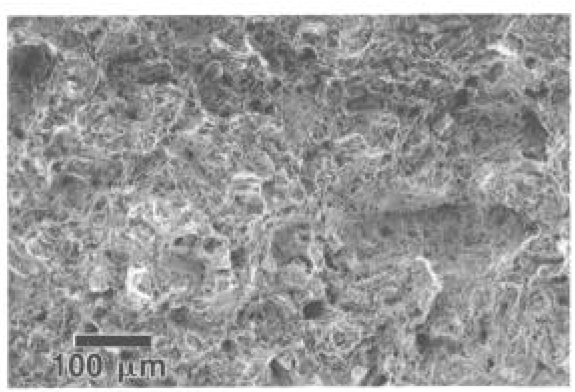
1 - Solution treated 1.5/WQ or AC + mid-aged as shown/AC + final-aged 1200°F/12h/AC.
2 - Fracture mode: DT = ductile transgranular; I = intergranular
3 - Relative amount: S = small; M = moderate; L = large; VL = very large
4 - Continuity: D = discontinuous; S = semi-continuous; C = continuous

Results of the microstructural evaluation of heat-treated tensile specimens using light and electron microscopy are listed in Table IV. Phase identities were the same as those reported for Type I treatments except that some eta phase (Ni_3Ti) was observed in the matrix and grain boundaries of specimens solution treated at 1800°F. Matrix precipitates in specimens with Type II treatments were banded to a greater extent than those of Type I treatments, particularly when a solution temperature of 1800°F was used.

Increasing the solution temperature from 1800°F to 1850°F resulted in lower strength and higher ductility when lower aging temperatures and times were used (underage treatments) as shown in Figure 7. Higher solution temperatures had little or no effect on the properties of material aged to peak hardness (1350°F/18h). The lower strength and higher ductility of material solution treated at 1850°F and underaged is attributed to coarser grain size and slower aging response. No effect on γ' size distribution was observed. Air cooling from the solution temperature resulted in lower strength than water quenching without benefit of higher ductility. As mentioned previously, stresses from water quenching from the solution

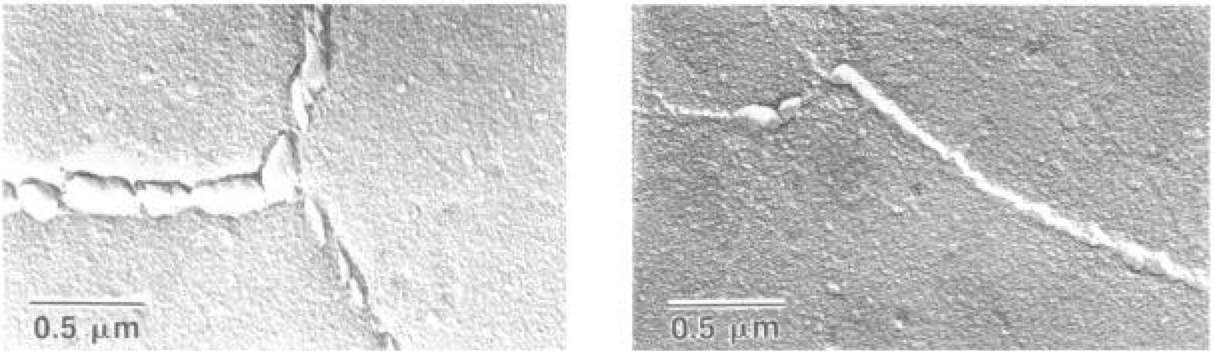


Air Cool - 122 ksi YS, 22% El
(25% Intergranular)



Water Quench - 129 ksi YS, 26% El
(100% Ductile Transgranular)

Figure 8. SEM micrographs of fracture surfaces of room-temperature tensile specimens with different ductility values.
Heat treatment: 1850°F/1.5h/AC or WQ + 1350°F/6h/AC + 1200°F/12h/AC



Air Cool Water Quench

Figure 9. Replica electron micrographs showing the effect of cooling rate from the solution temperature on grain boundary carbide precipitation. Heat treatment: 1825°F/1.5h/AC or WQ + 1325°F/12h/AC + 1200°F/12h/AC.

temperature may accelerate the aging process, particularly for underage-type treatments. SEM fractographs and replica electron micrographs of air-cooled and water-quenched specimens are shown in Figures 8 and 9. The lower-strength air-cooled specimens did not have improved ductility because of more extensive precipitation of larger grain boundary carbides resulting in a partially intergranular tensile fracture mode.

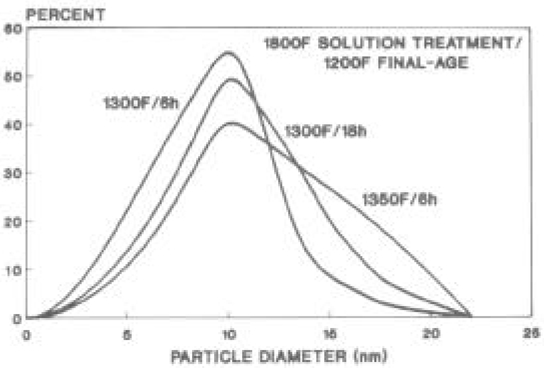


Figure 10. Effect of mid-age treatment on gamma-prime size distribution - Type II treatments.

The room-temperature tensile properties obtained using Type II treatments were primarily dependent on the size of the γ' particles precipitated during mid-aging and the resultant interparticle spacing (constant 1200°F/12h final age). The effect of mid-age treatment on γ' size distribution is shown in Figure 10. Yield strength increased and ductility decreased as mid-age time (6-18h) or temperature (1300-1350°F) increased because the size of the γ' particles increased and interparticle spacing decreased. Peak hardening was obtained only after about 18 hours at 1350°F, therefore most of the mid-age treatments evaluated resulted in underaging. Grain boundary precipitation had little effect on ductility as predominantly ductile transgranular tensile fractures were observed.

Type II treatments resulted in significantly higher room-temperature tensile ductility than Type I treatments at similar strength levels. The higher ductility was associated with ductile, microvoid-coalescence fracture modes (Figure 8) rather than the predominantly intergranular or mixed fracture modes characteristic of Type I heat treatments (Figure 4). The ductile transgranular fracture modes were associated with finer grain size and generally smaller amounts of finer and less continuous grain boundary precipitates.

Stress-Rupture Properties. Although there was considerable scatter in the data for Type II treatments, stress-rupture test results in Table III show that heat treatments which resulted in the best room-temperature tensile properties also resulted in 1200°F stress-rupture properties exceeding minimum requirements of common specifications. Type II heat treatments resulted in significantly lower stress-rupture lives than the Type I heat treatments previously discussed. The improvement in stress-rupture life obtained using a solution temperature of 1850°F is attributed to coarser grain size and smaller amounts of finer and less continuous grain boundary precipitates. Similar effects have been observed by others (3). The notch sensitivity of specimens air-cooled from the solution temperature was also related to an increase in the size and amount of grain boundary precipitates. The air-cooled specimens showed a primarily intergranular fracture mode in the fracture initiation zone while water-quenched specimens showed a mixed fracture mode.

Conclusions

Using heat treatments designed to optimize resistance of Pyromet alloy 901 to high-temperature creep and rupture (Type I), the best combinations of room-temperature strength and ductility were obtained using the minimum specified temperatures for solution treatment (1950-1975°F), mid-aging (1400-1425°F), and final-aging (1300°F). Excellent stress-rupture properties at 1200°F were also obtained using these combinations of solution and aging temperatures. Water quenching from the solution temperature generally resulted in better room-temperature tensile and stress-rupture properties than air cooling. Tensile properties were dependent on the size and spacing of γ' and grain boundary carbide precipitates.

Using heat treatments designed to optimize tensile and fatigue properties (Type II), room-temperature tensile properties were primarily dependent on mid-age treatment time and temperature. A mid-age treatment at 1350°F resulted in best properties after a minimum aging time of 6 hours. Type II heat treatments resulted in significantly higher tensile ductility but lower stress-rupture lives compared to Type I treatments. Higher solution temperatures improved stress-rupture life without reducing tensile properties. Air cooling from the solution temperature resulted in lower yield strengths and unacceptable stress-rupture properties. Tensile properties were primarily dependent on the size and spacing of the γ' hardening precipitates.

The lower tensile ductility resulting from Type I heat treatments was attributed to coarser grain size and partially or mostly intergranular tensile fracture modes. The amount of intergranular fracture increased with the size, amount and continuity of grain boundary precipitates. In contrast, Type II treatments resulted in finer grain size, finer and less continuous grain boundary precipitates and ductile, transgranular fracture modes.

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