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TECHNICAL REPORT

HEAT TREATMENT VERSUS PROPERTIES STUDIES ASSOCIATED WITH THE INCONEL 718 PBF ACOUSTIC FILTERS

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ABSTRACT

PBF acoustic filter Unit No. 1 cracked when heat treatment was attempted. Materials Technology Branch undertook investigations to facilitate the heat treatments of Units No. 1 and 2.

The effects of prior thermal cycling, solution anneal temperature, and cooling rate from solution anneals were investigated. The investigations concerned influences of the above variables upon both 1400 F stress rupture solution-annealed properties and room temperature age-hardened properties. 1400 F stress rupture properties were of interest to assist the prevention of cracking during heat treatments. Room temperature age-hardened properties were needed to ensure that design requirement would be provided. Prior thermal cycling was investigated to determine if extra thermal cycles would be detrimental to the repaired filter. Slow furnace cools were considered as a means of reducing thermal stresses. Effects of solution annealing at 2000 F and 1900 F were also determined.

Test results showed that slow cooling rates would not only reduce thermal stresses but also improve 1400 F ductility. A modified aging treatment was established which provided the required 145 ksi room temperature yeild strength for the slowly cooled material. Prior thermal cooling did not degrade final age-hardened room temperature tensile or impact properties.

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I. INTRODUCTION

Cracking occurred during heat treatments of the Inconel 718 PBF acoustic filters. The original failure occurred in Unit No. 1 during its aging cycle (Ref. 1). The filter had been solution annealed at 2000 F and cooled at a rate of 1200 F per hour to 1000 F and thereafter at a rate of 100 F per hour. Dye penetrant inspection did not reveal any cracks in the vessel. Failure then occurred during a 1400-1200 F duplex age. The crack passed near a fillet weld on the trunnion-shell interface, through two one-inch thick gussets, and 17 inches around the one-inch thick 16-inch diameter shell. It was concluded that this failure was caused by thermal stresses quenched into the vessel when cooled from the solution anneal and by the low ductility existing around 1400 F during aging.

The purpose of this program was to investigate methods of (1) decreasing the amount of thermal stresses quenched into the structure and (2) increase the ductility available during aging or when heated or cooled through the precipitation-hardening temperature region.

To meet the above program objective, the Materials Technology Branch initiated studies consisting of the following:

(1) A determination of the influences of prior thermal cycling, solution anneal temperatures, and cooling rates from the solution anneal temperature, see Section IV.

(2) A modified aging treatment to increase room temperature yield strength, of material slowly cooled from solution anneal temperatures, to the design requirement of 145 ksi, see Section V.

(3) Based on the results of the above two studies, tests were conducted on heat 87C0EK1 to see if the acoustic filter shell material would respond to the modifications in heat treatments, see Section VI.

(4) To determine potential thermal cycles required to enhance the ductility of heat-affected zones (HAZ). This was necessary since postweld heat-treat cracking of precipitation hardened nickel-base alloys most often occurs in HAZ, see Sections VII and VIII.

II. CONCLUSIONS

The information generated from these studies presents the following conclusions:

1. Stress Rupture Ductility and Life

1400 F stress rupture ductilities for 2000 F solution annealed conditions were progressively improved as the cooling rate was decreased from cooling rates of 400 to 1860 F per hour to a furnace cool of 100 F per hour.

The 1400 F stress rupture elongations for Inconel 718 base material (heats 83C2EK and 87C0EK1) was 20 percent when cooled from 2000 F at 100 F per hour as compared to 5 percent when cooled at rates ranging from 400 to 1860 F per hour.

2. Elevated Temperature Tensile Test Results

Base metal of heat 83C2EK slowly cooled (100 F/hr) from a 2000 F solution anneal provided better high temperature ductility between approximately 1300 and 1550 F than the same material cooled at a rate of 1200 F per hour. Some beneficial effect also existed from the slow cooling rate when the solution anneal was preceeded by a two or three second thermal cycle to 2250 F to simulate a HAZ. This better ductility indicates that welded components of Inconel 718 would have a reduced tendency to crack when slowly cooled from solution anneal temperatures.

A preliminary check on heat 2180-2-9247 showed that 1400 F tensile ductilities of simulated HAZ were improved when prior treatment involved a 100 F per hour cool rather than an air cool from 1725 F. This suggested that a slow cool would be effective as a preweld conditioning heat treatment which would decrease the probability of welded structures from cracking when heated to be solution annealed.

3. Room Temperature Tensile Test Results

Slowly cooling heat 83C2EK from 2000 F, changed its response to the 10 hour 1400 F plus 8 hour 1200 F duplex age. Cooling rates of 75 and 90 F per hour decreased the room temperature yield strength from 150 ksi (900 F per hour) to 128 ksi. The room temperature elongations and reductions in area were also slightly reduced by the slow cooling rates.

A modified lower temperature age consisting of 4 hours at 1325 and 16 hours at 1150 F was established which provided a 145 ksi room temperature yield strength for heats 83C2EK (gusset material) and 87COEK1 (sheel material) at cooling rates from solution anneals of approximately 100 F per hour. This met the design requirement for the PBF acoustic filters.

Heat 87COEK1 solution annealed at 2000 F and cooled at rates of 1200 and 400 F per hour then duplex aged at (1400-1200 F) and (1325-1150 F) respectively, exhibited room temperature yield strengths of 155 ksi. The room temperature tensile ductility for this above material cooled at rates of 1200 and 400 F per hour and then aged showed elongation values of 18 and 24 percent, respectively.

4. Room Temperature Impact Properties

Heat 87COEK1 (shell material) having received the 400 F per hour cool and (1325-1150 F) age possessed the best room temperature impact resistance of 60 ft-lb. Material cooled at rates of 100 and 1200 F per hour, respectively, and aged with the (1325-1150 F) and (1400-1200 F) treatments showed impact values of 51 and 41 ft-lb.

III. RECOMMENDATIONS

The results of these studies have promoted the following recommendations in the PBF acoustic filter program.

(1) The filters should be cooled at a rate of 150 ± 50 F after being solution annealed at 2000 F. This would not only result in lower thermal stresses but should also provide improved ductility between approximately 1300 to 1500 F.

(2) A duplex age consisting of 4 hours at 1325 F and 16 hours at 1150 F should be given to the acoustic filters following the 2000 F solution anneal outlined in the preceeding recommendation. This would provide the required 145 ksi room temperature yield strength.

(3) New material used in repairing the PBF acoustic filter Unit No. 1 should be given a 1750 F one hour treatment followed by a 100 ± 10 F per hour cool. Preliminary checks have shown the high temperature ductility of HAZ to be improved by such a thermal treatment. This should therefore reduce the probability of cracking when the unit is heated to the solution anneal.

IV. MATERIAL DEGRADATION TESTS (HEAT 83C2EK)

This program was planned to develop heat treating cycles for PBF acoustic filters Units No. 1 and 2 which would provide optimum stress rupture behavior for solution-annealed materials near the precipitation hardening temperatures. It was also designed to check the room temperature tensile and impact properties resulting from heat treating variables to ensure that the design requirement of 145 ksi for the 0.2% room temperature yield strength be maintained.

The program was designed to incorporate the following variables:

(a) Two starting material conditions - One was heat treated through the same cycles as the cracked PBF acoustic filter Unit No. 1 and the other was in the 1800 F stress relieved condition such as that of Unit No. 2. Cracked Unit No. 1 had experienced the following heat treating cycles: a 1800 F stress relief, a 2000 F solution anneal with a 1200 F per hour cool to 1000 F and a 100 F per hour cool to ambient temperature, a 1400-1200 F duplex age, and finally a 1750 F stress relief followed by a 100 F per hour cool to ambient temperature.

(b) Two solution anneal temperatures - One solution anneal temperature investigated was 2000 F, the solution anneal temperature previously given Unit No. 1. The other temperature selected was 1900 F.

(c) Two cooling rates from the solution anneal cycles - One of these rates, 1200 F per hour, represents the cooling conditions applied to Unit No. 1 and the other cooling rate of 100 F per hour represents the furnace cool proposed to reduce thermal stresses.

The test matrix for this program involved testing three (3) room temperature tensile and three (3) room temperature Charpy V-notch impact specimens which had been age-hardened from each of the above eight solution-annealed conditions. It also included one (1) tensile and fifteen (15) stress rupture tests at 1400 F for each of the eight solution-annealed conditions outlined above. The specifications for the specimens and testing procedures are detailed in the experimental procedures in Section IV.2.

1. MATERIAL IDENTITY AND REMOVAL

The specimens for this test program were prepared from excess material used for fabricating the gussets on the PBF acoustic filters. The heat identification for gusset material, 83C2EK, and the chemical composition per mill certifications are listed in Table I along with the other heats involved in the heat treat studies. This material was supplied in the 1700 to 1800 F mill-annealed condition per AMS Specification 5596C.

Sections were initially cut from the 1-inch by 31-inch by 17-inch plate with a band saw as illustrated in Figure 1. These sections were identified as shown and heat treated to the conditions listed in Table II.

TABLE I

CHEMISTRY OF THE INCONEL 718 HEATS USED FOR HEAT TREAT INVESTIGATIONS FOR THE PBF ACOUSTIC FILTERS

Heat								
87COEK1 VIM-ESR Shell Material	C	P	S	B	Cu	Si	Co	Mn
	0.07	0.011	0.007	0.0042	0.10	0.26	0.04	0.10
	Mo	Al	Ti	Cb+Ta	Cr	Ni	Fe	
	3.05	0.55	1.06	5.05	18.00	54.14	17.55	
83C2EK VIM-ESR Gusset Material	C	P	S	B	Cu	Si	Co	Mn
	0.06	0.006	0.007	0.0029	0.13	0.20	0.03	0.12
	Mo	Al	Ti	Cb+Ta	Cr	Ni	Fe	
	2.93	0.68	0.98	5.25	18.55	53.69	17.37	
2180-2-9247 VIM-VAR	C	P	S	B	Cu	Si	Co	Mn
	0.05	0.005	0.005	0.005	0.02	0.17	0.25	0.13
	Mo	Al	Ti	Cb+Ta	Cr	Ni	Fe	
	3.09	0.50	0.99	5.06	18.80	51.07	Balance	

VIM - Vacuum induction melted   VAR - Vacuum arc melted

ESR - Electroslag remelted

TABLE II

HEAT TREATMENTS FOR THE MATERIAL DEGRADATION STUDIES [heat 832C2EK]

CYCLE 1 Sections A, B, C, D, E, F, G, H and * are to be stress relieved at 1800 F.

CYCLE 2 Sections E, F, G, H and * are to be solution annealed at 2000 F.

CYCLE 3 Sections E, F, G, H and * are to be duplex aged at 1400 F for 10 hours and 1200°F for 8 hours.

CYCLE 4 Sections E, F, G, H and * are to be stress relieved at 1750 F.

CYCLE 5 Sections A, E, and * are to be solution annealed at 2000 F and cooled at a rate of 20 F per minute to 1000 F.

CYCLE 6 Sections B, F, and * are to be solution annealed at 2000 F and cooled at a rate of 100 F per hour to 1000 F.

CYCLE 7 Sections C, G, and * are to be solution annealed at 1900°F and cooled at a rate of 20 F per minute to 1000 F.

CYCLE 8 Sections D, H and * are to be solution annealed at 1900 F and cooled at a rate of 100 F per hour to 1000 F.

NOTE: All single lettered samples are complete in the solution annealed condition.

CYCLE 9 Sections AA, BB, CC, DD, EE, FF, GG, and HH are to be duplex aged at 1400 F for 10 hours and 1200 F for 8 hours.

NOTES:

(1) Solution anneals and stress reliefs shall be conducted in an air atmosphere.

(2) Heating and cooling rates unless otherwise specified shall be a maximum of 100 F per hour.

(3) Controlled heating and cooling rates are not required below 1000 F.

(4) The temperatures are to be recorded by thermocouples attached directly to the sections of plate material.

(5) The parts shall be held at the solution anneal and stress relief temperatures for (1) hour ± 5 minutes. The tolerances for the aging times shall be ± 5 minutes.

(6)a The solution anneals and stress reliefs shall be controlled within +25-0 F of the specified temperatures.

b During the duplex aging cycles the sections shall be controlled within ± 15 F of the specified temperatures.

* Include associated double lettered specimens i.e., AA, BB, CC, etc.

2. EXPERIMENTAL PROCEDURES

2.1 Heat Treatments of Plate Sections

The sections were shipped to Pyromet Industries and heat treated according to the schedule outlined in Table II. The heat treatments were performed in a bell furnace which was resistance heated. An air atmosphere was used for all of the thermal cycles. A continuous temperature record was maintained for all of the thermal cycles. These recordings were obtained from thermocouples directly attached to a plate section during each cycle.

The following instructions applied to the heat treatments in Table II:

(1) The sections identified with double letters shall accompany the respective single letter sections through all cycles except the final duplex age. Those sections having double letters shall then be aged 10 hours at 1400 F plus 8 hours at 1200 F. The section having single letters shall be retained in the solution-annealed condition.

(2) Controlled heating and cooling rates are not required below 1000 F.

(3a) Times at the solution anneal and stress relief temperatures shall be 1 hour ± 5 minutes.

(3b) The tolerance for the temperatures at the solution anneals and stress relief shall be $^{+25}_{-0}$F.

(4) The duplex age shall consist of 10 hours ± 5 minutes at 1400 ± 15 F, then a 100 F per hour cool to 1200 ± 15 F, and a hold at this temperature for an additional 8 hours ± 5 minutes.

(5) All heating and cooling rates not specified shall be a maximum of 100 F per hour.

2.2 Specimen Preparation

Fifteen (15) combination smooth-and-notched stress rupture specimens per AMS 5663 (specimen type -5) were prepared from each of the eight 5-1/2 inch by 15 inch sections representing variable solution-annealed conditions. These samples were machined such that their longitudinal axis paralleled the rolling direction, i.e.: the 5-1/2 inch dimension. They were also removed whereby this axis was taken along the midpoint of the one-inch thick dimension. The configuration and dimensions for this stress rupture specimen are given in Figure 2.

A single tensile specimen was removed from each of the eight plates adjacent to the stress rupture specimens. These specimens were standard one-quarter inch diameter samples per FTMS 151, Method 211.1 and were specified for 1400 F tensile tests to determine stress values for the initial stress rupture tests.

Three (3) one-quarter-inch diameter FTMS 151, Method 211.1 tensile specimens and three (3) Charpy V-notch impact specimens per ASTM E-23 (Type A) were prepared from each of the eight age-hardened 5-1/2 inch by 4-1/4 inch plates. These specimens were orientated and removed as shown in Figure 3. The tensile specimens were aligned with the rolling direction and located at the midpoints of the one-inch thick dimension for the plates. The Charpy V-notch specimens were also oriented in the longitudinal direction. The notches were directed through the thickness of the plate and located such that the notch fronts were at the center of the plate.

The configuration, dimensions, and tolerances for the FTMS 151, Method 211.1 (R3 Type) tensile specimens are shown in Figure 4. The configuration for the Charpy V-notch impact specimens is shown in Figure 5 along with the tolerances listed for ASTM E-23 (Type A).

2.3 Testing Procedures

Room temperature tensile tests were performed per ASTM E-8. Specimens tensile tested at 1400 F were conducted per ASTM E-21. Samples were inserted into a furnace operating at 1400 F for the high temperature tests. The temperature of the samples were monitored by three thermocouples wired onto each specimen. Five additional minutes were allowed for each sample to soak after it reached 1400 F. This ensured that the samples were heated uniformly throughout when the tests were initiated. The samples were strained at a rate of 0.005 min^{-1} to just beyond the yield point and thereafter at a rate of 0.05 min^{-1}.

The Charpy V-notch impact specimens were tested at room temperature according to ASTM Method E-23.

The stress rupture tests were conducted at 1400 F according to ASTM Recommended Practices E-139. This included placing the specimen into a furnace operating at 1400 F. Each specimen was then stressed with a constant load within five minutes from the time at which it was inserted into the furnace. Three thermocouples wired to each specimen were used to record its temperature. The specimen temperature was recorded immediately after insertion into the furnace in addition to the time to reach 1400 F and the time for failure to occur. Percent elongations and reductions in area were determined from the failed specimens.

2.4 Metallographic Procedures

Light microscopy, scanning electron microscopy, and transmission electron microscopy were used to examine solution-annealed material which had been either rapidly or slowly cooled. Failed stress rupture specimens representing some conditions of Section IV.1 were also examined by these three metallographic techniques.

Samples for examination by light microscopy were prepared by sectioning with a water cooled cut-off wheel, mounting in bakelite, grinding, polishing, and etching. Grinding was accomplished by successive stages using SiC papers of 120, 240, 400 and 600 grit. The samples were then polished with 9 micron diamond on nylon cloth. A final polish was attained using 0.3 micron alumina abrasive on a Syntron. Various etchants were used and are identified throughout this report as follows:

Etchant No. 1: composition - 25 gr CrO_3, 7 ml H_2O, and 130 ml acetic acid
method - electrolytic

Etchant No. 2: composition - 5 ml chromic acid (granular) dissolved in 100 ml H_2O
method - electrolytic

The scanning electron microscopy investigations were performed upon a Cambridge 600 instrument. Two types of investigations were performed. Fractography studies were made upon the failed PBF acoustic filter and a few of the stress rupture specimens. The microstructures of polished and etched samples of rapid and slowly cooled solution-annealed material were also checked. In this latter case, the samples were prepared by the process described for light microscopy. Etchant No. 2 was used to reveal the microstructure.

Transmission electron microscopy was applied as a means to reveal the precipitation phases characteristic of the rapid and slowly cooled material. Replication techniques involving Pt-20%Rd alloy shadowing on two-stage plastic and carbon replicas were used for polished and etched samples.

3. TEST RESULTS

The room temperature tensile properties for the aged samples having received the eight different solution annealing treatments shown in Table II are listed in Table III. All of these samples were given an initial 1800 F stress relief, cycle No. 1, and a final duplex age, cycle No. 9, which consisted of 10 hours at 1400 F and 9 hours at 1200 F. The intermediate heat treating cycles are also listed beneath Table III.

TABLE III

ROOM TEMPERATURE TENSILE DATA FOR HEAT 53C2EX OF INCONEL 718

Specimen	Heat Treatment Cycles (From Table I)	0.2% Yield Stress, ksi	Ultimate Tensile Strength, ksi	% Elong	% Red. in Area
AA - 1	1, 5,	149	192	20	25
2	and	152	193	20	26
3	9	151	194	19	25
EE - 1	1-5	150	192	20	23
2	and	148	191	18	24
3	9	150	193	17	23
BB - 1	1, 6	128	187	17	19
2	and	127	186	15	20
3	9	128	189	17	21
FF - 1	1-4, 6	127	187	17	22
2	and	129	187	18	22
3	9	132	189	17	21
CC - 1	1, 7	156	193	15	19
2	and	156	195	15	19
3	9	156	195	16	19
GG - 1	1-4, 7	157	194	17	19
2	and	155	194	15	20
3	9	158	194	16	20
DD - 1	1, 8	125	181	11	13
2	and	125	184	13	17
3	9	125	184	16	20
HH - 1	1-4, 8	128	185	14	16
2	and	130	185	12	15
3	9	129	183	13	15

Cycle 1 - 1800 F stress relief
Cycle 2 - 2000 F solution anneal
Cycle 3 - 1400-1200 F duplex age
Cycle 4 - 1750 F stress relief
Cycle 5 - 2000 F S.A.: 900 F/hr cool
Cycle 6 - 2000 F S.A.: 75 F/hr cool
Cycle 7 - 1900 F S.A.: 1860 F/hr cool
Cycle 8 - 1900 F S.A.: 90 F/hr cool
Cycle 9 - 1400-1200 F duplex age

The room temperature Charpy V-notch impact values for the corresponding aged conditions are shown below in Table IV. The samples have all recieved the initial 1800 F stress relief, cycle No. 1, and the final 1400-1200 F duplex age, cycle No. 9. The intermediate heat treating cycles specified are described in Table II and listed after Table IV.

TABLE IV

ROOM TEMPERATURE CHARPY V-NOTCH IMPACT TEST DATA
FOR INCONEL 718, HEAT 83C2EK

Specimen	Heat Treatment Cycles (from Table II)	Energy Absorbed, ft-lb	Lateral Deformation (inch)
AA - 20	1, 5	26.0	.015
21	and	26.0	.015
22	9	25.5	.014
EE - 20	1-5	27.0	.016
21	and	26.0	.016
22	9	26.0	.016
BB - 20	1, 6	23.0	.013
21	and	22.5	.013
22	9	22.5	.013
FF - 20	1-4, 6	26.0	.017
21	and	26.0	.016
22	9	26.5	.015
CC - 20	1, 7	17.5	.010
21	and	17.5	.009
22	9	18.0	.009
GG - 20	1-4, 7	20.0	.011
21	and	20.0	.014
22	9	19.5	.013
DD - 20	1, 8	17.0	.010
21	and	20.0	.014
22	9	16.5	.010
HH - 20	1-4, 8	17.0	.010
21	and	18.0	.010
22	9	18.0	.010

Cycle 1 - 1800 F stress relief
Cycle 2 - 2000 F solution anneal
Cycle 3 - 1400-1200 F duplex age
Cycle 4 - 1750 F stress relief
Cycle 5 - 2000 F S.A.: 900 F/hr cool
Cycle 6 - 2000 F S.A.: 75 F/hr cool
Cycle 7 - 1900 F S.A.: 1860 F/hr cool
Cycle 8 - 1900 F S.A.: 90 F/hr
Cycle 9 - 1400-1200 F duplex age

Results for the 1400 F tensile tests for the solution-annealed conditions are shown in Table V. All samples have received the initial 1800 F stress relief. The subsequent heat treat cycles, No. 2 through No. 8, are identified in Table II and listed beneath in Table V.

TABLE V

1400 F TENSILE TEST DATA, INCONEL 718 HEAT 83C2EK

Specimen	Heat Treatment Cycles (from Table II)	0.2% Yield Stress, ksi	Ultimate Tensile Strength, ksi	% Elong.	% Red. In Area
A-4	1, 5	92.6	111	5	7
A-4*	1, 5	95.1	114	8	10
B-4	1, 6	92.8	112	15	14
C-4	1, 7	82.9	92.4	7	8
D-4	1, 8	92.6	104	23	25
E-4	1-5	101	112	5	11
E-4*	1-5	98.1	116	7	9
F-4	1-4, 6	94.0	105	14	13
G-4	1-4, 7	73.2	80.4	8	9
H-4	1-4, 8	92.8	106	11	15

*Duplicate test

Cycle 1 - 1800 F stress relief
Cycle 2 - 2000 F solution anneal
Cycle 3 - 1400-1200 F duplex age
Cycle 4 - 1750 F stress relief
Cycle 5 - 2000 F S.A.: 900 F/hr cool
Cycle 6 - 2000 F S.A.: 75 F/hr cool
Cycle 7 - 1900 F S.A.: 1860 F/hr cool
Cycle 8 - 1900 F S.A.: 90 F/hr cool

Stress rupture data collected at 1400 F for the various solution-annealed conditions are listed in Table VI. This Table identifies the heat treat cycles, stress level, time to reach 1400 F and total time to failure, elongation, and reduction in area. It also identifies those specimens which failed in the notch. The heat treat cycles are shown in detail in Table II and also listed at the end of Table VI.

TABLE VI

1400 F STRESS RUPTURE TEST DATA, HEAT 83C2EK OF INCONEL 718

Specimen	Heat Treatment Cycles (from Table II)	Stress Level, ksi	Spec. Temp. at Insertion, F	Time: Hours to (1400 F)	- Failure	Temp. at Failure, F	% Elong.	% Red. in Area
A-5	1, 5	100	700	0.6 to failure		1370	3	(failed in the notch)
A-6		90	870	0.5	0.8	1400	4	17
A-7		79	780	0.6	1.9	1400	4	12
A-8		79	850	0.9	2.5	1400	4	12
A-9		70	770	1.2	6.4	1400	5	7
A-10		70	850	1.0	5.5	1400	3	15
A-11		60	700	1.0	24.1	1400	12	20
A-12		60	800	1.4	25.5	1400	10	19
A-13		60	750	1.2	24.9	1400	13	22
B-5	1, 6	88.5	750	1.1	1.3	1400	19	25
B-6		79	650	0.6	2.4	1400	20	29
B-7		79	900	0.8	2.4	1400	19	34
B-8		70	800	0.6	8.4	1400	21	30
B-9		70	800	0.7	9.3	1400	14	27
B-10		60	740	1.5	46.9	1400	15	24
B-11		60	600	1.3	14.3	1400	18	25
B-12		60	700	1.1	13.2	1400	16	29
B-13		60	670	1.0	17.4	1400	20	28
B-14		55	650	1.3	39.1	1400	15	27
C-5	1, 7	79	700	0.9	1.7	1400	6	17
C-6		70	900	1.0	9.1	1400	25	29
C-7		70	820	1.2	7.8	1400	17	20
C-8		60	870	0.9	34.1	1400	21	38
C-9		60	780	1.3	20.2	1400	27	36
C-10		60	770	1.2	19.9	1400	16	34
C-11		60	850	1.0	18.7	1400	17	36
C-12		79	750	0.8	1.3	1400	8	12

TABLE VI (CONT.)

Specimen	Heat Treatment Cycles (from Table II)	Stress Level, ksi	Spec. Temp. at Insertion, F	Time: Hours to (1400 F)	- Failure	Temp. at Failure, F	% Elong.	% Red. in Area
D-5	1, 8	88.5	760	0.7	1.1	1400	34	38
D-6		79	680	1.0	2.2	1400	35	41
D-7		79	700	1.1	1.8	1400	30	38
D-8		70	800	2.0	5.2	1400	25	43
D-9		70	900	1.0	4.1	1400	25	43
D-10		60	700	0.9	13.3	1400	24	39
D-11		60	780	1.0	14.4	1400	25	40
D-12		60	650	0.6	16.2	1400	22	40
E-5	1-5	105	575	0.6	0.6	1360	8	14
E-6		79	700	0.9	2.1	1400	3	7
E-7		70	720	0.7	12.4	1400	8	12
E-8		70	650	1.1	10.0	1400	5	9
E-9		60	770	0.4	40.8	1400	10	20
E-10		60	860	0.9	23.6	1400	13	15
E-11		60	920	0.5	24.8	1400	14	17
E-12		60	680	0.7	26.8	1400	13	12
E-13		79	700	0.8	2.4	1400	5	11
F-5	1-4, 6	88.5	800	1.2	1.4	1400	18	22
F-6		79	750	0.9	2.0	1400	22	27
F-7		79	750	1.1	1.8	1400	17	20
F-8		70	800	0.6	9.3	1400	20	22
F-9		70	650	1.0	5.7	1400	25	27
F-10		60	750	1.0	22.4	1400	14	26
F-11		60	750	1.1	19.9	1400	17	22
F-12		60	750	1.0	22.2	1400	17	27
G-5	1-4, 7	88.5	750	0.9 to failure		1355	9	22
G-6		79	850	0.9	0.9	1400	6	17
G-7		79	700	1.2 to failure		1390	6	18
G-8		70	730	0.6 to failure		1390	1	(failed in the notch)
G-9		70	700	0.9	2.7	1400	4	12

TABLE VI (CONT.)

Specimen	Heat Treatment Cycles (from Table II)	Stress Level, ksi	Spec. Temp. at Insertion, F	Time: Hours to (1400 F)	Time: Hours to Failure	Temp. at Failure, F	% Elong.	% Red. in Area
G-10	1-4, 7	60	870	0.7	23.7	1400	18	27
G-11		60	850	0.7	19.7	1400	17	17
G-12		60		Broke in the threads				
G-13		60	630	1.1	24.5	1400	18	22
G-14		79	830	0.8 to failure		1370	3	(failed in the notch)
G-15		70	800	0.9	1.8	1400	2	13
H-5	1-4, 8	88.5	800	1.0	1.2	1400	18	25
H-6		79	870	0.9	2.7	1400	24	27
H-7		69	850	0.9	2.7	1400	29	27
H-8		70	800	0.6	8.4	1400	25	33
H-9		70	900	0.8	8.1	1400	17	32
H-10		60	650	1.2	16.1	1400	21	29
H-11		60	700	1.0	17.1	1400	18	34
H-12		60	730	0.7	17.9	1400	19	33

Cycle 1 - 1800 F stress relief
Cycle 2 - 2000 F solution anneal
Cycle 3 - 1400-1200°F duplex age
Cycle 4 - 1750 F stress relief
Cycle 5 - 2000 F S.A.: 900 F/hr cool
Cycle 6 - 2000 F S.A.: 75 F/hr cool
Cycle 7 - 1900 F S.A.: 1860 F/hr cool
Cycle 8 - 1900 F S.A.: 90 F/hr cool

V. MODIFIED AGING STUDIES

The material degradation tests (Section IV) showed that slowly cooled material given the ten hour 1400 F plus eight hour 1200 F duplex age obtained room temperature yield strengths of approximately 125 ksi compared to 150 ksi for this material when rapidly cooled and aged. A program was initiated to develop adequate room temperature yield strength for Inconel 718 when furnace cooled at rates of 100 F and 400 F per hour. The 400 F per hour cooling rate was included in the event that 145 ksi could not be obtained from materials cooled at a rate of 100 F per hour. It was hoped that the intermediate cooling rate of 400 F would then provide a compromise giving lower thermal stresses and a room temperature yield strength of 145 ksi. This program consisted of checking the hardness of samples aged for various combinations of times and temperatures. An aging cycle which resulted in a promising hardness value would then be selected for combination smooth-notched specimens. These specimens would then be tensile tested at room temperature to determine if the required design strength of 145 ksi for the PBF acoustic filters was obtained.

1. MATERIAL SOURCE

This study employed two different heats of material The initial study utilized samples from heat 83C2EK. This was the heat used for the gussests on the PBF acoustic filters and in the material degradation tests. The material was taken from section B (see Figure 1) which was heat treated per cycles 1 and 6 in Table II. This represented solution-annealed material cooled at a rate of 75 F per hour from 2000 F. Once the hardness values were determined, a selected aging cycle was applied to two combination smooth-notched stress rupture specimens from the material degradation program. These specimens had also been machined from Section B.

A second study was made to determine the aging response for material cooled at a rate of 400 F per hour from a 2000 F solution anneal. This material was removed from the failed section of PBF acoustic filter Unit No. 1. The chemistry for this heat of material identified as 87C0EK1 is given in Table I. Hardness values were again recorded for various aging cycles. A cycle was selected and given to two more of the stress rupture specimens from heat 83C2EK after being solution annealed at 2000 F and cooled at a rate of 400 F per hour prior to aging.

2. EXPERIMENTAL PROCEDURES

2.1 Heat Treating Procedures

Blocks having dimensions of approximately 1/2 inch by 1/2 inch by 1/4 inch were cut from the materials described in Section V.1. These pieces were sealed in heat treating bags after being purged with argon and aged according to Aging Schedules I and II shown below. The heat treatments were performed in a resistance heated laboratory furnace which controlled within $^{+15}_{-0}$F of the desired aging temperature.

AGING SCHEDULE I

Samples per Package for Step #1	Step #1 1400 F (Removal time)	Step #2 Times at 1200 F Aging Step			
5	2 hr	4 hr	8 hr	16 hr	30 hr
5	4 hr	4 hr	8 hr	16 hr	30 hr
3	8 hr	----	8 hr	16 hr	-----
5	10 hr	4 hr	8 hr	16 hr	30 hr
3	16 hr	----	8 hr	16 hr	-----
3	24 hr	----	8 hr	16 hr	-----

AGING SCHEDULE II

Samples per package for Step #1	Step #1 1325 F (Removal time)	Step #2 Times at 1150 F Aging Step			
5	2 hr	4 hr	8 hr	16 hr	30 hr
5	4 hr	4 hr	8 hr	16 hr	30 hr
3	8 hr	----	8 hr	16 hr	-----
5	10 hr	4 hr	8 hr	16 hr	30 hr
3	16 hr	----	8 hr	16 hr	-----
3	24 hr	----	8 hr	16 hr	-----

A set of six heat treating bags was inserted for both the 1400 F or 1325 F steps as shown in Schedule I and II. Bags were then removed and air cooled after times of 2, 4, 8, 10, 16, and 24 hours. Samples for various conditions were then placed into another set of four bags which were aged at 1200 F or 1150 F for times of 4, 8, 16 and 30 hours.

2.2 Method for Recording Hardness Values

An edge of each block was polished following the aging cycles. Approximately 0.005 inch was ground off to remove any effects of oxidation. These surfaces were ground to a 600 grit finish. Hardness values were recorded with a Rockwell Hardness Tester using a sphero-conical diamond penetrator for the C-scale. The machine calibration was checked at intervals with a calibration block specified as having Rc values of 45.1 ± 0.1. These checks taken near the beginning, middle, and end of hardness measurements showed the following values:

1st check - 44.8, 45.2, and 45.0; 2nd check - 44.8, 44.7, and 44.8; and 3rd check - 45.2, 45.3, and 45.2.

2.3 Preparation and Testing of Tensile Specimens

Combination smooth-notched stress rupture specimens, heat 83C2EK, from the material degradation program, Section IV, were used for the room temperature tensile tests. These specimens were prepared per AMS specification 5663 (specimen type -5).

Testing was performed per ASTM E-8 at room temperature on a 60,000 lb. capacity screw driven Tinius Olsen testing machine. Elongation recordings for flow curves were obtained from a microformer extensometer. Elongation values for the failed specimens were established from one inch gage marks.

3. RESULTING HARDNESS AND TENSILE VALUES

The Rc hardness values for the various combinations of aging times and temperatures are shown in Table VII and Table VIII. Each value represents the average from three measurements.

TABLE VII

Rc HARDNESS VALUES DEVELOPED FOR HEAT 83C2EK OF INCONEL 718 COOLED AT A RATE OF 75 F PER HOUR FROM 2000 F AND AGED

Time at Initial 1400 F Age	Hardness	Time at Second Aging Step (1200 F) 4 hrs	8 hrs	16 hrs	30 hrs
Initial Value	37.0				
2 hrs	36.9	38.6	38.5	38.8	40.7
4 hrs	37.8	38.2	38.8	38.5	41.2
8 hrs	37.5		38.5	39.8	
10 hrs	35.4	37.3	37.8	39.2	40.6
16 hrs	35.4		37.3	38.5	
24 hrs	36.6		37.6	38.1	

Time at Initial 1325 F Age	Hardness	Time at Second Aging Step (1150 F) 4 hrs	8 hrs	16 hrs	30 hrs
Initial Value	37.0				
2 hrs	38.7	38.7	38.2	39.7	40.2
4 hrs	38.7	39.0	38.0	40.2	40.9
8 hrs	38.5		38.2	39.9	
10 hrs	38.0	39.3	38.7	39.8	40.5
16 hrs	38.7		39.3	39.7	
24 hrs	38.5		38.8	39.9	

TABLE VIII

Rc HARDNESS VALUES DEVELOPED FOR HEAT 87COEK1 OF INCONEL 718 COOLED AT A RATE OF 400 F PER HOUR FROM 2000 F AND AGED

Time at Initial 1400 F Age	Hardness	Time at Second Aging Step (1200 F) 4 hrs	8 hrs	16 hrs	30 hrs
Initial Value	34.6				
2 hrs	36.7	39.3	40.0	40.3	42.1
4 hrs	36.9	40.8	41.1	41.2	41.7
8 hrs	37.1		41.5	42.6	
10 hrs	37.6	39.8	41.5	41.4	42.4
16 hrs	37.2		40.2	41.8	
24 hrs	36.9		40.4	41.4	

Time at Initial 1325 F Age	Hardness	Time at Second Aging Step (1150 F) 4 hrs.	8 hrs.	16 hrs.	30 hrs.
Initial Value	34.6				
2 hrs	38.0	40.4	41.9	41.2	41.7
4 hrs	39.1	40.4	40.8	42.3	42.4
8 hrs	39.9		42.1	42.4	
10 hrs	39.1	40.6	41.7	42.3	43.0
16 hrs	39.4		41.5	42.5	
24 hrs	40.4		41.2	42.5	

From the hardness values collected in Table VII, a lower temperature (1325-1150 F) duplex age was selected. This duplex cycle consisted of four (4) hours at 1325 F plus sixteen (16) hours at 1150 F. Slightly higher hardness values were observed when the second aging step was extended to thirty (30) hours. However, a sixteen hour step was selected to limit the total aging cycle to approximately twenty (20) hours. A similar aging cycle was selected for the material cooled at 400 F per hour. The stress rupture specimens were aged accordingly and tensile tested at room temperature. The tensile data are given in Table IX.

TABLE IX

TENSILE PROPERTIES FOR SAMPLES STRENGTHENED WITH A MODIFIED 1325-1150 F DUPLEX AGE, HEAT 83C2EK OF INCONEL 718

Sample	Cooling Rate From Solution Anneal	0.2% Yield Stress, ksi	Ultimate Tensile Strength, ksi	% Elong.	% Red. in Area
BB5	100 F/hr	148*	197	16	14
BB6	100 F/hr	144	196	17	17
BB7	400 F/hr	*	189	24	27
BB8	400 F/hr	*	189	24	27

* The yield strength was questionable or nondeterminable as a result of the extensometer slipping.

VI. MATERIAL AND HEAT TREATING TESTS PERFORMED ON ACOUSTIC FILTER MATERIAL (HEAT 87COEK1)

Tests were conducted upon material removed adjacent to the crack in the shell of acoustic filter Unit No. 1. The purpose of these tests was twofold. To check the material properties at the location of failure and to confirm the response of this heat of material, 87COEK1, to the modified 1325-1150 F duplex age selected in Section V.3. The yield strength resulting from this modified age needed to be checked since the preceeding tensile tests were performed on a different heat. These checks would also eliminate the uncertainity caused by the slipage of the extensometer grips indicated in Table IX. Included in these checks were room temperature impact tests for the aged conditions as well as 1400 F stress rupture tests for the solution-annealed materials.

1. SOURCE OF MATERIAL REMOVED FROM THE CRACKED ACOUSTIC FILTER

Initially tests were conducted as part of a failure analysis to check the properties of the shell material heat treated to duplicate the conditions of Unit No. 1 during the heat treatments preceeding failure (Ref. 1). This involved performing 1400 F stress rupture tests on solution-annealed material cooled from 2000 F at an approximate rate of 1200 F per hour and room temperature tensile and impact tests on such material when subsequently aged ten (10) hours at 1400 F plus eight (8) hours at 1200 F. The material was sawed from a location directly beneath the trunnion near the failure site.

Latter checks were made upon this same material having modified heat treatments. These were to check the 1400 F stress rupture properties for material cooled at rates of 100 and 400 F per hour from 2000 F solution anneals. Room temperature tensile and impact properties were then determined for these conditions when aged four (4) hours at 1325 F plus sixteen (16) hours at 1150 F. The material for these checks was removed from locations near the trunnion positioned opposite the failure. Two sections measuring 6 inches by 8 inches were plasma cut from the failed unit and then sawed into 6 inch by 4 inch sections. An inch of material adjacent to each plasma cut was sawed off and discarded.

2. EXPERIMENTAL PROCEDURES

2.1 Heat Treatment of Filter Sections

The material for the failure analysis checks was sawed into two sections measuring 3 inches by 4 inches. These sections were inserted into a resistance heated laboratory furnace operating at 1800 F. They were then heated to 2000 $^{+25}_{-0}$F at a rate of 100 F per hour; held at this temperature for one (1) hour ± 5 minutes, cooled at a rate of approximately 1200 F per hour to 1200 F, and then air cooled to room temperature. The temperature was recorded and controlled by a thermocouple positioned between the two pieces. One section was then aged. It was placed into the furnace operating at 1400 $^{+15}_{-0}$F, held for ten (10) hours, cooled to 1200 $^{+15}_{-0}$F at a rate of 100 F per hour, held for an additional eight (8) hours, and then air cooled to room temperature.

Two (2) 3 inch by 4 inch sections were similarly heated to 2000 $^{+25}_{-0}$F and held for one (1) hour. These sections were then cooled to 1200 F at a rate of approximately 400 F per hour after which they were air cooled to room temperature. Two (2) additional sections were similarly solution annealed except they were cooled from 2000 F to 1200 F at a rate of 100 F per hour. A section from each of these two solution-annealed conditions was aged at 1325 F and 1150 F. They were placed into a furnace operating at 1325 $^{+15}_{-0}$F, held for four (4) hours, cooled to 1150 $^{+15}_{-0}$F at a rate of 100 F per hour, held for an additional sixteen (16) hours, and then air cooled to room temperature. All heat treating operations were performed in air.

2.2 Preparation and Testing of Specimens

Six (6) stress rupture specimens were prepared from each of the three solution-annealed conditions. Smooth bar specimens were used since no notch sensitivity was detected in prior stress rupture tests. These one-quarter inch diameter specimens were prepared per Figure 8 of ASTM E-8. Their longitudinal axes were along the rolling direction of the plate and located at the midpoint of the plate thickness.

Three (3) tensile specimens were prepared from each of the three aged conditions. These specimens were of the same type and had the same orientation and location as the stress rupture specimens described in the preceeding paragraph. In addition, three (3) impact specimens were prepared from each of the aged conditions. These were Charpy V-notch specimens per ASTM E-23 (Type A). The specimens were aligned longitudinally with the plate and the notches directed through the thickness of the plate. The root of the notches were located at the center of the plate thickness.

The stress rupture specimens were tested at 1400 F at stress levels of 79, 70 and 60 ksi. They were inserted into a furnace operating at 1400 F and then loaded within five minutes. Three thermocouples were wired to each specimen to monitor the temperature. The times were recorded for the samples to reach 1400 F and for failure to occur. Percent elongations and reductions in area were determined from the failed specimens.

The tensile and impact specimens were tested at room temperature according to ASTM Methods E-8 and E-23, respectively.

3. TENSILE, IMPACT AND STRESS RUPTURE DATA

The 1400 F stress rupture data for solution-annealed acoustic filter shell material cooled at rates of 1200, 400, and 100 F per hour are listed in Table X. Room temperature tensile and impact properties for material aged to the various conditions described in Section VI.2.1 are given in Tables XI and XII, respectively.

TABLE X

STRESS RUPTURE PROPERTIES FOR SOLUTION-ANNEALED CONDITIONS FOR HEAT 87COEK1 OF INCONEL 718

Specimen	Thermal Treatment	Stress Level, ksi	Spec. Temp. at Insertion, F	Time: Hours to (1400 F)	Time: Hours to Failure	Temp. at Failure, F	% Elong.	% Red. in Area
FS1		79	700	1.0	3.9	1400	19	26
FS2		70	500	1.9	11.4	1400	22	29
FS3		70	810	1.0	9.3	1400	23	30
FS4		79	700	1.3	3.6	1400	25	29
FS5	Solution annealed	60	700	0.9	29.3	1400	22	32
FS6	at 2000 F; cooled	Tensile Test		Y.S. - 95 ksi		U.T.S-110 ksi	14	18
FI1		79	675	1.3	2.2	1400	8	14
FI2		70	725	0.9	8.7	1400	25	28
FI3		70	750	1.2	8.4	1400	21	20
FI4		79	750	0.8	1.4	1400	5	16
FI5		60	690	0.8	26.0	1400	23	28
FI6		Tensile Test		0.2% Y.S. - 87 ksi		U.T.S.-103 ksi	10	16
FA1		79	800	0.7	0.7	1400	5	19
FA2	Solution annealed	70	830	0.9	5.3	1400	9	13
FA3	at 2000°F; cooled	70	770	0.7	2.8	1400	5	10
FA4		60	950	0.6	31.1	1400	25	27
FA5		60	800	0.6	30.0	1400	23	31
FA6		60	800	0.8	28.2	1400	25	34

Cooling rates: (FS)-100 F/hr , (FI)-400 F/hr , (FA)-1200 F/hr

TABLE XI

ROOM TEMPERATURE TENSILE PROPERTIES FOR AGE-HARDENED CONDITIONS OF HEAT 87COEK1 OF INCONEL 718

Specimen	Thermal Treatment	0.2% Yield Stress, ksi	Ultimate Tensile Strength, ksi	% Elong.	% Red. in Area
FS7	Solution annealed at 2000 F; cooled; aged 4 hr at 1325 F plus 16 hr at 1150 F	143	199	20	31
FS8		145	201	20	29
FS9		146	202	20	29
FI7		150	195	25	36
FI8		158	196	24	33
FI9		157	196	24	35
FA7	Solution annealed at 2000°F; cooled; aged 10 hr at 1400 F plus 8 hr at 1200 F	154	205	18	25
FA8		160	202	17	23
FA9		153	201	18	24

Cooling rates: (FS)-100 F/hr, (FI)-400 F/hr, (FA)-1200 F/hr

TABLE XII

ROOM TEMPERATURE IMPACT PROPERTIES FOR AGE-HARDENED CONDITIONS OF HEAT 87COEK1 OF INCONEL 718

Specimen	Thermal Treatment	Energy Absorbed, ft-lb	Lateral Deformation (inch)
FS10	Solution annealed at 2000 F; cooled; aged 4 hr at 1325 F plus 16 hr at 1150 F	49.0	0.027
FS11		51.5	0.028
FS12		53.5	0.029
FI10		59.0	0.031
FI11		60.0	0.032
FI12		61.0	0.032
FA10	Solution annealed at 2000 F; cooled; aged 10 hr at 1400 F plus 8 hr at 1200 F	40.0	0.017
FA11		40.5	0.020
FA12		44.0	0.018

Cooling rates: (FS)-100 F/hr, (FI)-400 F/hr, (FA)-1200 F/hr

VII. HIGH TEMPERATURE DUCTILITY STUDIES PERFORMED ON BASE METAL AND SIMULATED HEAT AFFECTED ZONE SAMPLES

Slow cooling rates have been reported to cause overaging and reduced postweld heat treat cracking in Rene' 41, another nickel-base precipitation hardenable alloy (2,3). Berry and Hughes (2) propose that overaging allows base metal to be weaker and more ductile than heat affected zones. This will allow stress relaxation to occur in base metal rather than HAZ. Fawley and Prager (4) related tensile ductility to the cracking occurrence observed in circular weld patch specimens by Thompson et al, (3). Prager and Sines (5) then propose an explanation for the reduced cracking tendency with overaged Rene' 41. Their explanation involved the influence of precipitate size upon deformation characteristics. Small precipitates would be sheared by dislocations resulting in heterogeneous planar deformation with stress concentrations developing at grain boundaries. In contrast, coarse precipitates will cause homogeneous intragranular slip and a reduced cracking tendency. These studies have led to the utilization of slow cooling rates or an overaged condition as a preweld heat treatment to reduce postweld heat-treat cracking in Rene' 41.

It was desirable to see if cooling Inconel 718 slowly would induce a similar reduction in cracking tendency. Two programs were conducted to determine the high temperature tensile ductility of Inconel 718 thermally cycled to represent heat affected zones. One study involved checking the tensile ductility of base metal and simulated heat affected zones of Inconel 718 which had been subjected to postweld thermal cycles consisting of solution annealing at 2000 F and cooling rates of 1200 and 100 F per hour. The material degradation studies (Section IV) have shown that 1400 F ductility of solution-annealed base metal was improved by slow cooling. If this benefit persisted for heat affected zones, the possibility of cracking as a result of residual thermal stresses such as that reported for Incident No. PBF-73-9 (Ref. 1) would be lessened. A second study checking the high temperature ductility of heat affected zones, subjected to various preweld thermal cycles, was conducted and reported in Section VIII.

1. MATERIAL AND SPECIMEN IDENTITY

The samples used for this study were forty-one (41) excess combination smooth-notched stress rupture specimens from the material degradation program. These specimens were derived from heat 83C2EK and were machined according to specimen Type-5 of AMS specification 5663, see Figure 2. The samples had been solution-annealed to various conditions outlined in Table II. However, in this study the samples were cycled to 2250 F with an induction coil, to simulate a HAZ, and subsequently solution annealed at 2000 F. It was assumed, at the time these tests were conducted, that these steps would obscure any previous thermal history.

2. EXPERIMENTAL PROCEDURES

2.1 Thermal Cycling Specimens with Induction Coil

Twenty-one (21) of the specimens were heated with an induction coil to produce simulated heat affected zones. The samples were thermally cycled by positioning the gage sections of the specimens within the hot zone of the coil. The samples were then heated to temperatures ranging from 2250 F to 2270 F. They were held at this peak temperature for two to three seconds and then cooled to 1400 F in approximately twenty-five seconds. This entire operation was performed in an argon purge to keep oxidation to a minimum and assist in rapidly cooling the samples. The temperatures were recorded by thermocouples spot welded onto the gage sections.

2.2 Heat Treatment of Specimens

Samples in this study were given two different solution anneals. In one case, ten (10) parent metal samples and ten (10) simulated HAZ samples were heated to 2000 F, held for one hour, cooled at a rate of 100 F per hour to 1200 F, and then air cooled. In the other case, ten (10) parent metal samples plus eleven (11) simulated HAZ samples were similarly heated to 2000 F and held for one hour. These samples were, however, cooled from 2000 F to 1200 F at a rate of approximately 1200 F per hour.

Two samples from each of the four conditions were aged. The samples cooled at a rate of 100 F per hour were given the modified age of four (4) hours at 1325 F plus sixteen (16) hours at 1150 F. The samples having experienced the more rapid cool were aged for ten (10) hours at 1400 F plus eight (8) hours at 1200 F.

All of the heat treatments were performed in a laboratory resistance heated furnace which controlled within ± 7.5 F. The samples were placed in heat treating bags, purged with argon and sealed. The cooling rates were maintained by controlling the furnace temperature. Forced air was used to reduce the furnace temperature during the 1200 F per hour cool.

2.3 Method for High Temperature Tensile Testing

The high temperature tensile tests were conducted in the same manner as those in the material degradation studies. Samples were placed into a furnace operating at the selected temperature. The samples were allowed to heat to this temperature after which they were held for five minutes to obtain equilibrium. They were then strained at a rate of 0.005 in. per min beyond the yield points and thereafter at a rate of 0.05 in. per min. Sample temperatures were monitored by thermocouples wired onto the gage sections.

3. RESULTS OF HIGH TEMPERATURE TENSILE TESTS

The tensile data for the parent metal and simulated HAZ samples having cooling rates of 100 and 1200 F per hour from 2000 F solution anneals are shown in Table XIII.

TABLE XIII

ELEVATED TEMPERATURE TENSILE DATA FOR INCONEL 718
SIMULATED HEAT-AFFECTED ZONE AND PARENT METAL SAMPLES, HEAT 83C2EK

Specimen	Test Temperature, F	0.2% Yield Stress, ksi	Ultimate Tensile Strength, ksi	% Elong.	% Red. in Area
HS7	1000	104	150	8	14
HS6	1100	103	153	9	14
HS1	1200	101	142	13	16
HS3*	1200	116	148	6	9
HS8*	1200	120	150	5	9
HS2	1300	100	127	15	17
HS4	1300	104	134	12	16
HS9	1300	102	133	5	9
HS10	1400	94	114	12	15
HS5	1500	73	80	18	19
PS8	1000	100	158	16	17
PS10	1100	103	159	14	15
PS3*	1200	118	151	16	18
PS4*	1200	118	155	13	16
PS7	1200	100	152	15	17
PS1	1300	101	138	15	17
PS5	1400	93	103	17	19
PS9	1500	70	81	16	18
HF6	1100	84	108	9	21
HF5	1200	87	120	13	20
HF10**	1200	131	157	7	8
HF11**	1200	130	159	8	12
HF8	1250	91	121	11	20
HF1	1300	92	124	15	22
HF2	1400	85	98	7	9
HF7	1400	86	103	7	14
HF3	1500	73.	85	7	14
HF9	1550	61	65	18	44***
HF4	1600	46	47	41	56***
PF1**	1200	130	159	14	19
PF2**	1200	129	158	16	16
PF8	1200	78	118	30	33
PF10	1250	90	127	25	30
PF3	1300	90	122	22	35
PF5	1400	86	98	7	16
PF6	1400	80	110	10	16
PF4	1500	71	83	9	15
PF9	1550	63	67	19	22
PF7	1600	45	46	41	61

NOMENCLATURE:

HS: Heat Affected Zones with 100 F/hr furnace cool from solution anneal temperature.
PS: Parent Metal with 100 F/hr furnace cool from solution anneal temperature.
HF: Heat Affected Zones with 1200 F/hr cool from solution anneal temperature.
PS: Parent Metal with 1200 F/hr cool from solution anneal temperature.

Notes: 1) HAZ simulated by induction heating to 2250-2270 F.
2) All material subsequently solution anneal at 2000 F.

*Aged at 1325 F for 4 hours plus 16 hours at 1150 F.
**Aged at 1400 F for 10 hours plus 8 hours at 1200 F.
***Those specimens failed outside the simulated Heat Affected Zones.

VIII. PREWELD IMMUNIZATION CHECK

Several authors[2,3] have promoted the use of slow cooling rates or overaged conditions as a preweld condition to reduce postweld heat-treat cracking in Rene' 41 (see Section VII). It was desirable to check if such a preweld condition would be beneficial in Inconel 718. An investigation was therefore conducted to check 1400 F ductilities of samples given simulated HAZ cycles after having been stress relieved at 1725 F and rapidly or slowly cooled.

1. SPECIMEN IDENTITY AND THERMAL HISTORY

Hourglass specimens were prepared per ANC drawing No. 403350 from heat 2180-2-9247 (see Table I). The configuration of the specimens was such that the minimum diameter was 0.2 inch and the radius of curvature at the test section was one inch. The specimens had been stress relieved at 1750 F for 2 hours and then air cooled after which they were duplex aged at 1325 F for 8 hours and 1150 F for 8 hours.

2. EXPERIMENTAL PROCEDURES

Four specimens were again stress relieved at 1725 F for one hour. Two specimens were air cooled from 1725 F and two were furnace cooled at a rate of 100 F per hour. A single specimen, which was used for comparison, was solution annealed at 2000 F for one hour and then air cooled. One specimen for each cooling rate from the 1725 F stress relieves were then tensile tested at 1400 F. The other specimen for each cooling rate was heated with an induction coil to 2250 F to simulate a HAZ produced during welding. The gage sections of these specimens were heated from 1400 F to 2250 F in 15 to 20 seconds. After a one second hold the gage sections were cooled to 1400 F in approximately 35 seconds. These specimens were then tensile tested at 1400 F. The control specimen was tested at 1400 F in the 2000 F solution-annealed condition. Diametrical strains were recorded from which the 0.2% yield strengths were determined.

3. RESULTS OF TENSILE TESTS AND METALLOGRAPHIC INVESTIGATIONS

The values calculated for the 0.2% yield strengths along with the reductions in area and ultimate tensile strengths are listed in Table XIV.

TABLE XIV

SIMULATED HEAT AFFECTED ZONE AND BASE METAL TENSILE PROPERTIES AT 1400 F FOR INCONEL 718, HEAT 2180-2-9247

Specimen	0.2% Offset Yield Strength, ksi	Tensile Strength, ksi	Reduction in Area, %
-36 (2000 F, A.C.(1))	57.3	100.3	23.4
-38 (1725 F, A.C.(1))	100.0	113.0	43.8
-37 (1725 F, A.C.(1), HAZ(3))	85.5	115.1	13.5
-46 (1725 F, F.C.(2))	85.0	122.6	39.9
-45 (1725 F, F.C.(2), HAZ(3))	94.6	116.2	24.3

(1) A.C. indicates an air cool from the solution anneal

(2) Furnace cool was at a rate of 100 $\pm$ 10 F per hour.

(3) 2250 F cycle was selected to represent weld HAZ.

Light microscopy was used to check the microstructure of the rapidly and slowly cooled base metal in addition to those of the HAZ formed on rapidly and slowly cooled material. These specimens are listed as 46 and 38, and 37 and 45 in Table XIV. Matrix precipition was apparent in the slowly cooled base metal but not in the rapidly cooled base material. The simulated HAZ cycle dissolved all matrix precipition which could be resolved by light microscopy. Another item noted was that no grain growth occurred in the samples given the simulated HAZ cycles. The grain boundaries were also almost completely delineated with a phase which was probably some form of carbide.

IX. DISCUSSIONS AND SUMMARIES OF HEAT TREATING STUDIES

1. MATERIAL DEGRADATION TESTS (HEAT 83C2EK)

1.1 1400 F Tensile Tests for Solution-Annealed Material

Tensile data for 1400 F tests of the various solution-annealed conditions from Table II are shown in Table V. These tests were primarily performed to aid in establishing stress levels for subsequent stress rupture tests. Cooling rates affected the 1400 F tensile properties in two ways. The yield and tensile strengths were less for specimens rapidly cooled from 1900 F solution anneals (C and G). Conditions C and G possessed yield strengths of 82.9 and 73.2 ksi, respectively, while the ultimate tensile strengths were 92.4 and 80.4 ksi. The minimum yield and tensile strengths for all other conditions (1900 F slow cool and 2000 F fast and slow cool) were 92.6 and 104 ksi, respectively. Higher strengths were noted for the samples rapidly cooled from 2000 F (samples A and E) than those rapidly cooled from 1900 F (samples C and G). This indicates that there is a discrepancy. An examination of the heat treat charts offers one explanation for these differences. Samples C and G were cooled from 1900 F to 1200 F at a rate of 1860 F per hour whereas the actual cooling rate for samples A and E over this temperature range was 900 F per hour. This indicates that precipitation was avoided during the rapid 1860 F per hour cool but partial aging occurred for the slower cooling rates of 75 and 90 F per hour and possibly during the 900 F per hour cool. The heating rates to test temperature and soaking times were not rigorously controlled. This may have also contributed to high tensile values for conditions A and E.

A second and more important influence of cooling rate was an improvement shown in the 1400 F ductility. Comparisons of samples A to B, C to D, E to F, and G to H show that both the percents elongation and reductions in area are in each case better for the samples cooled at slow rates of 75 or 90 F per hour than at rates of 900 or 1860 F per hour. This improvement was by as much as a factor of 3 in some cases.

1.2 Solution-Annealed Material Stress Rupture Tested at 1400 F

Data for 1400 F stress rupture tests of the eight solution-annealed conditions per Table II are listed in Table VI. This data is plotted and is shown in Figures 6 through 11. Figures 6 and 7 show the rupture lives for these various solution-annealed conditions loaded to initial stresses with constant weights. Solution anneal temperatures (2000 F or 1900 F), cooling rate (75 to 1860 F per hour) and prior thermal history did not show any distinguishable differences (less than a factor of two) in rupture life when stressed at 60 ksi. The majority of the specimens tested at this level failed between approximately fifteen and twenty-five hours. Rapid cooling appeared to reduce the rupture lives at higher stresses for a material solution annealed at 1900 F. This is particularly apparent for condition G in Figure 7 and condition C appears to show this same trend in Figure 6. This reduction in rupture life may be attributed to rapid cooling since conditions C and G were cooled at a rate of 1860 F per hour which was much more rapid than any other thermal cycle. The low rupture lives for conditions C and G corresponded with the low 1400 F yield strengths of 82.9 and 73.2 ksi shown for these conditions in Section VIII.1.3. Stress rupture specimens would undergo age hardening during extended tests and at 60 ksi the rupture life for conditions C and G were comparable to other conditions. Many of the G specimens failed before the samples reached the 1400 F test temperature. Two of these premature failures occurred in the notches machined in the gage section and a third occurred in the threaded section of the specimen. This indicated that prior thermal cycling may have caused material solution annealed at 1900 F and cooled at a rate of 1860 F per hour to be somewhat notch sensitive. In contrast, the only other failure to occur in a notch was specimen A-5 which was stressed at a very high level of 100 ksi.

The rupture life for all of the solution-annealed conditions may be summarized as follows; stresses of 79 ksi caused failures in less than two hours while at 70 ksi rupture occurred in less than ten hours. As previously mentioned, stresses of 60 ksi generally caused failures between approximately fifteen and twenty-five hours.

1400 F stress rupture ductilities were improved by slowly cooling from the solution anneals. This is shown in Figure 8 where percent elongations are plotted for stresses ranging from 60 to 90 ksi for conditions A, B, C, and D. Substantially better elongations are noted for slowly cooled material (conditions B and D) at higher stress levels. Specifically, conditions A and B represent materials solution annealed at 2000 F. Condition A was cooled at a rate of approximately 900 F per hour and condition B at 75 F per hour. The plots in Figure 8 show that condition A possessed elongation values of 5 percent or less at stresses of 70 ksi or greater. Condition B had elongation values near 20 percent for equivalent stresses. At 60 ksi the elongation values became more similar, i.e., 12 and 18 percent, respectively, for conditions A and B. The samples rapidly and slowly cooled from 1900 F, conditions C and D, showed the same trend. Condition D, which was cooled at a rate of 90 F per hour, had elongation values of 30 to 35 percent for stresses of 79 ksi and greater. Condition C, cooled at a rate of 1860 F per hour, showed elongation values of only 7 percent at 79 ksi. At stresses of 70 and 60 ksi conditions C and D displayed similar elongation values; i.e., 21 percent for condition C and 24 percent for condition D.

Better 1400 F stress rupture ductilities were obtained from 1900 F solution anneals as compared to 2000 F solution anneals when the cooling rates were comparable and no prior thermal cycling had been experienced. Compare the plots for conditions A and C (rapidly cooled) and conditions B and D (slowly cooled) in Figure 8. Condition C the material having the lower solution anneal (1900 F) showed better elongation values than condition A which was solution annealed at 2000 F when stressed at 60 and 70 ksi. This shows that the lower solution anneal temperature offers better ductility over a wider range of stress levels. Condition D, the material solution annealed at 1900 F and cooled at a rate of 90 F per hour, had better elongation values than condition B over the 60 to 90 ksi stress range . Condition B was the material solution annealed at 2000 F and cooled at a rate of 75 F per hour.

The ductility expressed in terms of percent reduction in area shows the same trends as the percent elongation (see Figure 9). These trends are; 1) improved reduction in area as a result of slowly cooling from a comparable solution anneal temperature (1900 or 2000 F), and 2) better ductilities for 1900 F solution anneals than

2000 F solution anneals when the cooling rate is comparable.

Prior thermal history involving a 2000 F solution anneal, 1400-1200 F duplex age, and a 1800 F stress relief did not affect 1400 F stress rupture ductilities for material solution annealed at 2000 F but did reduce the ductilities when the solution annealed was performed at 1900 F. This can be observed from elongations plotted in Figures 8 and 10 and reductions in area plotted in Figures 9 and 11. Prior thermal history also caused the material solution annealed at 1900 F and rapidly cooled to be somewhat notch sensitive at 1400 F.

The elongation plots in Figure 10 for conditions E and F which represent materials having received prior thermal history and a 2000 F solution anneal are very similar to the corresponding solution-annealed conditions of A and B in Figure 8. Figures 9 and 11 show that plots of reductions in area are also quite similar when one compares conditions A to E and conditions B to F. The samples solution-annealed at 1900 F, conditions C and D in Figures 8 and 9, may be compared with those having the same annealing temperature, conditions G and H, in Figures 10 and 11 to show the detrimental effects of prior cycling. The elongation values for condition C, having a 1900 F solution anneal and fast cools are 20 percent at 70 ksi. Condition G which was also solution annealed at 1900 F and rapidly cooled but having received extra thermal cycles had an average elongation of only 4 percent at this stress level. The elongation values are between 25 and 35 percent for 1900 F solution anneal with a slow cool (condition D) in Figure 8. In comparison the elongation values were reduced to values ranging from 17 to 29 percent for similarly solution annealed material having received extra thermal cycles (condition H) in Figure 10.

1.3 Age-Hardened (1400-1200 F) Room Temperature Tensile Properties

Room temperature tensile properties for aged samples have been reported in Table III. These samples had been heat treated to various conditions to establish the effects of prior thermal cycling, solution anneal temperature (2000 F or 1900 F), and cooling rates from the solution anneal. In checking the effects of prior thermal cycling, material solution annealed and aged directly from a mill-annealed condition was compared with material solution annealed and aged after a previous 2000 F solution anneal, age, and stress relief. The latter condition would be representative of acoustic filter Unit No. 1 after repair.

The most apparent effect of the variables mentioned above was a reduction in 0.2% yield and ultimate tensile strengths resulting from slow cools from solution anneals. A comparison of the rapidly cooled specimens (AA, CC, EE, and GG) with the slowly cooled specimens (BB, DD, FF, and HH) in Table III will show this trend. The specific cooling rates were as follows; conditions AA and EE - 900 F per hour, conditions CC and GG - 1860 F per hour, conditions BB and FF - 75 F per hour, and conditions DD and HH - 90 F per hour. All specimens were then aged 10 hours at 1400 F plus 8 hours at 1200 F. Yield strengths ranged from 148 to 158 ksi for the rapidly cooled specimens and from 125 to 132 ksi for the slowly cooled specimens. Accordingly, the ultimate tensile strengths ranged from 191 to 195 ksi and from 181 to 189 ksi, respectively, for the rapidly and slowly cooled specimen.

Slow cooling from the solution anneal slightly degraded the room temperature elongation and reduction in area. Samples solution annealed at 2000 F and rapidly cooled (AA and EE) showed an average elongation value of 19 percent compared to 17 percent for the specimens slowly cooled from the same temperature (BB and FF). The reductions in area were 25 and 21 percent, respectively. Annealing at 1900 F resulted in elongation values averaging 16 and 13 percent, respectively, for the rapidly and slowly cooled samples, (compare CC and GG specimens to DD HH specimens). These same specimens showed average reductions in area of 19 and 16 percent, respectively.

Solution annealing at 2000 F gave consistently better ductility in terms of percent elongation and reduction in area than 1900 F solution anneals. This effect of solution anneal temperature may be checked by comparing the AA to CC specimens (rapidly cooled) and the BB to DD specimens (slowly cooled) in Table III. Specimens rapidly cooled from 2000 F showed ductility measurements which were a factor of 1/3 better than specimens solution annealed at 1900 F. For example, the average elongation was 20 percent for AA specimens compared to 15 percent for CC specimens. The corresponding averages for the reductions in area were 26 percent and 19 percent. The average elongation value was 16 percent (BB specimens) for the 2000 F solution anneal compared to 13 percent (DD specimens) for 1900 F solution anneals when cooled at rates of 75 and 90 F per hour, respectively.

Yield strength was not noticeably and consistently effected by the solution anneal temperature. Yield stresses were slightly greater for rapidly cooled specimens solution annealed at 1900 F rather than 2000 F. The average yield stress was 156 ksi when solution annealed at 1900 F (CC specimens) compared to 151 ksi when solutions annealed at 2000 F (AA specimens). This may also be the effect of cooling rate since the samples solution annealed at 1900 F were cooled at a rate of 1860 F per hour while those solution annealed at 2000 F were cooled at a rate of 900 F per hour. This trend was reversed for specimens cooled at rates of 75 and 90 F per hour. The yield strengths were 125 and 128 ksi (see specimens DD and BB), respectively, for 1900 and 2000 F solution anneals. The ultimate tensile strength also was not greatly effected by the solution anneal temperature. When rapidly cooled, this property was nearly the same, i.e.; 193 ksi compared to 194 ksi for 2000 and 1900 F solution anneal. A 75 F per hour cool from 2000 F caused the ultimate tensile strength to be slightly higher than a 90 F per hour cool from 1900 F. This improvement was from 183 to 187 ksi.

Prior thermal history did not influence either the yield strength or ultimate tensile strength for specimens having received the same solution anneal temperature and approximately the same cooling rates. Extra cycles appear to have reduced the ductility of samples solution annealed at 2000 F and rapidly cooled. The elongation was reduced from 20 to 18 percent and the reduction in area from 26 to 23 percent as a result of the extra heat treat cycles (see samples AA and EE). This trend was reversed when slowly cooled from 2000 F. For example, elongation averages were increased from 16 to 17 percent and percent reduction in area averages from 20 to 22 percent as a result of the extra cycles. No changes in ductility were noted as a result of prior thermal history for solution anneals performed at 1900 F.

There are three noteworthy observations from the room temperature tensile properties of the aged specimens. Yield strengths and ultimate tensile strengths were reduced as a result of 75 or 90 F per hour cools from the solution anneals. The average yeild strength of approximately 130 ksi for these conditions was unsatisfactory in regard to 145 ksi requirement for the acoustic filters. Secondly, solution annealing to 2000 F rather than 1900 F showed an improvement in room temperature base metal ductility for comparable cooling rates. Finally, an extra thermal cycle of a 2000 F solution anneal, age and stress relief will not impair base metal room temperature tensile properties.

1.4 Room Temperature Impact Properties for Materials Aged at 1400-1200 F

An improvement in impact values was observed when material was solution annealed at 2000 F rather than 1900 F. This is shown by reviewing the charpy impact values reported in Table IV for base materials age-hardened 10 hours at 1400 F plus 8 hours at 1200 F as outlined in Table II. A comparison shows impact values of 26 ft-lb for specimens rapidly cooled from 2000 F solution anneals (AA) and values of 18 ft-lb for samples solution annealed at 1900 F and similarily cooled (CC). The samples of conditions BB and DD show a similar trend. These slowly cooled samples had impact values of 23 ft-lb and 18 ft-lb, respectively, for 2000 and 1900 F solution anneals. The higher solution anneal temperature, 2000 F, continued to provide better impact values even when material had received a prior thermal history of a 2000 F solution anneal, a 1400-1200 F age and a 1800 F stress relief. For example, impact values for samples EE were 26 ft-lb compared with 18 ft-lb obtained for GG. A comparison of samples FF and HH which were slowly cooled showed that the 2000 F solution anneal resulted in average impact values of only 18 ft-lb.

Additional prior thermal history involving a 2000 F solution anneal, a 1400-1200 age and a 1800 F stress relief, did not impair room temperature impact properties. In fact, it improved impact resistance in two cases. Compare the 23 ft-lb average for the BB specimens with the 26 ft-lb average for the FF specimens. Samples CC and GG show this same trend of 18 and 20 ft-lb. It should be noted, however, that the prior thermal history did not increase impact values for samples annealed at 1900 F to the levels observed for 2000 F solution anneals. This is shown by comparing the values of 20 ft-lb reported for the GG to those of 26 ft-lb for AA and 23 ft-lb for BB. This indicated that grain size is not the controlling factor.

The microstructural phases either dissolved or precipitated between 1900 F and 2000 F may have an influence on the differences in impact values.

Cooling rates from the solution anneal did not affect the impact properties by more than approximately 10 percent. For instance, samples AA and EE both solution annealed at 2000 F and rapidly cooled showed average impact values of 26 ft-lb. Samples designated as BB and FF were also solution annealed at 2000 F but cooled at a rate of 75 F per hour. Values for BB were slightly lower, 23 ft-lb, but those for FF were equal to the 26 ft-lb shown for the AA and EE conditions.

The above discussion showed that solution annealing at 2000 F will provide better room temperature impact properties for base metal than 1900 F solution anneals. Also, additional thermal cycling involving a 2000 F solution anneal, age and stress relief did not impair the room temperature impact properties.

1.5 Microstructural and Fractographic Findings

The fracture surfaces of failed 1400 F stress rupture specimens were examined with a table microscope and a scanning electron microscope. Specimens tested at 70 ksi or greater for materials rapidly cooled from solution anneals had distinct intergranular crescent-shaped fracture origins. Such an intergranular origin for stress rupture specimen A-6 is shown in Figure 12. This specimen had been solution annealed at 2000 F and then cooled at a rate of 900 F per hour to 1000 F prior to being tested at 1400 F at an initial stress of 90 ksi. Failure occurred in 0.2 hour with an elongation of only 4 percent. Such large distinct intergranular crescent-shaped fracture origins were not observed on this material when tested at stresses below 70 ksi. The fractures were more irregular than those exhibited by specimens stressed above 70 ksi and showed a mixed intergranular and transgranular mode with approximately 20-30 percent of the surface displaying grain boundaries. This irregular mixed fracture mode also occurred for slowly cooled material tested at all stress levels.

Scanning electron fractography verified that the crescent-shaped origin for specimen A-6 was predominantly intergranular (see Figure 13). The intergranular facets were very smooth and did not show any signs of dimpling. In contrast, dimpling resulting from microvoid coalescence along grain boundaries was observed at fracture origins for slowly cooled material. Examples of dimples are identified by arrows in Figure 14 for specimen B-5. This specimen had been solution annealed at 2000 F and then cooled at a rate of 75 F per hour. The elongation was 19 percent when stressed at 88.5 ksi. Dimpling was observed on grain boundaries for all solution-annealed conditions at levels of less than 70 ksi. Figure 15 illustrates the dimpling which occurred on specimen A-11. This specimen, which had received a 900 F per hour cool, showed an elongation of 12 percent when tested at 60 ksi. In summary, dimpling on grain boundaries appeared to be associated with specimens having better 1400 F stress rupture ductility. Specimens having such a fracture characteristic, which exhibited dimpling, were ones which had been slowly cooled from solution anneals or rapidly cooled and tested at stress levels of less than 70 ksi.

Metallographic sections of failed 1400 F stress rupture specimens were examined with a light metallograph. The only noteworthy item was hetrogenous precipitation which became very apparent in material which had been slowly cooled (conditions B and D). These material conditions differentially etched to reveal a banding effect. This banding is shown at a magnification of 500X for condition B in Figure 16. The same sample was then examined with the scanning electron microscope. Observations at approximately 4200X showed the shapes of some precipitates present in the bands (see Figure 17). The bands which appeared light colored in light micrographs contained small round precipitates in addition to larger elongated precipitates. No precipitates were resolved by this technique for the bands which etched darker.

Material conditions A and B were examined by transmission electron microscopy on replicas taken from polished and etched samples. Conditions A and B represent heat 83C2EK cooled from 2000 F at rates 900 and 75 F per hour, respectively. Replicas photographed at 15,000X and shown in Figure 18 illustrate the precipitate forms detected for conditions A and B. The electron micrograph in Figure 18a shows that the matrix precipitation for the rapidly cooled material is very fine whereas there are three forms of precipitates shown for condition B in Figure 18b. The precipitate forms include small round or blocky-shaped precipitates, and intermediate and large elongated or lenticular-shaped precipitates. These shapes agree with those observed with the SEM.

Inconel 718 differs from most other nickel-base superalloys in that it contains a 5 weight percent alloy addition of niobium. This causes a more complex precipitation process. Most nickel-base precipitation hardenable alloys form a face-centered-cubic precipitate of Ni_3 (Al, Ti) which is coherent with the matrix. Inconel 718, however, can undergo a process which involves three different phases. There is a face-centered cubic (fcc) coherent precipitate which is generally associated with temperatures below 1200 F. This phase rich in aluminum and titanium has been reported to have a round or cubic shape[6-9]. Between 1200 and 1600 F a coherent body-centered-tetragonal (bct) phase is observed which is rich in niobium. This phase often shown as Ni_xNb and referred to as gamma double prime (γ'') has been reported as having disc-shaped or platelet forms (Ref. 6, 7, 9-11). Paulonis, et al[6], and Cozar and Pineau[7] have reported these precipitates to coarsen to diameters of over 3000A°. The times reported to cause γ'' growth to 3000A° at 1400 F was 100 and 500 hours, respectively. The bct γ'' precipitate is metastable. It will transform to Ni_3Nb which has an orthorhombic structure. This orthorhombic phase has been reported as appearing as large platelets (Ref. 9, 12 and 13). In fact, Wagner and Hall[12] have referred to it as having a Widmanstätten pattern. The ratio of Ti + Al/Nb has been shown to influence the extent of γ' or γ'' precipitation[7].

The electron micrograph in Figure 18b shows small round precipitates and intermediate-sized lenticular precipitates. According to the above discussion, it would appear that these would be the (fcc) γ' and (bct) γ'', respectively. The lengths of the lenticular precipitates in Figure 18b have been calculated to be from 2000 to 5000A°. This agrees with the size of the coarsened γ'' reported by Paulonis, et al[6], and Cozar and Pineau[7]. The longer platelets in Figure 18b may be orthorhombic Ni_3Nb since it appears to be formed by consuming smaller adjacent precipitates. These electron microscopy studies have shown that cooling heat 83C2EK at a rate of 75 F per hour caused coarse higher temperature precipitates to form. In contrast, very fine precipitates formed when rapidly cooled. This detection of coarsened precipitates with slow cooling rates supports the observations of reduced strain-age cracking for overaged conditions of Rene' 41 according to references 2 and 3 in Section VII. It also agrees with the proposal of Prager and Sines[5] in that coarse precipitates will cause homogeneous intergranular slip and result in a reduced cracking tendency.

2. MODIFIED AGING STUDIES

In an investigation to improve room temperature yield strength, hardness measurements were recorded for heat 83C2EK cooled at a rate of 75 F per hour and then aged for various combinations of times at 1400-1200 F and 1325-1150 F (see Table VII). The changes in hardness when aged at initial temperatures of 1400 and 1325 F are plotted on the left in Figure 19. The hardness of samples having exposures of 2 (A), 4 (B) and 10 (C) hours at 1400 and 1325 F are then plotted with respect to the times aged at the second stages of either 1200 or 1150 F. Hardnesses obtained from these second stage exposures are indicated on the right in Figure 19 for times extending to 30 hours. A similar set of hardness plots for heat 87C0EK1 cooled at a rate of 400 F per hour (see Table VII) is shown in Figure 20.

The initial hardness for heat 83C2EK cooled at a rate of 75 F per hour was Rc 37. The material hardened to between Rc 38 and 39 when initially aged at 1325 F and remained fairly constant for times up to 24 hours. The 1400 F aging step caused the hardness to decrease after 8 hours possibly indicating overaging. Additional hardening occurred during the second aging steps at 1200 and 1150 F. The lower temperature ages at 1150 resulted in slightly greater hardnesses after 16 hours. Table VII shows hardness values ranging from 38.1 to 39.8 Rc for the higher temperature ages and from 39.7 to 40.2 Rc for the lower temperature ages for exposures limited to 16 hours at the second aging step. The recorded hardness values were nearly the same for all conditions when the second aging steps were extended to 30 hours. This was true even for those samples which appeared to be overaged by 10 hour exposures at 1400 F.

The initial hardness of heat 87COEK1 cooled at a rate of 400 F per hour was 34.6 Rc. The hardness values for various exposures to initial ages at 1400 F and 1325 F are shown in Figure 20. Aging at 1325 F caused greater increases in hardness than 1400 F for times extending to 24 hours. Rc hardnesses were increased to values ranging from 39 to 40.4 for the 1325 F age but only 37.2 for 1400 F exposures. There is again evidence of decreasing hardness after 10 hours at 1400 F. The lower temperature second stage age at 1150 F again showed better hardness values after 16 hours than the exposures at 1200 F. Except for the cycle aged for 2 hours at 1325 F the Rc hardness was consistently greater than 42 whereas those aged at 1400--1200 F were generally below Rc 42 after the 16 hour second stage ages. Holding the samples for 30 hours at the second aging steps caused most of the conditions to obtain comparable hardnesses.

Heat 87COEK1 cooled at 400 F per hour had a lower initial hardness than heat 83C2EK cooled at a rate of 100 F per hour; i. e., 34.6 Rc compared to 37. This may be the result of either or both the different heats or cooling rates. Samples cooled at the more rapid rate were often harder after aging. This corresponds with higher room temperature yield strength values for rapidly cooled material in the Material Degradation Tests, (Section IV).

The combination of 4 hours at 1325 F plus 16 hours at 1150 F provided good hardness values for samples of heat 83C2EK cooled at a rate of 75 F per hour and for heat 87COEK1 cooled at a rate of 400 F per hour. Room temperature tensile data are listed in Table IX. The 0.2% yield strengths for the samples cooled at a rate of 75 F per hour were 144 and 148 ksi. The yield strengths for samples cooled at 400 F per hour could not be determined because the extensometer slipped. This raised some question about the resulting yield strengths, however, the required design stress of 145 ksi for the PBF acoustic filter appeared to be satisfied since the hardness was better than the samples cooled at a rate of 75 F per hour.

The studies have shown that a 10 hour 1400 F plus 8 hours 1200 F duplex age on heat 83C2EK cooled at a rate of 75 F per hour resulted in a hardness of 37.8 Rc. This agrees with the low room temperature yield strengths of 125 to 132 ksi reported for conditions BB, FF, DD, and HH in Table III which represent the same heat and similar cooling rates. This heat having the same solution anneal and a similar cooling rate showed a hardness of 40.2 Rc after a 4 hour 1325 F plus 16 hours 1150 F age. The yield strength was accordingly increased to around 145 ksi.

The results of the hardness studies may be summarized as follows: 1) Higher hardnesses may be obtained for slowly cooled material by using a modified duplex age of 4 hours at 1325 F plus 16 hours at 1150 F rather than the 10 hour - 1400 F plus 8 hour - 1200 F age previously used on the PBF acoustic filters. This increase in hardness accompanied an improvement of room temperature yield strength to values around 145 ksi. 2) Material solution annealed at 2000 F and cooled at a rate of 400 F per hour rather than 100 F per hour had lower hardness values when solution annealed but aged to higher hardnesses.

3. MATERIAL AND HEAT TREATING TESTS PERFORMED ON ACOUSTIC FILTER MATERIAL (HEAT 87COEK1)

3.1 1400 F Stress Rupture Properties of Solution Annealed Acoustic Filter Material

1400 F stress rupture data for heat 87COEK1 cooled from 2000 F solution annealed at rates of 100, 400, and 1200 F per hour are listed in Table X. The rupture ductility in terms of the percent elongation is plotted with respect to the initial applied stresses in Figure 21. Rapidly cooling the material at a rate of 1200 F per hour from a 2000 F solution anneal resulted in elongation values of 5 to 9 percent when stressed at 79 and 70 ksi. When the cooling rate was reduced to 400 F per hour, elongations were still from 5 to 8 percent at 79 ksi but increased to values of over 20 percent at 70 ksi. Lowering the cooling rate to 100 F per hour caused an additional increase in ductility. The average elongation was 22 percent for specimens having received a 100 F per hour cool and tested at a 79 ksi stress level. At the lowest stress level of 60 ksi elongation values were nearly the same for all conditions. This improvement in stress rupture ductility with progressively slower cooling rates for heat

87COEK1 corresponds with that shown for heat 83C2EK in Figure 8.

Rupture life at higher stress levels progressively improved as the cooling rate is decreased for heat 87COEK1. This observation of reduced rupture life with rapid cooling rates agrees with the observations of conditions C and G from heat 83C2EK in Figures 6 and 7, respectively. It can be observed in Table X that rupture life was nearly the same for all solution-annealed conditions for heat 87COEK1 when stressed at 60 ksi. This is also in agreement with the behavior of heat 83C2EK in Figures 6 and 7.

It has been calculated that thermal stresses of approximately 85 ksi were quenched into the filter Unit No. 1 when cooled from the 2000 F solution anneal with a 1200 F per hour cool. Specimens FA in Table X represent the solution-annealed condition of PBF acoustic filter Unit No. 1 prior to the aging cycle during which failure occurred(1). These data show that specimens initially stressed at 79, 70 and 60 ksi at 1400 F would fail in 0, 3.3, and 29 hours, respectively. Stresses on a constant load stress rupture specimen would increase as specimen diameters decrease whereas residual thermal stresses in the filter would decrease due to relaxation. Nevertheless, it seems that the residual thermal stresses were instrumental in causing failure during the 1400-1200 F duplex aging cycle.

3.2 Age-Hardened Room Temperature Tensile Properties

The room temperature tensile properties for filter shell material, heat 87COEK1, aged to various conditions are given in Table XI. The specimens for material solution annealed at 2000 F, cooled 100 F per hour, aged 4 hours at 1325 F and 16 hours at 1150 are identified as FS. They possessed an average 0.2% yield strength of 145 ksi, the required design stress for the PBF acoustic filter. Specimens cooled from a 2000 F solution anneal at a rate of 400 F per hour and similarly aged are identified as FI specimens. These specimens showed an average yield strength of 155 ksi. This was comparable to the average of 156 ksi listed for the FA specimens which represent the age-hardened condition of the failed PBF acoustic filter Unit No. 1.

Room temperature tensile ductilities are also listed in Table XI. The percent elongations were 20, 24 and 18 percent, for cooling rates of 100, 400 and 1200 F per hour, respectively. This manner in which 100 and 1200 F per hour cooling rates affected ductility is a reversal from the observations discussed in Section IX.1.3. This indicated the modified 1325-1150 F age provides better room temperature ductility for slowly cooled material than the 1400-1200 F aging cycle.

3.3 Impact Properties for Age-Hardened Acoustic Filter Material

Room temperature impact properties are listed in Table XII for the FS, FI, and FA conditions. Material cooled at 400 F per hour (FI condition) and aged had the highest average impact resistance of 60 ft-lb. The next highest impact values (51 ft-lb) were shown by the (FS) specimens cooled at a rate of 100 F per hour. The FA specimens cooled at 1200 F per hour and 1400-1200 F duplex aged showed an average impact value of 41 ft-lb. The impact values of Table XII (heat 87COEK1) may be compared to values given in Table IV for heat 83C2EK. The FA specimens of Table XII were solution annealed and aged the same as the AA specimens listed in Table IV. The average respective impact values were 41 and 26 ft-lb. This shows that heat 87COEK1 has significantly better impact resistance. Heat treat conditions having slower cooling rates, 400 F and 100 F per hour, and the modified 1325-1150 F duplex age showed better impact values than the FA specimens (see Table XII). In contrast, slowly cooled and 1400-1200 F duplex aged BB specimens in Table IV showed worse impact resistance than the AA specimens. This shows that either the effect of cooling rate upon the impact properties is different for the two heats or that the modified 1325-1150 F duplex age has significantly improved the impact properties.

3.4 Review of Stress Rupture, Tensile and Impact Data for Acoustic Filter Material

1400 F stress rupture tests for various solution-annealed conditions for heat 87COEK1 have shown:

1. The stress rupture ductility in terms of elongation and reduction in area is progressively improved as the cooling rate from 2000 F solution anneals is reduced from 1200 F to 400 F to 100 F per hour.

2. Progressively slower cooling rates from 2000 F solution anneals (1200, 400 and 100 F per hour) caused an increase in 1400 F rupture life for stress levels of 70 ksi or greater.
3. This material cooled from a 2000 F solution anneal at a rate experienced by the acoustic filter will fail in less than 5 hours when stressed above 70 ksi at 1400 F.

The room temperature tensile tests have shown:

1. Shell material for the acoustic filters (heat 87C0EK1) may be age-hardened with a 1325 F - 4 hour plus 1150 F - 16 hours cycle after being cooled at a rate of 100 F per hour from a 2000 F solution anneal to provide a room temperature 0.2% yield strength of 145 ksi.
2. No discrepancy was noted in the tensile properties of material removed adjacent to the failure site and solution annealed and aged like the acoustic filter (condition FA).
3. The tensile ductility was better for material solution annealed at 2000 F, cooled at rates of 400 F (FI) and 100 F (FS) per hour; and then aged 4 hours at 1325 F and 16 hours at 1150 F than material solution annealed at the same temperature, cooled at a rate of 1200 F per hour, and aged 10 hours at 1400 F and 8 hours at 1200 F (condition FA).

The room temperature impact tests have shown:

1. The impact resistance for samples aged after being cooled at rates of 100, 400 and 1200 F were 51, 60, and 41 ft-lb, respectively. This has shown that the impact resistance for specimens cooled at rates of 400 and 100 F per hour could be improved by the modified 1325-1150 F age to values greater than the previously used FA condition.
2. The impact resistance of heat 87C0EK1 heat treated to condition FA to represent the condition of the acoustic filter was 41 ft-lb. This is superior to the average of 26 ft-lb for heat 83C2EK shown in Table IV. It is also superior to two heats (2180-2-9251 and 2180-2-9247) and equal to a third heat (HT52C9EK) tested for the Alloy 718 code case submit-

tal (AEC Inconel 718 Materials Program, Contract 41-1375-I-94). This shows there was no weakness in the room temperature impact properties of aged material near the failure site.

4. HIGH TEMPERATURE TENSILE STUDIES PERFORMED ON BASE METAL AND SIMULATED HEAT AFFECTED ZONE SAMPLES

Section VII describes a program investigating high temperature tensile ductility for solution-annealed base metal and simulated heat affected zone specimens. This program specifically checked the effects of cooling rates of 1200 F and 100 F per hour from 2000 F solution anneals upon tensile ductility between 1000 and 1600 F. These tensile data for specimens taken from heat 83C2EK (acoustic filter gusset material) are listed in Table XIII. Elongations and reductions in area for the solution-annealed conditions are plotted as a function of test temperature in Figures 22 and 23. Both parent metal and simulated HAZ conditions having received the rapid cool of 1200 F per hour display minima in ductility of around 7 to 8 percent at 1400 or 1450 F (see Figure 22). In contrast, when specimens were cooled from 2000 F solution anneals at 100 F per hour fairly constant elongation percents were exhibited from 1000 F to 1500 F. This resulted in the slowly cooled specimens providing better elongation values between approximately 1300 and 1550 F than the specimens which were rapidly cooled. At temperatures below 1300 F, rapid cooling provided better elongation values for parent metal and simulated HAZ specimens, respectively. Reduction in area plots in Figure 23 also illustrate ductility minima for the rapidly cooled specimens and fairly constant values for the slowly cooled specimens. Plots in Figures 22 and 23 show that ductilities were generally worse for HAZ compared to base material having the same solution anneal and subsequent cooling rate.

Yield strengths for the various solution-annealed conditions are presented in Figure 24. These plots of yield strengths may explain the ductility behavior for the rapidly and slowly cooled conditions. Yield strengths of around 100 ksi which extended to 1300 F show that the slowly cooled conditions are partially aged. This has also been illustrated by the electron microscopy investigations discussed in

Section IX.1.5. The ductilities are therefore fairly constant over this temperature range indicating that coarse precipitates cause uniform deformation. At 1400 F and higher the yield strengths decreased and the ductilities accordingly showed some improvement. Rapidly cooled specimens possessed yield strengths of around 85 ksi at temperatures of 1200 F or less. Maxima then occur in the yield strengths as the test temperature is increased to 1300 F. This shows that the material underwent some age hardening as the specimens were heated to test temperatures. Rapidly cooled conditions showed good ductilities at 1300 F or less where the grains are not hardened. Grain matrixes then strengthened with increasing age hardening causing the material to fail by an intergranular mode. This type of fracture mode has been identified with the ductility minima near 1400 F (see Section IX.1.5). At temperatures of 1550 F or greater, yield strengths are markedly lower and the ductilities vastly improved.

Two age-hardened specimens from each of the four solution-annealed conditions in Table XIII were tested at 1200 F. Ductilities at this temperature were approximately equal, displaying elongations of 15 percent, for base metal regardless of the prior cooling rate from the solution anneal. Aged simulated HAZ specimens showed elongations of only 5 to 8 percent. Cooling rates from solution anneals again showed little influence upon the 1200 F ductilities.

The important observation from these high temperature tensile tests has been an improvement shown in ductility of both base metal and simulated HAZ between 1300 and 1550 F as a result of slow cools from solution anneals. Rapidly cooled specimens showed minima in elongation of around 7 to 8 percent between 1300 and 1550 F. Slowly cooled base metal and simulated HAZ showed elongation values between 12 and 18 percent for this temperature range. Slow cooling rates appear to increase the capability of Inconel 718 components to accommodate stresses by relaxation and to reduce the magnitude of residual thermal stresses formed. This would be beneficial when a welded component is cooled from a solution anneal.

5. PREWELD IMMUNIZATION CHECK

A study which checked the possibility of a preweld condition to reduce the susceptibility of PBF acoustic filter Unit No. 1 from cracking during subsequent repair and heat treating was described in Section VIII. The effects of an air cool and a 100 F per hour furnace cool after a 1725 stress relief upon 1400 F tensile ductility were checked to determine if the slow cool would be an effective preweld conditioner.

The information presented in Table XIV indicates the simulated HAZ on the specimen slowly cooled from 1725 F showed better ductility than the HAZ on the rapidly cooled specimen. Specifically, the air cooled specimen, No. 37, had a reduction in area of 13.5 compared to 24.3 percent for specimen 45 which was cooled at a rate of 100 F per hour. This indicated that HAZ would be more able to accommodate relaxation strains to relieve welding stresses when heated to be solution annealed. A 1750 F stress relief with a 100 F per hour cool was therefore recommended as a preweld heat treatment for the shell and new material used in repairing PBF acoustic filter Unit No. 1.

Metallographic investigations showed that grain growth had not taken place in the simulated HAZ on either the rapidly or slowly cooled specimen. This is contrary to the grain growth normally observed near actual weldments. The grains were also surrounded by a second phase which were probably carbides.

It is possible that the second phase pinned the grain boundaries and prevented grain growth during the rapid exposures to 2250 F. The thermal history responsible for the formation of the second phase at the grain boundaries is not known. It may be attributed to the 2 hour 1725 F stress relief plus 1325-1150 F duplex age or a combination of the mill processes and the specific thermal cycle. This is possible since various carbides (MC and M_6C), orthorombic Ni_3Nb precipitates, and Laves phase (M_2Ti) are not completely dissolved at 1725 F.

IX. REFERENCES

1. G. R. Smolik, "Analyses of Failure which Occurred During Heat Treatments of the Inconel 718 PBF Acoustic Filter" ANCR-117P (October 1974).

2. T. F. Berry and W. P. Hughes, "A Study of the Strain-Age Cracking Characteristics in Welded Rene' 41 - Phase II", Welding Journal, 48 (November 1969) pp. 5053-5133.

3. E. G. Thompson, S. Nunez, and M. Prager, "Practical Solutions to Strain-Age Cracking of Rene' 41", Welding Journal, 47, (July 1968) pp. 299503135.

4. R. W. Fawley and M. Prager, "Evaluating the Resistance of Rene' 41 to Strain-Age Cracking", Welding Research Council, Bulletin No. 150 (May 1970) pp 1-12.

5. M. Prager and G. Sines, "A Mechanism for Cracking During Postwelding Heat Treatment of Nickel-Base Alloys", Welding Research Council, Bulleting No. 150 (May 1970) pp. 24-32.

6. D. F. Paulonis, S. M. Oblak, and D. S. Duvall, "Precipitation in Nickel-Base Alloy 718", Transactions Quarterly; American Society for Metals, 62 (September 1969) pp. 611-622.

7. R. Cozar and A. Pineau, "Morphology of Y' and Y" Precipitates and Thermal Stability of Inconel 718 Type Alloys", Metallurgical Transactions, 4 (January 1973) pp. 47-59.

8. W. J. Boesch and H. B. Canada, "Precipitation Reactions and Stability of Ni_3Cb in Inconel Alloy 718", Journal of Metals, (October 1969) pp. 34-38.

9. J. F. Barker, E. W. Ross, and J. F. Rodavich, "Long Time Stability of Inconel 718", Journal of Metals, (January 1970) pp. 31-41.

10. E. L. Raymond, "Effect of Grain Boundary Denudation of Gamma Prime on Notch-Rupture Ductility of Inconel Nickel-Chromium Alloys X-750 and 718", Transactions Met. Soc. of AIME, 239 (September 1967) pp. 1415-1422.

11. P. S. Kotval, "Identification of the Strengthening Phase in Inconel Alloy 718", Trans. Met. Soc. AIME, 242 (August 1968) pp. 1764.

12. H. J. Wagner and A. M. Hall, Physical Metallurgy of Alloy 718, DMIC Report 217 (June 1965).

13. H. L. Eiselstein, Metallurgy of a Columbium-Hardened Nickel-Chromium-Iron Alloy, ASTM Spec. Tech. Pub. No. 369, (1965).

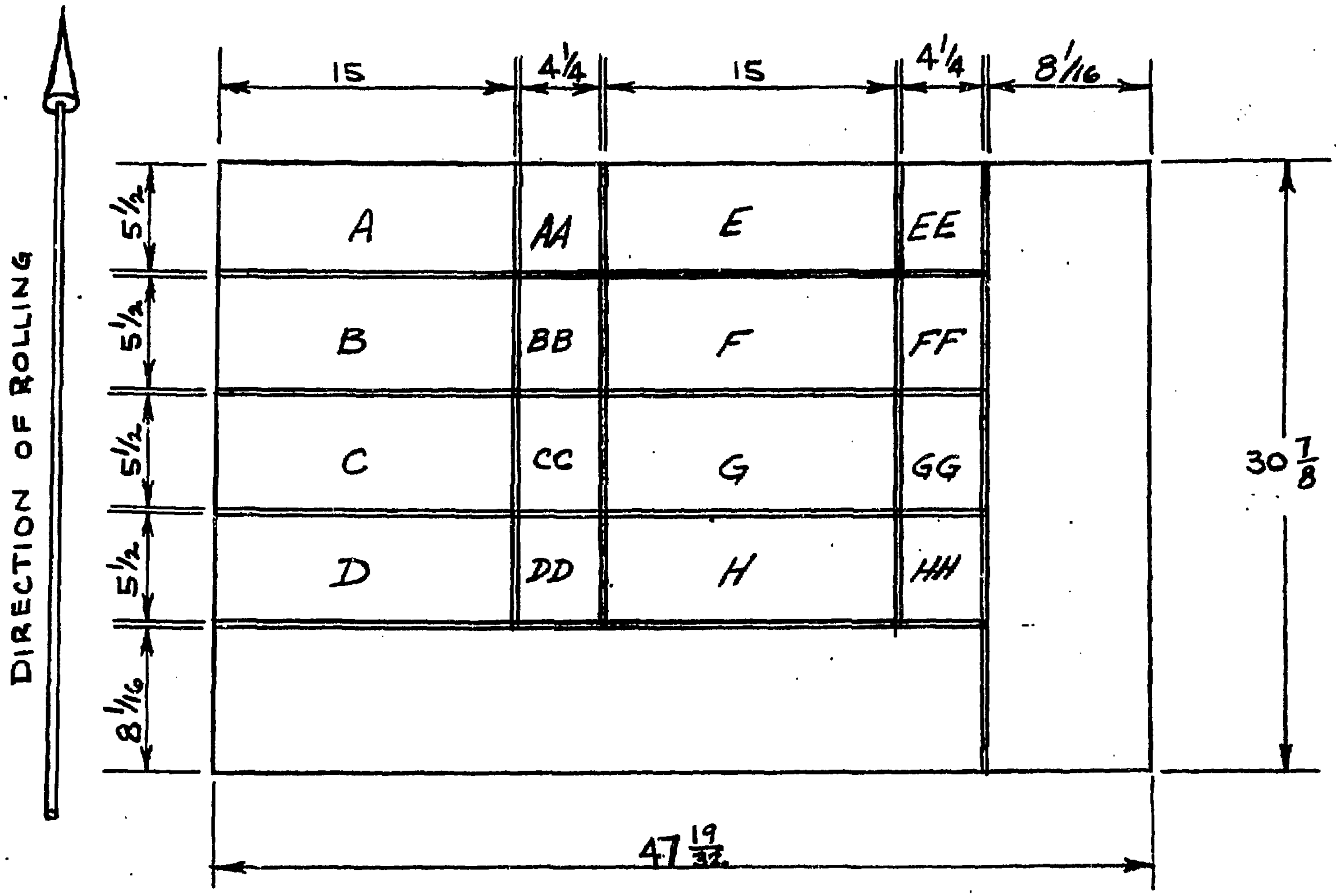


Fig. 1 Sectioning layout for one-inch thick Inconel 718 plate : Ht 83C2EK.

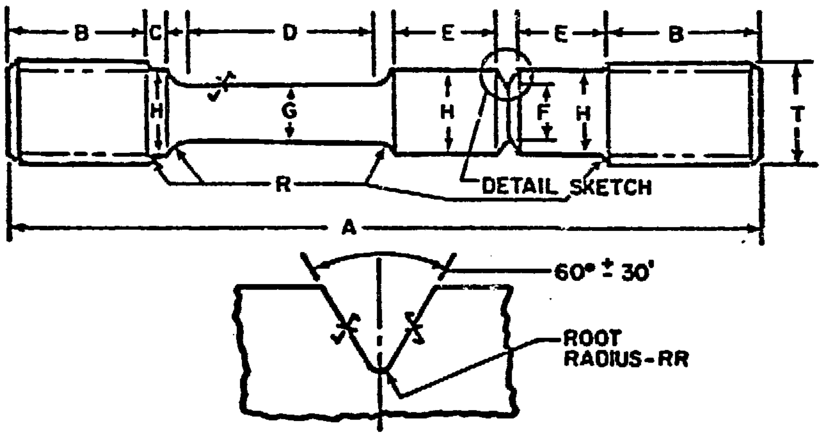


Fig. 2 COMBINATION SMOOTH-AND-NOTCHED TEST STRESS-RUPTURE SPECIMEN

AMS 5663 (Specimen type 5)

Specimen Number	Center Gage Diameter G, Inch	C inch	D, min inch	E, min inch	F inch	H inch	RR inch
5	0.252	0.125	1.000	0.375	0.252	0.357	0.009
Tolerance	±0.001	±0.062	----	----	±0.001	±0.003	±0.[illegible]5

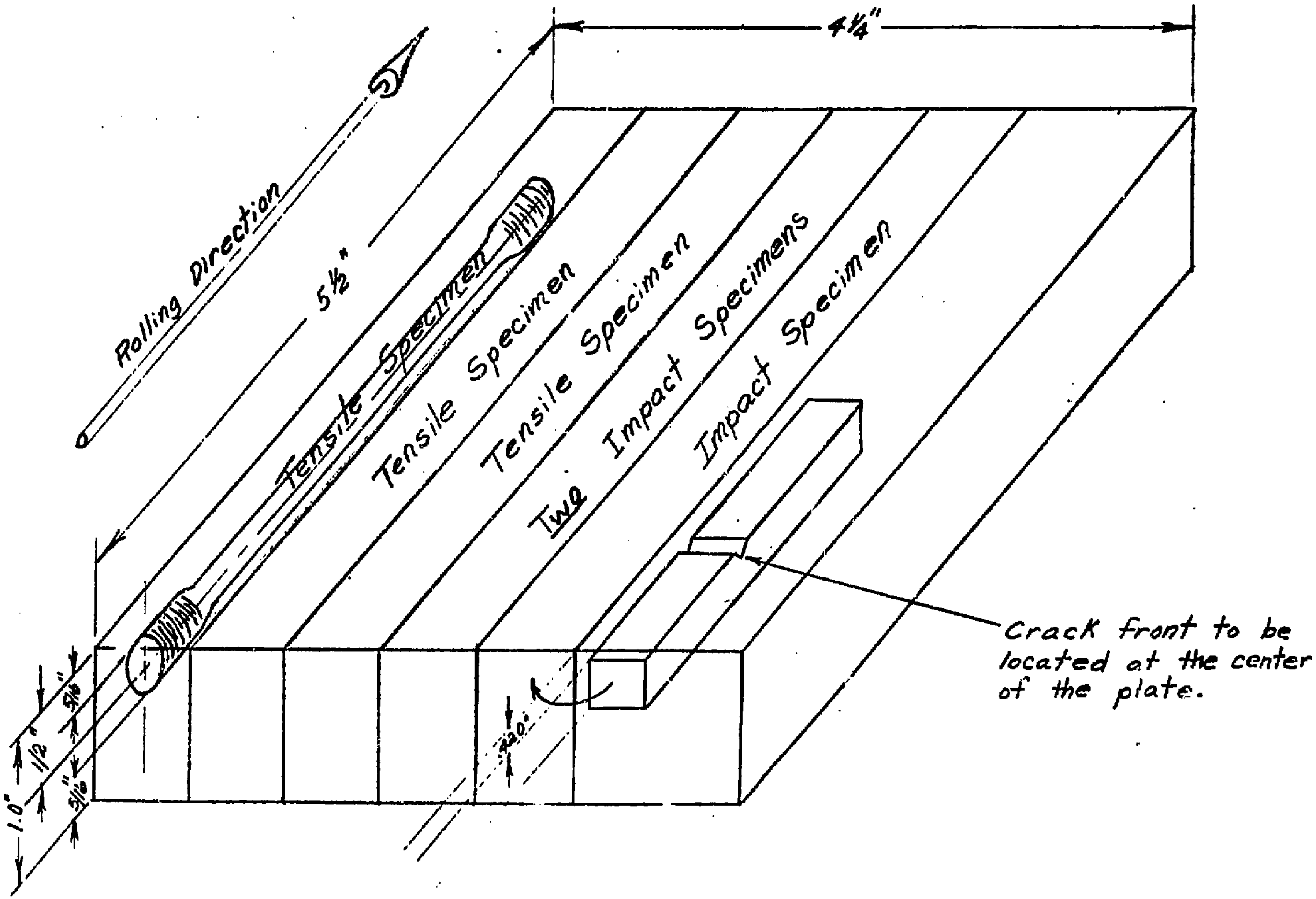
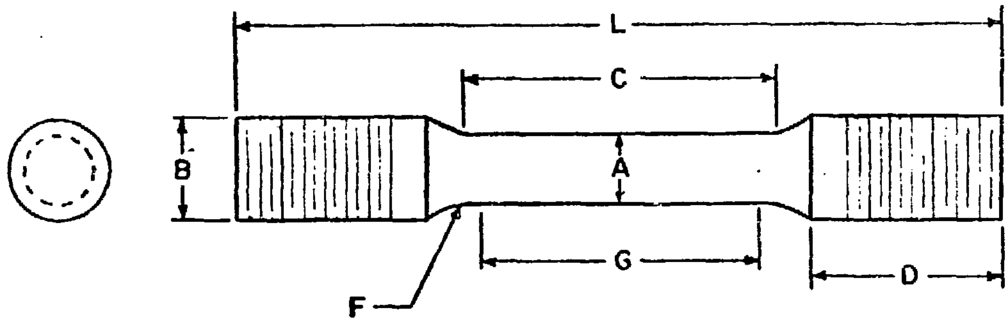


Fig. 3 Orientation and location of tensile and impact specimens.



All dimensions in inches

DIMENSIONS	R1 (FORMERLY TYPE 1)	R2 (FORMERLY TYPE 4)	(R3) (FORMERLY TYPE 4)	R4	R5 (FORMERLY TYPE 4)
A–DIAMETER AT CENTER[1]	0.505±0.010	0.357±0.007	0.252±0.005	0.160±0.003	0.113±0.002
B–DIAMETER OF GRIP END APPROX.[3]	3/4	1/2	3/8	5/16	1/4
C–LENGTH OF REDUCED SECTION, MIN.[2]	2 1/4	1 3/4	1 1/4	3/4	5/8
D–GRIP LENGTH, APPROX.[3]	1	5/8	5/8	1/2	3/8
F–FILLET RADIUS, MIN.[4]	3/8	0.25	0.18	0.15	0.09
G–GAGE LENGTH	2.0±0.005	1.4±0.005	1.0±0.005	0.64±0.005	0.45±0.005
L–TOTAL LENGTH, APPROX.[2 3]	5 1/2	3 1/2	3	2	1 5/8
AREA OF CROSS SECTION, SQ. IN., APPROX.	1/5	1/10	1/20	1/50	1/100

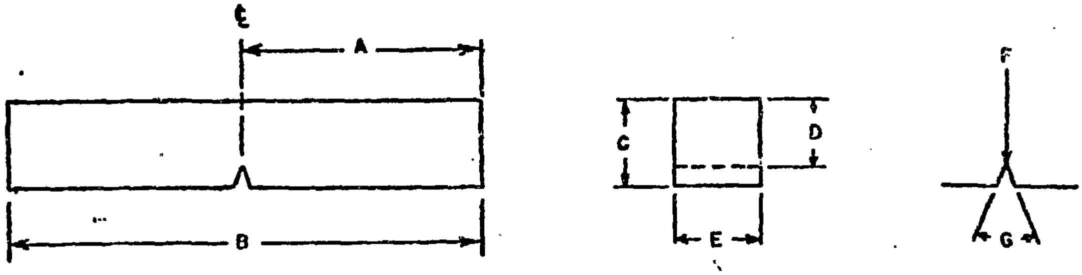
1. DIAMETER OF THE REDUCED SECTION MAY BE SMALLER AT CENTER THAN AT ENDS. DIFFERENCE SHALL NOT EXCEED 1 PERCENT OF DIAMETER AT CENTER.

2. WHEN AN EXTENSOMETER IS REQUIRED TO DETERMINE ELASTIC PROPERTIES, DIMENSIONS C AND L MAY BE MODIFIED TO PROVIDE THEREFOR. IN ALL CASES THE PERCENTAGE OF ELONGATION SHALL BE BASED ON DIMENSION G.

3. ENDS MAY BE OF ANY SHAPE TO FIT THE HOLDERS OF THE TESTING MACHINE IN SUCH A MANNER THAT THE LOAD IS AXIAL.

4. A GENEROUS FILLET IS DESIRABLE.

Fig. 4 FTMS 151, Method 211.1 (R_3) tensile specimen.



DIMENSION	MM.	TOLERANCE (MM.) +	−	INCHES	TOLERANCE (INCHES) +	−
A — CENTER TO END	1/2 B		—	1/2 B[1]		—
B — LENGTH	55.000	0	−2.50	2.165	0	.100
C — CROSS SECTION (DEPTH)	10.000	.025	.025	.394	.001	.001
D — NOTCH TO BASE	8.000	.025	.025	.315	.001	.001
E — CROSS SECTION (WIDTH)	10.000	.025	.025	.394	.001	.001
F — RADIUS OF NOTCH	.250	.025	.025	.010	.001	.001
G — ANGLE OF NOTCH	45° PLUS OR MINUS 1°					
OUT OF SQUARE (OF CROSS SECTION)	0.001 INCH IN 0.394 INCH OR 0.15° (9 MINUTES)					

NOTE—Permissible variations shall be as follows:

Adjacent sides shall be at	90 deg ± 10 min
Cross-section dimensions	±0.001 in. (±0.025 mm)
Length of specimen (*L*)	+0, −0.100 in. (2.5 mm)
Centering of notch (*L*/2)	±0.039 in. (1 mm). When an end-centering device is necessary to center the specimen in the anvil, see 8.3, it is necessary that the notch be accurately centered to ensure compliance with 4.2.1.
Angle of notch	±1 deg
Radius of notch	±0.001 in. (0.025 mm)
Dimensions to bottom of notch:	
Type A specimen	0.315 ± 0.001 in. (8 ± 0.025 mm)
Types B and C specimen	0.197 ± 0.002 in. (5 ± 0.05 mm)
Finish requirements	63 μin. (2 μm) on notched surface and opposite face; 125 μin. (4 μm) on other two surfaces

Fig. 5 ASTM E-23 (Type A) Charpy V-notch impact specimen.

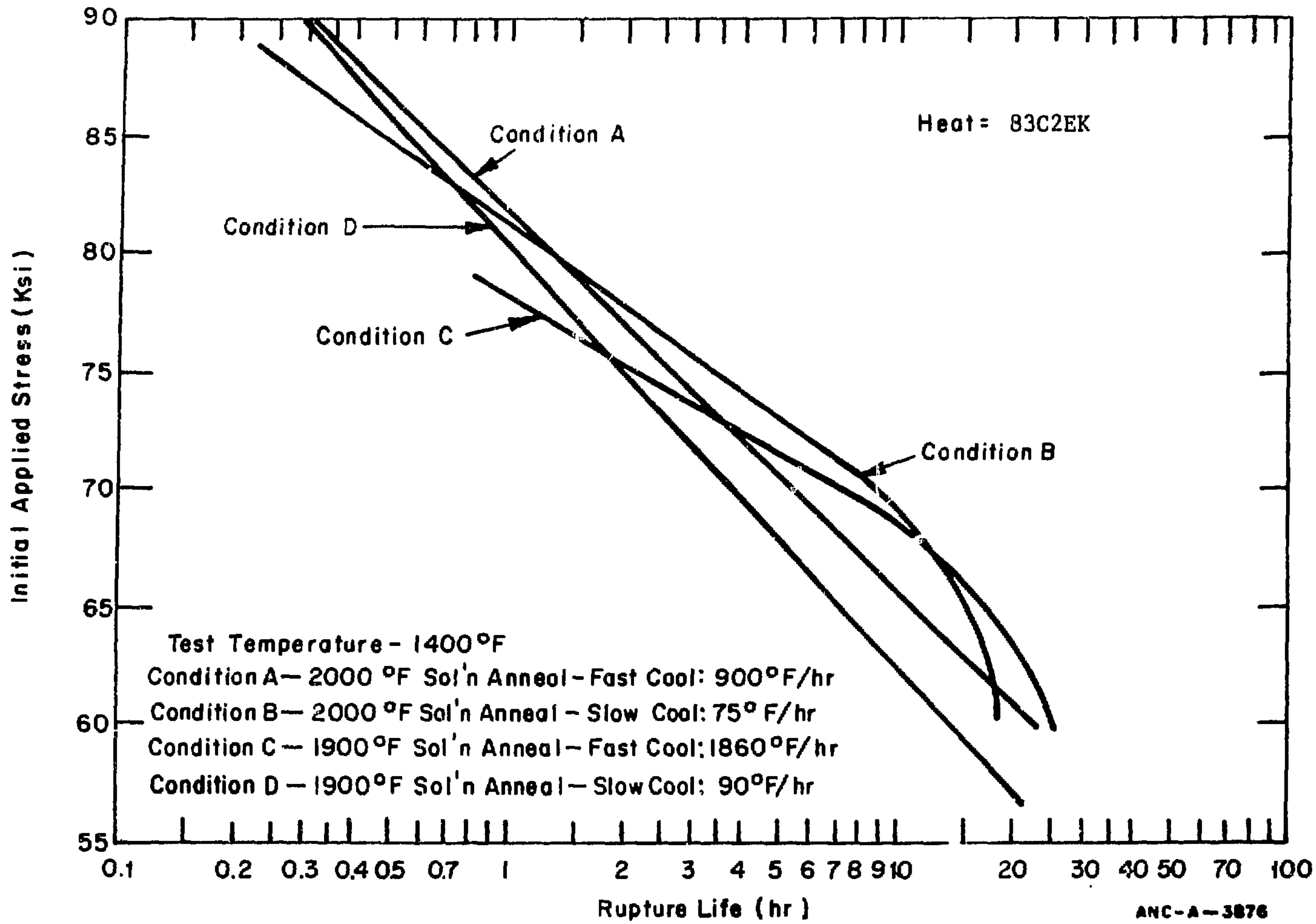


Fig. 6 Rupture life at 1400 F for solution-annealed 718 alloy with varying solution anneal conditions.

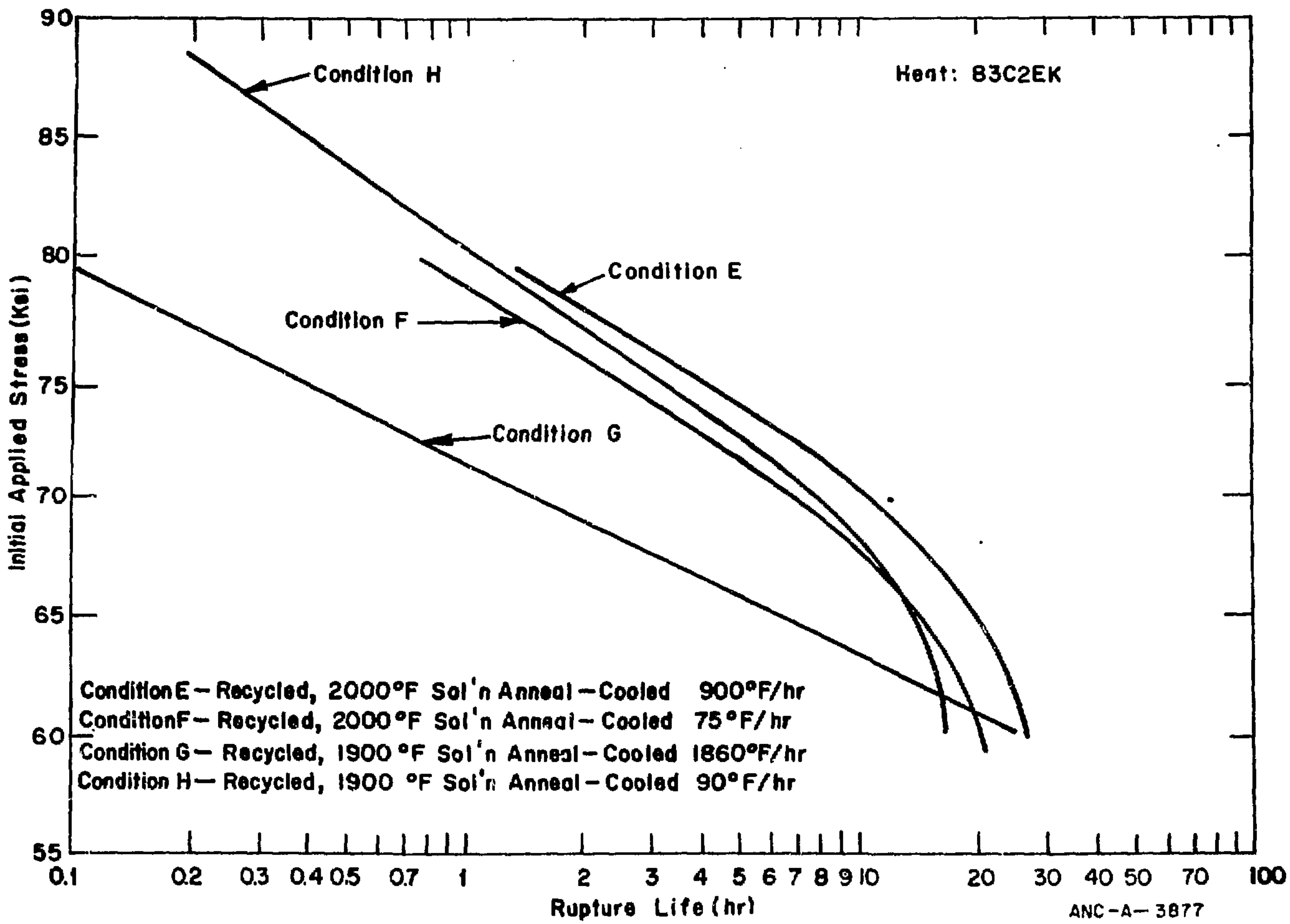


Fig. 7 Rupture life at 1400 F for Inconel 718 solution annealed to various conditions following a prior thermal history involving a 2000 F solution anneal, 1400-1200 F age, and a 1800 F stress relief.

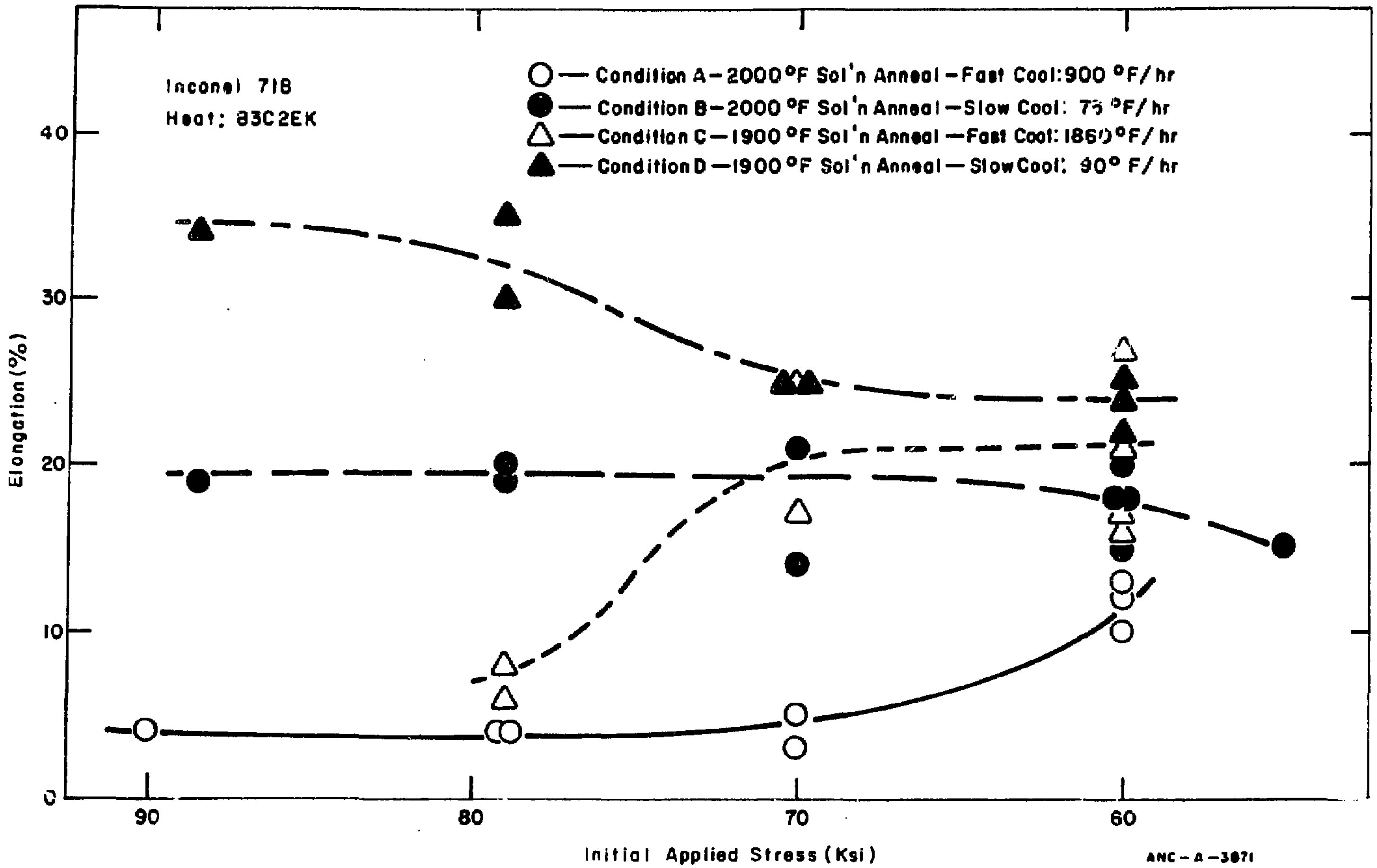


Fig. 8 Stress rupture elongations resulting from initial stress levels at 1400 F on various solution-annealed conditions.

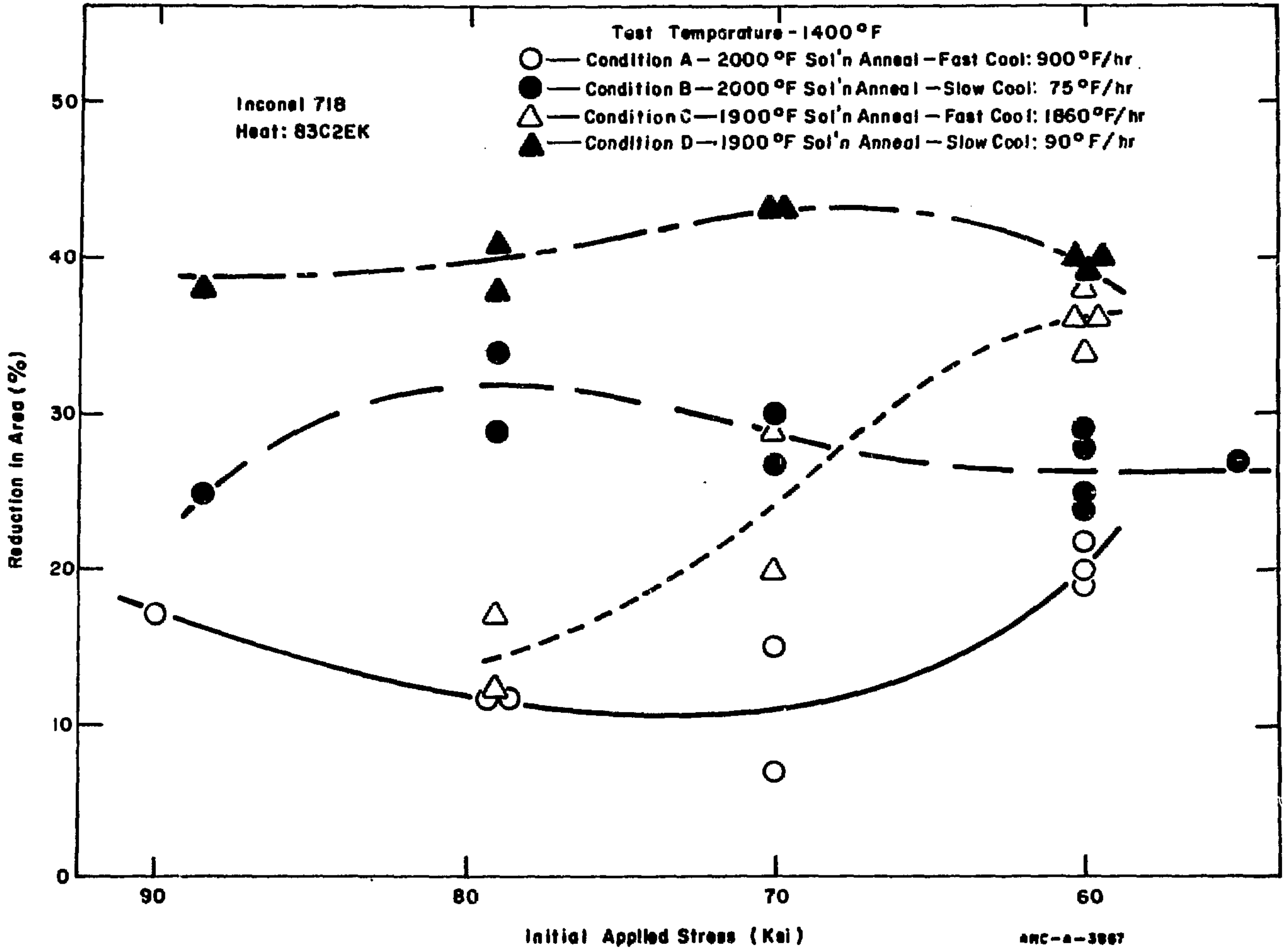


Fig. 9 Reductions in area for 1400 F stress rupture tests of various solution-annealed conditions.

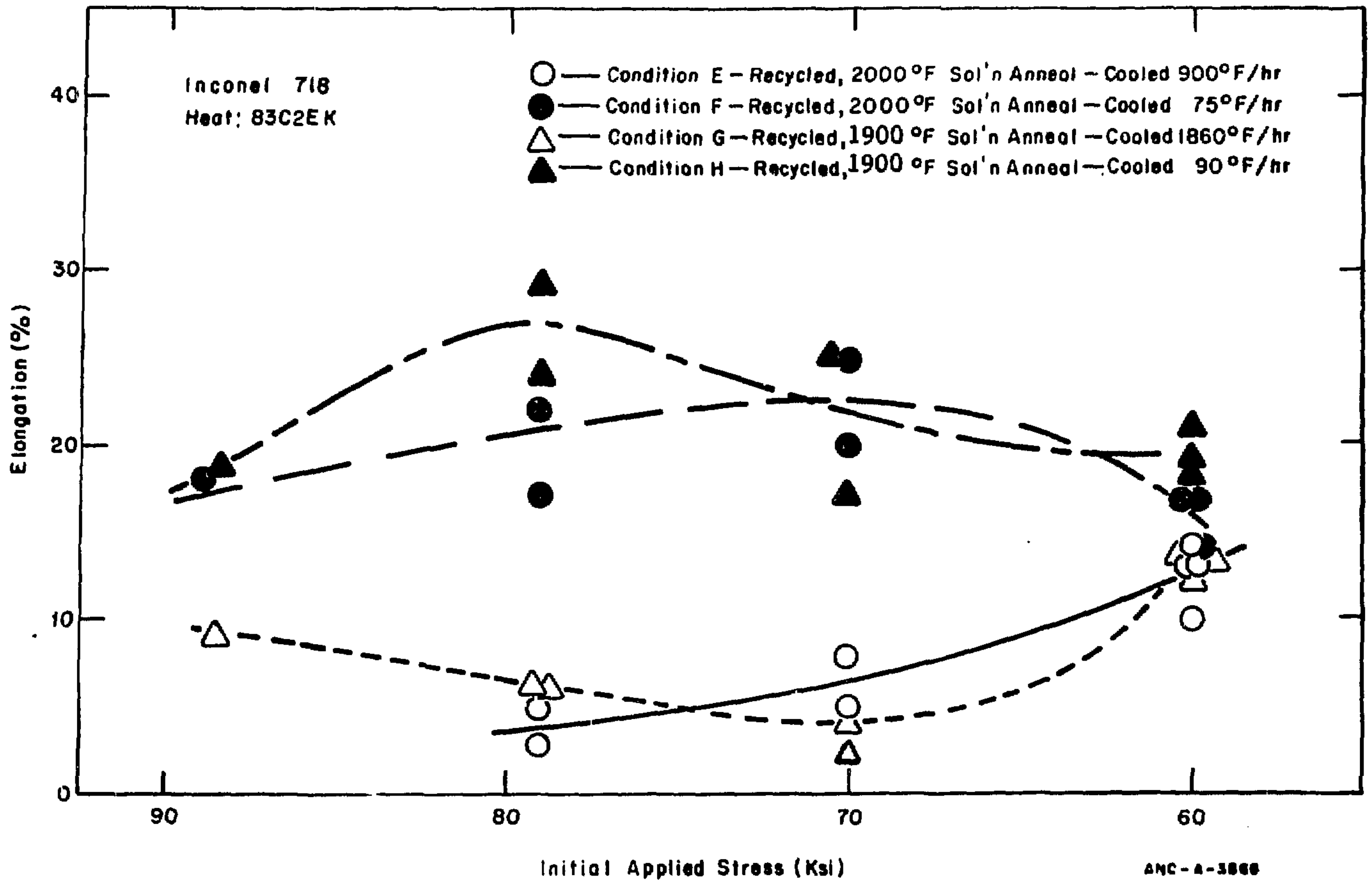


Fig. 10 Stress rupture elongations obtained at 1400 F for various solution-annealed conditions which had received prior thermal cycling.

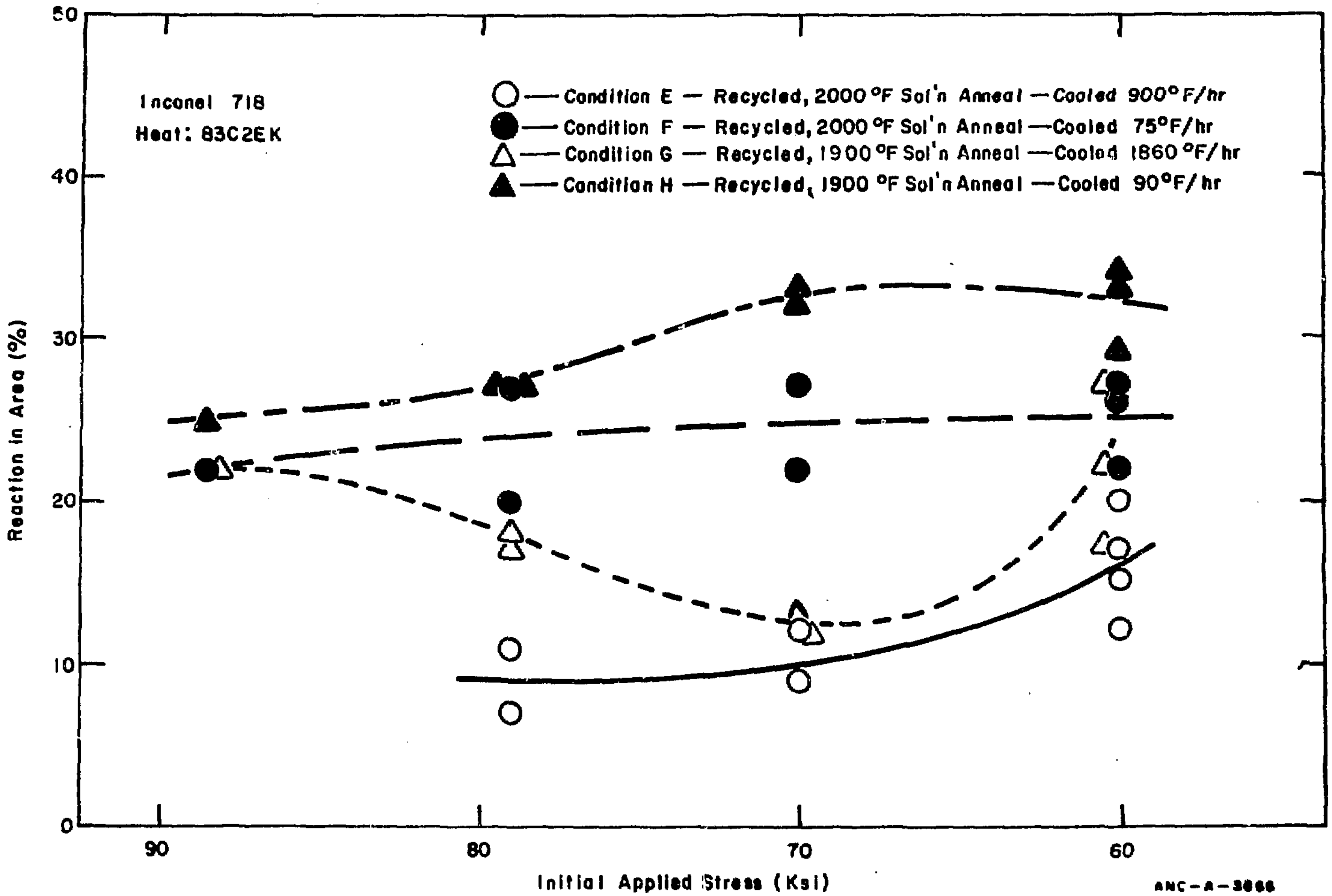


Fig. 11 Reductions in area for 1400 F stress rupture tests for various solution-annealed conditions which had received prior thermal cycling.

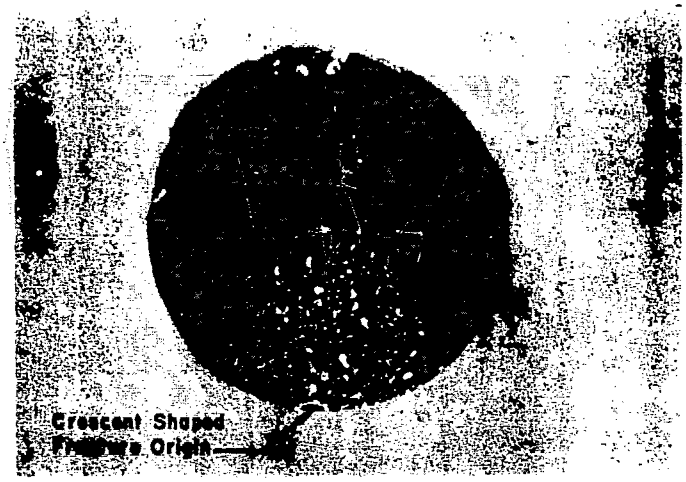


Fig. 12 Fracture of Inconel 718 1400^{o}F stress rupture specimen A-6 showing an intergranular crescent shaped origin. The fracture mode of this specimen stressed at 90 ksi is characteristic of specimens rapidly cooled from solution anneals and tested at stresses of 70 ksi or greater. 10X

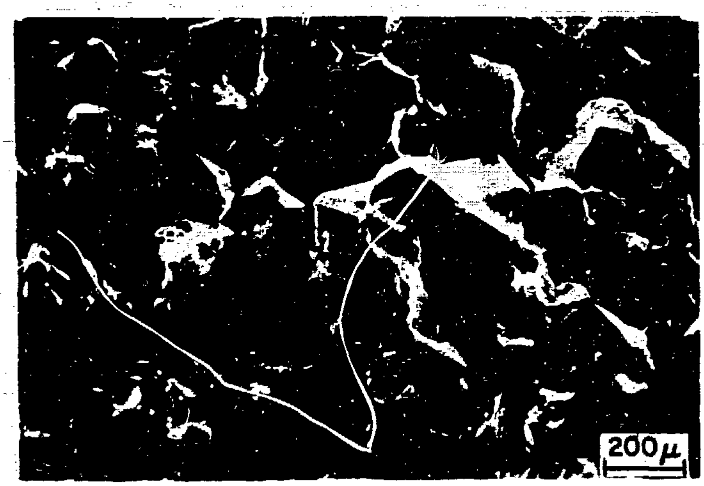


Fig. 13 Scanning electron fractograph displaying intergranular cracking and fracturing at fracture origin of stress rupture specimen A-6 tested at 90 ksi. The grain boundaries are very smooth. 80X

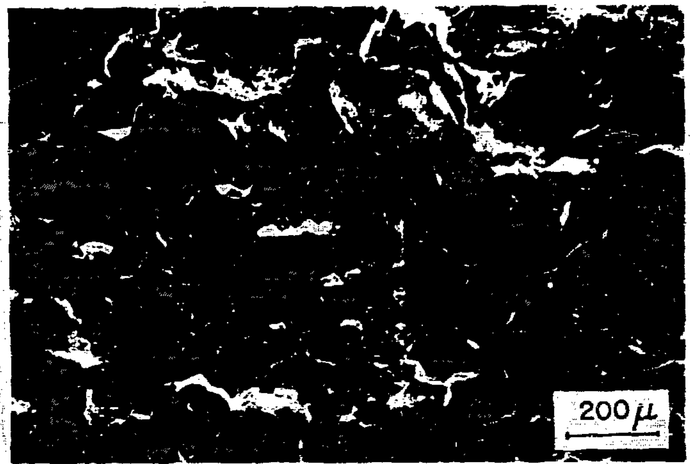


Fig. 14 Dimpling as indicated by the arrows may be observed on many grain boundaries at the intergranular origin of a stress rupture specimen cooled from a 2000°F solution anneal at a rate of 75°F per hour. This specimen identified at B-5 was stressed at 88.5 ksi at 1400°F. 65X

Fig. 15 Extensive dimpling observed on grain boundaries of stress rupture specimen A-11 cooled from 2000°F at a rate of 900°F per hour then tested at 1400°F at a stress of 60 ksi. 325X

Fig. 16 Microstructure illustrating banding in Inconel 718 (Heat 83C2EK) cooled from a 2000^{o}F solution anneal at a rate of 75^{o}F per hour (condition B). Etchant No 1 500X

Fig. 17 SEM micrograph showing the precipitates in the banded structure of Inconel 718 (Heat 83C2EK) heat treated to condition B. Precipitates at the top of the micrograph represent those in the light colored bands of Fig. 16. Precipitates were not distinctly resolved in the dark bands. Etchant No. 1 4200X

18a. Replication of the polished and etched microstructure for Heat 83C2EK cooled from 2000°F at a rate of 900°F per hour (condition A). Etchant No. 2 15,000X

18b. Polished and etched microstructure shown by a TEM replica for Heat 83C2EK cooled from 2000°F at a rate of 75°F per hour (condition B). Etchant No. 2 15,000X

Fig. 18 Effects of cooling rate upon Inconel 718 microstructures observed by TEM replication processes.

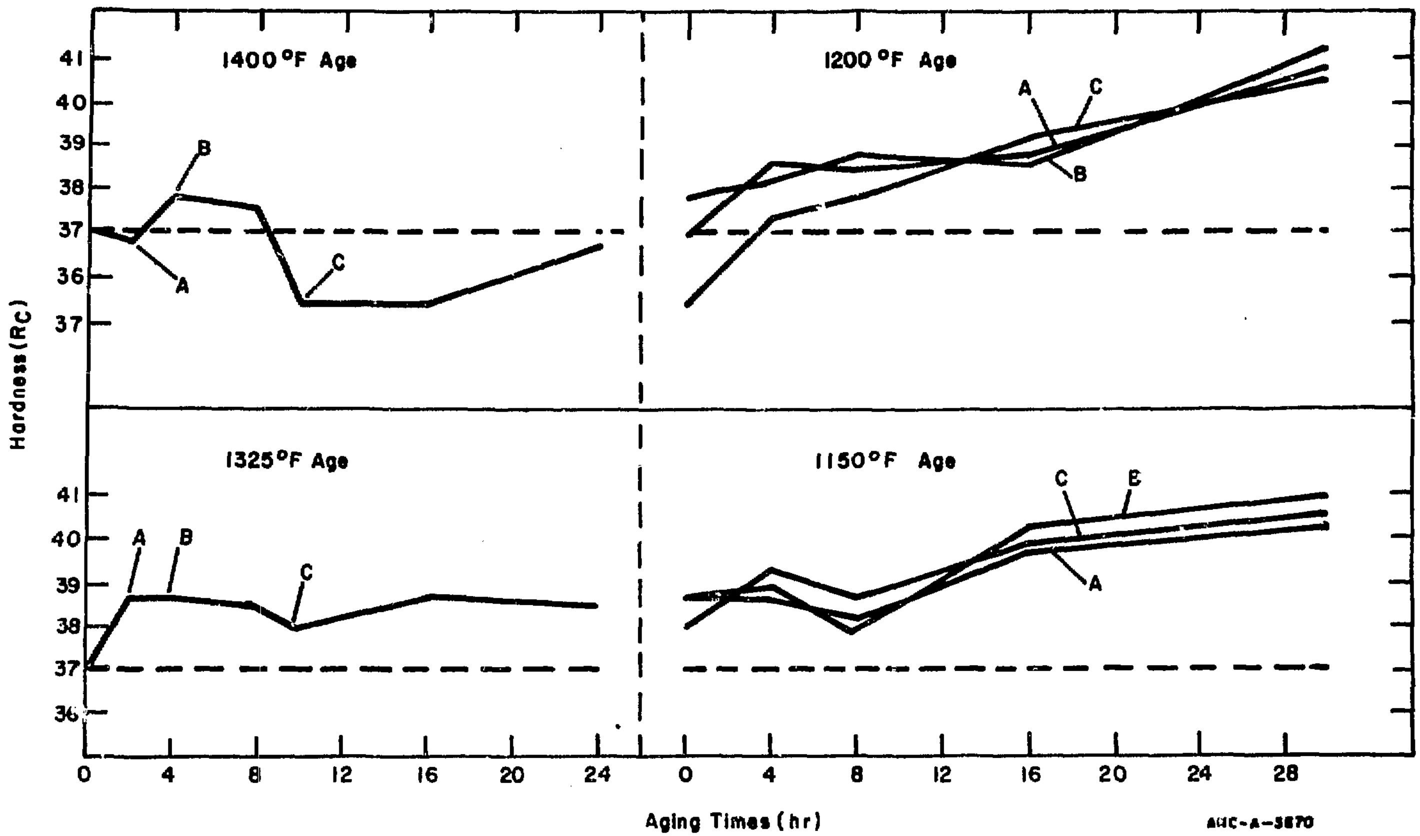


Fig. 19 Hardness curves for heat 83C2EK of Inconel 718 cooled from 2000 F at a rate of 75 F per hour and aged.

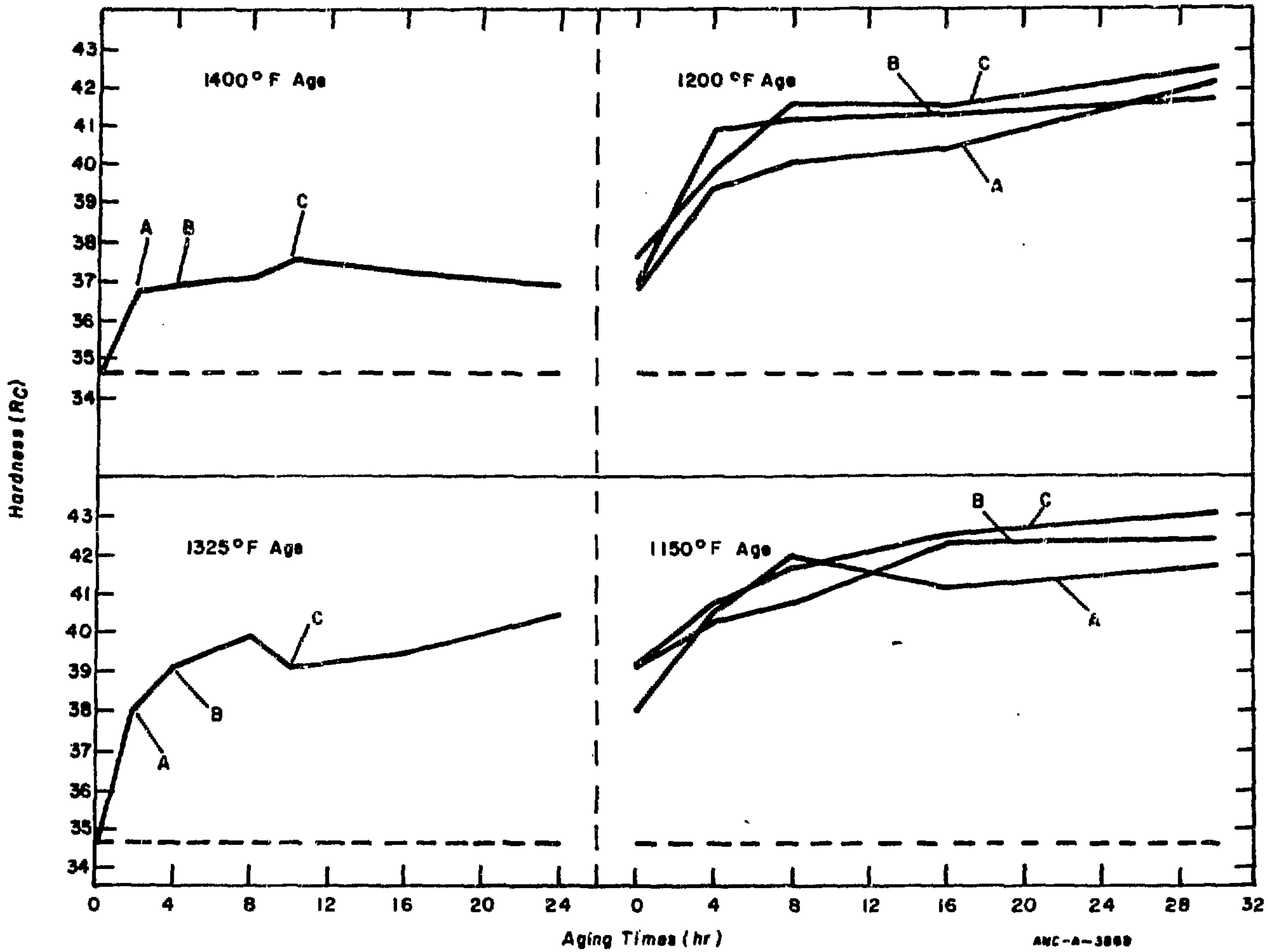


Fig. 20 Hardness curves for heat 87COEK1 of Inconel 718 cooled from 2000 F at a rate of 400 F per hour and aged.

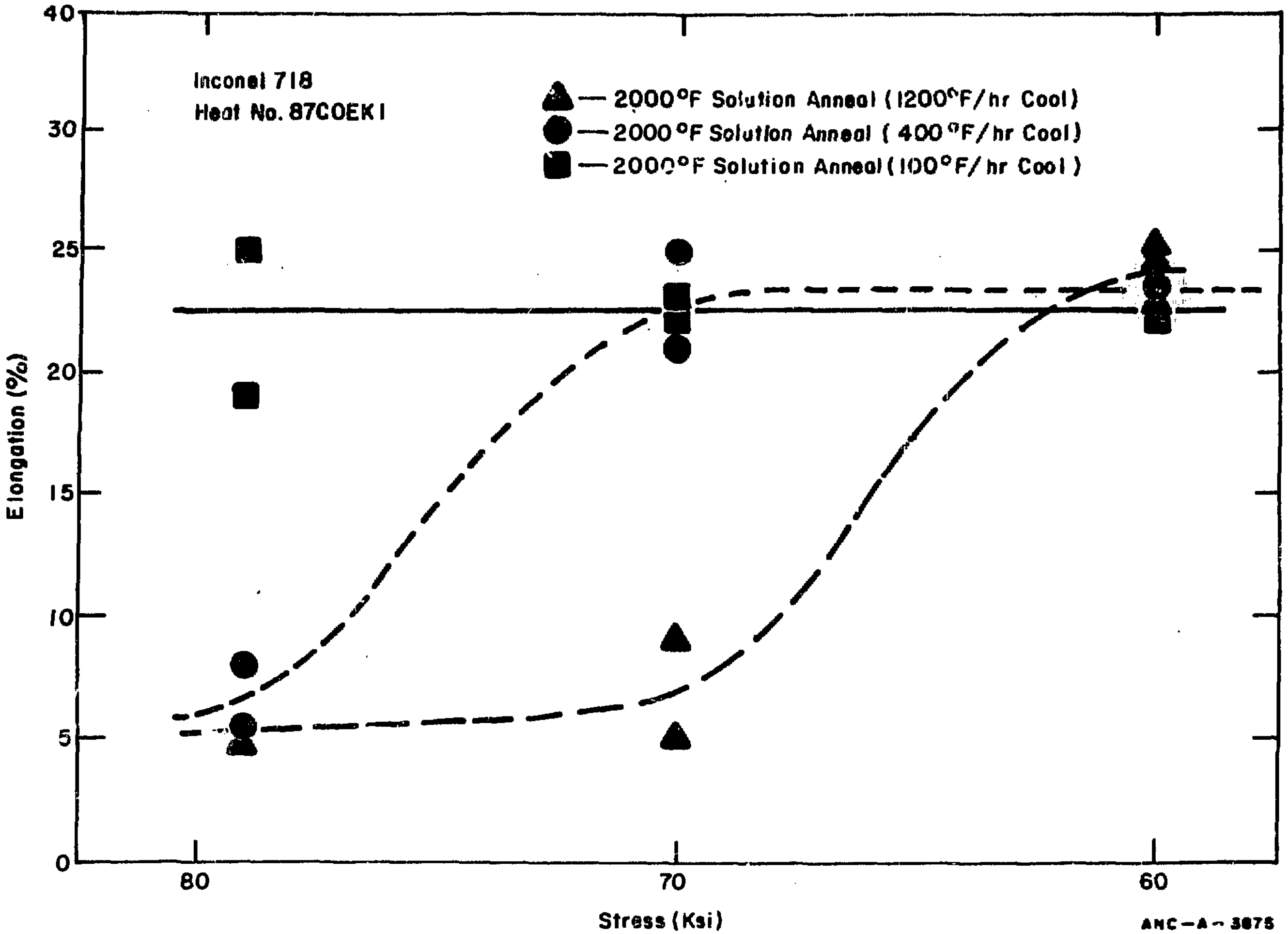


Fig. 21 Effects of cooling rates upon the rupture ductility of solution-annealed PBF acoustic filter shell material.

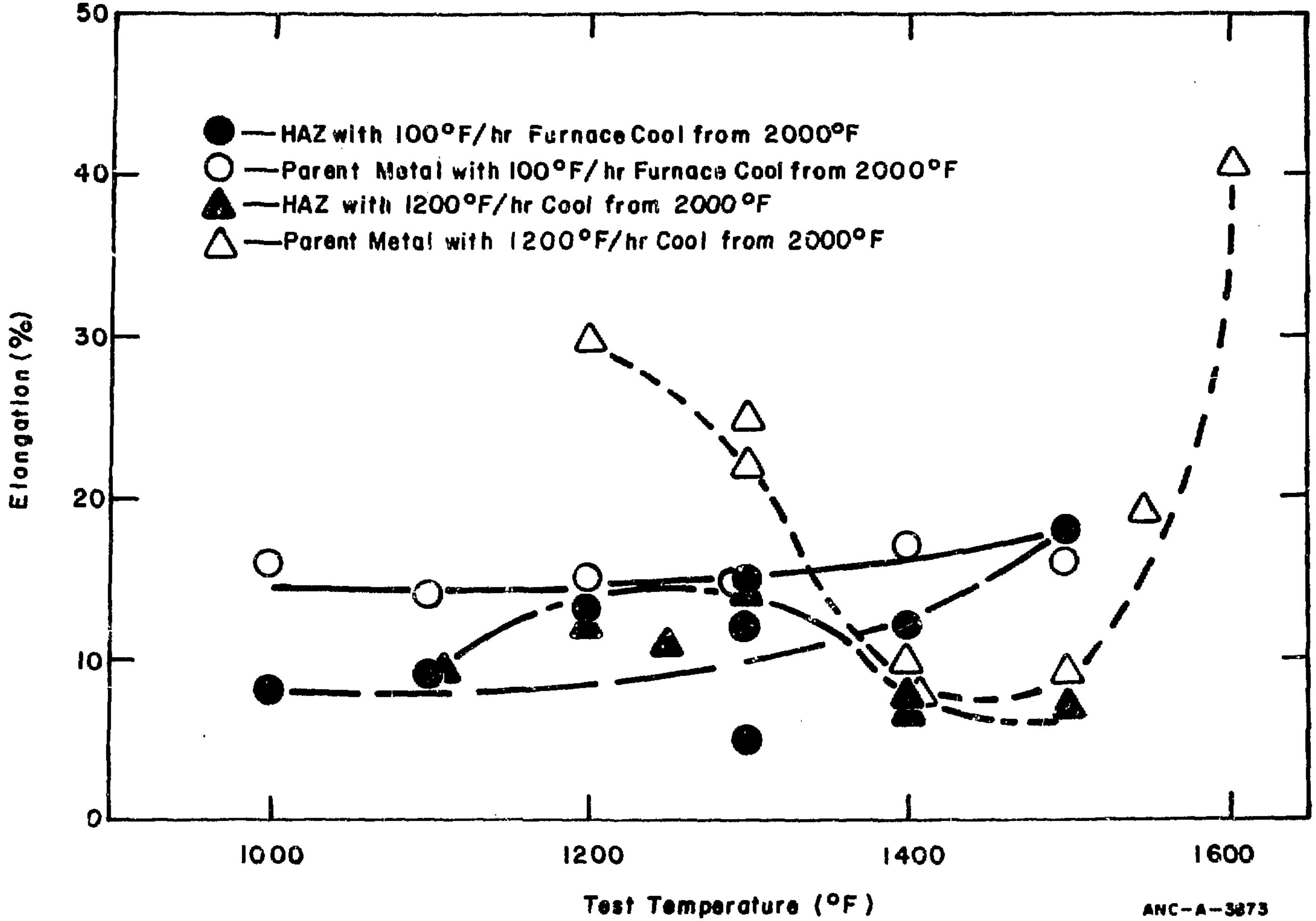


Fig. 22 Plots of percent elongation versus tensile test temperature for solution-annealed base metal and simulated HAZ of Inconel 718, heat 83C2EK.

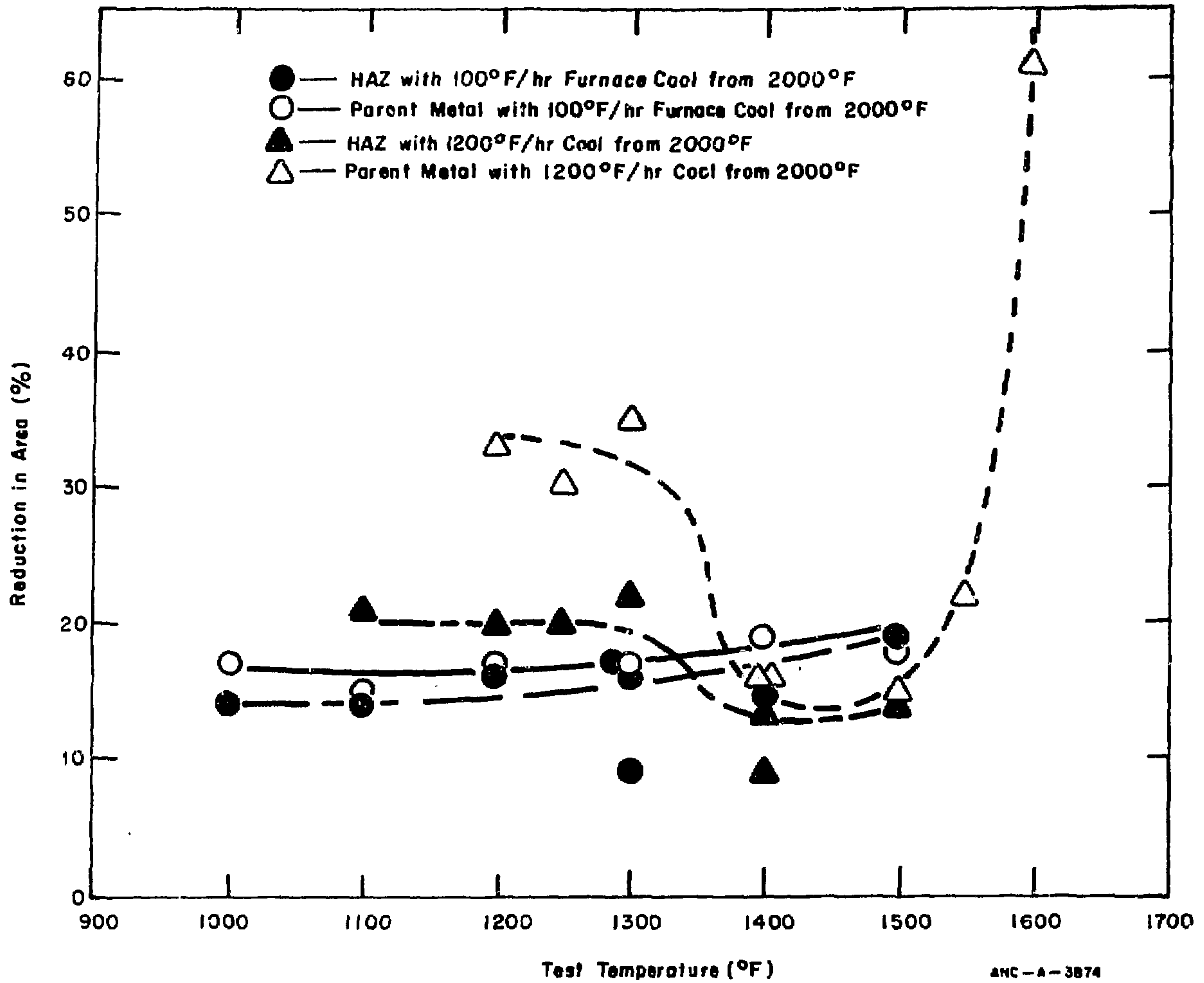


Fig. 23 Plots of reductions of area versus tensile test temperature for solution-annealed base metal and simulated HAZ of Inconel 718, heat 83C2EK.

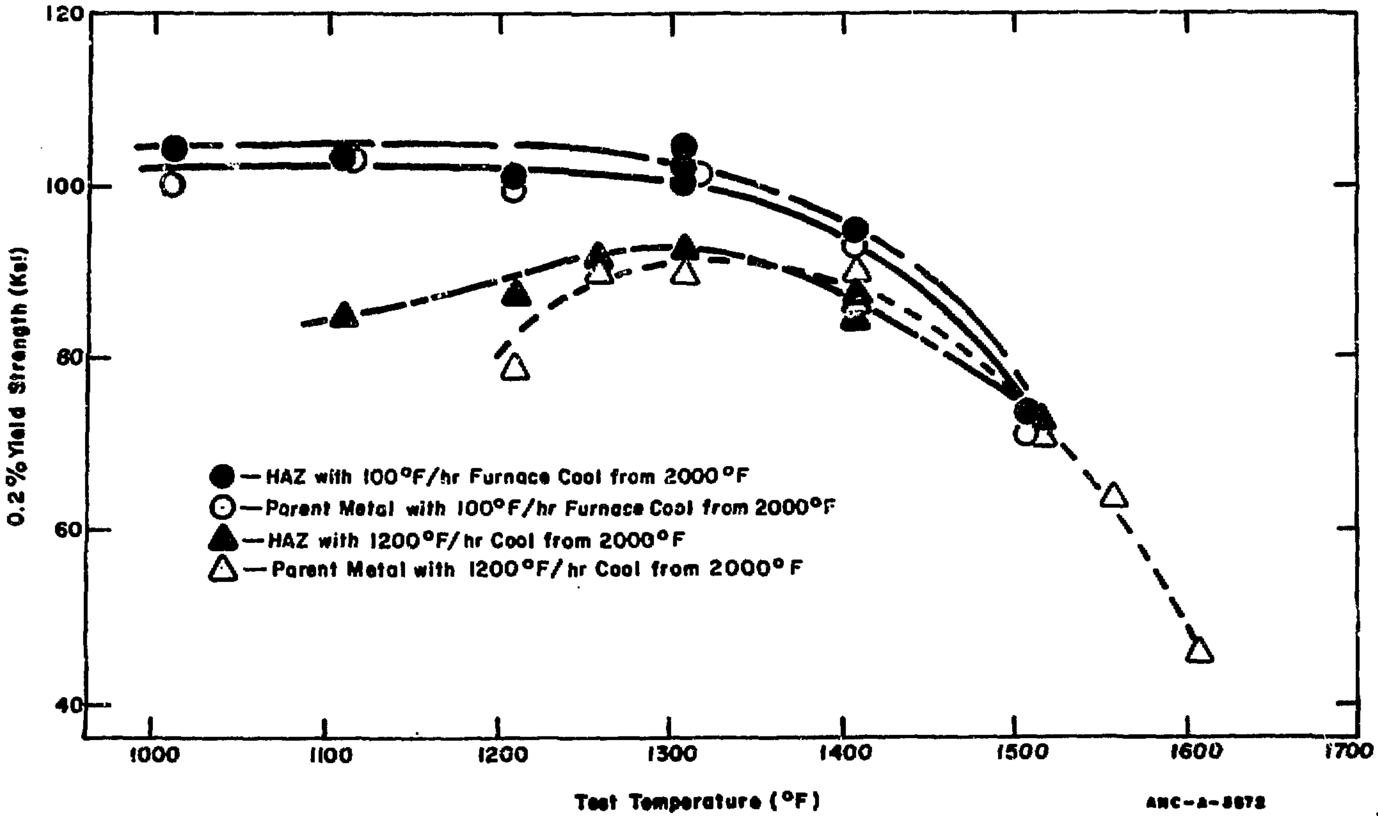


Fig. 24 Yield strengths as a function of test temperature for solution-annealed base metal and simulated HAZ of Inconel 718, heat 83C2EK.