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RELATIONSHIP OF MECHANICAL CHARACTERISTICS AND MICROSTRUCTURAL FEATURES TO THE TIME-DEPENDENT EDGE-NOTCH SENSITIVITY OF INCONEL 718 SHEET

by

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ABSTRACT

Time-dependent notch sensitivity of Inconel 718 sheet was observed at 900°F to 1200°F (482 - 649°C). It occurred when edge-notched specimens were loaded below the yield strength and smooth specimen tests showed that small amounts of creep consumed large rupture life fractions. The severity of the notch sensitivity was reduced by decreasing the solution temperature, increasing the time and/or temperature of aging and increasing the test temperature to 1400°F (760°C). Elimination of time-dependent notch sensitivity correlated with a change in dislocation motion mechanism from shearing to by-passing precipitate particles.

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INTRODUCTION

The results presented were derived from a study of the severe-time-dependent edge-notch sensitivity that has been observed for nickel-base superalloy sheet materials at temperatures from 900° to 1300°F (482 - 704°C). The research was carried out at The University of Michigan, Ann Arbor, Michigan, under sponsorship of the National Aeronautics and Space Administration.

Extensive research (1, 2) established the scope and the cause of the problem for Waspaloy. In addition, heat treatments were defined which eliminated the time-dependent notch sensitivity. Continuing research is directed at broadening the applicability of the concepts developed for Waspaloy. To achieve this, the study is being extended to include other alloys. The experiments reported deal with results obtained for Inconel 718. The composition of this nickel-base superalloy differs considerably from that of Waspaloy. Of major significance, columbium is present in Inconel 718 but not in Waspaloy. This element has a marked influence on phase relationships and hence, on the microstructures and mechanical behavior.

As part of an evaluation of potential usefulness of various superalloys in sheet form for construction of the Supersonic Transport, Inconel 718 in three heat treated conditions was shown to exhibit time-dependent notch sensitivity at 1000° and 1200°F (538 and 649°C) (3). The research presently reported was designed to extend the scope of these initial results. Heat treatments were selected to provide a wide range of microstructural features. These were also expected to produce considerable variation in mechanical characteristics. Tensile and creep-rupture tests of smooth and sharp-edged (K_t>20) notched specimens were conducted at temperatures from 900° to 1200°F (482 - 649°C) where severe time-dependent notch sensitivity can occur. Testing was also carried out at 1400°F (760°C) where the notch sensitivity has not been observed. The microstructural features, particularly the dislocation mechanisms in the tested specimens, were evaluated.

EXPERIMENTAL DETAILS

The commercially produced Inconel 718 used in the investigation had the following reported composition (weight percent):

Ni	C	Mn	Fe	S	Si
53.97	0.05	0.12	16.50	0.007	0.22

Cr	Al	Ti	Co	Mo	Cb	B
18.98	0.52	1.04	0.05	3.15	5.25	0.002

The material was received as 0.030-inch (.75mm) thick cold reduced sheet. Specimen blanks were cut in the longitudinal direction prior to heat treatment as follows:

	Solution Treatment		Aging Treatment
1.	10 hours at 1950°F(1066°C)	+	48 hours at 1350°F (732°C)
2.	1 hour at 1950°F(1066°C)	+	48 hours at 1350°F (732°C)
3.	1 hour at 1950°F(1066°C)	+	2 hours at 1550°F (843°C)
4.	1 hour at 1950°F(1066°C)	+	24 hours at 1550°F (843°C)
5.	10 hours at 1800°F(982°C)	+	48 hours at 1350°F (732°C)
6.	10 hours at 1700°F(927°C)	+	3 hours at 1325°F (718°C)
7.	10 hours at 1700°F(927°C)	+	48 hours at 1350°F (732°C)
8.	1 hour at 1700°F(927°C)	+	3 hours at 1325°F (718°C)
9.	1 hour at 1700°F(927°C)	+	2 hours at 1550°F (843°C)

It should be noted that although the higher temperature exposures are referred throughout this paper as "solution treatments", the use of this designation does not necessarily signify complete solution of all constituent phases. The heat treatment practice and the testing procedures were the same as those used previously (1).

Conventional methods were employed for microstructural examination. Samples for optical microscopy and replica electron microscopy were etched electrolytically in "G" etch (4). Samples approximately 0.5 inches (1.3cm) wide by 0.7 inches (1.8cm) long for transmission electron microscopy of the tested specimens were cut from the gauge lengths, ground on wet silicon carbide papers and electropolished. This was carried out using 20 volts potential in conjunction with a chilled mixture of 83 percent Methanol, 7.5 percent Sulphuric Acid, 3 percent

Nitric Acid, 4.5 percent Lactic Acid and 2 percent Hydrofluric Acid. The thin films were studied in a JEM electron microscope operated at 100KV.

X-ray diffraction analysis was carried out on precipitate particles electrolytically extracted from the as-heat treated materials using a platinum cathod at 3-4 volts potential and a 10 percent Phosphoric Acid in Water Solution. X-ray exposures were conducted in a 144.6mm diameter Debye camera with nickel-filtered copper radiation.

EXPERIMENTAL OBSERVATIONS

Influence of Heat Treatment on the Mechanical Characteristics

The heat treatments evaluated resulted in a wide range of mechanical characteristics (Tables I through III). In particular, the severity of the time-dependent notch sensitivity varied considerably with changes in both the solution and aging treatments. The results for the material heat treated 10 hours at 1950°F (1066°C) plus 48 hours at 1350°F (732°C) (Fig. 1, 2) provide an example of severe time-dependent notch sensitivity. At the shorter times at the lower test temperatures, the notched specimen rupture curves were somewhat below those for smooth specimens. The notched to smooth rupture strength ratios (N/S) were the same order as determined by tensile tests (Table III). At varying time periods, the notched specimen rupture curves exhibited drastic increases in steepness so that the N/S ratios decreased with time to values considerably below those obtained in tensile tests, i.e. the material exhibited time-dependent notch sensitivity. At intermediate temperatures, upward breaks occurred in the rupture curves. This resulted in increases in the N/S ratios or decreases in notch sensitivity. The rupture strength ratios increased with increasing test temperature and longer rupture times, until, at the highest temperature a value of about 1.0 was obtained.

The results for the material heat treated 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C) are in contrast to the above behavior and are typical of the heat treated materials not susceptible to time-dependent notch sensitivity (Figs. 3, 4). The N/S rupture strength ratios were generally higher than those obtained in tensile tests (Table III).

The severity of the time-dependent notch sensitivity for the heat treated conditions studied, including those evaluated previously (3) are tabulated below. Ratings of 2 and 1 correspond to severe and limited time-dependent notch sensitivity respectively. Test conditions for which no time-dependent notch sensitivity was apparent were rated 0.

Solution Treatment	Aging Treatment(s)	Test Temperature				
		°F 900	1000	1100	1200	1400
		°C 482	538	593	649	760
10hr. 1950°F(1066°C)	+ 48hr. 1350°F(732°C)	2	2	2	2	0
1hr. 1950°F(1066°C)	+ 48hr. 1350°F(732°C)	2	2	2	2	0
	+ 2hr. 1550°F(843°C)	2	2	2	1	0
	+ 24hr. 1550°F(843°C)		0	0	?	0
10hr. 1800°F(982°C)	+ 48hr. 1350°F(732°C)	2	2	1	?	0
10hr. 1700°F(927°C)	+ 3hr. 1325°F(718°C)	2	1	0	0	0
	+ 48hr. 1350°F(732°C)	0	0	0	0	0
1hr. 1700°F(927°C)	+ 3hr. 1325°F(718°C)	2	2	?	0	0
	+ 2hr. 1550°F(843°C)		0	0	0	0
Cold Worked 20%	+ "Multiple" 1*		2		2	
1hr. 1950°F(1066°C)	+ "Multiple" 2		2		2	
1hr. 1750°F(954°C)	+ "Multiple" 1	2	2	1	?	

*"Multiple" 1 - 1325°F(718°C)/8 hours, F.C. to 1150°F (621°C) in 10 hours, A.C.

"Multiple" 2 - 1350°F(732°C)/8 hours, F.C. to 1200°F (649°C) in 12 hours, A.C.

The principle features evident are as follows:

(1) Decreasing the solution temperature decreased the susceptibility to time-dependent notch sensitivity.

(2) Increasing the severity of the aging treatment (increasing time and/or temperature) reduced the susceptibility to time-dependent notch sensitivity.

(3) The materials with the commonly used "multiple" aging treatments exhibited time-dependent notch sensitivity.

For all heat treated conditions, the rupture strengths decreased rapidly with increasing time and/or temperature for parameter values above about 37,000 (Fig. 1 and 3). Extensive use is made of Inconel 718 for "high temperature" applications for which the parameter values are relatively low. Under these conditions, the smooth specimen rupture strengths are high, however, severe time-dependent notch sensitivity can occur. At the lower parameter values, the notched rupture strengths were below those for smooth specimens and varied considerably with heat treatment. (In contrast, the range of smooth specimen rupture strengths was relatively small.) Particularily important, was whether or not the material exhibited time-dependent notch sensitivity. Those which did not had similar notched rupture strengths while those which were notch sensitive had considerably lower strengths. It should also be noted that the commonly used "multiple" aging treatments evaluated previously, resulted in lower notched rupture strengths than obtained for a number of the heat treatments used in the present investigation. This occurred even though the heat treatments used were not selected to maximize notched rupture strengths.

Rupture ductility has often been used to indicate the susceptibility of a material to notch sensitivity. Results reported for Waspaloy (1) indicated that no such relationship occurred which was generally applicable. The same conclusion was drawn from analysis of the elongation and reduction of area values at rupture for Inconel 718 (Table II). For both alloys, the results indicated that factors which contribute to ductility rather than the ductility per se, control the time-dependent notch sensitivity. This was evident from analysis of the deformation-time characteristics (reported in a later section for Inconel 718).

Fracture Characteristics

The fracture characteristics of Inconel 718 were similar to those established for Waspaloy (1). Consequently, this aspect is not reported in depth. Both smooth and notched rupture tested specimens failed by initiation and relatively slow growth of intergranular cracks followed by transgranular fracture. The latter fracture occurred when the increase in stress on the load bearing area, due to growth of the intergranular crack, exceeded that necessary to cause rapid shear. In consequence, the lengths of the intergranular cracks (expressed as a percentage of specimen width in Tables II and III) increased with decreasing test stress and thus with increasing time. The occurrence of time-dependent notch sensitivity was due to more rapid initiation of intergranular cracks in notched than in smooth specimens. This was associated with local deformation at the base of the notches accompanying the relaxation of the stress concentrations.

Stress Relaxation

Relaxation of stress concentrations can occur by "yielding" on loading (time-independent deformation) and by subsequent creep (time-dependent deformation). For notched specimen tests loaded above the approximate 0.2 percent offset yield strengths (established by smooth specimens tests) no time-dependent notch sensitivity was observed (Figs. 1, 3). This occurred since "yielding" on loading reduced the stresses across the specimens at the base of the notches to approximately the nominal stresses.

For Waspaloy time-dependent notch sensitivity occurred in notched specimens loaded below their yield strength when tests of smooth specimens showed that small amounts of creep consumed large fractions of creep-rupture life (1). In other words, when the creep deformation necessary to relax stress concentrations caused excessive damage resulting in premature initiation of intergranular cracks. The smooth specimen deformation characteristics of Inconel

718 were examined to determine whether a similar correlation existed. Iso-creep strain curves were constructed for each temperature on plots of life fraction versus test stress. These curves were derived for 0.1; 0.2 (the order of deformation necessary to ensure relaxation of elastic stresses from the approximate yield stress to the nominal stresses), and also for 0.5, 1, and 2 percent creep strain.

Again, a correlation was evident between the time-dependent notch sensitivity and the characteristics of the iso-creep strain curves. For the material heat treated 10 hours at 1950°F (1066°C) plus 48 hours at 1350°F (732°C), the life fractions for small amounts of creep strain at 1000°F (538°C) increased drastically as the test stress decreased (Fig. 5). For notched tests loaded to nominal stresses below about 110 ksi, the relaxation of the stresses (from the approximate yield stress to the nominal) would consume considerable, if not all, of the creep-rupture life of the material at the base of the notch. Thus, as observed experimentally (Figs. 1, 2), time-dependent notch sensitivity would be expected. At 1200°F (and presumably 1100°F) as the test stress decreased, the life fractions for 0.1 and 0.2 percent strain increased to relatively high levels and subsequently decreased. This is consistent with the observed increase followed by a decrease in notch sensitivity. At 1400°F (760°C) the life fractions were at relatively low levels and no time-dependent notch sensitivity occurred.

For the material heat treated 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C), the life fractions for small amounts of creep strain remained at low levels for all test conditions (fig. 6). In accordance with this, no time-dependent notch sensitivity was observed (Figs. 3, 4).

Data for other heat treated materials showed similar correlations between the nature of the iso-creep strain curves and the time-dependent notch sensitivity behavior.

Microstructural Features Contributing To Time-Dependent Notch Sensitivity

Many nickel-base superalloys (including Waspaloy) age harden by precipitation of an LI_2 - ordered fcc phase, $Ni_3(Al, Ti)$, called gamma prime. Inconel 718 differs in that age hardening occurs primarily due to precipitation of Ni_3Cb (5). This phase, designated γ'', has a DO_{22} - ordered bct structure. The γ'' phase is metastable so that it is replaced on thermal exposure by the β phase. This phase is also Ni_3Cb, but it has a Cu_3Ti - ordered orthorhombic structure.

In the study of Waspaloy (2), a correlation was established between the dislocation motion mechanism operative and the time-dependent notch sensitivity. Dislocations sheared γ' particles smaller than a critical size. Particles larger than the critical were by-passed by dislocations. The former mechanism promoted the deformation-characteristics that gave rise to the time-dependent notch sensitivity. The microstructural features of Inconel 718 were studied primarily to determine whether a similar correlation occurred.

Original Microstructures

Increasing the solution treatment time at 1950°F (1066°C) from 1 to 10 hours resulted in an increase in grain size (0.03mm to 0.06mm). This reflects the absence of large amounts of precipitate particles which would act to restrain growth. Ti(C, N) is the only precipitate expected to be present at 1950°F (1066°C). Presumably, these are the large particles evident in the optical micrograph for the material aged 48 hours at 1350°F (732°C) after solution treatment at 1950°F (1066°C) (fig. 7a). X-ray diffraction of extracted residues indicated the presence of Cb, Ti(C, N), γ' and/or* γ'' in these materials (Table IV). These latter phases were not resolvable in the

*The γ' and γ'' phases are difficult to distinguish using X-ray diffraction. The "d" values are similar. Differences do occur in the superlattice reflections but these are not readily resolvable. (5).

electron microscope using replica techniques. In thin films, the carbide particles observed were primarily present as "plate-like" grain boundary precipitates while the γ' and/or γ'' were intragranular precipitates about 300 Å in diameter (fig. 7b). The presence of γ'' was demonstrated by electron diffraction (6). The presence of γ' was inferred from subsequently reported metallographic observations for materials aged at 1550°F(843°C). (In cases such as this where the γ' and the γ'' could not be distinguished visibly, the precipitate will be referred to as γ'/γ''.)

Aging 2 or 24 hours at 1550°F (843°C) after solution treatment at 1950°F (1066°C) resulted in γ' and γ'' particles large enough to be resolvable using replica techniques (Fig. 8).

The γ' was present as spherical particles with a relatively low volume fraction. The average size was about 450 Å and 1100 Å for the 2 and 24 hour treatments respectively. The majority of the precipitate particles were plates of γ'' (precipitates coherently with the c-axis normal to the plane of the plates and along any of the three <110> fcc directions - ref. 5). The approximate average thickness and length of the plates were respectively 200 Å and 1000 Å for the 2 hour treatment and 500 Å and 4000 Å for the 24 hour treatment.

X-ray diffraction indicated the presence of small amounts of β phase for the materials aged at 1550°F(843°C) after solution treatment at 1950°F (1066°C)(Table IV). In micrographs, this phase was evident as needles, predominately alongside grain boundaries. The areas adjacent to grain boundaries and β precipitate particles were depleted of γ'' (Fig. 8).

The majority of the precipitate present in the optical micrographs of the material heat treated 10 hours at 1800°F (982°C) plus 48 hours at 1350°F (732°C) (Fig. 9a) was β phase. This phase formed during the 1800°F (982°C) solution treatment (Fig. 9b). X-ray diffraction showed that the aged material also contained Cb,Ti(C, N), γ' and/or γ'' (Table IV).

All of the materials solution treated at 1700°F (927°C) and aged contained Cb,Ti(C, N), γ'/γ'' and the β phase (Table IV). Ni_3Cb needles precipitated during the 1700°F (927°C) treatments (fig. 10 a). A much larger amount of β phase was present after the 10 hour exposure than the 1 hour treatment (Fig. 10 b). Aging 3 hours at 1325°F (718°C) or 48 hours at 1350°F (732°C) after solution treatment at 1700°F (927°C) resulted in small γ'/γ'' particles. Even in thin films (Fig. 10 c), resolution was difficult because the γ'/γ'' particles were only about 60 Å and 200 Å in diameter for the 1325°F (718°C) and 1350°F (732°C) treatments, respectively. These particle sizes are smaller than those produced by similar aging treatments after solution treatment at 1950°F (1066°C). {This also occurred for the 2 hour at 1550°F (843°C) aging treatment.}

Microstructures of Tested Specimens

Examination of tested specimens by transmission electron microscopy was carried out primarily to determine the dislocation structures present. The observations were made for selected heat treatments and test conditions. However, they were expected to be representative of all tested materials.

(1) Material heat treated 1 hr. at 1950°F (1066°C) plus 48 hrs. at 1350°F (732°C)

For the specimen tested at 1100°F (593°C) {at 120 ksi (827MN/m^2) ruptured in 1.4 hours} the most obvious feature was {111} planar slip banding (Fig. 11 a). This reflects shearing of the γ'/γ'' coherent precipitates by dislocations (5, 7). Similar dislocation structures would be present in other specimens tested at temperatures low enough so that little or no growth or γ'/γ'' occurred.

It was evident from microstructures of the specimen tested at 1400°F (760°C) {at 30 ksi (207 MN/m^2) ruptured in 384 hours } that structural changes had occurred during the test exposure. Ni_3Cb needles precipitated and the γ' particles increased in size to about 750 Å. The γ'' also grew so that it was clearly resolvable as plates approximately 500 Å thick and 4000 Å long. Contrast effects associated with coherency (8) were observed for both γ' and γ'' precipitate particles. Because of the presence of large precipitates, the dislocations by-passed the particles and the deformation was homogeneous (Fig. 11b). Dislocations were observed entangled with the γ'' particles and in some cases as loops around the γ'. It must be assumed that in the early part of the test when the particles were small, the dislocations sheared the particles, i.e. the microstructure would have been similar to Figure 11a.

The above observations are analogous to those reported for Waspaloy (2). In this case, dislocations sheared γ' particles smaller than a critical size. When the particles were larger than the critical, they were by-passed by dislocations. These mechanisms resulted in localized and homogeneous deformation respectively.

(2) <u>Material heat treated 1 hour at 1950°F (1066°C) plus 2 hours at 1550°F (843°C):</u>

The deformation that occurred in the specimen tested at 1100°F (593°C) {at 100 ksi (690MN/m^2) ruptured in 385 hours} was localized in slip bands. Presumably, dislocations sheared the γ' particles (about 450 Å in diameter) and also the γ'' (200 Å thick and 1000 Å long). The previously described results would indicate that growth of the precipitates during higher temperature tests would cause dislocations to by-pass the particles and thus the deformation would become homogeneous.

One additional feature was evident from the study of the specimen tested at 1100°F (593°C). In a number of micrographs, a fine precipitate (about 70 Å in diameter) was detected (Fig. 12). Presumably, this is γ'/γ'' that formed subsequent to the 1550°F (843°C) aging treatment and developed during the test exposure.

(3) <u>Material heat treated 1 hour at 1950°F (1066°C) plus 24 hours at 1550°F (843°C):</u>

The γ'/γ'' in the aged material was larger than "critical" size.

Even in a low temperature test specimen {at 1000°F(538°C) and 115 ksi (793MN/m^2), ruptured in 1857 hours} the dislocations were homogeneously distributed.

A fine dispersion of γ'/γ'' (about 100 Å in diameter) was observed in the specimen tested at 1100°F (593°C) {at 100 ksi (690MN/m^2) ruptured in 3528 hours}. This feature was similar to that described for the material aged 2 hours at 1550°F (843°C).

(4) <u>Material heat treated 1 hour at 1700°F (927°C) plus 3 hours at 1325°F (718°C):</u>

The deformation in the specimen tested at 1000°F (538°C) {at 130 ksi (896MN/m^2) ruptured in 5613 hours} was localized (Fig. 13). In the majority of cases, the dislocations in pile ups, were dissociated to form stacking fault ribbons. This type of deformation was not expected, because in the presence of γ'', it requires coplanar motion of multiple dislocations. Four whole dislocations must move along the same plane to restore order for all three orientations of γ'' (7). (It should be noted that the presence of γ'' was confirmed by electron diffraction.)

Growth of γ'/γ'' occurred during exposure of the specimen tested at 1200°F (649°C) {at 65 ksi (448MN/m^2) ruptured in 937 hours}. As a result, the dislocations by-passed the γ'/γ'' particles (about 250 Å in diameter). This would indicate a much smaller "critical" size than for the materials aged after solution treatment at 1950°F (1066°C). This probably occurred due to differences in the volume fraction of precipitate. Reducing the volume fraction of the γ'/γ'' precipitate should lower the "critical" size (9). Although not determined as part of the investigation, less γ'' was probably present for materials aged after solution treatment at 1700°F (927°C) than for those treated at higher temperatures, e.g. 1950°F (1066°C). This could be expected because precipitation of Ni_3Cb needles during the 1700°F (927°C) treatment must reduce the amount of Cb in solid solution available to form γ'' during aging. Further research is necessary to clarify these effects.

(5) Material heat treated 10 hours at 1700°F (927°C) plus 3 hours at 1325°F (718°C):

Limited study of a specimen tested at 1000°F (538°C) {at 130 ksi (896MN/m^2) ruptured in 391 hours} did not reveal any inconsistencies from the results described above for the material solution-treated 1 hour at 1700°F (927°C) and aged at 1325°F (718°C).

(6) Material heat treated 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C):

For the specimen tested at 1000°F (538°C) {at 120 ksi (827MN/m^2) ruptured in 1382 hours} the deformation was homogeneous. The results indicated that the γ'/γ'' produced by the aging treatment (about 200 Å in diameter) was larger than the "critical" size.

Correlation of the Time-Dependent Notch Sensitivity with the Dislocation Structure

The results indicate that a correlation exists between the predominant dislocation mechanism and the time-dependent notch sensitivity. The relationship was the same as evident for Waspaloy (2). Shearing the precipitate particles by dislocations resulted in greater susceptibility to time-dependent notch sensitivity than when they were by-passed. For the materials heat treated 1 hour at 1950°F (1066°C) plus 48 hours at 1350°F (732°C), 1 hour at 1950°F (1066°C) plus 2 hours at 1550°F (843°C), 1 hour at 1700°F (927°C) plus 3 hours at 1325°F (718°C) and 10 hours at 1700°F (927°C) plus 3 hours at 1325°F (718°C), time-dependent notch sensitivity was observed at the lower test temperatures. During these tests, the γ'/γ'' particles were sheared by dislocations and the deformation was localized. During the higher temperature tests, growth of the γ' and γ'' precipitates occurred. This resulted in a change of dislocation motion to "by-passing" so that a homogeneous distribution of dislocations resulted. This correlates with the elimination of time-dependent notch sensitivity with increasing test temperature.

For the materials heat treated 1 hour at 1950°F (1066°C) plus 24 hours at 1550°F (843°C) and 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C), the dislocations were homogeneously distributed and no time-dependent notch sensitivity was observed.

There was no evidence to indicate that the other heat treated materials, for which tested specimens were no studied, would not follow the above correlation.

It is of interest to compare the behavior of materials with a given aging treatment, e.g. 48 hours at 1350°F (732°C). The time-dependent notch sensitivity was severe for the material solution treated at 1950°F (1066°C). Decreasing the solution temperature to 1800°F (982°C) decreased the notch sensitivity, until for the 1700°F (927°C) treatment, none was observed. This is consistent with the metallographic observation that the "critical" size decreased with decreasing solution temperature. As suggested previously, this probably occurred because lowering the solution temperature reduced the volume fraction of γ'/γ''. This would also explain why heat treatment 1 hour at 1700°F (927°C) plus 3 hours at 1325°F (718°C) resulted in more severe time-dependent notch sensitivity (or occurred at high test temperatures) than for the material heat treated 10 hours at 1700°F (927°C) plus 3 hours at 1325°F (718°C).

The influence of variations in the grain boundary characteristics on the time-dependent notch sensitivity were not evident from the results. Nor was a relationship evident from the study of Waspaloy (2). In both cases, a correlation was evident between the dislocation mechanism and the time-dependent notch sensitive behavior. This, as previously discussed (2), suggests that the influence of the γ'' and/or γ' precipitates on the notch sensitivity overshadows effects from variations in grain boundary characteristics.

Hardness Testing

Results for Waspaloy indicated that room temperature hardness tests could be used to monitor γ' size relative to the critical and hence to predict time-dependent notch sensitive behavior (2). Consequently, hardness tests were also conducted for a range of heat treatments of Inconel 718, including those used in the test program, to determine whether these could be similarly utilized. The results showed the following:

(1) For the heat treatments for which the hardness indicated that the γ'/γ'' size was definitely above or below the critical size (i. e. maximum hardness), the time-dependent notch sensitive behavior agreed with the hardness results. Heat treatment 1 hour at 1950°F (1066°C) plus 24 hours at 1550°F (843°C) resulted in γ'/γ'' larger than the critical size (Fig. 14a) and no time-dependent notch sensitivity occurred. Materials heat treated 1 or 10 hours at 1700°F (927°C) plus 3 hours at 1350°F (732°C) exhibited time-dependent notch sensitivity and the hardness tests indicated that the γ'/γ'' were smaller than the critical size (Fig. 14b).

(2) For many heat treated materials, the hardness tests indicated that the γ'/γ'' sizes were near the critical and therefore, it would not be possible to predict the time-dependent notch sensitive behavior. Again, as was the case for Waspaloy, there was no instance where hardness tests indicated the incorrect notch sensitive behavior.

(3) The times at which the maximum in hardness occurred at each temperature decreased as the solution temperature decreased. This corresponds to the observed decrease in "critical size".

CONCLUSIONS

(1) Time-dependent notch sensitivity of 0.030-inch (.75mm) thick Inconel 718 sheet was observed at temperatures from 900° to 1200°F (482 - 649°C). No reasons were evident why similar behavior could not be expected at prolonged times at lower temperatures. At 1400°F (760°C) ratios of about 1.0 were obtained, i.e. no notch sensitivity was observed.

(2) Time-dependent notch sensitivity occurred when (i) the notched specimen loads were below the approximate 0.2 percent smooth specimen offset yield strength; and (ii) test data from smooth specimens indicated that small amounts of creep used up large fractions of creep-rupture life.

(3) Decreasing the solution temperature or increasing the time and/or temperature of the aging treatment decreased the susceptibility to the time-dependent notch sensitivity. No time-dependent notch sensitivity was observed for materials heat treated 1 hour at 1950°F (1066°C) plus 24 hours at 1550°F (843°C), 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C) and 1 hour at 1700°F (927°C) plus 2 hours at 1550°F (843°C).

Commonly used aging treatments: 1350°F (732°C)/8 hours, F.C. to 1200°F (649°C) in 12 hours, A.C. {after solution treatment at 1950°F (1066°C)} and 1325°F (718°C)/ 8 hours, F.C. to 1150°F (621°C) in 10 hours, A.C. {after solution treatment at 1750°F (954°C)} result in notch sensitive behavior.

(4) Variations in notch sensitive behavior were correlated with changes in the dislocation motion mechanism. Ni_3Cb(bct) particles (and gamma prime) smaller than a "critical size" were sheared by dislocations. This gave rise to localized deformation and time-dependent notch sensitive behavior. Larger particles were by-passed by dislocations and the deformation was homogeneous. Under these conditions, no time-dependent notch sensitivity was observed.

(4) Room temperature hardness tests indicate particle size relative to the "critical" and are, therefore, useful in the prediction of notch sensitive behavior.

(5) Most important, the results showed that the time-dependent notch sensitivity of Inconel 718 could be correlated with the same mechanical characteristics and similar microstructural features as evident for Waspaloy (1, 2). This would suggest even wider applicability of the results.

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LIST OF FIGURES

TABLE 1

TENSILE PROPERTIES OF 0.030 INCH (.75mm) THICK INCONEL 718 SHEET AT 1000° AND 1200°F (538° and 649°C)

Heat Treatment	Test Temperature (°F)	Test Temperature (°C)	Tensile Strength (ksi)	Tensile Strength (MN/m^2)	0.2% Offset Yield Strength (ksi)	0.2% Offset Yield Strength (MN/m^2)	Y.S./T.S.	Elong. (%)	R.A. (%)	Notched Tensile Strength (ksi)	Notched Tensile Strength (MN/m^2)	N/S Tensile Strength Ratio
	Smooth Specimen Properties											
10 hrs. at 1950°F (1066°C)	1000	538	145.4	1002	118.2	817	0.82	15.4	23	149.0	1027	1.02
+ 48 hrs. at 1350°F (732°C)	1200	649	142.4	982	116.5	803	0.82	8.1	14	135.3	933	0.95
1 hr. at 1950°F (1066°C)	1000	538	160.1	1104	129.5	893	0.81	22.5	30	152.4	1051	0.95
+ 48 hrs. at 1350°F (732°C)	1200	649	157.5	1086	130.5	900	0.83	8.4	14	147.5	1017	0.94
1 hr. at 1950°F (1066°C)	1000	538	134.9	930	90.5	624	0.67	27.0	34	113.6	783	0.84
+ 2 hrs. at 1550°F (843°C)	1200	649	133.4	920	99.5	686	0.75	11.3	18	116.0	800	0.87
1 hr. at 1950°F (1066°C)	1000	538	134.2	925				32.8	30	102.1	704	0.76
+ 24 hrs. at 1550°F (843°C)	1200	649	132.6	914	83.5	576	0.63	14.6	17	108.0	745	0.81
10 hrs. at 1800°F (982°C)	1000	538	162.0	1117	122.0	841	0.75	19.6	23	135.9	937	0.84
+ 48 hrs. at 1350°F (732°C)	1200	649	147.6	1018	120.5	831	0.82	16.7	15	140.8	971	0.95
10 hrs. at 1700°F (927°C)	1000	538	157.5	1086	116.0	800	0.75	15.5	24			
+ 3 hrs. at 1325°F (718°C)												
10 hrs. at 1700°F (927°C)	1000	538	166.0	1144	116.5	803	0.71	15.9	23	125.1	863	0.75
+ 48 hrs. at 1350°F (732°C)	1200	649	143.7	991	115.5	786	0.80	16.8	26	138.1	952	0.96
1 hr. at 1700°F (927°C)	1000	538	165.2	1139	135.0	931	0.82	12.9	27	154.9	1068	0.94
+ 3 hrs. at 1325°F (718°C)	1200	649	160.2	1105	136.0	938	0.85	8.3	22	140.3	967	0.88
1 hr. at 1700°F (927°C)	1000	538	149.0	1027	101.5	700	0.68	19.8	27	121.9	841	0.82
+ 2 hrs. at 1550°F (843°C)	1200	649	135.4	934	103.0	710	0.76	14.2	20	121.2	836	0.89

TABLE 2

SMOOTH SPECIMEN CREEP-RUPTURE PROPERTIES AT

900° TO 1400°F (482 -760°C)

Heat Treatment	Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress (MN/m²)	Rupture Time (hrs.)	Elong. (%)	R.A. (%)	Min. Creep Rate (% / hr.)	Intergranular Crack Length (%)
10hr. 1950°F(1066°C) + 48hr. 1350°F(732°C)	1000	538	145.4	1002	Tensile	15.4	23		0
			130	896	21.5	7.4	13		8
			115	793	333.7	3.4	9	0.00033	11
			105	724	3332.2ph			-ve	
	1100	593	100	690	431.3	1.5	4	0.00041	18
			85	586	9691.4	1.0	3	-ve	26
	1200	649	142.4	982	Tensile	8.1	14		0
			90	621	216.1	1.1	8	0.00067	26
			80	552	282.4ph			0.00055	
			70	483	924.8	1.1	5	0.00014	34
			60	414	3564.0	1.3	4	0.000038	44
	1400	760	30	207	292.1	1.9	3	0.00091	60
1hr. 1950°F(1066° C) + 48hr. 1350°F(732°C)	1000	538	160.1	1104	Tensile	22.5	30		0
			140	965	385.0	2.7	11	0.00043	10
			130	896	2506.1	1.6	7	0.000022	14
	1100	593	120	827	1.4	4.2	10		8
			110	758	1930.1	1.3	7	0.00015	22
			100	690	6833.6	1.6	6	0.000013	23
	1200	649	157.5	1086	Tensile	8.4	14		0
			90	621	707.6	1.3	8	0.00028	24
			80	552	228.9ph			0.00025	
			70	483	2469.5	0.8	2	0.000073	40
			60	414	6261.4	0.8	2	-ve.	40
			60	414	8168.8		2	0.000026	48
	1400	760	30	207	384.0	2.1	5	0.00077	24
+ 2hr. 1560°F(843°C)	1000	538	134.9	930	Tensile	27.0	34		0
			130	896	63.5	18.7	19		6
			120	827	154.1	11.9	14	0.00020	11
			115	793	4402.3	7.3	13	0.000005	17
	1100	593	110	758	138.2	7.8	13	0.00030	20
			100	690	385.2ph			<0.00047	
			85	586	18,470.5	1.1	5	-ve	30
	1200	649	133.4	920	Tensile	11.3	18		1
			100	690	72.6	3.2	9	0.0025	26
			80	552	938.6	1.8	6	0.00021	39
			70	483	2094.6	1.6	5	0.00011	44
	1400	760	60	414	16.9	3.5	8		43
			30	207	508.0	2.9	5	0.00091	98
+ 24 hr. 1550°F(843°C)	1000	538	134.2	925	Tensile	32.8	30		0
			125	862	40.9	24.9	23	0.090	3
			115	793	1856.7	10.2	12	0.00014	10
	1100	593	110	758	90.5	9.9	12	0.0034	15
			100	690	3528.1	4.7	6	0.000006	20
	1200	649	132.6	914	Tensile	14.6	17		1
			100	690	11.6	6.9	13	0.17	19
			80	552	926.5	3.2	6	0.00037	23
			70	483	2480.9	2.0	5	0.000080	32
	1400	760	30	207	558.1	8.7	12	0.0011	92
10 hr. 1800°F(982°C) + 48 hr. 1350°F(732°C)	1000	538	162.0	1117	Tensile	19.6	23		0
			130	896	219.7	7.0	10	0.0175	9
			115	793	13437.5	3.1	6	0.000046	13
	1100	593	115	793	336.9	4.5	10	0.0052	12
			100	690	1266.5	3.0	8	0.00035	19
			90	621	5793.2	3.0	10	0.00010	26
10hr. 1800°F(982°C) + 48hr. 1350°F(732°C)	1200	649	147.6	1018	Tensile	16.7	15		1
			100	690	34.0	7.2	12	0.095	5
			80	552	410.4	6.2	11	0.027	26
			70	483	1420.0	6.0	13	0.00070	42
	1400	760	30	207	231.8	17.4	42	0.0056	52
10hr. 1700°F(927°C) + 3hr. 1325°F(718°C)	1000	538	157.5	1086	Tensile	15.5	24		0
			145	1000	26.5	21.6	22		1
			130	896	391.1	9.0	11	0.0048	5
	1100	593	120	827	95.0	8.4	13	0.022	8
			90	621	2138.2	7.8	13	0.00024	11
	1200	649	85	586	79.4	6.0	18	0.012	65
			60	414	1061.8	12.6	16	0.0016	
			50	345	3766.9	15.8	16	0.00038	
	1400	760	20	137	506.9	32.3	50	0.0068	
+ 48hr. 1350°F(732°C)	1000	538	166.0	1144	Tensile	15.9	23		0
			130	896	125.7	11.7	14	0.050	2
			120	827	1382.8	5.6	8	0.0022	5
			115	793	1163.2	6.1	10	0.0023	6
	1100	593	100	690	192.2	11.7	14	0.036	10
			85	586	1574.0	2.7	6	0.00073	22
	1200	649	143.7	991	Tensile	16.8	26		0
			85	586	42.5	9.2	20	0.087	15
			60	414	530.7	12.0	18	0.0045	30
			50	345	3378.9	12.4	21	0.00072	47
	1400	760	30	207	96.9	18.0	43	0.040	32
			20	137	439.0	32.7	45	0.013	
1 hr. 1700°F(927°C) + 3 hr. 1325°F(718°C)	1000	538	165.2	1139	Tensile	12.9	27		0
			145	1000	168.9	12.1	14	0.0048	2
			130	896	5613.4	3.5	12	0.000086	13
	1100	593	125	862	184.8	1.65	11	0.001	12
			115	793	406.6	1.6	6	0.00072	13
			100	690	2727.0	3.4	11	0.00017	26
	1200	649	160.2	1105	Tensile	8.3	22		0
			100	690	53.0	1.75	9	0.0096	24
			85	586	291.1		23		18
			85	586	117.1	5.0	15	0.0061	23
			65	448	937.4	3.0	7	0.00096	36
			65	448	1601.2		9		32
	1400	760	30	207	144.0	10.8	27	0.013	60
+ 2 hr. 1550°F (843°C)	1000	538	149.0	1027	Tensile	19.8	27		0
			145	1000	2.0	24.3	32	1.7	1
			130	896	18.1	19.4	25		1
			120	827	785.0	9.3	12	0.0016	7
			115	793	4385.2	2.6	6	0.000082	7
	1100	593	110	758	486.3	5.8	9	0.0018	9
			100	690	703.6	1.5	7	0.00056	19
			85	586	4078.6	2.2	4	0.00010	23
	1200	649	135.4	934	Tensile	14.2	20		1
			100	690	21.9	9.3	15		12
			85	586	154.3	4.0	9	0.0085	20
			65	448	1011.8	6.2	7	0.0013	20
	1400	760	30	207	160.1	21.4	34	0.030	33
			20	137	535.2	25.4	35	0.0053	

ph = failed at pinhole

TABLE 3

NOTCHED SPECIMEN (Kt>20) RUPTURE PROPERTIES AT 900° TO 1400°F (482 - 760°C)

Heat Treatment	Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress (MN/m²)	Rupture Time (hrs.)	Intergranular Crack Length (%)	N/S Strength Ratio
10hr. 1950°F(1066°C) + 48hr. 1350°F(732°C)	900	482	120	827	360.0	16	
			90	621	1075 1	27	
	1000	538	149.0	1027	Tensile	0	1.02
			120	827	18.7	6	0.92
			90	621	100.7	26	0.74
			60	414	682.7	54	0.53
			50	345	2342.4	38	0.47
	1100	593	80	552	31.6	35	0.71
			60	414	212.8		0.58
			50	345	2121.8	39	0.55
	1200	649	135.3	933	Tensile	1	0.95
			60	414	3.1	32	0.46
			50	345	1.5	52	0.37
			40	276	5774.6	73	0.72
	1400	760	30	207	451.8	79	1.13
1hr. 1950°F(1066°C) + 48hr. 1350°F(732°C)	900	482	130	896	34.6	6	
			90	621	1745.1	41	
	1000	538	152.4	1051	Tensile	0	0.95
			90	621	86.2	29	0.60
			70	483	221.2	41	0.49
			60	414	9262 Discontinued		>0.49
	1100	593	80	552	10.8	30	0.55
			60	414	184.1	44	0.47
			50	345	8902 Discontinued		>0.51
	1200	649	147.5	1017	Tensile	1	0.94
			50	345	46.3	45	0.41
			45	310	330.2	70	0.45
	1400	760	30	207	419.8	75	1.02
+ 2hr. 1550°F(843°C)	900	482	80	552	5763.6	30	
	1000	538	113.6	783	Tensile	0	0.84
			90	621	173.8	29	0.72
			75	517	752.9	34	0.62
			65	448	1138.9	49	0.55
	1100	593	80	552	20.0	26	0.67
			60	414	704.8	59	0.61
	1200	649	116.0	800	Tensile	3	0.87
			80	552	4.1	29	0.66
			60	414	225.0	44	0.65
			50	345	5003.7	51	0.79
	1400	760	50	345	2.8	44	0.58
			40	276	148.3	59	1.05
			30	207	408.5	71	0.95
			30	207	493.8	73	1.00
+ 24hr. 1550°F(843°C)	1000	538	102.1	704	Tensile	0	0.76
			90	621	253.5	14	0.76
			80	552	6862 Discontinued		>0.73
			60	414	11000 Discontinued		>0.55
	1100	593	85	586	40.6	23	0.70
			75	517	788.5	43	0.74
			60	414	9044 Discontinued		>0.61
	1200	649	108.0	745	Tensile	4	0.81
			60	414	18.9	26	0.61
			60	414	40.3	30	0.63
			50	345	5588.8	48	0.81
	1400	760	30	207	373.8	53	0.88
10hr 1800°F(982°C) + 48hr 1350°F(732°C)	900	482	90	621	980.3	17	
	1000	538	135.9	937	Tensile	0	0.84
			90	621	131.4	20	0.68
			75	517	238.4	26	0.58
			65	448	305.8	30	0.51
	1100	593	90	621	8.2	19	0.67
			75	517	105.2	31	0.64
			60	414	12867 Discontinued		>0.68
	1200	649	140.8	971	Tensile	0	0.95
			90	621	14.0	32	0.85
			75	517	5.1	39	0.68
			60	414	1406.5	58	0.87
	1400	760	30	207	215.4	70	0.99
10hr. 1700°F(927°C) + 3hr. 1325°F (718°C)	900	482	90	621	2639.1	12	
	1000	538	90	621	598.6	26	0.72
			75	517	7680 Discontinued		>0.95
	1100	593	90	621	77.4	26	0.75
			80	552	951.2		0.83
	1200	649	85	586	45.2	31	0.91
			60	414	1024.2	55	1.00
	1400	760	30	207	103.9	70	1.00
+ 48hr. 1350°F(732°C)	900	482	90	621	13140 Discontinued		
	1000	538	125.1	863	Tensile	0	0.75
			90	621	299.2	20	0.72
			75	517	7656 Discontinued		>0.70
	1100	593	90	621	167.8	28	0.89
			80	552	847.9	33	0.90
	1200	649	138.1	952	Tensile	1	0.96
			80	552	53.2	40	0.98
			60	414	468.2	52	0.94
			50	345	2422.0	65	0.97
	1400	760	35	241	57.0ph	54	1.00
			30	207	103.0	57	1.00
			20	137	483.0	67	1.02
1hr. 1700°F(927°C) + 3hr. 1325°F (718°C)	900	482	90	621	2547.2	30	
	1000	538	154.9	1068	Tensile	0	0.94
			130	896	17.5	7	0.86
			90	621	152.6	23	0.62
			75	517	239.4	39	0.53
			65	448	1561.3	51	0.49
	1100	593	90	621	30.7	28	0.60
			80	552	131.4	26	0.59
	1200	649	140.3	967	Tensile	1	0.88
			85	586	2.6	35	0.63
			75	517	109.8	37	0.83
			65	448	1031.9	52	0.99
	1400	760	30	207	105.9	61	0.94
+ 2hr. 1550°F(843°C)	1000	538	121.9	841	Tensile	0	0.82
			90	621	3205.7	23	0.78
			75	517	5133 Discontinued		>0.66
	1100	593	90	621	748.4	28	0.92
	1200	649	121.2	836	Tensile	0	0.89
			65	448	223.3	42	0.80
			50	345	1677	59	0.82
	1400	760	30	207	92.5	61	0.85

ph = failed at pin hole

TABLE 4

X-RAY DIFFRACTION DATA OF EXTRACTED RESIDUES OF INCONEL 718 IN THE HEAT TREATED CONDITIONS

Solution Treatment	10 hrs. at 1950°F (1066°C)	1 hr. at 1950°F (1066°C)	1 hr. at 1950°F (1066°C)	1 hr. at 1950°F (1066°C)	10 hrs. at 1800°F (982°C)	10 hrs. at 1700°F (927°C)	10 hrs. at 1700°F (927°C)	1 hr. at 1700°F (927°C)	1 hr. at 1700°F (927°C)	
Aging Treatment	48 hrs. at 1350°F (732°C)	48 hrs. at 1350°F (732°C)	2 hrs. at 1550°F (843°C)	24 hrs. at 1550°F (843°C)	48 hrs. at 1350°F (732°C)	3 hrs. at 1325°F (718°C)	48 hrs. at 1350°F (732°C)	3 hrs. at 1325°F (718°C)	2 hrs. at 1550°F (843°C)	Indicated Phases
	d(A) I	d(A) I	d(A) I	d(A) I	d(A) I	d(A) I	d(A) I	d(A) I	d(A) I	
	3.25 vw	3.26 vvw	3.25 vw	3.25 w	3.25 vw	3.26 vvw	3.26 vvw		3.25 w	γ''
					2.643vvw	2.646w	2.659vvw		2.644w	?
	2.552w	2.57 vw	2.564w	2.561w	2.555w		2.567vvw	2.552w	2.553w	MC
					2.385vvw	2.394vvw		2.388vvw		L?
								2.330w		?
			2.266vvw		2.257vw	2.272w	2.272vw		2.264vvw	β
						2.231m	2.232m	2.232m		β
	2.214w	2.226vw	2.223w	2.217w	2.219w		2.222vw		2.221w	MC, L?
	2.116 to*	2.123 to*	2.111vs	2.108vs	2.104vs	2.117s	2.116s	2.118m	2.108vs	β
	2.078vs	2.087vs	2.080vw	2.080m	2.083vs	2.08 vw	2.081s	2.080w	2.079s	γ'/γ''
				2.040vvw	2.033vw	2.033vvw		2.043w	2.038vw	L?
			1.994w	1.992vw	1.992m	2.000m	2.000s	2.003m	1.996m	β, L?
			1.969w	1.948w	1.965s	1.977s	1.973s	1.975s	1.968s	β
	1.850vw		1.854vw	1.846w					1.855w	γ'/γ''
	1.809s	1.813m	1.818w	1.808s	1.805s		1.807m	1.807vvw	1.809m	γ'/γ''
	1.568w	1.575vvw	1.569vw	1.567w	1.567vw	1.573vvw	1.573vvw	1.574w	1.566vw	MC
					1.545vvw	1.548vvw	1.549vw		1.546vvw	?
			1.535vw		1.530w	1.536m	1.534w	1.536w	1.535w	?
	1.337w	1.339vw	1.338vw	1.337w	1.338w		1.337vvw	1.344vvw	1.338w	MC
	1.292w	1.296vvw	1.298vw	1.294m	1.300w	1.302w	1.302w	1.303vvw	1.300w	β, γ'/γ''
	1.278m	1.280w	1.280vw	1.280m	1.278m	1.281w	1.280w	1.281vvw	1.280w	MC, ?
					1.195w	1.195w		1.199w		L?
			1.193vw	1.188vvw	1.189w	1.191w	1.193w	1.193vvw	1.192w	?
	1.110vw	1.110w	1.107vw	1.110m	1.106w	1.108w	1.109w	1.111w	1.108w	β, ?
Phases										
MC	w	w	w	w	w	w	w	w	w	
γ'/γ''	s	s	s	vs	s	w	m	w	s	
β	0	0	w	w	m	s	s	s	m	

* Values at extremes of wide reflection
Intensities I: s = Strong, m = Medium, w = Weak, v = Very
Phases: MC = Cb, Ti(C, N), γ' = $Ni_3(Al, Ti)$, γ'' = Ni_3Cb(bct), β = Ni_3Cb(orthorhombic)
L = "Laves", $Fe_2(Ti, Cb)$.

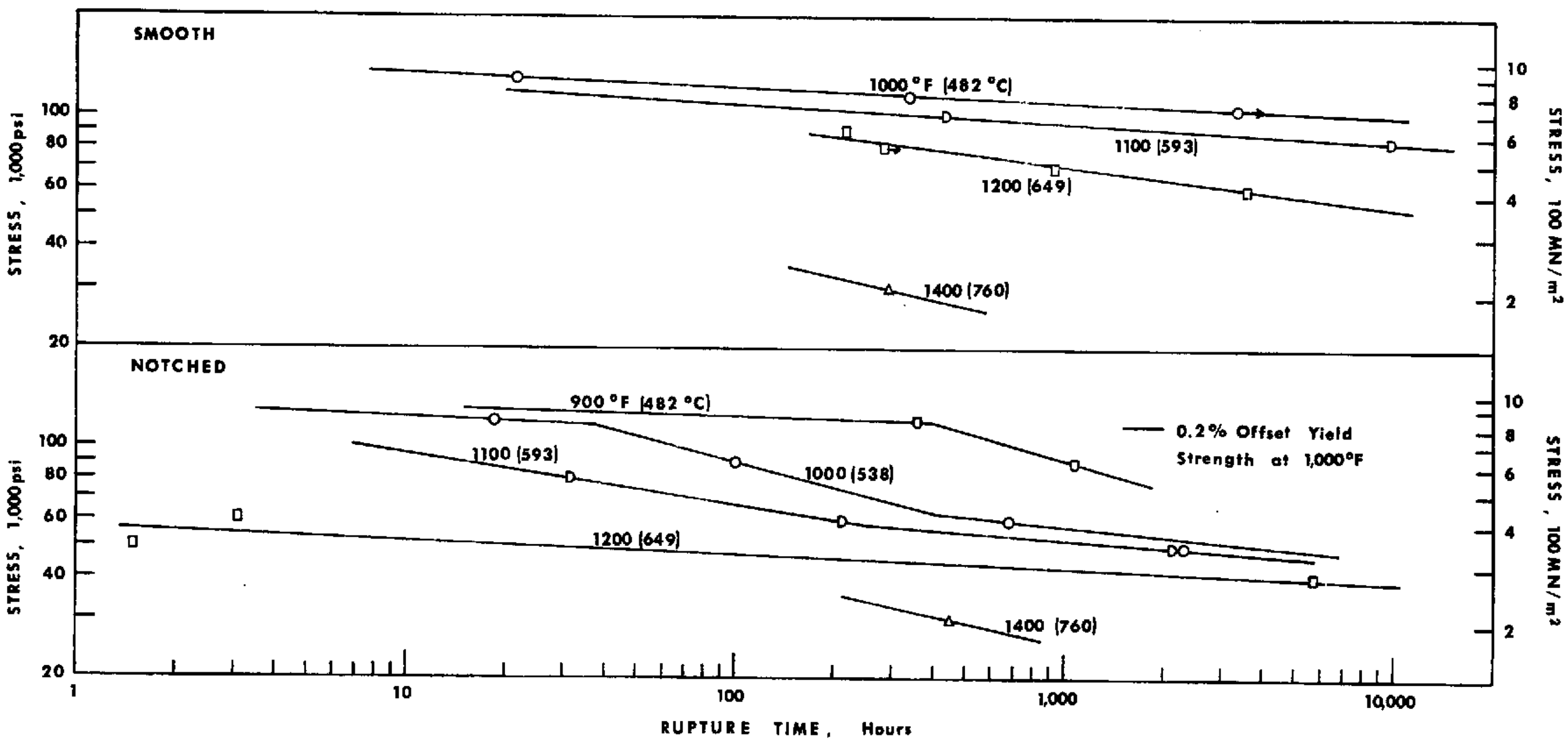

Figure 1. Stress versus rupture time data at temperatures from 900°F to 1400°F (482-760°C) obtained from smooth and notched specimens of Inconel 718 sheet heat-treated 10 hours at 1950°F (1066°C) plus 48 hours at 1350°F (732°C). Time-dependent notch sensitivity was evident at temperatures from 900°F to 1200°F (482 -649°C) but not at 1400°F (760°C).

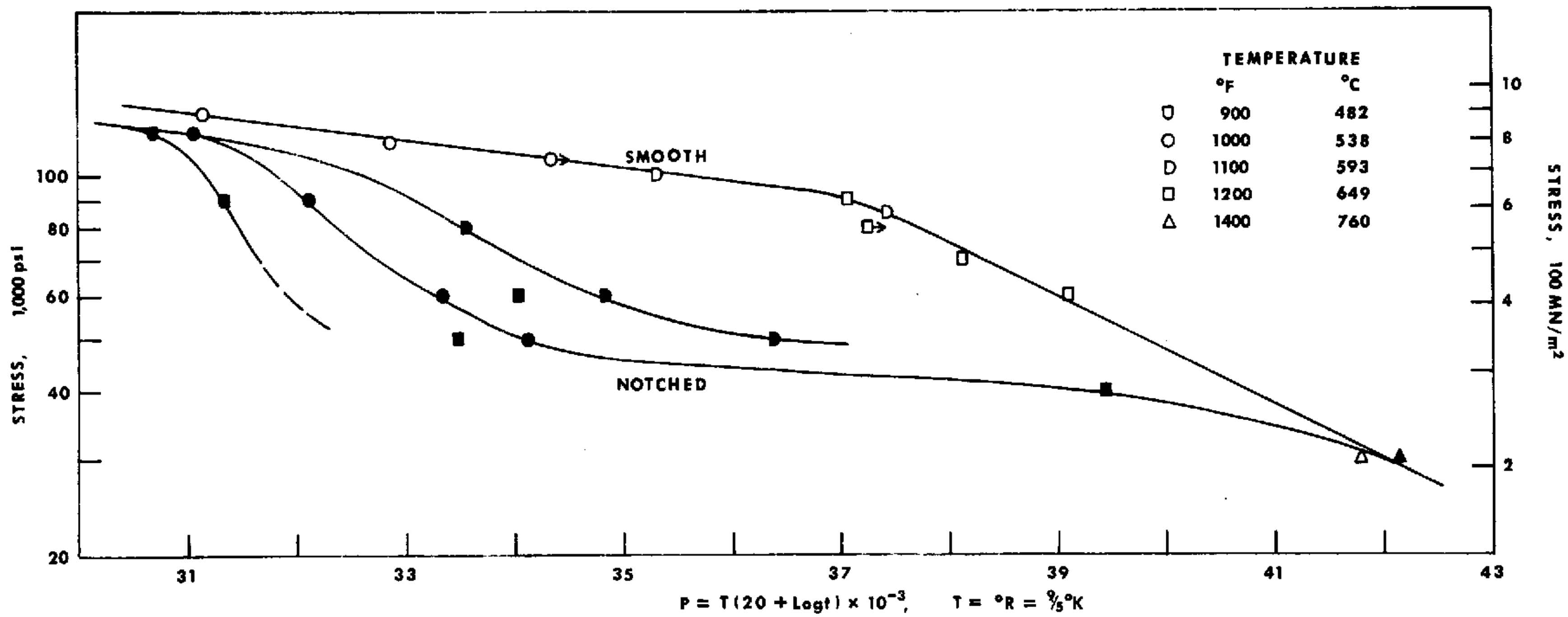

Figure 2. Time-temperature dependence of the rupture strengths of smooth and notched specimens of Inconel 718 sheet heat treated 10 hours at 1950°F (1066°C) plus 48 hours at 1350°F (732°C).

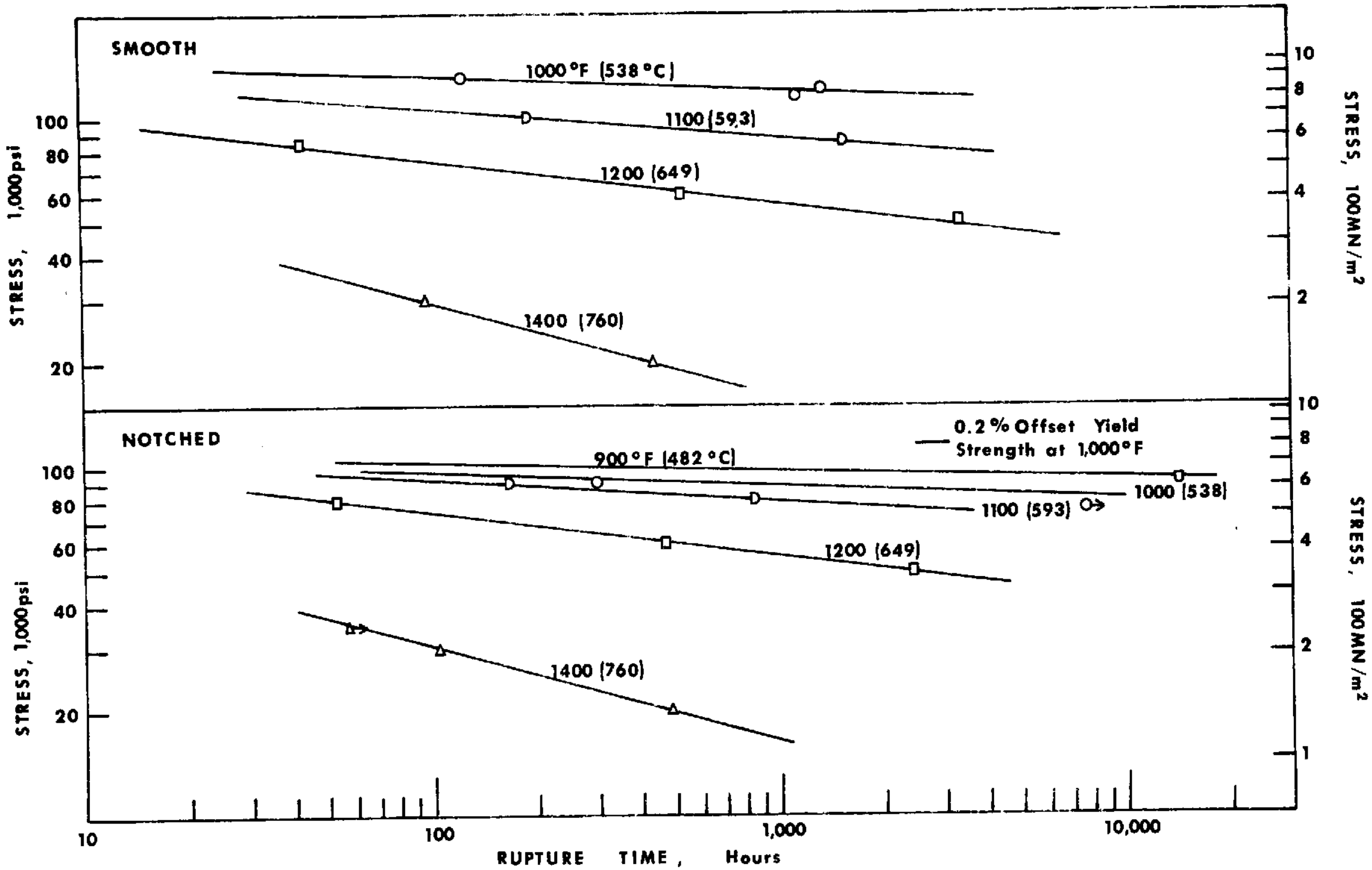

Figure 3. Stress versus rupture time data at temperatures from 900°F to 1400°F (482 - 760°C) obtained from smooth and notched specimens of Inconel 718 sheet heat treated 10 hours at 1700°F (954°C) plus 48 hours at 1350°F (732°C). The tests showed no time-dependent notch sensitivity.

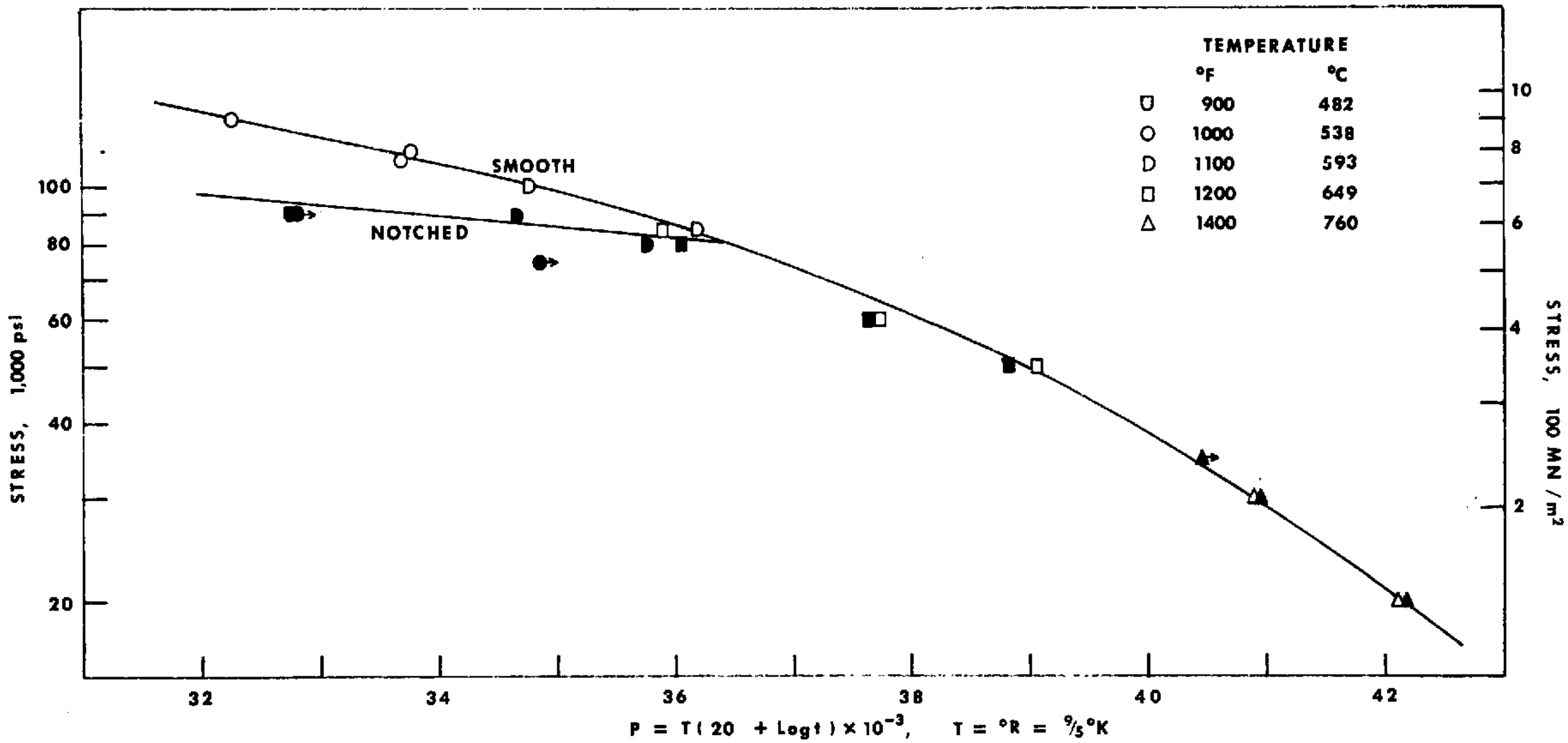

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Figure 4. Time-temperature dependence of the rupture strengths of smooth and notched specimens of Inconel 718 sheet heat treated 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C).

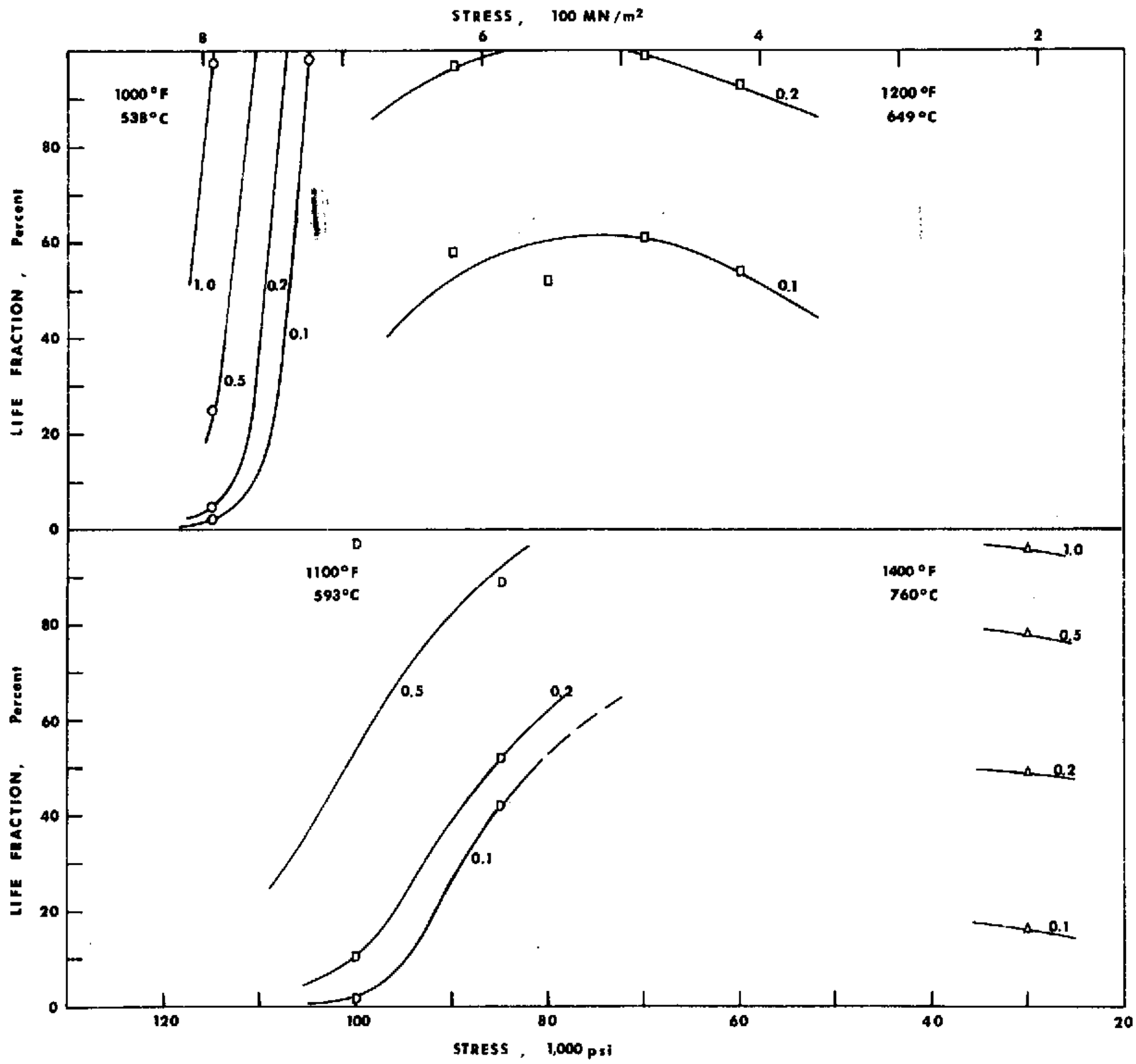


Figure 5. Iso-creep strain curves of life fraction versus stress at temperatures from 1000°F to 1400°F (538 - 760°C) for Inconel 718 heat treated 10 hours at 1950°F (1066°C) plus 48 hours at 1350°F (732°C). Time-dependent notch sensitivity occurred under test conditions where large amounts of rupture life were utilized for small creep strains at test temperatures 1000°F, 1100°F and 1200°F (538, 593 and 649°C).

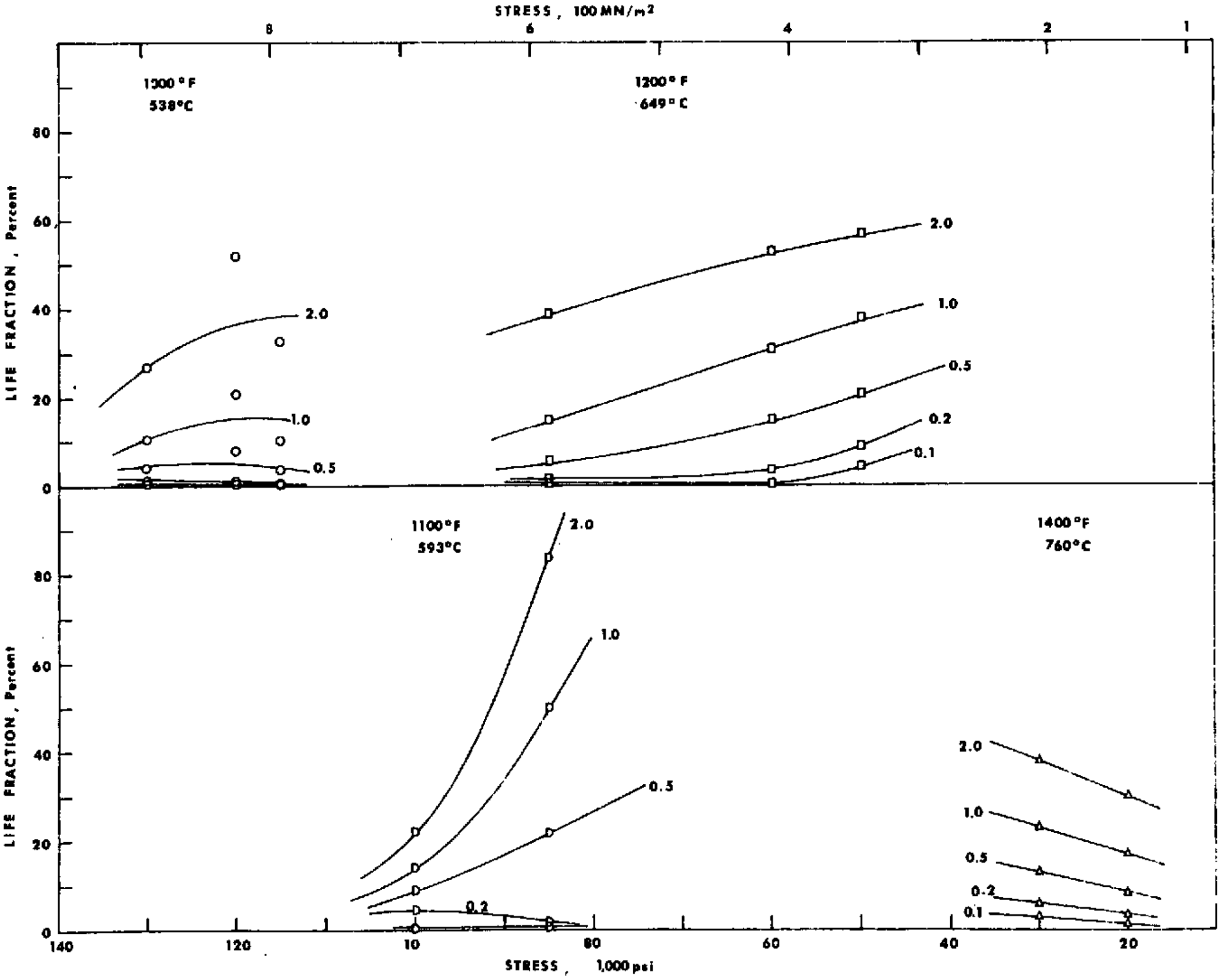
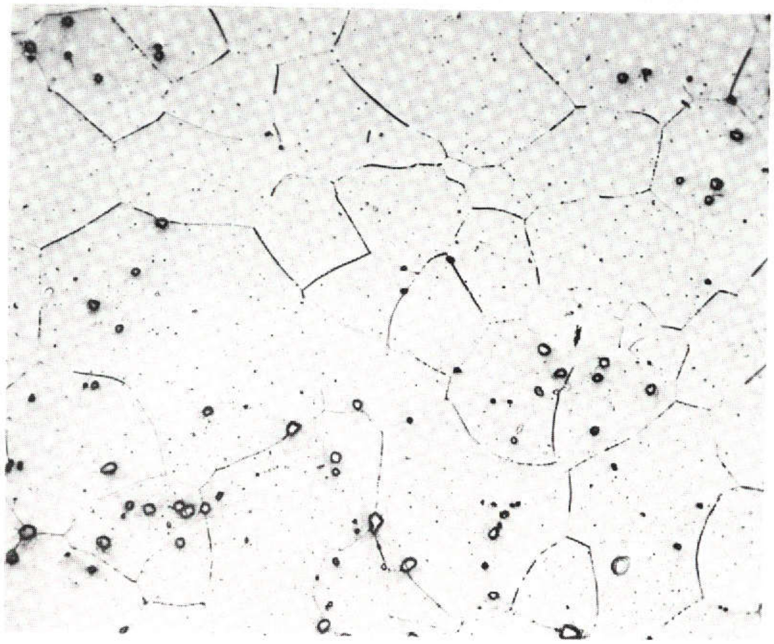
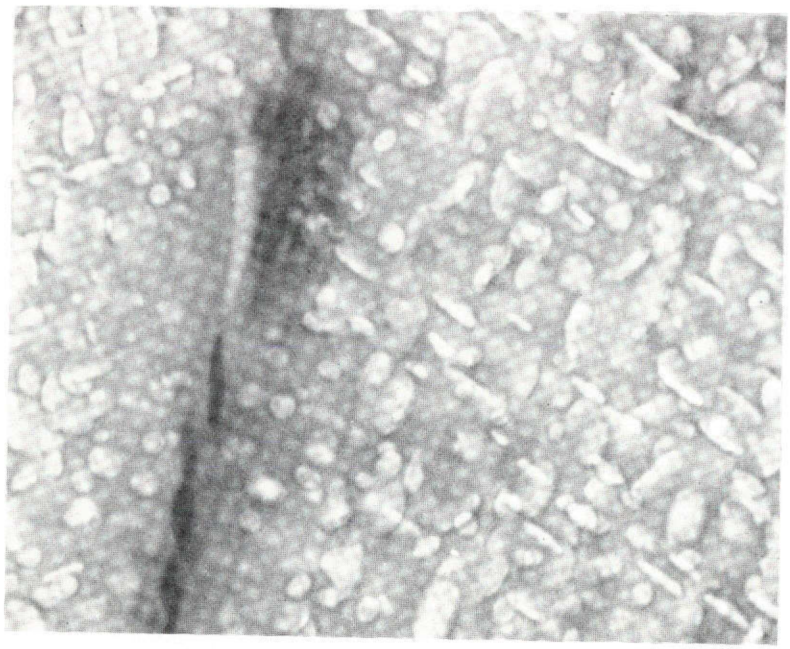


Figure 6. Iso-creep strain curves of life fraction versus stress at temperatures from 1000°F to 1400°F (538 - 760°C) for Inconel 718 heat treated 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C). Little rupture life was utilized for small amounts of creep under all test conditions and no time-dependent notch sensitivity was observed.

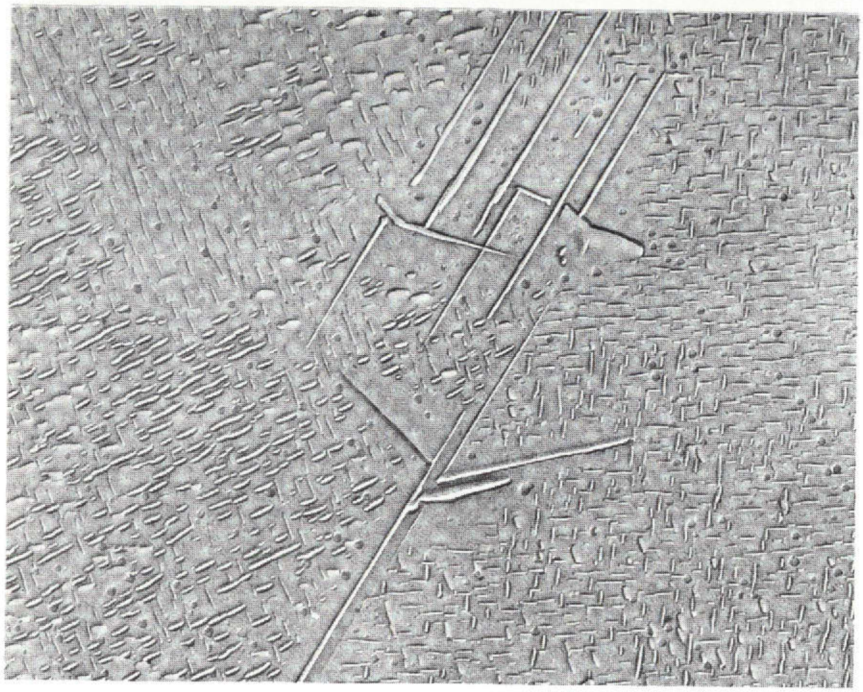


(a) 250x
10hr. 1950°F (1066°C) + 48hr. 1350°F (732°C)



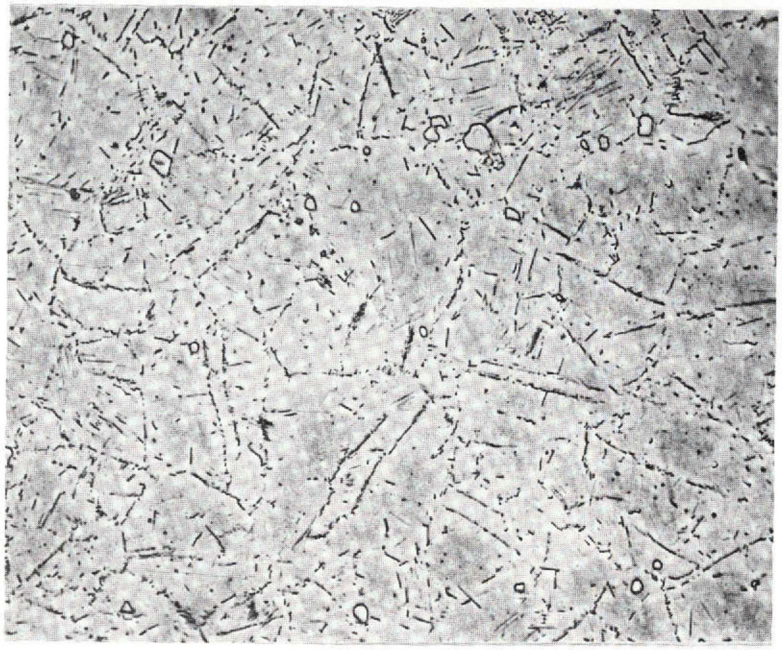
(b) 100,000x
1hr. 1950°F (1066°C) + 48hr. 1350°F (732°C)

Figure 7. Optical and transmission electron micrographs showing microstructural features of Inconel 718 solution treated at 1950°F (1066°C) and aged at 1350°F (732°C).



6000x

Figure 8. Electron micrograph of a replica of Inconel 718 heat treated 1hr. 1950°F (1066°C) + 24hrs. at 1550°F (843°C). The phases present are γ' (spherical particle), γ'' (plates) and β (needles).



(a) 250x

10 hr. 1800°F (982°C) + 48 hr. 1350°F (732°C)

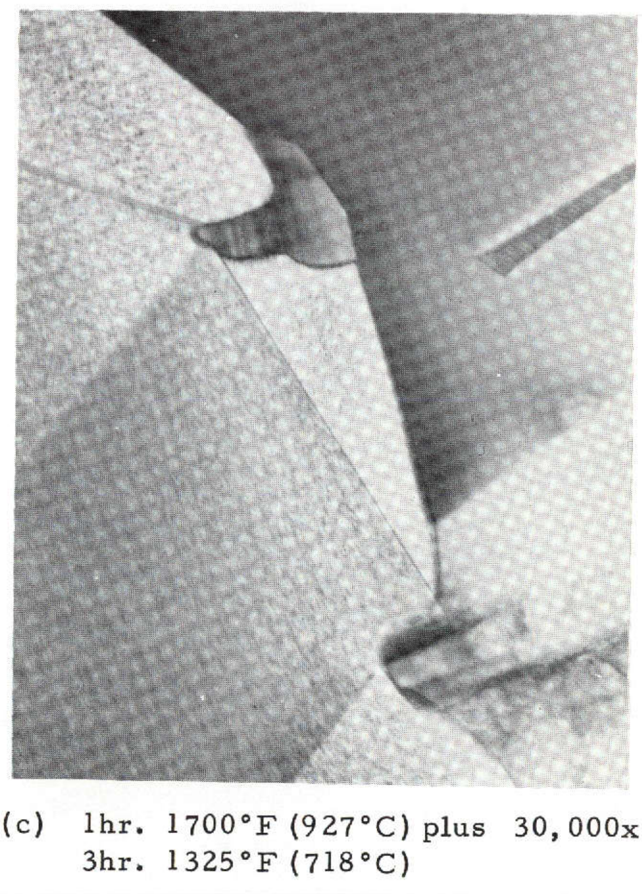
(b) 250x

10 hr. 1800°F (927°C)

Figure 9. Optical micrographs of heat treated Inconel 718. β phase precipitated during the 1800°F (927°C) "solution treatment". Fine γ'/γ'' (not resolved) formed during aging.

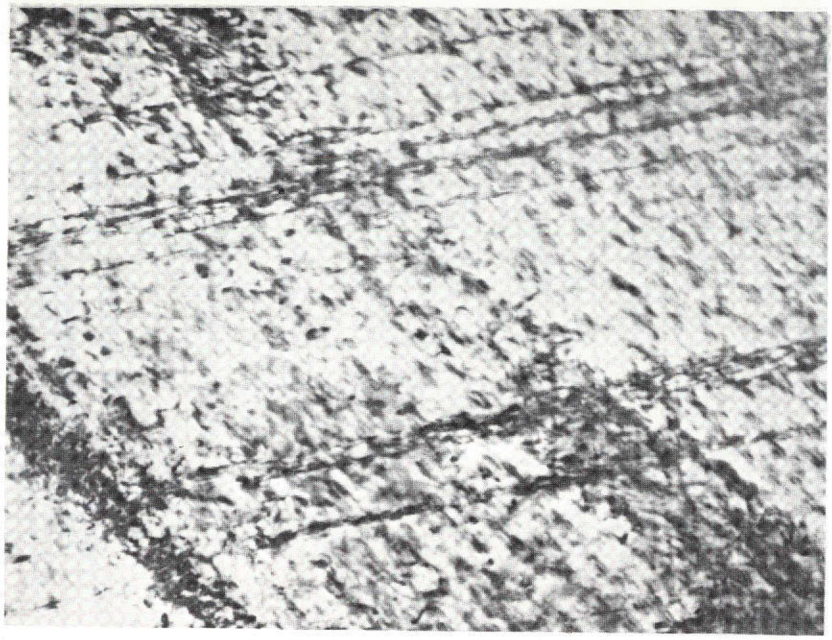
(a) 10hr. 1700°F (927°C) 250x

(b) 1hr. 1700°F (927°C) 250x

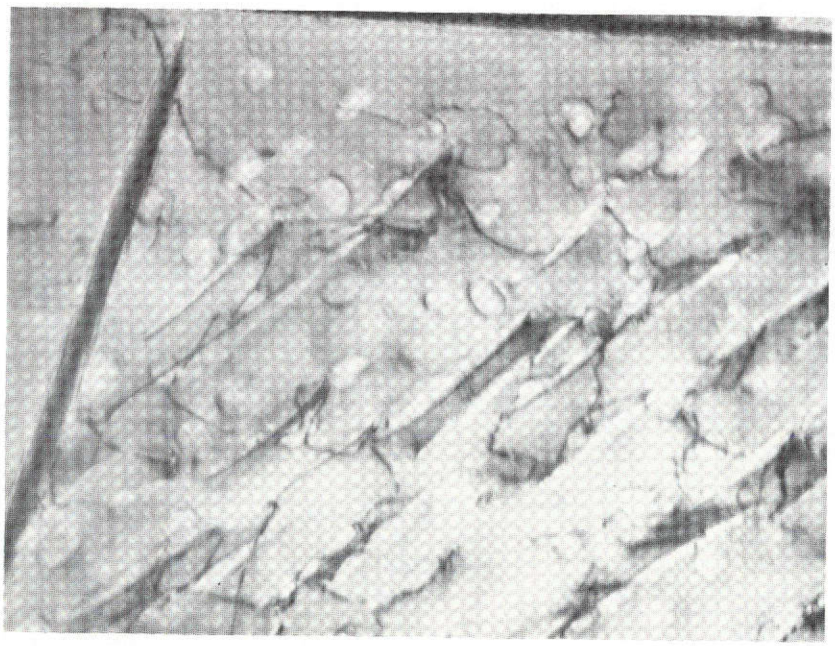


(c) 1hr. 1700°F (927°C) plus 3hr. 1325°F (718°C) 30,000x

Figure 10. Optical and transmission electron micrographs showing microstructural features of Inconel 718 after solution at 1700°F (927°C) and after aging.



(a) 50,000x

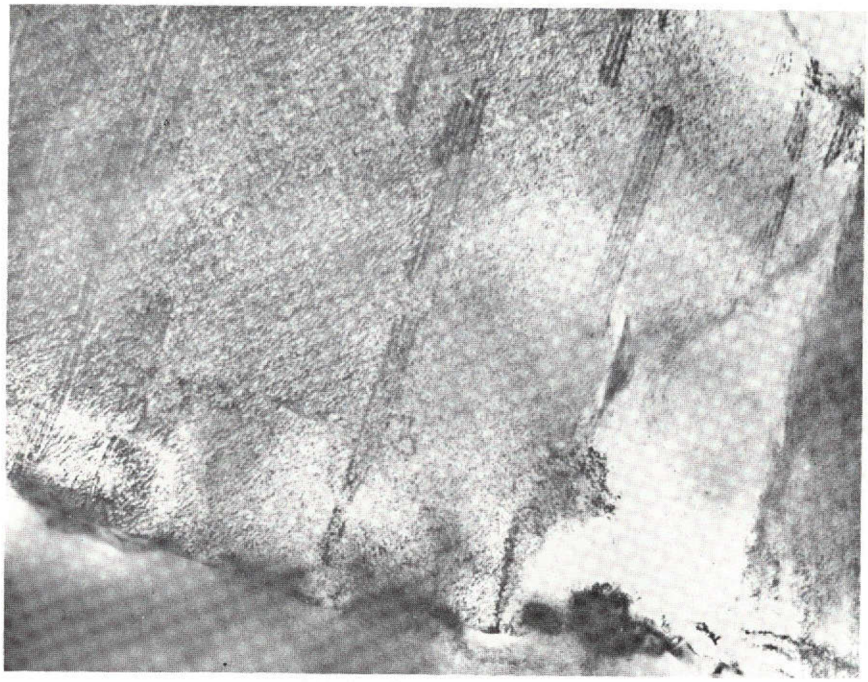


(b) 45,000x

Figure 11. Transmission electron micrograph of Inconel 718 heat treated 1 hr. at 1950°F (1066°C) plus 48 hrs. at 1350°F (732°C) and creep-rupture tested (a) at 120ksi (827MN/m^2) at 1100°F (593°C) (ruptured in 1.4 hrs. at 4.2% elongation), (b) at 30ksi (207MN/m^2) at 1400°F (760°C) (ruptured in 384 hrs. at 2.1% elongation). In the lower temperature tests the γ'/γ'' were sheared by dislocations and the deformation was localized. In the test at 1400°F (760°C) γ'/γ'' growth occurred causing the dislocation to by-pass the particles and the deformation to be homogeneous.

100,000x

Figure 12. Transmission electron micrograph of Inconel 718 heat treated 1 hr. at 1950°F (1066°C) plus 2 hrs. at 1550°F (843°C) and creep-rupture tested at 100ksi (690MN/m^2) at 1100°F (593°C) (ruptured in 385 hrs.) The fine dispersion of γ'/γ'' formed subsequent to the 1550°F (843°C) treatment and developed during the test exposure.



25,000x

Figure 13. Transmission electron micrograph of Inconel 718 heat treated 1 hr. at 1700°F (927°C) plus 3 hrs. at 1325°F (718°C) and creep-rupture tested at 130ksi (896MN/m^2) at 1000°F (538°C) (ruptured in 5613 hours at 3.5% elongation). The deformation was localized. The dislocation sheared the γ'/γ'' particles and were in many cases extended to form stacking fault ribbons.

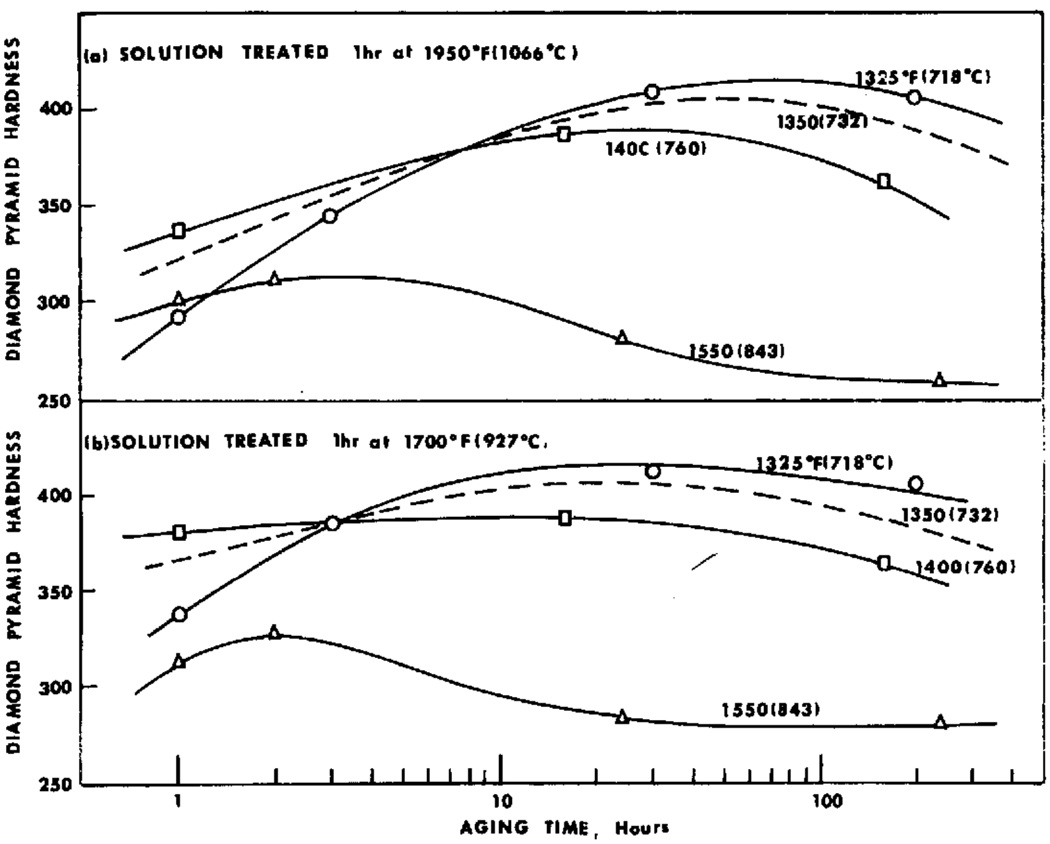


Figure 14. Effect of aging exposures at 1325°, 1400° and 1550°F (718, 760 and 843°C) on the Diamond Pyramid Hardness of Inconel 718 sheet solution treated at 1950°F and at 1700°F (1066° and at 927°C). Increasing the aging time increased and subsequently decreased the hardness. This corresponds to an increase followed by a decrease in time-dependent notch sensitivity.