IMPROVING THERMAL STABILITY OF ALLOY 718

VIA SMALL MODIFICATIONS IN COMPOSITION

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Abstract

Expanding upon the effort to redesign Alloy 718 in order to provide microstructural and mechanical stability beyond 650°C, six modified compositions have been studied after a precipitation hardening treatment and then after aging for 534 hr at 730°C (1350°F). The variations in Al, Ti and Nb content provided an (Al+Ti)/Nb ratio between 0.66 and 0.92, an Al/Ti ratio between 0.88 and 1.69 and total hardener (Al+Ti+Nb) content between 5.80 and 6.80 at pct; two of the alloys also contained 0.76 or 1.77 at pct W as a solid solution strengthening element. Even though the rate of γ' coarsening was faster after the aging treatment, the coarsening rate of γ'' and the decline in strength occurred more slowly in the four alloys with a higher (Al+Ti)/Nb ratio, and with less transformation to δ phase, than was the case in the two lower (Al+Ti)/Nb ratio alloys. The two alloys having the highest (Al+Ti)/Nb and Al/Ti ratios and a W addition provided increasing strength and a slower rate of γ'' coarsening. Hardness and tensile strength after the conventional heat treatment exceeded the published results for high strength processed 718 and, after 534 hours at 730°C and testing at room temperature, these mechanical properties were retained to the highest degree in these same alloys. Under stress rupture conditions at 700°C (1300°F), these alloys survived from 35 to 53 hours under a stress of 92.5 ksi whereas conventional 718 has a published rupture life of 18 hours when stressed at 85 ksi. At this stress level, one of the alloys having both ratios and total hardening element content on the high side of the range survived for 74.3 hours. These preliminary results suggest that the basic composition of Alloy 718 can be modified to a minor degree in order to improve temperature capability beyond 650°C.

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Introduction

The initial study by Cozar and Pineau[1] revealed that a more stable microstructure may be obtained in Alloy 718 for long periods of time at the high aging temperature of 750°C (1380°F) by employing an (Al+Ti)/Nb ratio of ~1, on an atomic percent basis. As a result, a stable plateau in hardness was achieved for long periods of time whereas a significant drop in hardness occurred in regular 718 in a relatively short time because of the formation of deleterious δ phase. The (Al+Ti)/Nb ratio was raised from ~0.7 in regular 718 by removing niobium, the most significant strengthening element. Although a stable plateau was realized, the hardness level would be considered low for most applications. Cozar and Pineau also demonstrated that relatively small modifications in composition, providing a larger (Al+Ti)/Nb ratio, and heat treatment lead to a so-called compact morphology in which small (~200Å) γ' precipitates are fully embedded by γ'' precipitates and greater resistance to thermal overaging.

The goal of any study to improve upon the thermal stability of Alloy 718 above 650°C (1200°F) is to reduce the amount of overaging of the γ' and γ'' phases and to increase the transformational time to δ phase. In a promising redesign effort, Collier and Tien[2,3] have systematically varied the content of the three critical elements, Al, Ti and Nb. They have extended Cozar and Pineau's work[1] by modifying the alloy in the direction of higher (Al+Ti)/Nb (0.90) and Al/Ti (1.60-1.75) ratios, with 3.4 or 3.6 at pct Nb, and they were able to reduce the amount of overaging of the γ'' particles and also to increase the amount of γ' in order to stabilize mechanical properties. This preliminary study extends the results obtained by Collier and Tien by providing additional data on the microstructural and mechanical stability of six equivalent compositions, two of which also contain tungsten to provide additional solid solution strengthening at elevated temperatures. It is confined to an examination of the alloys after a standard precipitation hardening treatment and then after aging for 534 hours at 730°C (1350°F). Also, the rupture lives of the as heat treated alloys were obtained at 700°C (1300°F) under high stress levels.

Materials and Procedure

The modifications of Alloy 718 in this study were vacuum induction melted and cast into 5 kg ingots. After homogenizing for 24 hours at 1100°C and for 1 hour at 1160°C, the ingots were hot forged into 32 mm bars. Chemical analysis samples cut from the bars provided the compositions listed in Table I. The translation of weight percent to atomic percent and the corresponding total (Al+Ti+Nb) hardener content, (Al+Ti)/Nb and Al/Ti ratios are listed in Table II. The six alloys were given the standard 718 heat treatment of 955°C for 1 hr, air cool, 720°C, hold 8 hr then furnace cool to 620°C, hold 8 hr and air cool. Then, specimens for hardness testing and metallographic examination were held for nine intervals between 2 and 534 hours at 730°C and air cooled. Tensile test specimens were also prepared after the above precipitation hardening treatment and after the exposure to 534 hours at 730°C. Thin foils were prepared and observed by transmission electron

Alloy	C	Cr	Mo	Ti	Al	Nb	Fe	W	B
1	0.053	17.56	2.85	1.33	0.66	5.51	16.78	--	0.0037
2	0.055	17.56	2.85	1.01	0.64	5.57	16.92	--	0.0026
3	0.056	17.58	2.85	0.97	0.86	5.51	17.00	--	0.0033
4	0.058	17.58	2.93	1.00	0.50	5.57	17.46	--	0.0031
5	0.048	16.60	3.09	0.98	0.93	5.57	13.71	2.30	0.0019
6	0.046	16.48	--	0.96	0.95	5.57	13.66	5.46	0.0014

Table I. Chemical Composition of Alloys wt.%

Alloy	Cr	Mo	Ti	Al	Nb	Fe	W	Ni	Al+Ti+Nb	Al/Ti	Al+Ti/Nb
1	19.60	1.73	1.62	1.42	3.45	17.45	--	54.72	6.49	0.88	0.88
2	19.64	1.73	1.22	1.38	3.49	17.62	--	54.92	6.09	1.13	0.74
3	19.60	1.72	1.18	1.85	3.44	17.65	--	54.90	6.47	1.57	0.88
4	19.69	1.78	1.22	1.08	3.50	18.21	--	54.53	5.80	0.88	0.66
5	18.87	1.90	1.21	2.04	3.55	14.52	0.76	57.14	6.80	1.69	0.92
6	18.91	--	1.19	2.10	3.58	14.60	1.77	57.85	6.87	1.76	0.92

Table II Chemical Composition of Alloys at.%

microscope. Using electron diffraction and dark field techniques with [001] axis of the matrix parallel to the electron beam, γ' and γ'' phases were identified and the γ' diameters and γ'' lengths were measured in each alloy after aging for 534 hours at 730°C. Also, the weight percentage of δ, $\gamma' + \gamma''$ and NbC + TiC was obtained by means of chemical analysis of extraction phase. To provide additional material for stress rupture testing, larger 23 kg ingots of the same compositions were prepared, then forged and rolled to bar stock. The specimens were solution treated at 1030°C for 1 hr instead of 955°C and then given the same precipitation hardening treatment. Stress rupture testing was conducted at 700°C under stresses of 70, 65, 63 and 60 kg/mm^2 which correspond to 99.5, 92.5, 89.6 and 85.4 ksi.

Results

Optical micrographs of the three alloys that produced the best thermal stability on the basis of retained hardness and tensile properties at room temperature after the precipitation hardening treatment and then after exposure to 534 hours at 730°C are depicted in Figure 1. A grain size of ASTM8-9 was measured in Alloys 3 to 6 and ASTM10 in Alloys 1 and 2 arising from the forging practice employed on the small ingots. This would correspond to the fine grain size averaging ASTM8 in conventional 718 attained through larger forge reductions at or near the δ phase solvus temperature which results in higher tensile strengths and LCF properties.[4] In general, the grain size of the six alloys was small and the grain boundaries contained a certain amount of δ and MC phases. Transmission electron micrographs identified the γ' and γ'' phases and their morphology after the dual heat treatment, and Figure 2 illustrates such structures in the better performing Alloys 1, 3, 5 and 6. It is apparent that Cozar and Pineau's compact morphology of cubic shaped γ' particles coated on their six faces with a shell of γ'' precipitates was not attained in any of the six alloys. However, only Alloys 5 and 6 possessed the minimal (Al+Ti)/Nb ratio between 0.9 and 1.0 which would allow the compact morphology to be attained according to Cozar and Pineau[1] plus the fact that their alloy that did so contained 7.0 wt pct Mo.

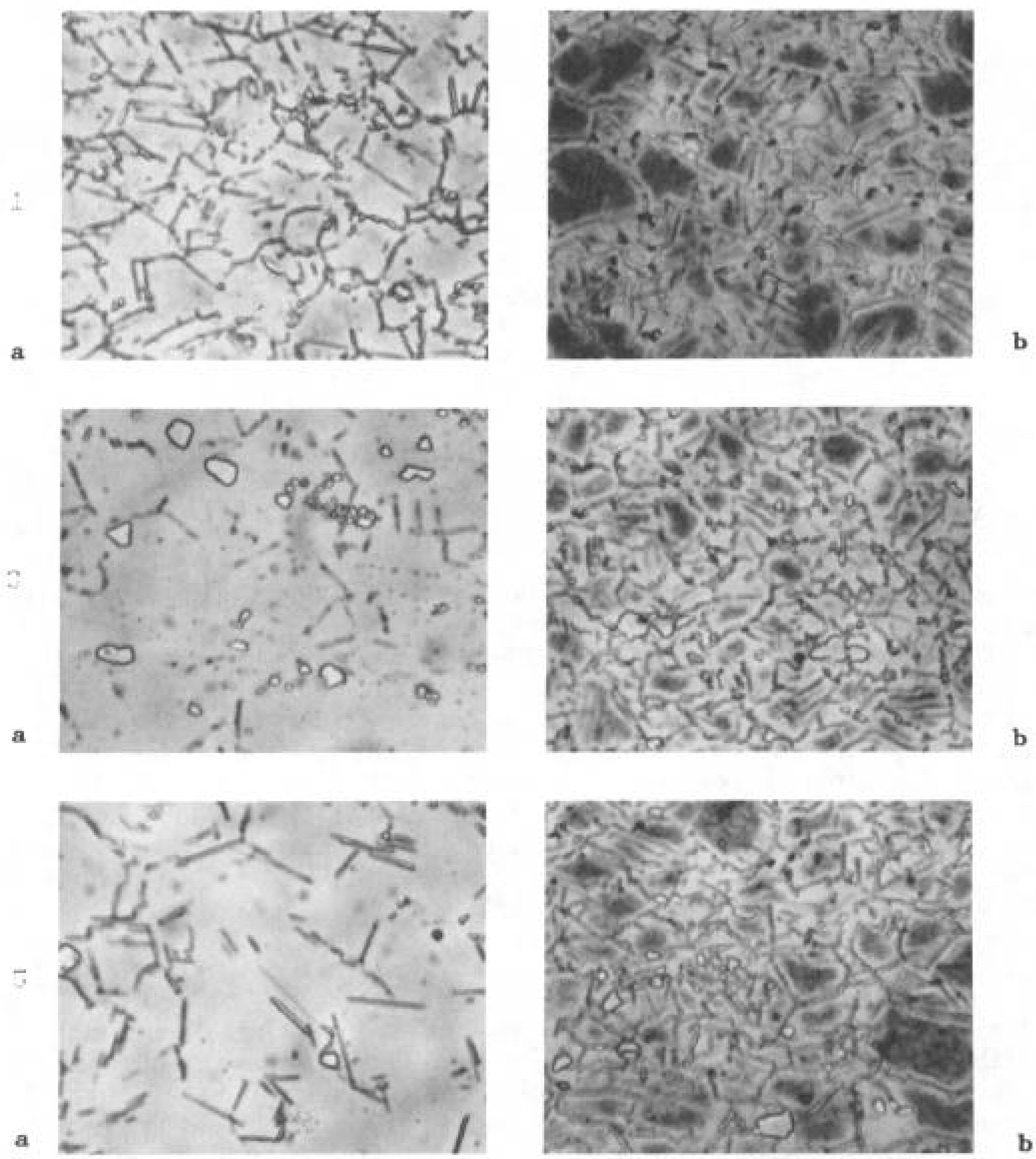

Figure 1 - Optical micrographs of Alloys 1, 3 and 5 after (a) normal heat treatment. (b) plus 534 hr at 730^{0}C. X100.

The size variations of the γ' and γ'' precipitates determined by the TEM analysis are presented in Figure 3. In Alloys 1 to 4, the γ' diameter was larger in Alloys 1 and 3 vs. Alloys 2 and 4 but the γ'' length was shorter after the dual heat treatment. Alloys 5 and 6 with W addition produced the same trends as Alloys 1 and 3. The γ'' was longest in Alloys 2 and 4 and shortest in Alloy 5, although one could consider that the γ'' lengths of Alloys 3, 5 and 6 are comparable. The average length of the γ'' particles was reduced by approximately 30 to 60 pct in their comparison with Alloys 2 and 4. The γ'' lengths of the alloys appear to be related to the γ' diameter; the smaller the γ' diameter, the longer is γ''. The comparison of the γ' and γ'' size variations with the weight percentage of the deleterious δ phase is shown in Table III.

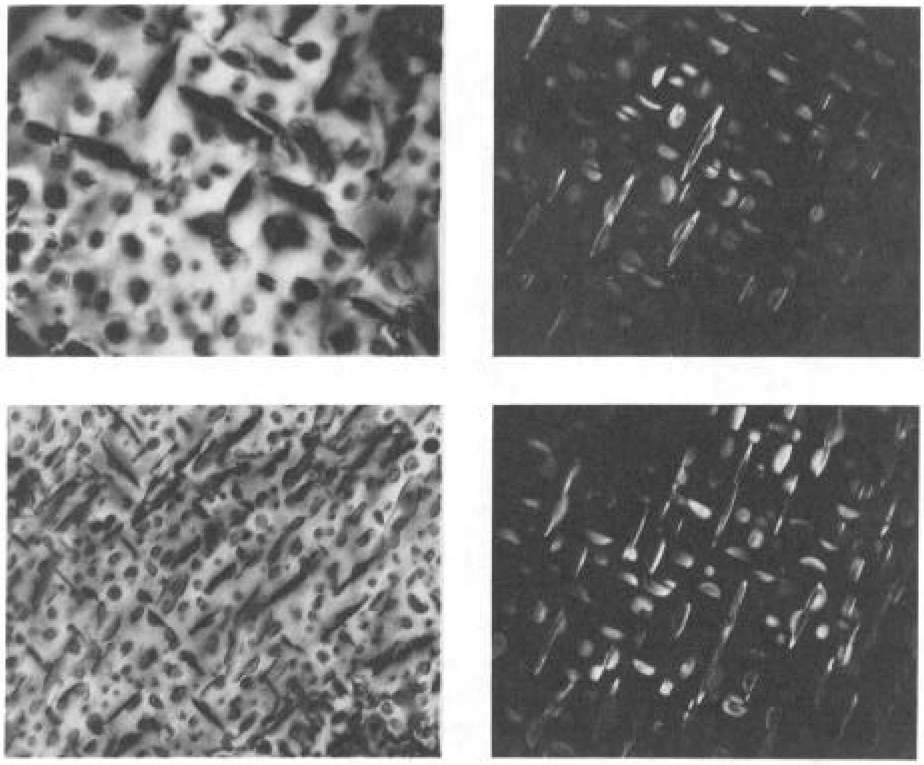

Figure 2 - Transmission electronmicrographs of Alloys 1, 3, 5 and 6 after normal heat treatment plus 534 hr at 730°C. (a) Alloy 1, bright field, X48000. (b) Alloy 3, dark field, X37000. (c) Alloy 5, bright field, X28000. (d) Alloy 6, dark field, X37000.

It can be seen that Alloys 1 and 3 had significantly less δ phase than Alloys 2 and 4 while Alloys 5 and 6 had the least amount of δ phase in concert with their smaller γ'' lengths. Because of the crystal symmetry between the close packed (112) plane of the γ'' phase and the close packed (010) plane of the δ phase, it is believed that the γ'' particles act as nucleation sites for the δ phase.[5] Although the extraction method could not separate them, the larger amounts of $\gamma' + \gamma''$ precipitation in Alloys 1, 3, 5 and 6 are noteworthy. The highest values in Alloys 5 and 6 can be attributed to the W addition being able to change the disposition of refractory elements in each phase, thereby modifying lattice mismatches between γ–γ', γ'–γ'' and γ–γ''.

The variations in hardness after nine holding times up to 534 hours at 730°C are shown in Figure 4. It is noteworthy that higher hardness was obtained in the as heat treated Alloys 1 to 4 compared to Alloys 5 and 6 containing W but with a reduced Fe content. The high initial hardness of Rc 48.5 to 50, which correlates with the tensile strength, is advantageous and then the gradual decline in hardness of Alloys 1 and 3 to Rc 44 compared to Alloy 2 and 4 which decreased to Rc 39-40 after 534 hours at 730°C should be noted. Also, Alloys 5 and 6 possessing the lowest (but still acceptable) as heat treated hardness of Rc 45-46 surprisingly increased in hardness linearly to the highest Rc 48-50 level after 534 hours at 730°C. When coupled with the

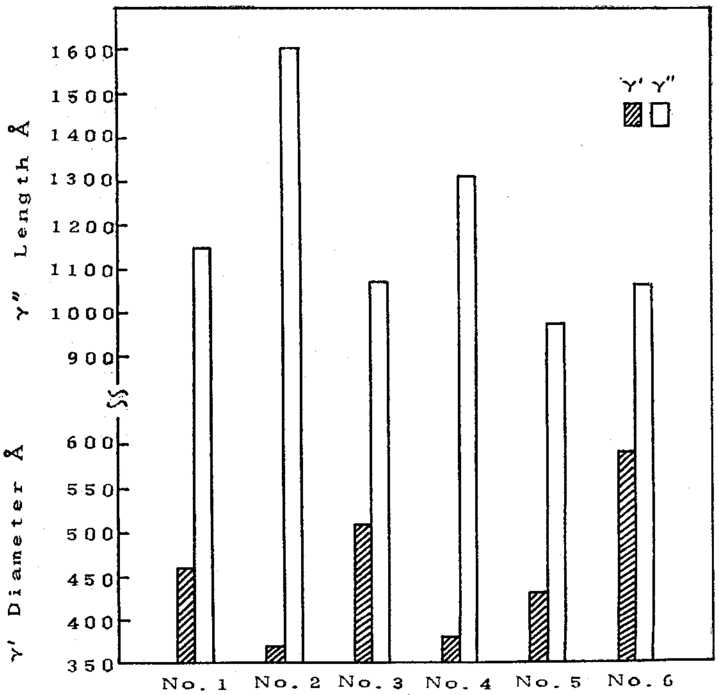

Figure 3 - γ' and γ'' sizes aged for 534 hr at 730^{0}C.

observed smaller γ'' particles and reduced amounts of δ phase, these results indicate that a higher amount of the high temperature, heat resistant Nb (5.6 wt pct) can be added into the alloy without corresponding loss in stability, provided that the composition is properly balanced towards higher (Al+Ti)/Nb and Al/Ti ratios.

The as heat treated tensile and yield strength and the decline in these strength values after aging for 534 hours at 730°C are shown in Figure 5. Alloys 1 and 3 maintained their initial tensile strength significantly better than Alloys 2 and 4 and the hardness trends were duplicated. This was also the case for Alloys 5 and 6 which produced the lowest tensile strength values as heat treated which then rose to the highest values after the aging treatment, and this is attributed to the solid solution W in these two alloys which provided a more significant component of the overall strength at 730°C than at room temperature. The as heat treated tensile strength range of 1300 to 1445 MPa (188.5 to 210 ksi) for all six alloys compares favorably with the

Table III. Size Variations in γ' and γ'' Phases and Weight Percentage of δ Phase, $\gamma' + \gamma''$ Phases and NbC + TiC.

Alloy	γ' Diameter,Å	γ'' Length,Å	δ, Wt Pct		$\gamma'+\gamma''$ Wt Pct		NbC + TiC, Wt Pct	
	A+B	After A+B	After A	A+B	After A	A+B	After A	A+B
1	465	1150	0.36	1.38	16.57	17.50	0.46	0.47
2	352	1600	0.67	2.15	13.24	13.29	0.47	0.50
3	520	1060	0.29	1.42	15.67	17.42	0.53	0.54
4	363	1300	0.46	3.25	14.18	12.74	0.52	0.53
5	435	980	0.36	1.32	16.46	19.27	0.48	0.44
6	590	1060	0.51	0.84	15.78	19.45	0.50	0.46

Heat Treatment A = 935°C, 1 hr, AC, 720°C, FC to 620°C, 8 hr, AC.
Thermal Treatment B = Heat Treatment A + 730°C for 534 hr, AC.

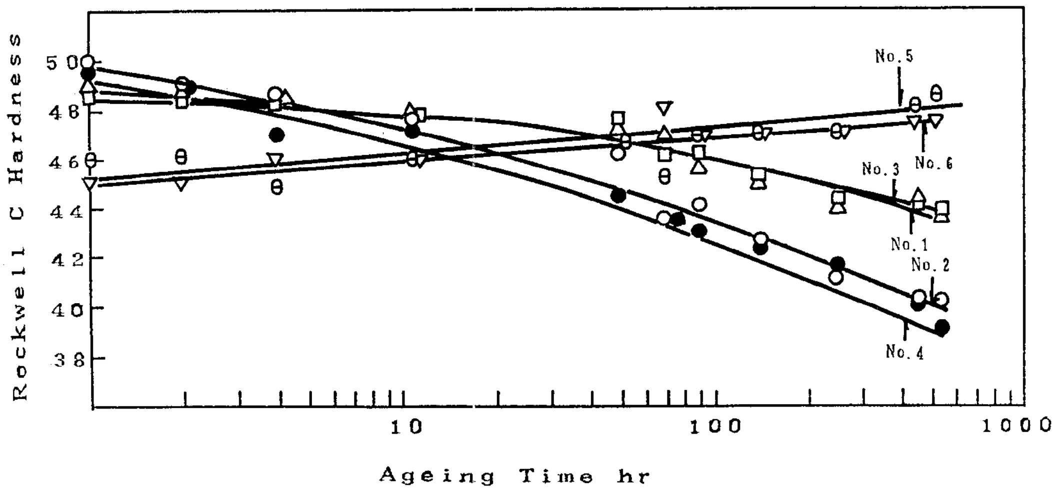

Figure 4 - Hardness variation with aging time at 730°C.

typical 1344 MPa (195 ksi) reported by Barker et al.[4] for high strength processing of conventional 718. There is a larger variation in yield strength among the alloys in the as heat treated condition with the values coming closer together when tested at room temperature after long time aging, and with predominantly the same trends noted for tensile strength. The corresponding elongation values depicted in Figure 6 show a variation between 22 and 30 pct as heat treated and a decrease within a range of 20 to 23 pct for all six alloys which is satisfactory.

Fine grain size, which also provided a finer δ phase, will provide higher strengths under lower temperature-higher stress conditions, but at high temperature, coarse grain size will reduce grain boundary sliding and be beneficial to creep strength. Thus, the stress rupture properties were obtained from specimens given a higher solution treatment of 1030°C instead of 955°C and an ASTM5-6 grain size was produced. The rupture life and ductility values for tests conducted at 700°C under high stresses of 70, 65, 63 and 60 kg/mm^2 (99.5, 92.5, 89.6 and 85.4 ksi) are listed in Table IV. Most of the alloys would not be expected to meet a 25 hour life at 700°C under a high stress of 70 kg/mm^2 (99.5 ksi) when the 0.2 pct yield strength at 700°C is around 80 kg/mm^2 (113.8 ksi). However, it is noteworthy that Alloy 1 did survive 23.3 hours. When the alloys are compared under a stress of 65 kg/mm^2 (92.5 ksi) at 700°C, the best performers were Alloys 1, 3 and 5 in that they survived from 35 to 53 hours. Again, they are the compositions with higher (Al+Ti)/Nb ratio and total hardening element content and usually higher Al/Ti ratio. The only published result on Alloy 718 from the 1970 study by Barker et al.[6] lists a rupture life of 18 hours at 700°C under a lower stress of 85 ksi (~60 kg/mm^2). At this stress level and temperature, Alloy 5 survived for 74.3 hours, and it is recognized that it contained tungsten as an additional component of the overall strength.

Discussion

This preliminary investigation on modifying Alloy 718 in the direction of increasing (Al+Ti)/Nb, Al/Ti and total hardener (Al+Ti+Nb) content indicates the good possibility of improving upon the thermal stability of

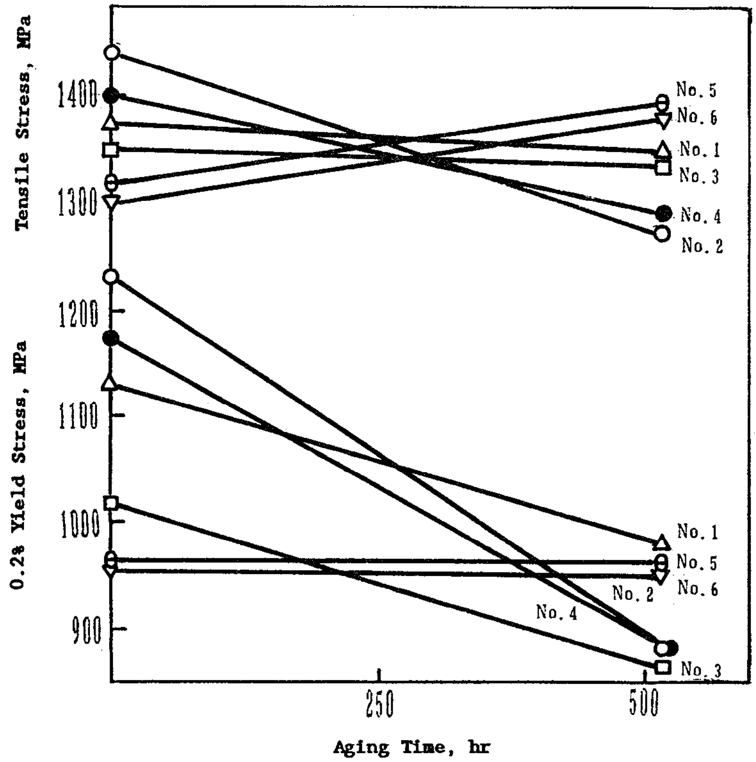

Figure 5 - Tensile and yield strength after precipitation hardening treatment plus aging for 534 hr at 730°C.

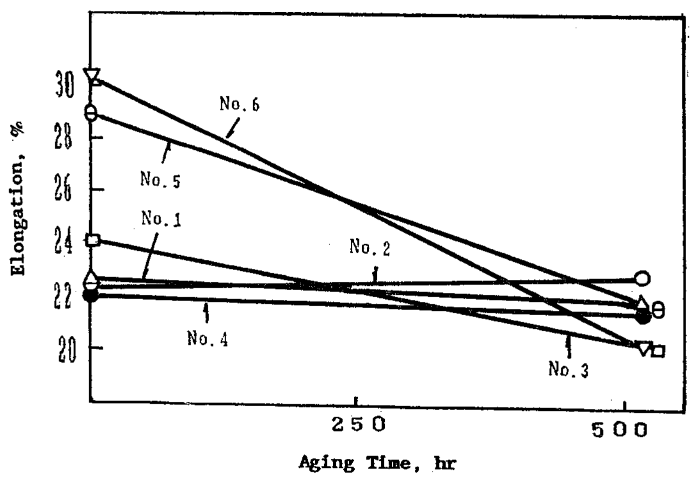

Figure 6 - Tensile elongation after precipitation hardening treatment and then after aging for 534 hr at 730°C.

conventional 718 above 650°C. The direction that appears promising on the basis of a commercial heat treatment plus aging at 730°C for 534 hours is a (Al+Ti)/Nb ratio approaching unity and a higher total (Al+Ti+Nb) content of 6.5 to 6.8 at pct. Alloys 1, 3, 5 and 6 are the basis for this conclusion as well as the beneficial effect of a higher Al/Ti ratio between 1.6 and 1.7 in three of these alloys. The compositional balance producing a high Al/Ti and (Al+Ti)/Nb ratio appears logical even though there are those who favor

Table IV. Stress Rupture Properties at 700°C (1300°F).

Alloy	Stress kg/mm²	Stress ksi	Rupture Life, hr	Ductility El.,%	Ductility R.A.,%
1	70	99.5	23.3	3.0	17.4
4	70	99.5	8.9	5.7	13.1
1	65	92.5	46.0	2.0	8.6
2	65	92.5	36.0	21.5	57.3
2	65	92.5	35.6	12.2	58.9
3	65	92.5	26.8	23.6	49.7
3	65	92.5	38.7	8.0	19.4
4	65	92.5	28.6	5.9	19.0
4	65	92.5	19.6	2.1	15.0
5	65	92.5	40.1	13.6	50.2
5	65	92.5	52.9	16.7	58.0
4	63	89.6	44.8	14.4	59.4
4	63	89.6	65.7	25.1	57.0
5	63	89.6	52.2	14.0	55.4
5	63	89.6	48.5	22.7	55.4
4	60	85.3	43.7	14.2	61.0
5	60	85.3	74.3	13.6	51.0

titanium as the choice to promote a γ' precipitate.[7] These alloys provided the lowest ratio of γ'' coarsening and the least amount of transformation to δ phase. The base compositions of Alloys 3, 5 and 6 are nearly the same as Alloys 11 and 12 which were the best performers in Collier and Tien's study while Alloy 1 resembles their Alloy 4 which was also a good performer. In a microstructural comparison, recognizing the differences in the aging treatments (534 hr at 730°C vs. 100 hr at 760°C), the γ' diameters are practically identical but the γ'' lengths are increased by about 50 pct because of the five times longer interval at nearly the same temperature plus the fact that the subject alloys were in the precipitation hardened state when subjected to this treatment.

The objectives of reducing the amount of overaging of the γ'' particles and reducing the driving force to form δ above 650°C which will degrade alloy properties have been realized in this preliminary study. By maintaining their as heat treated hardness and tensile properties after exposure to 534 hours at 730°C, Alloys 1, 3, 5 and 6 show the beneficial effect of an increased (Al+Ti)/Nb ratio and three of these alloys also confirm the beneficial effect of an increased Al/Ti ratio. Although Alloy 1 has mechanical stability with a high (Al+Ti)/Nb but a low Al/Ti ratio, it should be noted that Alloy 4, which contains the conventional amounts of Al and Ti for Alloy 718 but a higher Nb content, thereby resulting in a lower Al/Ti and (Al+Ti)/Nb ratio, had inferior properties and contained the greatest amount of δ phase. It should also be recognized that the better performing compositions had a higher total amount of hardening elements, Al+Ti+Nb, resulting in the proper combination of γ' and γ'' effects plus the solid solution strengthening effect of W in the case of Alloys 5 and 6.

Collier and Tien maintain that alloys with increased (Al+Ti)/Nb and/or Al/Ti ratios possess enhanced mechanical stability because the loss in γ'' volume fraction is accompanied by an increase in γ' fraction. These measurements were not made in this preliminary study. What has been found is that, even though the rate of γ' coarsening was faster, the coarsening rate of γ'' and the decline in strength occurred more slowly in the four alloys with a higher (Al+Ti)/Nb ratio and with less transformation to δ phase than was the case in the two lower (Al+Ti)/Nb ratios alloys.The two alloys having both high (Al+Ti)/Nb and Al/Ti ratios and a tungsten addition for additional solid solution strengthening produced increasing strength and a slower rate of γ'' coarsening after 534 hours at 730°C. A larger γ' size and smaller γ'' length in the four best alloys could mean that more γ' is being produced while stabilizing γ'' at the same time. Finally, when subjected to a high stress of 65 kg/mm^2 (92.5 ksi) at 700°C, Alloys 1, 3 and 5 had rupture lives between 36 and 53 hours compared to 18 hours reported for conventional 718 at 700°C under a lower stress of 60 kg/mm^2 (85 ksi). Although more stress rupture testing is necessary, these are promising indications that the microstructural changes which occur more rapidly under stress at elevated temperature are similar to those occurring solely on the basis of the temperature sojourn and then tensile testing at room temperature.

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