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FATIGUE-CRACK PROPAGATION BEHAVIOR OF INCONEL 718

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HANFORD ENGINEERING DEVELOPMENT LABORATORY
Operated by Westinghouse Hanford Company
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Lee A. James

September, 1975

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ABSTRACT

The techniques of linear-elastic fracture mechanics were used to characterize the effect of several variables (temperature, environment, cyclic frequency, stress ratio, and heat-treatment variations) upon the fatigue-crack growth behavior of Inconel 718 base metal and weldments. Relevant crack growth data on this alloy from other laboratories is also presented.

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FIGURES (CONT'D)

FIGURES (CONT'D)

FATIGUE-CRACK PROPAGATION BEHAVIOR OF INCONEL 718

I. INTRODUCTION

The techniques of linear-elastic fracture mechanics are enjoying ever-increasing usage in design and safety analyses of reactor structural components(1). With these techniques it is possible to assume the existence of a hypothetical crack or crack-like flaw in a component and, knowing the crack growth behavior of the material and the expected stress history, to verify the "Safe Life" for the expected lifetime of the component.

Inconel 718* is being utilized, or considered for use, in several LMFBR designs. Therefore, it is necessary to know the effect of various parameters (e.g. temperature, cyclic frequency, stress ratio, environment, metallurgical variables, etc.) upon the fatigue-crack propagation behavior of this material. Some of this information is available from the aerospace industries, where Inconel 718 has been extensively applied(3-8). However, many of the environmental and operating conditions for LMFBRs are relatively unique to these systems and much of the crack growth data needed for design and safety analyses has not been available. Therefore, crack propagation studies of Inconel 718 have been underway at HEDL. This report will present the results of that work along with that relevant information available from the aerospace industries(5-8) which will be useful in the analysis of reactor components.

II. EXPERIMENTAL PROCEDURE

Several heats of Inconel 718 were employed in this study and these are identified in Table I. Several heat-treatments were also utilized, and these are detailed in Table II. The chemical composition and mechanical properties are identified in Tables III and IV, respectively. Weldment specimens were tested as well as base metal specimens, and the welds were produced using the gas-tungsten-arc (GTA) process.

* Inconel 718 is a registered trademark of the International Nickel Company. However, the general specifications covering this material may be found in Reference 2.

TABLE I

IDENTIFICATION OF MATERIAL HEATS

Heat Identification	Producer/Heat	Purchase Specification
A	Haynes/2180-9-9104	AMS 5597
B	Huntington/52C9EK	AMS 5596C
C	GTA Weld: Haynes/2180-1-9345 (base) and Huntington/60C7E (filler)	AMS 5832A
D	GTA Weld: Huntington/52C9EK (base) and Huntington/52B0E (filler)	AMS 5832A

TABLE II

HEAT TREATMENTS

Treatment No.	
1.	Solution annealed at 1950°F, air cooled to room temperature.
2.	Solution annealed at 1950°F, air cooled to room temperature. Aged 8 hours at 1325°F, furnace cooled to 1150°F and held at 1150°F for a total aging time of 18 hours, air cooled to room temperature.
3.	Annealed at 1750°F, air cooled to room temperature. Aged 8 hours at 1325°F, furnace cooled to 1150°F and held at 1150°F for a total aging time of 18 hours, air cooled to room temperature. (Equivalent to the heat-treatment specified in Reference 9).
4.	Solution annealed 1 hour at 2000°F, cooled to 1325°F at 100°F/hour, aged 4 hours at 1325°F, cooled to 1150°F at 100°F/hour, aged 16 hours at 1150°F, air cooled to room temperature.
5.	Mill annealed. GTA welded. After welding given heat treatment No. 3.
6.	Mill annealed. GTA welded. After welding, annealed 1 hour at 2050°F, cooled to 1370°F at 101°F/hour, aged 4 hours at 1370°F, cooled to 1190°F at 102°F/hour, aged 16 hours at 1190°F, air cooled to room temperature.
7.	Mill annealed. GTA welded. After welding given heat treatment No. 4.

TABLE III

CHEMICAL COMPOSITION (Percent by weight)

Producer Heat	C	Mn	Fe	S	Si	Cu	Ni	Cr	Al	Ti	Co	P	Mo	B	Cb & Ta	Cb	Ta
Haynes 2180-9-9104	.05	.19	Bal.	.003	.09	<.01	53.00	17.95	.46	.95	.07	.003	3.16	.003	5.17	*	*
Haynes 2180-1-9345	.06	.08	Bal.	<.005	.07	.04	52.88	18.06	.51	1.11	.22	<.005	2.89	.004	5.04	*	*
Huntington 52C9EK	.07	.08	18.19	.007	.17	.11	53.44	18.11	.59	1.08	.03	*	3.00	*	5.10	*	*
Huntington 60C7E	.04	.09	17.97	.007	.18	.10	53.84	18.04	.55	1.01	.03	.010	3.00	.0042	5.12	5.11	.01
Huntington 52B0E	.04	.04	17.60	.007	.19	.05	54.58	17.91	.49	.99	.04	.011	2.96	.0031	5.08	5.07	.01

* Not determined

TABLE IV

MECHANICAL PROPERTIES

Heat Identification	Heat Treatment	Test Temp. (°F)	Strain Rate (sec.$^{-1}$)	Strength (KSI) P.E.L.	Strength (KSI) 0.2% Y.S.	Strength (KSI) U.T.S.	Elongation(%) Total	Elongation(%) Uniform	Red. of Area (%)	Hardness	ASTM Grain Size	Site Sequence Number***
A	1	72	*	*	45.6	111.0	60.5	*	*	R_B=86.0	4.5	*
A	3	72	*	*	161.1	198.2	20.5	*	*	R_C=43.6	*	*
B	3	72	3x10^{-5}	113.8	154.5	198.0	16.4	15.5	30.9	R_C=44.9	5	H04781
B	3	72	3x10^{-5}	121.9	144.2	183.7	20.0	18.8	30.9	*	*	H04782
B	3	72	3x10^{-5}	118.4	152.7	191.1	19.0	18.2	28.5	*	*	H04794
B	3	72	3x10^{-5}	118.0	150.5	192.3	19.5	18.0	38.2	*	*	H04796
B	3	600	3x10^{-5}	96.3	134.0	167.1	18.9	16.4	31.7	*	*	H04783
B	3	800	3x10^{-5}	112.2	135.1	166.4	17.4	13.5	30.9	*	*	H04784
B	3	800	3x10^{-5}	108.9	133.3	164.1	17.9	14.4	30.9	*	*	H04785
B	3	1000	3x10^{-5}	97.6	130.1	162.4	17.9	15.5	30.1	*	*	H04786
B	3	1000	3x10^{-5}	104.1	129.0	160.1	13.6	12.2	25.2	*	*	H04787
B	3	1200	3x10^{-5}	89.4	119.8	136.2	6.8	6.3	17.1	*	*	H04792
B	3	1200	3x10^{-5}	97.6	124.0	139.5	7.2	6.8	15.4	*	*	H04793
B	4	72	3x10^{-5}	123.0	146.7	192.2	11.2	10.8	18.0	R_C=41.4	*	H04910
B	4	800	3x10^{-5}	111.3	129.0	175.4	13.0	11.9	17.7	*	*	H04909
B	4	1000	3x10^{-5}	105.7	125.6	171.5	14.8	12.9	24.4	*	*	H04908
B	4	1200	3x10^{-5}	96.8	120.2	131.6	10.3	6.2	17.7	*	*	H04907
C**	5	72	3x10^{-5}	105.7	147.5	184.1	14.9	14.5	17.1	*	*	H08334
C**	5	800	3x10^{-5}	100.6	135.4	162.4	17.6	12.7	23.6	*	*	H08309
C**	5	1000	3x10^{-5}	104.1	134.7	163.8	14.6	13.3	22.8	*	*	H08310
C**	5	1200	3x10^{-5}	95.9	126.3	141.0	6.9	6.4	16.3	*	*	H08335

* Not Determined.
** Longitudinal specimens. Specimens were entirely weld metal.
*** HEDL LMFBR Fuel Cladding Information Center.

The ASTM "Compact Specimen" design[10] was used in the fatigue tests. The specimens were tested on an MTS servo-controlled electro-hydraulic testing machine using load as the control parameter. Various test temperatures, cyclic frequencies, stress ratios and heat-treatments were used depending upon which parameter was being investigated, and the specimen dimensions and test parameters are listed in Table V.

With the exception of one specimen which was tested in a liquid sodium environment, all specimens were tested in air where the relative humidity ranged between 65% and 85% at room temperature. Elevated temperature tests were conducted in an air-circulating furnace, and temperatures were controlled to within 2°F (1°C). The specimen tested in a liquid sodium environment was tested in an environmental test chamber which has been described elsewhere[11].

Crack lengths were determined periodically throughout each test using a travelling microscope. The crack growth rate (da/dN) was calculated by dividing each increment of crack extensions (Δa) by the number of cycles producing that increment (ΔN). The stress intensity factor (K) was calculated using the average crack length for each growth increment, and the relationships

$$K = \frac{P}{B\sqrt{W}} \left[29.6 \left(\frac{a}{W}\right)^{0.5} - 185.5 \left(\frac{a}{W}\right)^{1.5} + 655.7 \left(\frac{a}{W}\right)^{2.5} - 1017.0 \left(\frac{a}{W}\right)^{3.5} + 638.9 \left(\frac{a}{W}\right)^{4.5}\right]$$

$$0.3 < \frac{a}{W} < 0.7 \quad \text{(Ref. 10)} \qquad \text{Eq. [1]}$$

and

$$K = \frac{P}{2B} \left[\frac{w+a}{(w-a)^{1.5}}\right] \left[4.0 + \frac{w-a}{w+a}\right]$$

$$\frac{a}{W} > 0.7 \quad \text{(Ref. 12)} \qquad \text{Eq. [2]}$$

where a = crack length, B = specimen thickness, P = applied load, and W = specimen width.

TABLE V

TEST PARAMETER AND SPECIMEN DIMENSIONS

Spec. No.	Mat. Cond.	Heat No.	Orientation**	Test Temp. (°F)	Test Freq. (cpm)	Stress Ratio	Environment	Max. Load (lb.)	"W" (in.)	"B" (in.)
126	1	A	L-T	75	500	0.050	Air	1550	1.9997	0.4915
127	2	A	L-T	600	40	0.050		1750	1.9967	0.4896
128	2	A	L-T	75	500	0.050		1650	1.9920	0.4912
129	1	A	L-T	1000	40	0.050		1500	2.0004	0.4897
130	2	A	T-L	1000	40	0.050		1450	1.9984	0.4904
131	2	A	T-L	1000	40	0.050		1500	1.9961	0.4901
132	1	A	T-L	75	500	0.050		1650	1.9988	0.4893
133	1	A	T-L	600	40	0.050		1650	1.9993	0.4877
154	3	B	L-T	1000	40	0.050		1350	1.9978	0.4760
155	3	B	L-T	1000	40	0.050		1300	1.9981	0.4761
156	3	B	L-T	1200	40	0.050		1400	1.9968	0.4767
157	3	B	L-T	800	400	0.333		1300	1.9955	0.4765
158	3	B	L-T	75	500	0.050		1500	1.9984	0.4765
159	3	B	L-T	1000	40	0.050	Air	1300	1.9972	0.4777
160	3	B	L-T	800	40/400	0.05/0.50	Sodium	1550	1.9969	0.4781
161	3	B	L-T	800	400	0.500	Air	1200	1.9959	0.4784
162	3	B	L-T	800	40	0.050		1400	1.9962	0.4764
163	3	B	L-T	800	400	0.050		1300	1.9974	0.4772
164	3	B	L-T	1000	400	0.667		1700	1.9934	0.4789
165	3	B	L-T	1000	40	0.050		1500	1.9978	0.4757
208	3	B	*	1000	400	0.333		550	1.1520	0.2977
209	3	B	*	1000	4	0.050		600	1.1511	0.2998
210	3	B	*	600	40	0.050		550	1.1521	0.2988
212	3	B	*	1000	0.083	0.050		650	1.1516	0.2991
213	3	B	*	1000	400	0.600		650	1.1514	0.2977
214	3	B	*	1000	400	0.050		500	1.1514	0.2978
215	3	B	*	1000	400	0.500		600	1.1516	0.2986
250	4	B	*	1000	40	0.050		1300	1.9868	0.5110
251	4	B	*	1000	40	0.050		1450	1.9795	0.5083
252	4	B	*	1200	40	0.050		1300	1.9819	0.5064
253	4	B	*	75	500	0.050		1400	1.9843	0.5060
254	4	B	*	600	40	0.050		1450	1.9797	0.5123
255	4	B	*	800	40	0.050		1400	1.9848	0.4995
367	5	C	AB	1000	40	0.050		1350	1.9966	0.4664
368	5	C	BA	1000	40	0.050		1300	2.0034	0.4539
369	5	C	BA	1000	40	0.050		1400	1.9992	0.4693
468	6	D***	BA	1000	40	0.050		1300	2.0126	0.4248
486	7	D***	BA	1200	40	0.050		850	1.4993	0.4514
487	7	D***	BA	1000	40	0.050	Air	900	1.5007	0.4508

* Not determined

** See Ref. 10 for crack orientation system for base metal specimens, and Ref. 37 for crack orientation system for weldment specimens

*** The welds in Specimens 468, 486 and 487 were produced by Aerojet Nuclear Co.

The results were then plotted in the form of fatigue-crack growth rate as a function of the stress intensity factor range (ΔK) or alternatively (as will be explained later) as a function of the "effective" stress intensity factor (K_{eff}).

III. RESULTS AND DISCUSSION

A. EFFECT OF TEMPERATURE

The effect of temperature upon material receiving the "conventional" heat-treatment (i.e. heat-treatments 2 or 3, which differ only in the annealing temperature) was determined for test temperatures of 75°C (24°C), 600°F (316°C), 800°F (427°C), 1000°F (538°C) and 1200°F (649°C). With the exception of 800°F (427°C) and 1200°F (649°C), tests were conducted on both Heat A material and Heat B material. The results are given in Figures 1-5 (the results for Heat A material have also been reported previously in Reference 13). Also shown in Figure 1 are the results of Shahinian et al[14] and Speidel[15] for tests conducted under similar conditions at room temperature. In general, the agreement between the results of Reference 14 and 15, and those of the present study are quite good. Coles et al[8] have tested heat-treated (similar to Heat-Treatment 3) Inconel 718 at 1000°F (538°C) at a frequency and stress ratio similar to those employed in the present study. Their data is also shown in Figure 4 for comparison purposes, and the agreement between the two sets of data is quite good.

In Figures 1, 2, and 4 where both heats of material were tested, there is no observable difference in the crack growth rate behavior between Heat A material and Heat B material. These results, although obviously quite limited, are in agreement with the results of Reference 16 which suggested that, for two austenitic stainless steels, heat-to-heat variations are at most a second-order variable in crack growth testing.

Comparing the results of Figures 1-5, it will be seen that, at a given value of ΔK, the fatigue-crack growth rate increases with increasing test temperature. The mean data trends of Figures 1-5 are summarized in Figure 6. The trend of crack growth rate increasing with increasing test temperature in an air environment has been previously observed in a wide variety of alloy

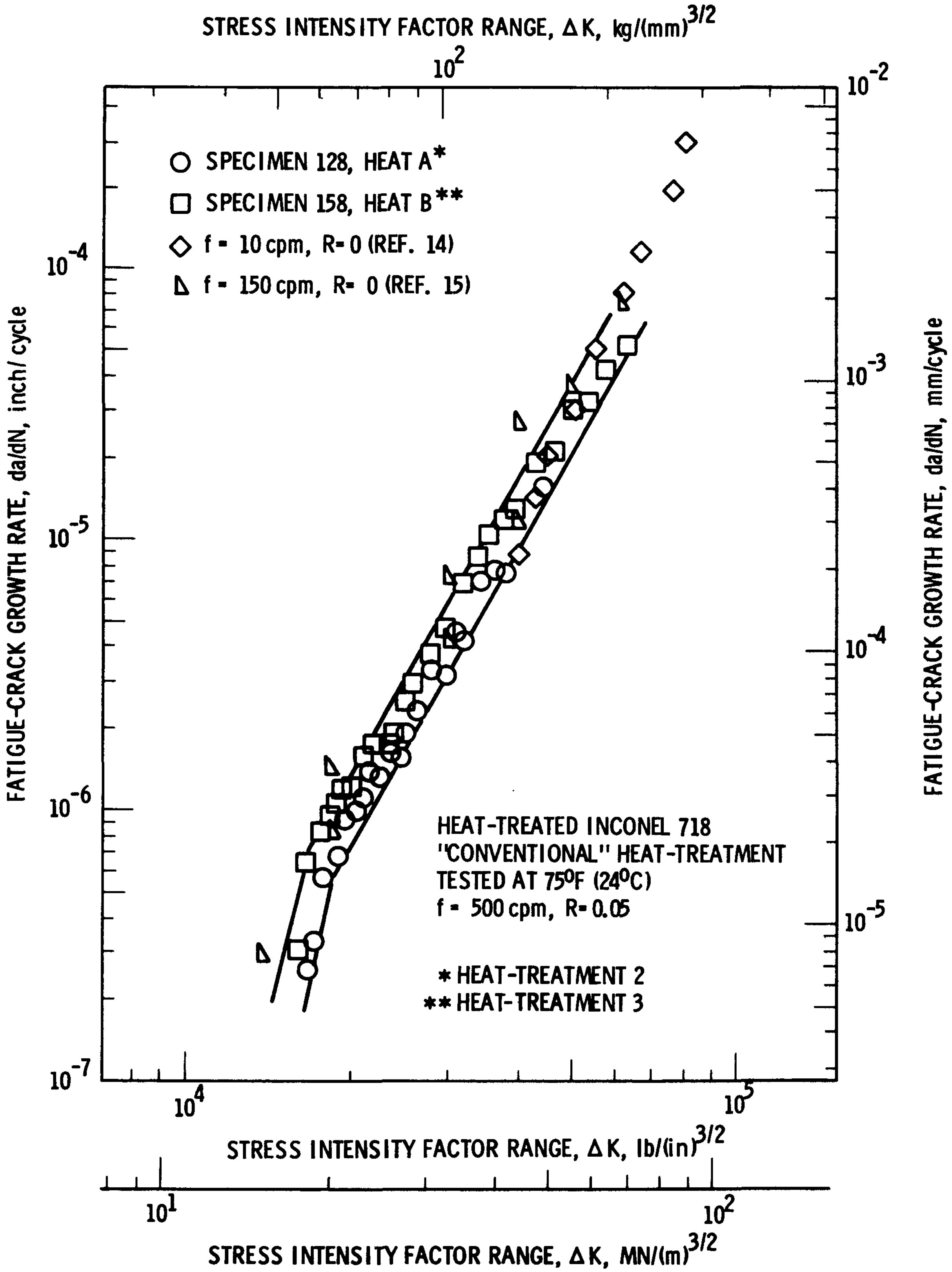

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FIGURE 1. Fatigue-Crack Growth Behavior of "Conventionally" Heat-Treated Inconel 718 in an Air Environment at Room Temperature.

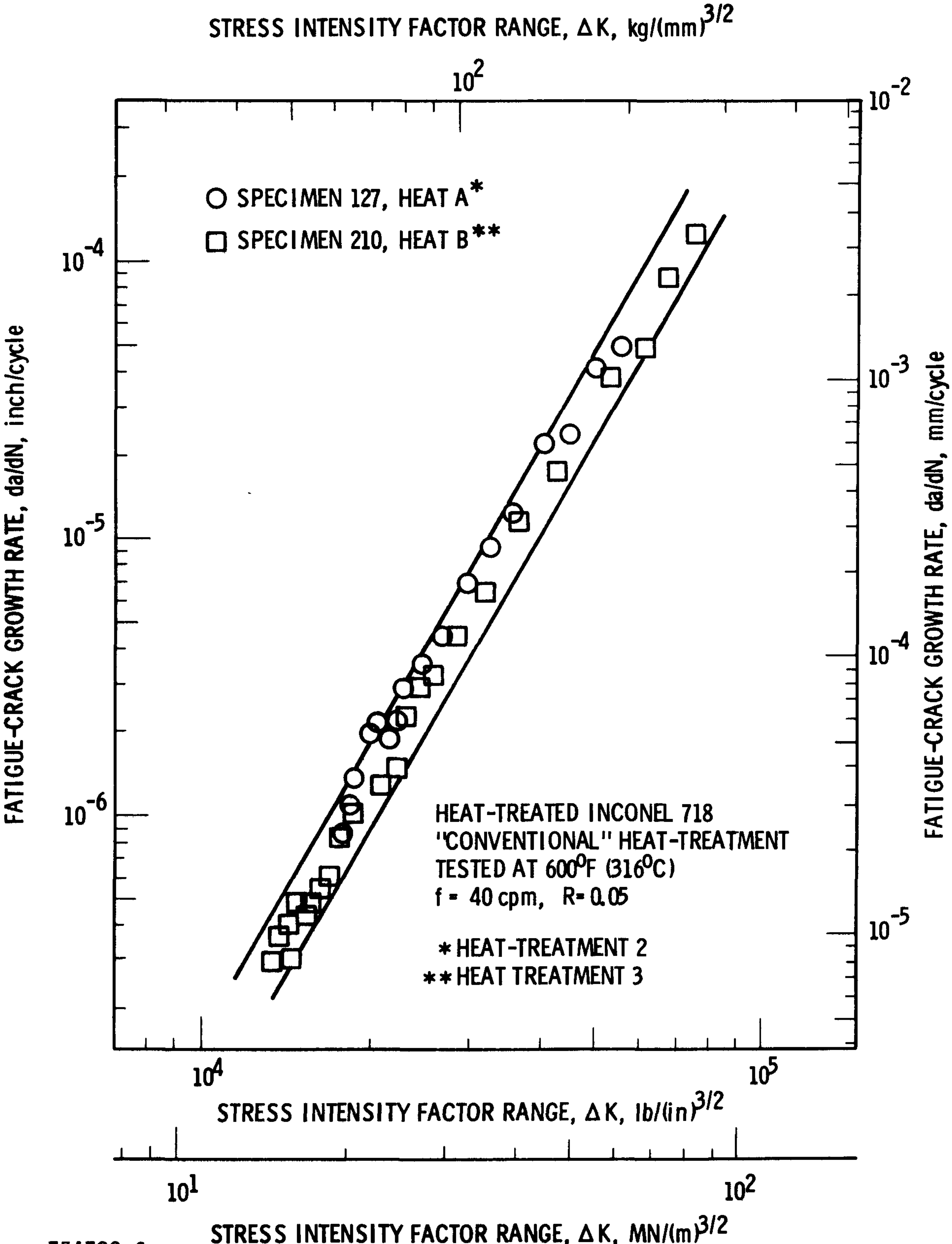

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FIGURE 2. Fatigue-Crack Growth Behavior of "Conventionally" Heat-Treated Inconel 718 in an Air Environment at 600°F (316°C).

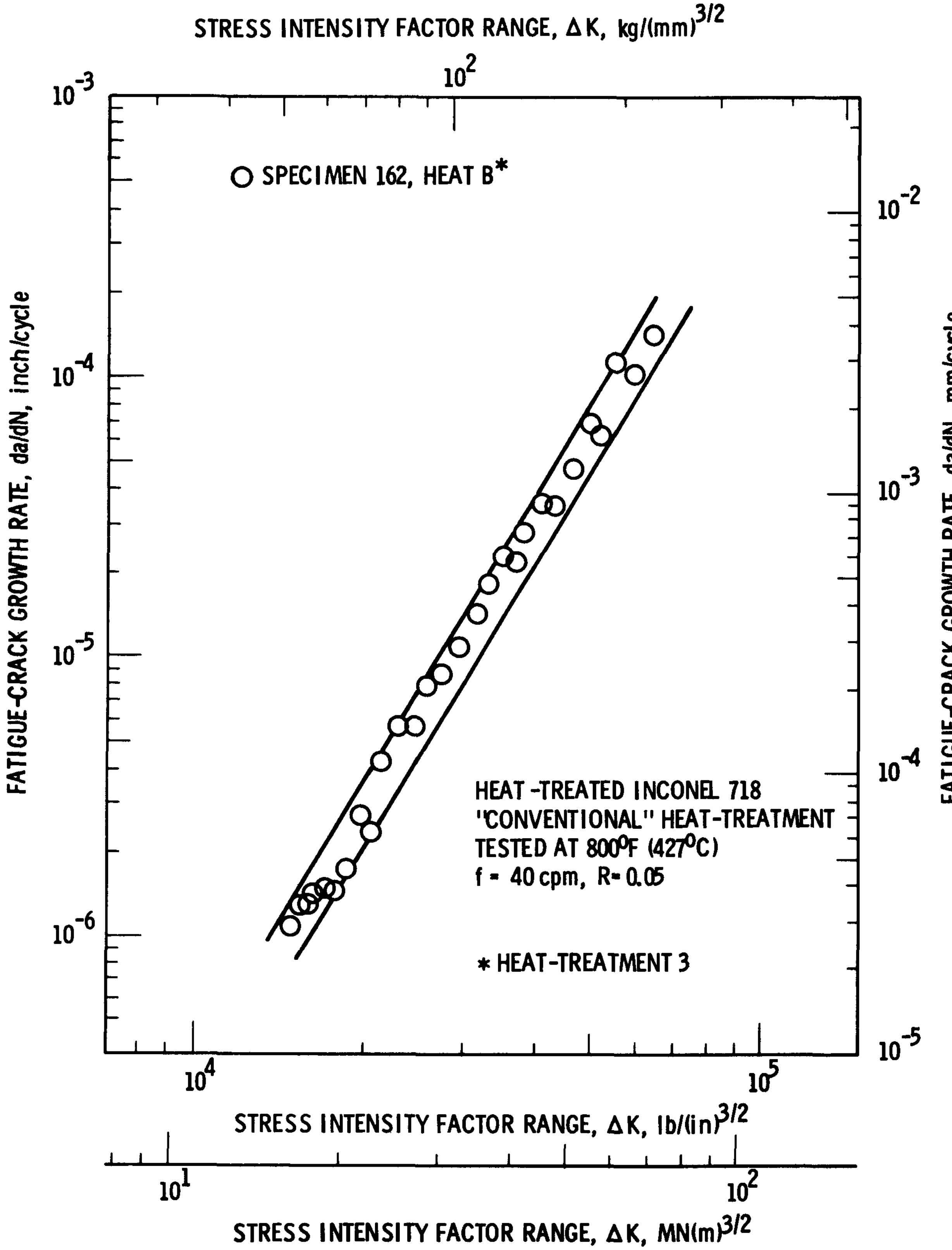

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FIGURE 3. Fatigue-Crack Growth Behavior of "Conventionally" Heat-Treated Inconel 718 in an Air Environment at 800°F (427°C).

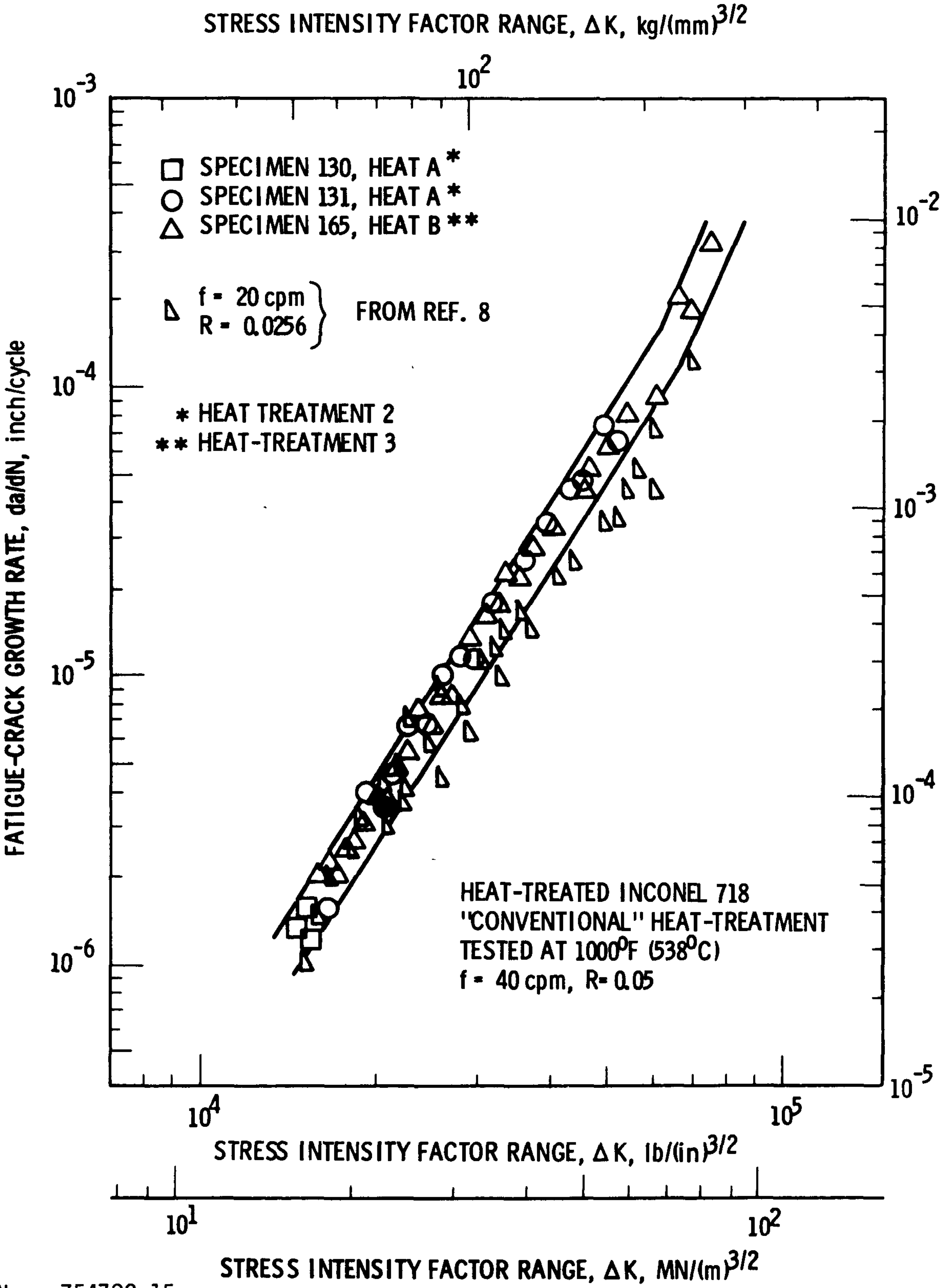

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FIGURE 4. Fatigue-Crack Growth Behavior of "Conventionally" Heat-Treated Inconel 718 in an Air Environment at 1000°F (538°C).

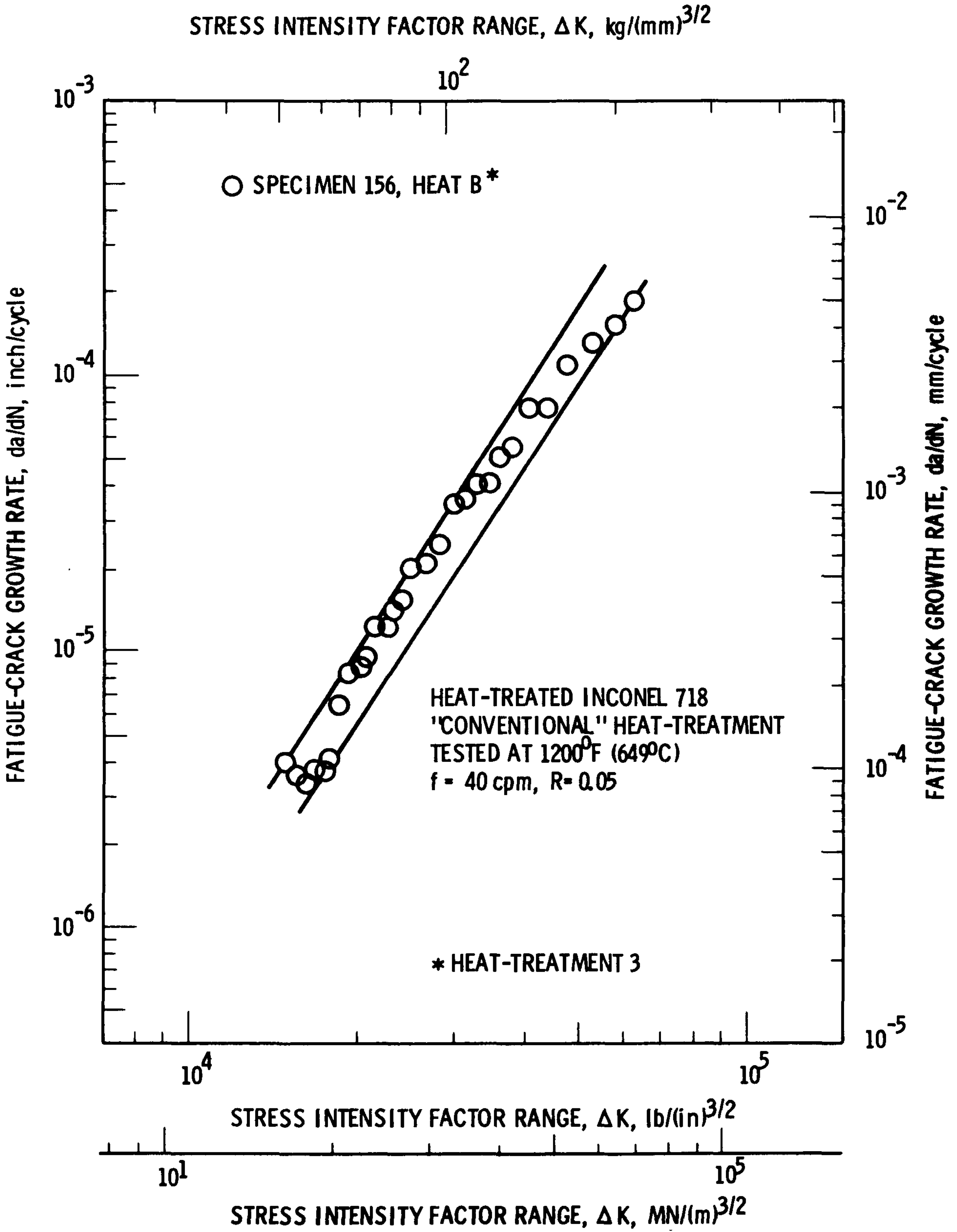

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FIGURE 5. Fatigue-Crack Growth Behavior of "Conventionally" Heat-Treated Inconel 718 in an Air Environment at 1200°F (649°C).

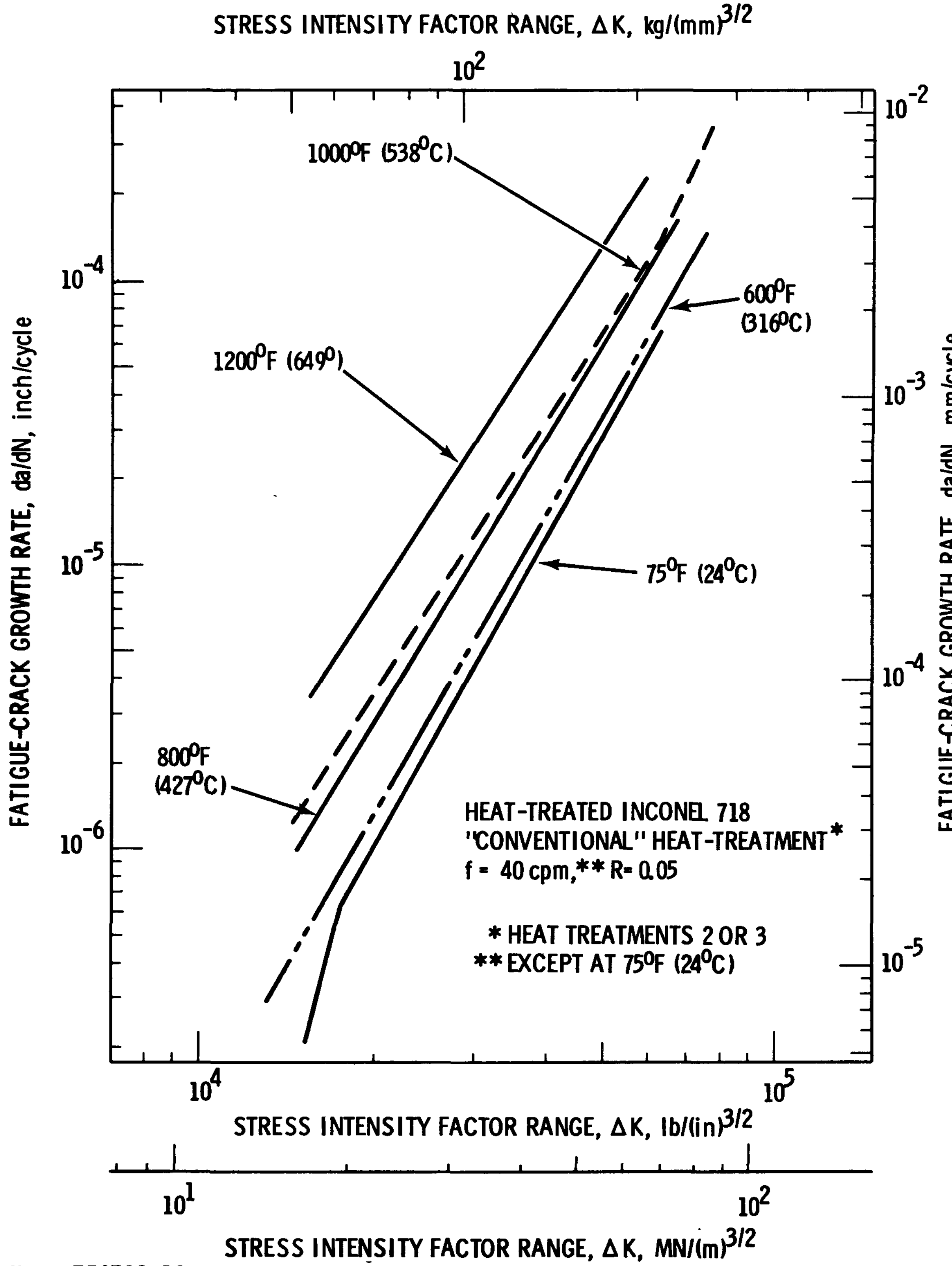

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FIGURE 6. The Effect of Temperature Upon the Fatigue-Crack Growth Behavior of "Conventionally" Heat-Treated Inconel 718 Tested in an Air Environment.

systems including aluminum alloys[17], ferritic steels[18], austenitic steels[14,19], titanium alloys[20], nickel alloys[13], cobalt alloys[21] and zirconium alloys[22]. As will be discussed later, there are reasons to believe[11,23] that a large percentage of the increase in crack growth rate with increasing temperature is due more to the aggressive air environment than to a creep effect.

B. EFFECT OF STRESS RATIO

The cyclic stress ratio ($R = \sigma_{min}/\sigma_{max}$) can have a significant effect upon the fatigue-crack growth behavior[24,25]. In general, at a given value of ΔK crack growth rates increase with increasing values of R and therefore the stress intensity factor range (ΔK) is not the best parameter to use in an analysis if the stress ratio in service differs appreciably from that used in the laboratory to generate the crack growth data. It has been shown[24,25] that the "effective" stress intensity factor (K_{eff})

$$K_{eff} = K_{max} [1-R]^m \qquad \text{Eq. [3]}$$

produces an excellent correlation of crack growth rates over a wide range of stress ratios. The exponent "m" in Equation [3] is empirically determined and is dependent upon both the material and the test temperature[25]. Limited results also suggest that "m" is independent of the test environment[25].

The use of K_{eff} to correlate the crack growth behavior of heat-treated Inconel 718 (heat-treatment 3) is shown in Figure 7 for the range $0.05 > R > 0.50$ at a test temperature of 800°F (427°C), and in Figure 8 for the range $0.05 > R > 0.667$ at a test temperature of 1000°F (538°C). These data were previously reported in Reference 25. It will be noted that the value of "m" at 800°F (427°C) is 0.5, and the value of "m" at 1000°F (427°C) is 0.6.

Coles et al[8] have also investigated the effect of stress ratio upon crack growth behavior of Inconel 718 at an elevated temperature, and their results are shown in Figure 9. Their tests were conducted at a temperature of 940°F (504°C), and they obtained a value of m = 0.55 which agrees well with the present results of m = 0.5 and m = 0.6 for test temperatures of 800°F (427°C) and 1000°F (538°C), respectively. The results of Figure 9 contain

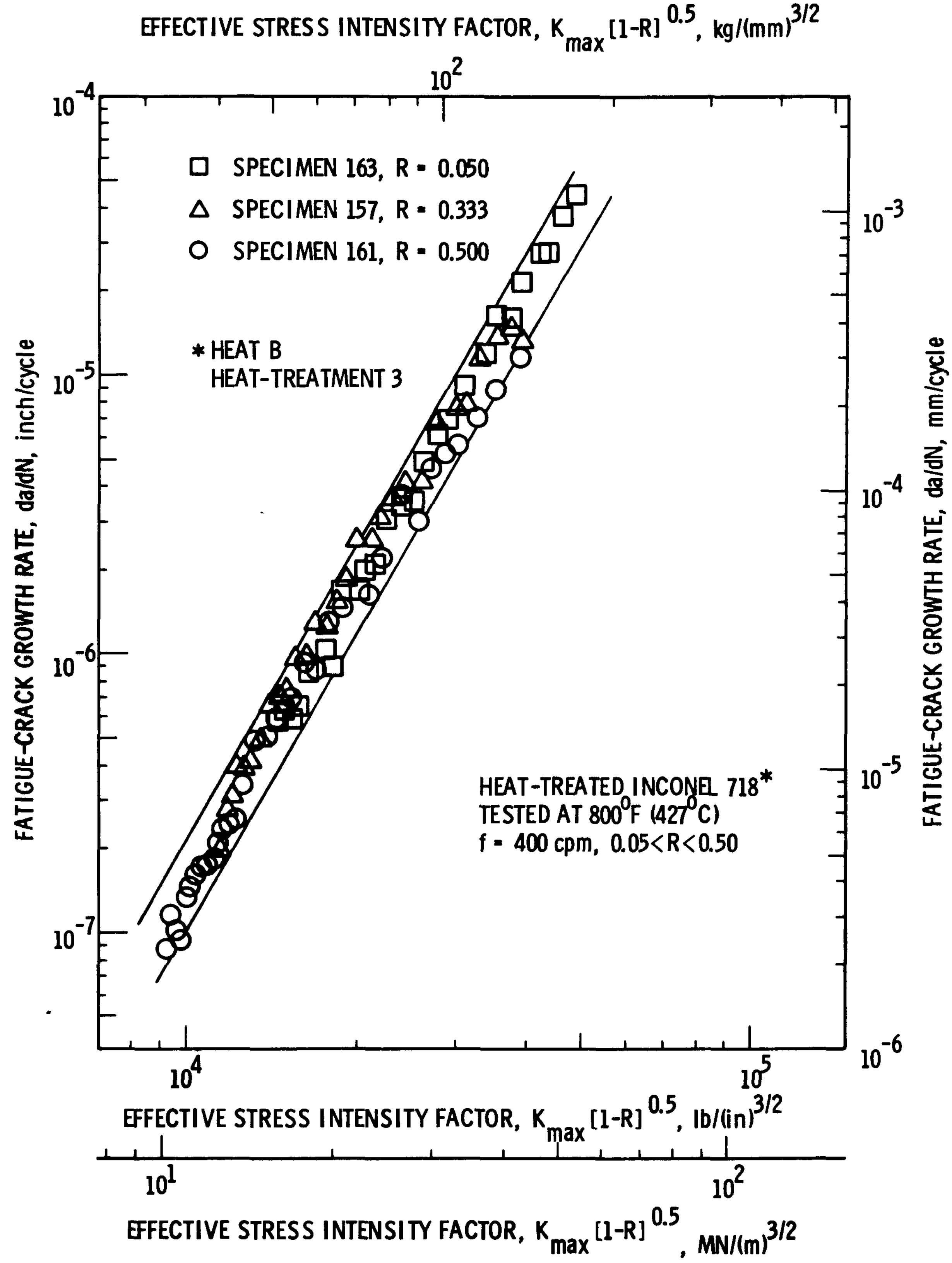

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FIGURE 7. Use of the Effective Stress Intensity Factor to Correlate the Crack Growth Behavior of Inconel 718 Over a Range of Stress Ratios at 800°F (427°C).

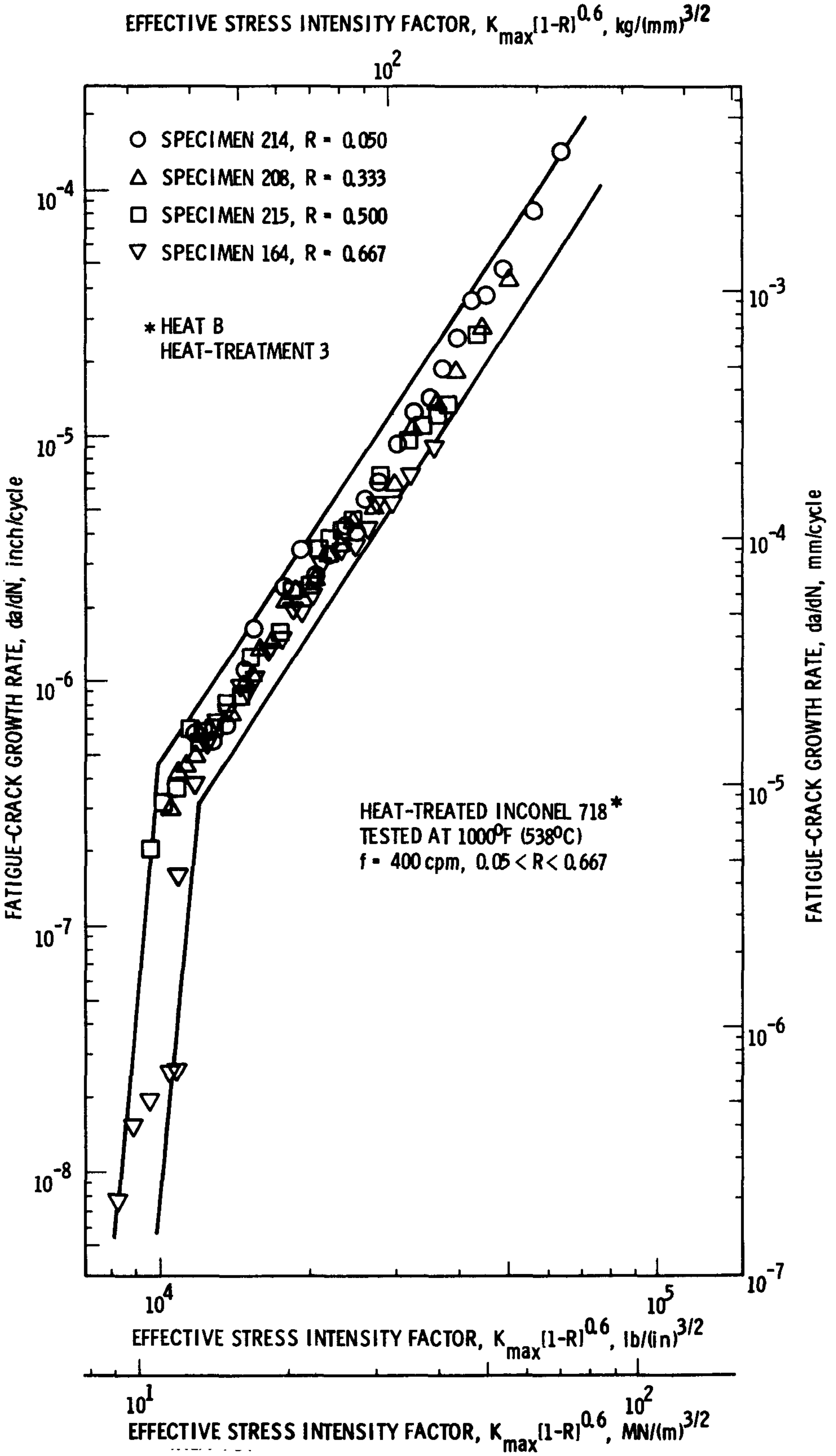

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FIGURE 8. Use of the Effective Stress Intensity Factor to Correlate the Crack Growth Behavior of Inconel 718 Over a Range of Stress Ratios at 1000°F (538°C).

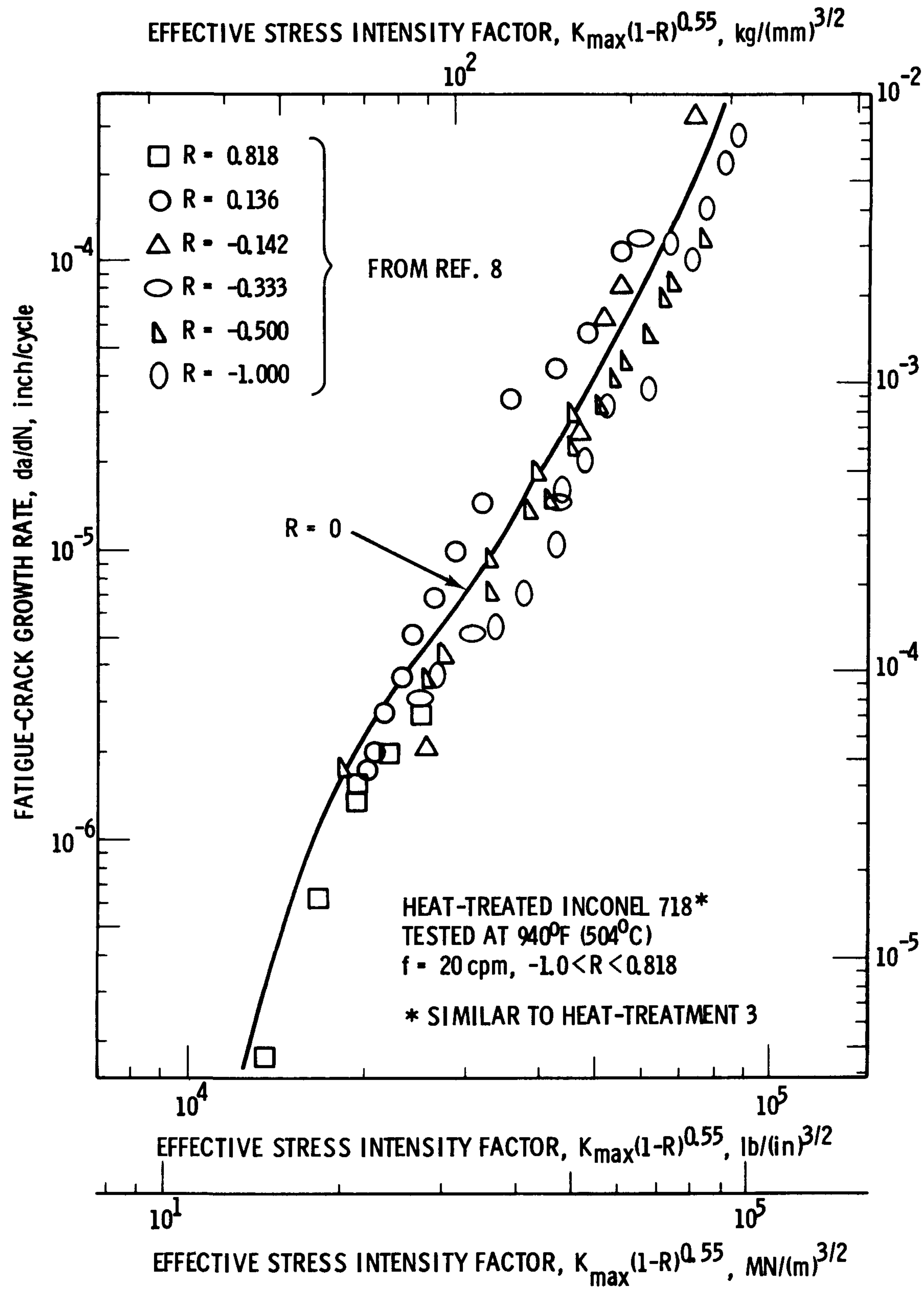

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FIGURE 9. Use of the Effective Stress Intensity Factor to Correlate the Crack Growth Behavior of Inconel 718 Over a Range of Stress Ratios at 940°F (504°C). From Reference 8.

data for negative values of R (i.e. tests in which the minimum cyclic stress was compressive), and it should be noted that the use of K_{eff} continues to correlate the crack growth rates down to a value of R = -1.0 (complete stress reversal). The effect of negative stress on fatigue-crack growth behavior is not yet fully understood and results are sometimes contradictory. For example, in contrast to the results of Figure 9, Hudson and Scardina(26) found that for an aluminum alloy tested at room temperature over a similar range of stress ratios (-1.0 > R > 0.8), the tests conducted at negative stress ratios produced essentially the same results as those conducted at R = 0. Therefore, under conditions such as those observed by Hudson and Scardina(26), negative values of R would be treated as R = 0 in the formulation of Equation [3]. However, in view of the results of Coles et al.(8) showing that negative values of R can influence crack growth behavior in Inconel 718 at at least one temperature, it is advisable to use the negative values in the calculation of K_{eff} for applications at any temperature involving this material. In fact, use of K_{eff} is to be preferred over use of ΔK in structural analyses since it has been shown(1) that use of ΔK can lead to unconservative results.

C. EFFECT OF CYCLIC FREQUENCY

Previous studies of several alloy systems including ferritic steels(27), austenitic steels(28,29) and cobalt-base alloys(21), have shown that in general, when fatigue-crack growth tests are conducted in an air environment at elevated temperatures, the fatigue-crack growth rate increases with decreasing cyclic frequency. In order to survey such effects in Inconel 718, tests were conducted over a limited frequency range at 800°F (427°C) and over a somewhat greater frequency range at 1000°F (538°C). These results (Figures 10 and 11) show that in an air environment at both temperatures, the fatigue-crack growth rate increases with decreasing frequency. However, with the exception of the test conducted at 0.083 cpm (1.39×10^{-3} Hz) at 1000°F (538°C), the effect of frequency at the higher frequencies (f > 4 cpm) is not large and is only slightly greater than the data scatter. The relative increase in the crack growth rate at 1000°F (538°C) for Inconel 718 between the frequencies of 400 cpm and 0.083 cpm is slightly less than that noted for annealed Type 304 stainless steel(28) over the same frequency range in an air environment at the

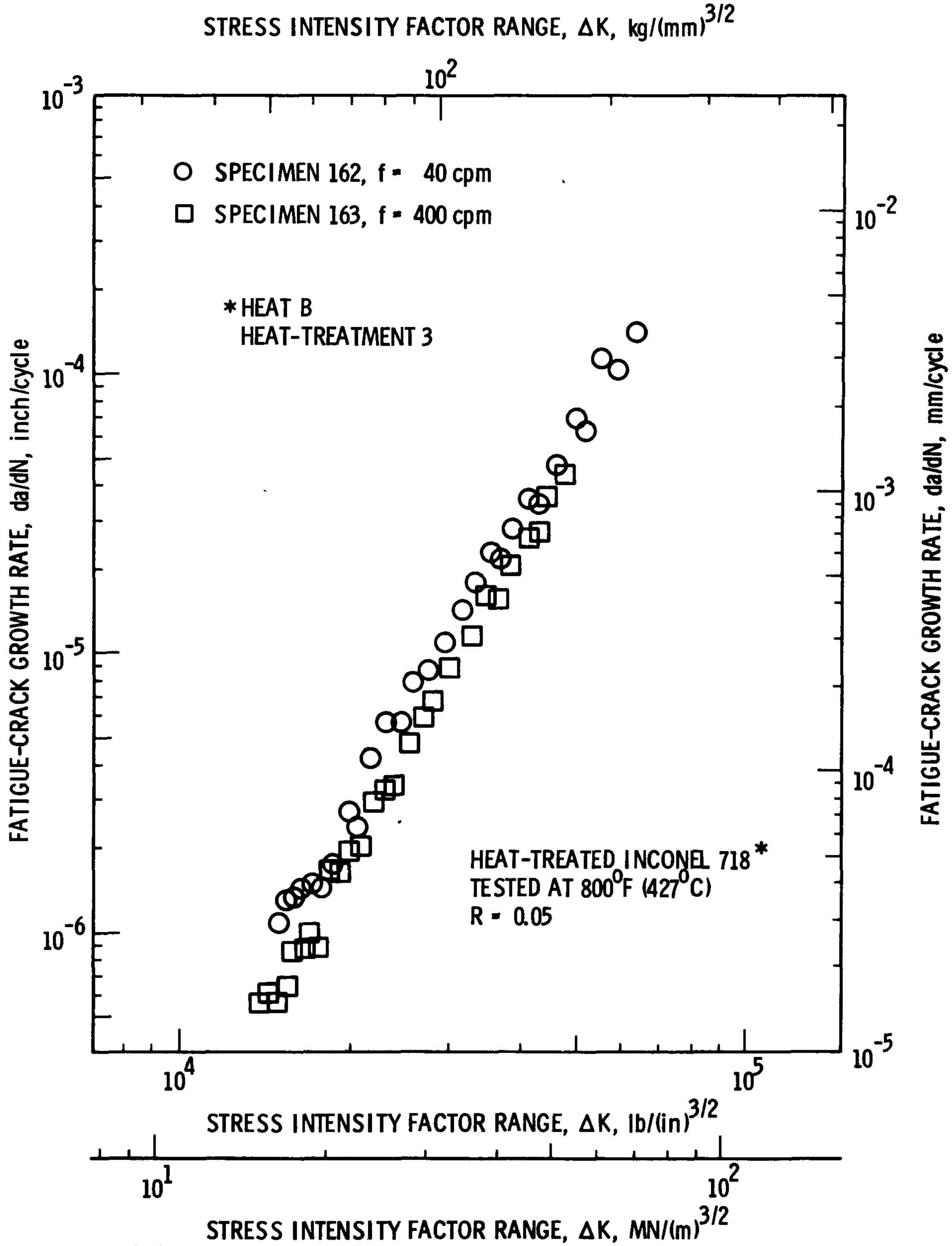

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FIGURE 10. The Effect of Cyclic Frequency Upon the Crack Growth Behavior of Inconel 718 Tested in an Air Environment at 800°F (427°C).

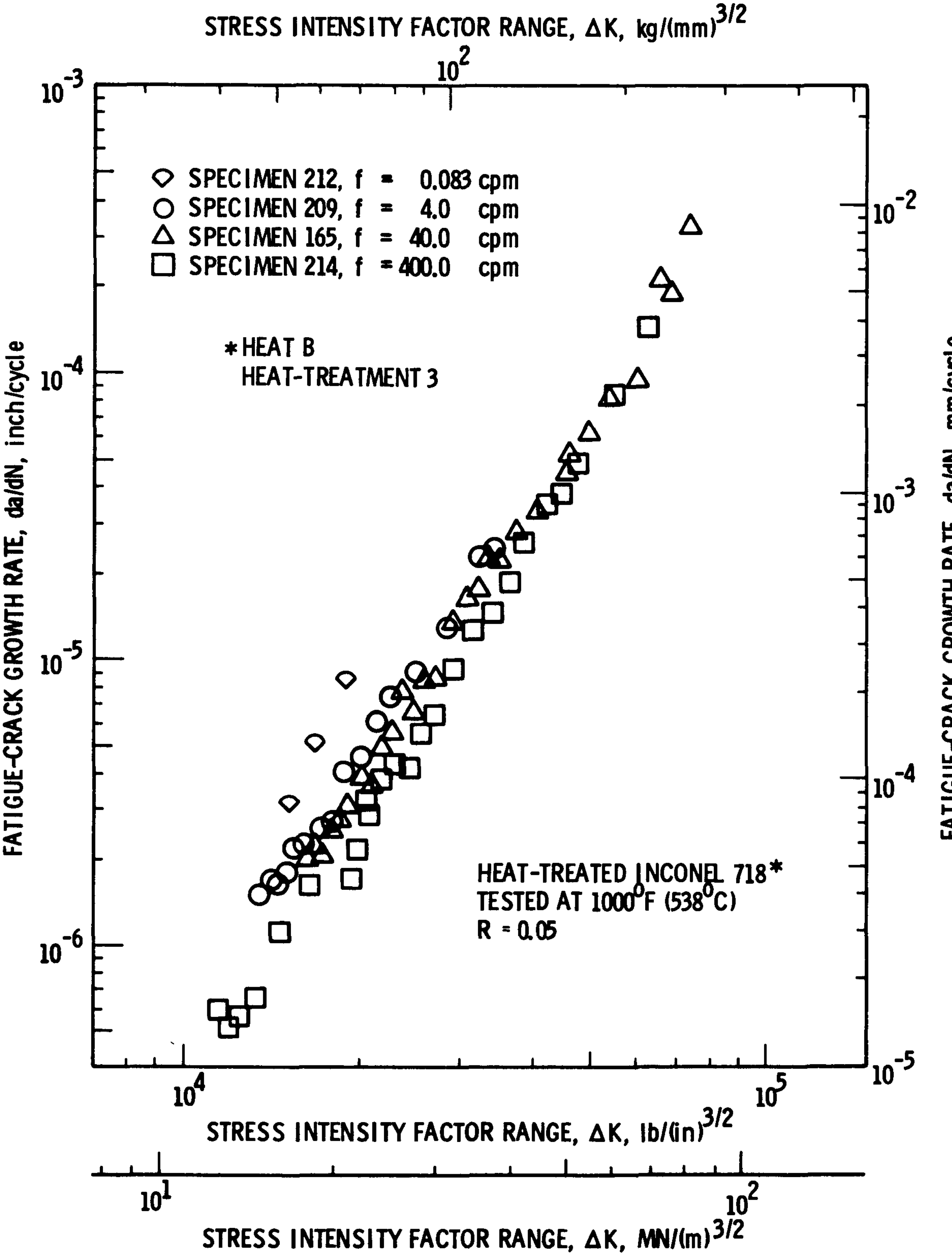

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FIGURE 11. The Effect of Cyclic Frequency Upon the Crack Growth Behavior of Inconel 718 Tested in an Air Environment at 1000°F (538°C).

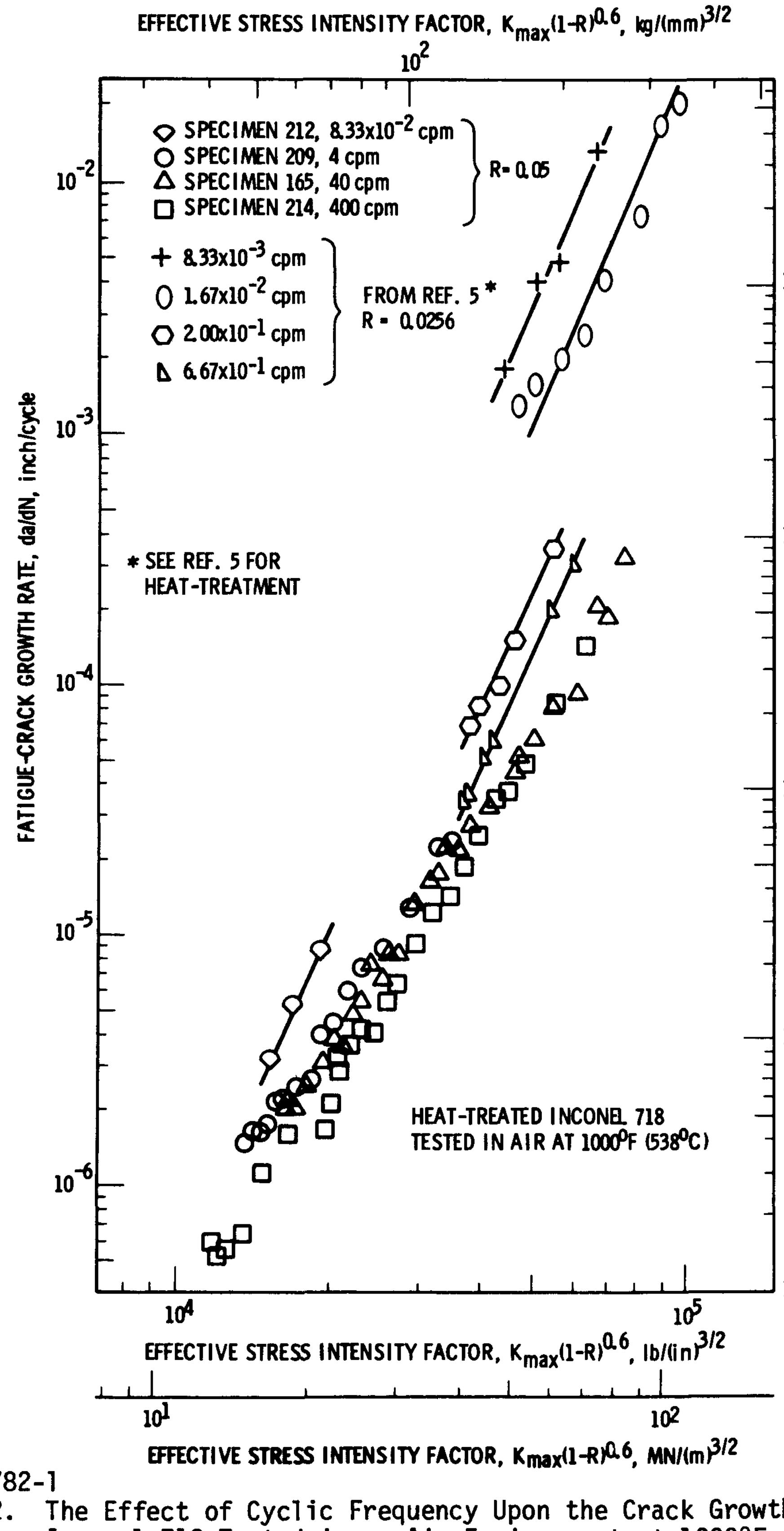
same temperature.

Popp and Coles[5] have also examined the effect of frequency in an air environment at 1000°F (538°C), and their results are shown in Figure 12. The results from Figure 11 are also plotted in Figure 12 for comparison purposes, and because of the slight difference in stress ratios between the two sets of data, the results are plotted in terms of K_{eff} (see the previous section). In general, the agreement between the two sets of data is quite good. Popp and Coles[5] used a "square" loading waveform with a tensile hold-time, while the present results were generated using either a sinusoidal or a "sawtooth" waveform. It has been shown that, for Type 304 stainless steel tested in air at 1000°F (538°C) over the range 0.083 > f > 4 cpm, that the "square" wave and "sawtooth" wave produced essentially the same fatigue-crack growth behavior[30]. Although the data is quite limited and extrapolation relating to waveform should be viewed with caution, the same conclusion regarding the lack of a significant waveform effect might be reached by extending the trend line from Specimen 212 in Figure 12 up into the region covered by Reference 5. The trend line for Specimen 212 (at 8.33×10^{-2} cpm) would fall between the data from Reference 5 for 1.67×10^{-2} cpm and 2.00×10^{-1} cpm although, of course, the waveforms were quite different.

Popp and Coles[5] also conducted one test where a crack extended under static loadings in an air environment at 1000°F (538°C). They were able to describe the crack extension per unit time (da/dt) as a function of the stress intensity factor. This data is shown in Figure 13, along with the fatigue data from Figure 12 for comparison purposes. The fatigue-crack growth rates were converted to a time-base using

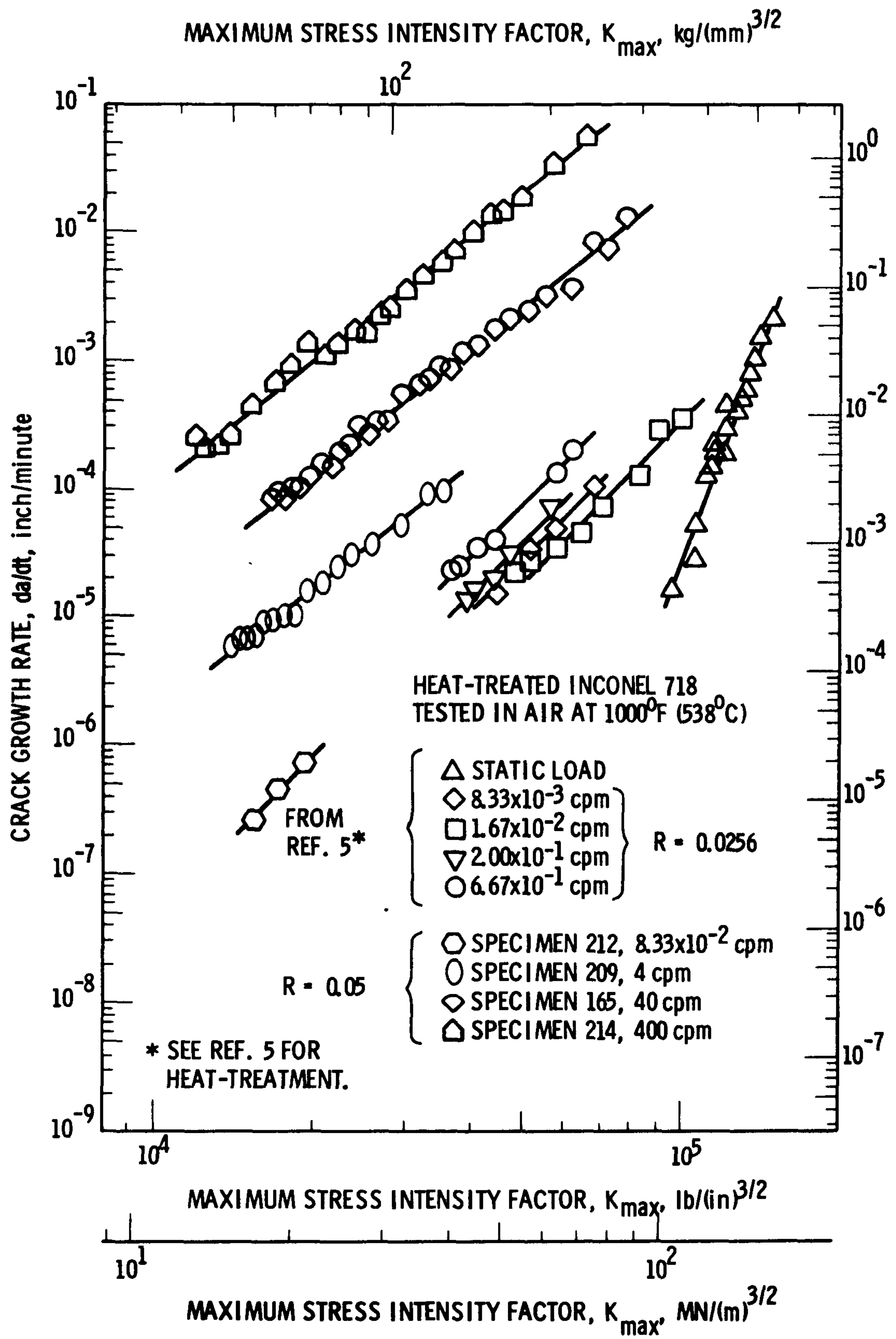
$$\frac{da}{dt} = \left(\frac{da}{dN}\right) \text{(frequency)} \qquad \text{Eq. [4]}$$

and Equation [3] was used to convert K_{eff} to K_{max}. The trends observed in Figure 13 are similar to those observed earlier[31] in comparing crack growth under static and cyclic loadings in cold-worked Type 316 stainless steel tested in an air environment at 1000°F (538°C).



Neg. 754782-1

FIGURE 12. The Effect of Cyclic Frequency Upon the Crack Growth Behavior of Inconel 718 Tested in an Air Environment at 1000°F (538°C).



Neg. 754782-4

FIGURE 13. Time-Base Crack Growth Rates of Inconel 718 Tested in an Air Environment at 1000°F (538°C).

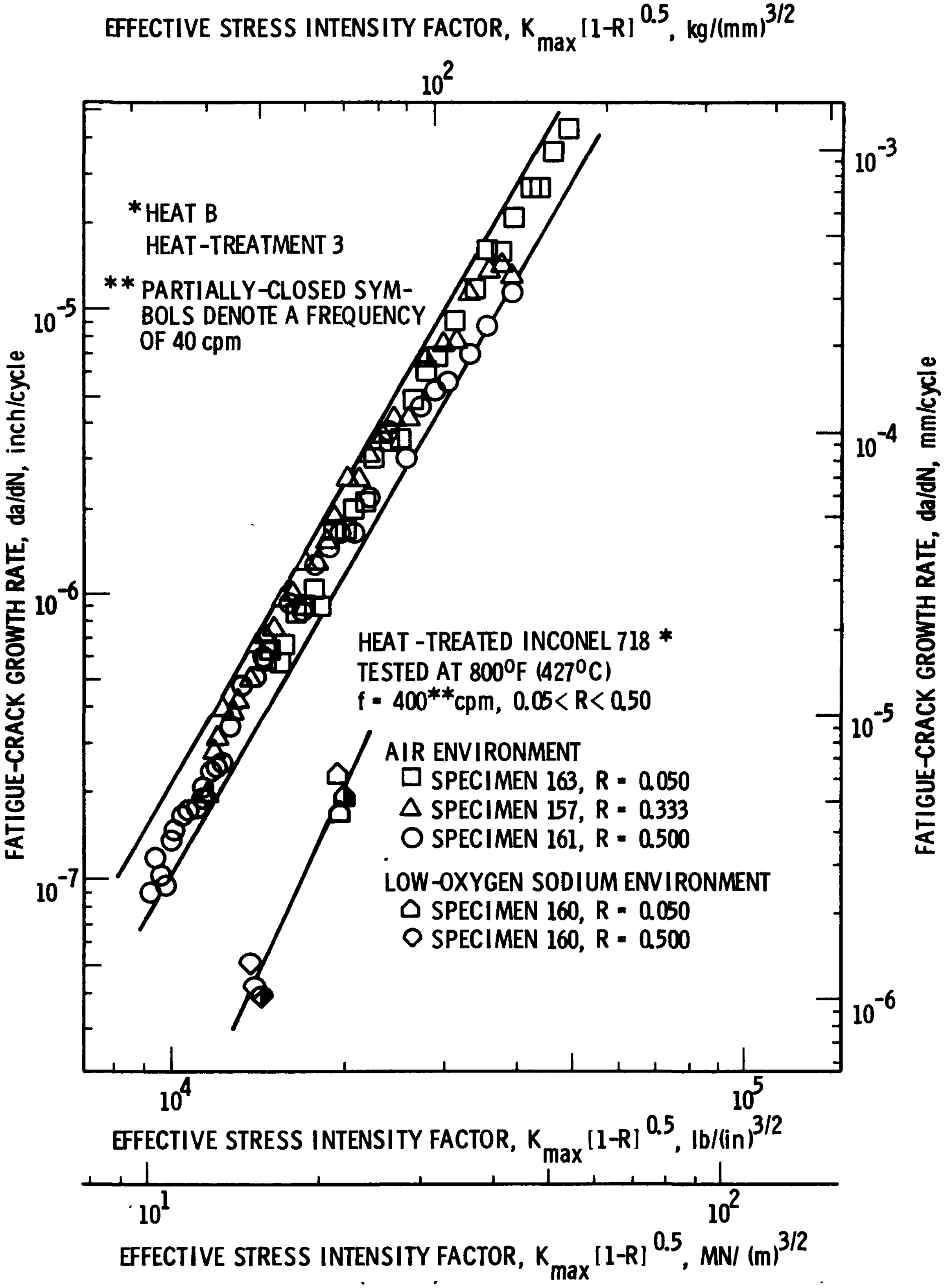
D. EFFECT OF ENVIRONMENT

The environment surrounding the test specimen or the structural component can have a profound influence upon the fatigue-crack propagation behavior, and this appears to be especially true at elevated temperatures[23]. Air can be a relatively aggressive environment at elevated temperatures, and decreases in the fatigue-crack growth rate of approximately an order of magnitude or more relative to the behavior in air have been noted at elevated temperatures when austenitic stainless steels were tested in the relatively inert environments of argon[32], nitrogen[32,33], liquid sodium[11], or vacuum[11,34].

One test has been conducted at HEDL on Inconel 718 in a low-oxygen (1-3 ppm oxygen) sodium environment at 800°F (427°C), and the results are shown in Figure 14. Reference 11 should be consulted for a description of the sodium testing apparatus and testing procedures. At a test temperature of 800°F (427°C), the fatigue-crack growth rate in a sodium environment is approximately an order of magnitude lower than in an air environment under similar conditions. This is in agreement with previous results[11] for annealed Type 304 stainless steel tested in a sodium environment. Reference 11 also showed that, insofar as the crack growth behavior of Type 304 S.S. is concerned, sodium environments at 800°F (427°C) and 1000°F (538°C) were approximately as inert as a vacuum (6.7 - 8.6 x 10^{-6} torr) at the same temperature.

The data shown in Figure 14 was generated using a cyclic frequency of 400 cpm, with the exception of a limited amount of data at 40 cpm in the sodium environment. This limited data suggests that there was no observable effect of frequency over this very narrow frequency range. It should be noted, however, that there is only a very small frequency effect over the same frequency range in an air environment at 800°F (427°C) (e.g. see Figure 10).

The data from the specimen tested in sodium represents a relatively narrow range of values of K_{eff}. This is because the crack grew at a rate lower than expected, and hence the growth increments upon which the crack growth rates were based were smaller than optimum (see Reference 11 for a discussion on crack growth calculation methods). Therefore, although the data points for



Neg. 754782-26

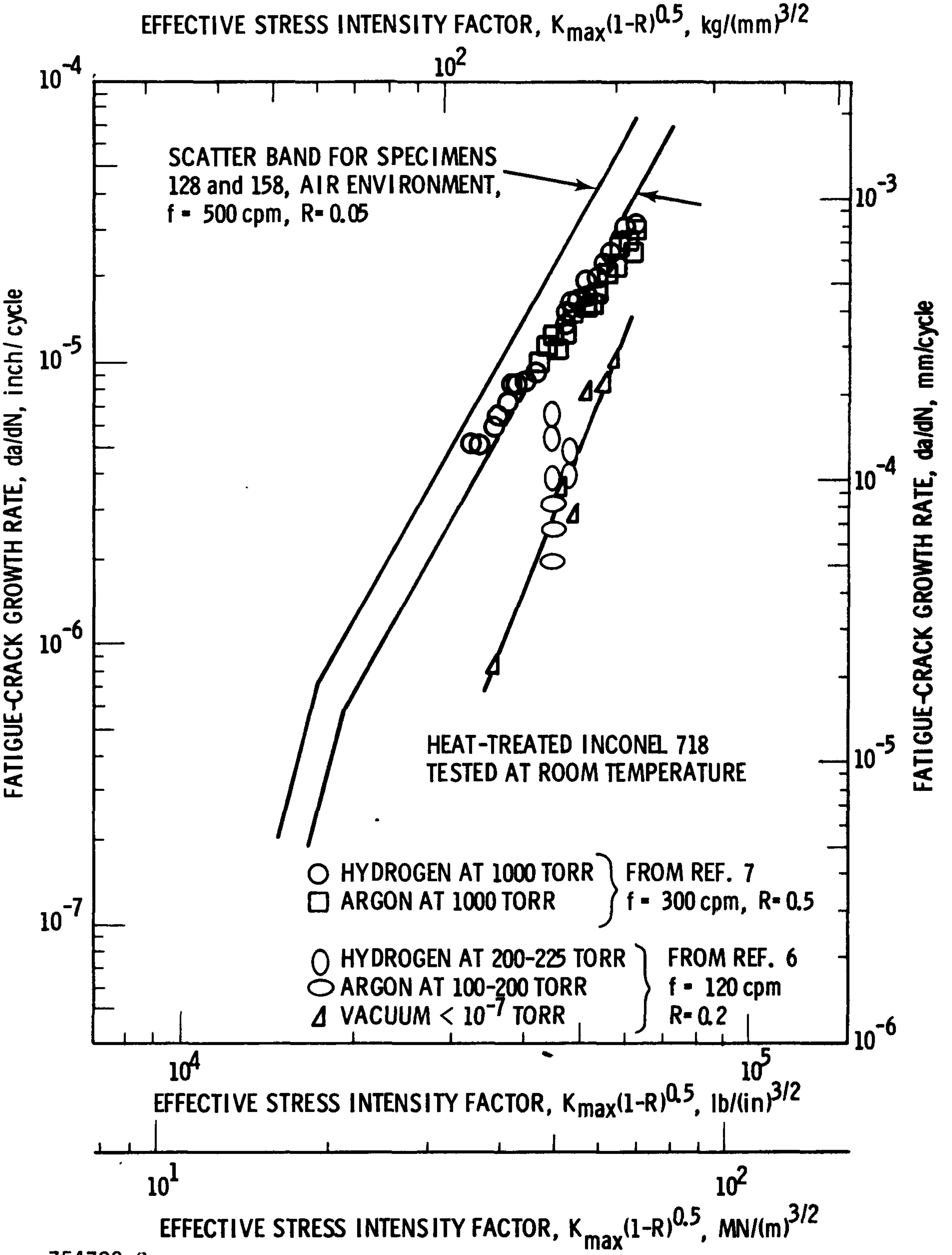
FIGURE 14. Comparison of the Crack Growth Behavior of Inconel 718 Tested in Air at 800°F (427°C) and in Sodium at 800°F (427°C).

Specimen 160 in Figure 14 do not exhibit excessive scatter and are considered valid, some errors may have been introduced due to the relatively small growth increments.

Fatigue-crack growth tests have been conducted on Inconel 718 in gaseous environments at other laboratories. For example, Wei et al [7] have studied the crack growth behavior at room temperature in argon and hydrogen environments, and Frandsen et al[6] have tested in argon and hydrogen environments as well as <u>in vacuo</u>. Walter and Chandler[39] studied crack growth under both static and cyclic loads at temperatures ranging from -100°F (-73°C) to room temperature in a hydrogen environment at pressures up to 10,000 psig. The results from References 6 and 7 as well as results from the present study are shown in Figure 15*. The results of Frandsen et al[6] indicate that the crack growth <u>in vacuo</u> at room temperature is considerably lower than those observed in an air environment in the present study. There is, however, an apparent discrepency between the results of References 6 and 7 for argon and hydrogen environments, Wei et al[7] observed identical crack growth behavior between argon and hydrogen environments, and these, in turn, are essentially the same as that observed in the present study in air. Frandsen et al[6], on the other hand, observed an increase in crack growth rate in a hydrogen environment (relative to <u>in vacuo</u>) of factors of 1.4-2.3 at K_{max} = 48,750 psi $\sqrt{in}$ and factors of 1.0 - 1.2 at K_{max} = 52,500 psi $\sqrt{in}$. The tests at the two different levels of K_{max} were conducted on specimens which had received different heat-treatments. Frandsen et al[6] observed normalized crack growth rates in an argon environment of factors of 0.7 to 1.1, relative to <u>in vacuo</u> at K_{max} = 48,750 psi $\sqrt{in}$.. Walter and Chandler[39] found crack growth rates increased by a factor of up to 4 in a room temperature hydrogen environment (5000 psig), relative to a helium environment at the same temperature and pressure. They found that, in general, in a hydrogen environment crack growth rates increased with increasing

---

* Because of the different stress ratios between References 6 and 7, the data are plotted as a function of K_{eff} for comparison purposes. The value of "m" in Equation [3] has not been determined for Inconel 718 tested at room temperature, and is therefore assumed to be 0.5. The errors in such an assumption are small: for example, if "m" was actually 0.6 instead of 0.5, the error in K_{eff} would be 6.7% at R = 0.5, 2.2% at R = 0.2, and only 0.5% at R = 0.05.



Neg. 754782-9

FIGURE 15. Comparison of the Crack Growth Behavior of Inconel 718 in Air, Argon, Hydrogen and Vacuum Environments at Room Temperature.

pressure and/or decreasing cyclic frequency. The reasons for the differences between the observations of References 6, 7, and 39 are not known at this time, but may be related to the hydrogen pressure.

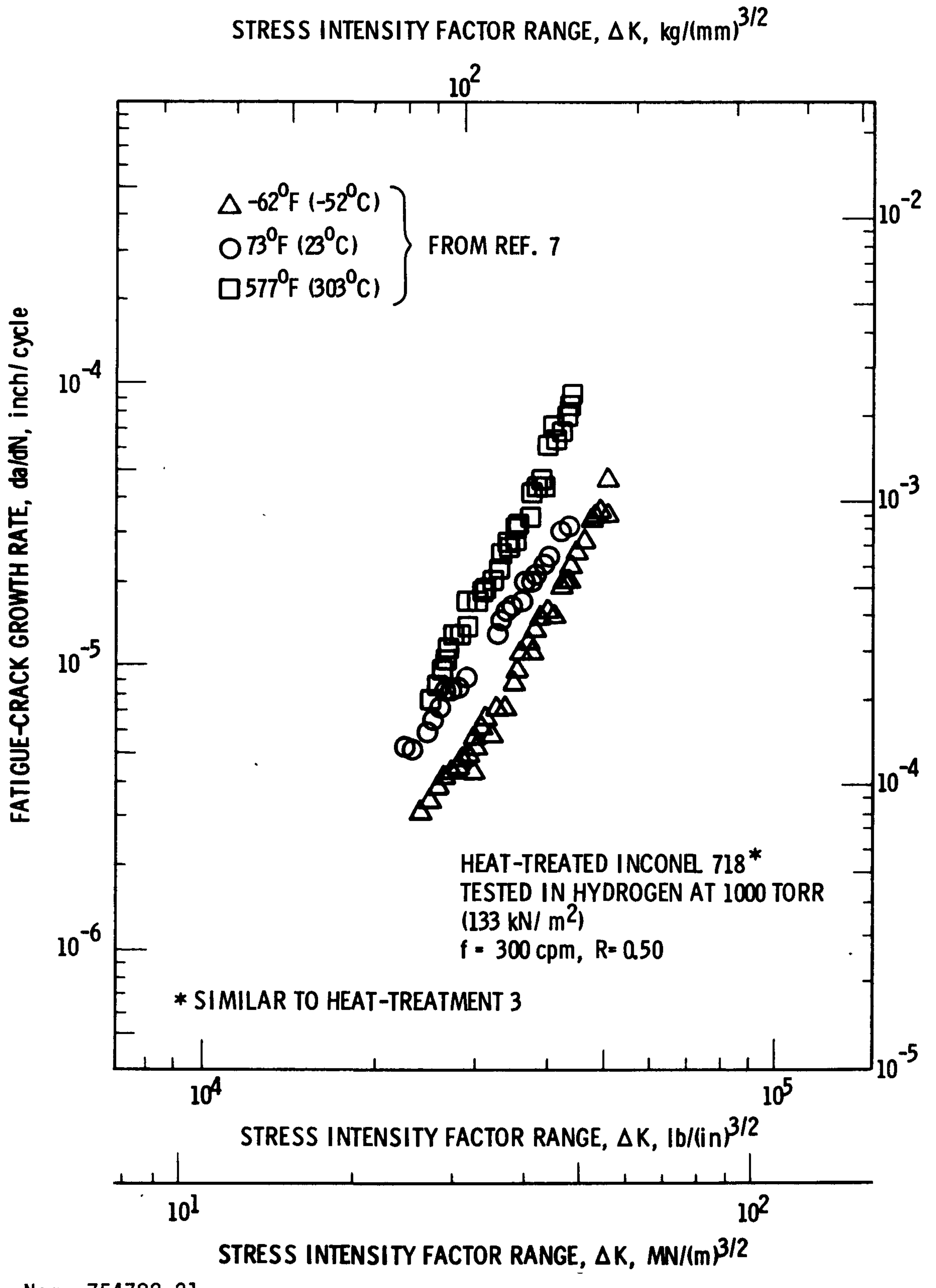
Wei et al [7] also studied the effect of temperature in a hydrogen environment at three test temperatures: -62°F (-52°C), 73°F (23°C) and 577°F (303°C). Their results are shown in Figure 16. These data show the same trend as previously noted in an air environment (e.g. see Figure 6): fatigue-crack growth rates increasing with increasing test temperature. However, comparison of the increase in an air environment between 75°F (24°C) and 600°F (316°C) in Figure 16 suggests that the effect of temperature may be slightly more pronounced in the hydrogen environment, especially at the higher values of ΔK.

E. EFFECT OF HEAT-TREATMENT

Until recently, the majority of Inconel 718 utilized in structural applications at temperatures below 1200°F (649°C) was precipitation heat-treated using the "conventional" treatment (i.e. Heat-Treatment 3 in Table II). However, it has been found[35] that weldments given the "conventional" heat-treatment following welding (i.e. Heat Treatment 5) did not display the desired level of ductility. Therefore, Aerojet Nuclear Company conducted an investigation to develop a more satisfactory post-weld heat-treatment[35], and developed what will be called in this report the "modified" heat-treatment (Heat-Treatment 7 in Table II). The corresponding "modified" heat-treatment for base metal would be Heat-Treatment 4 in Table II. The fatigue-crack growth behavior of the weldment specimens will be covered in the following section, and this section will deal with the effect of heat-treatment variations upon the crack growth behavior of base metal.

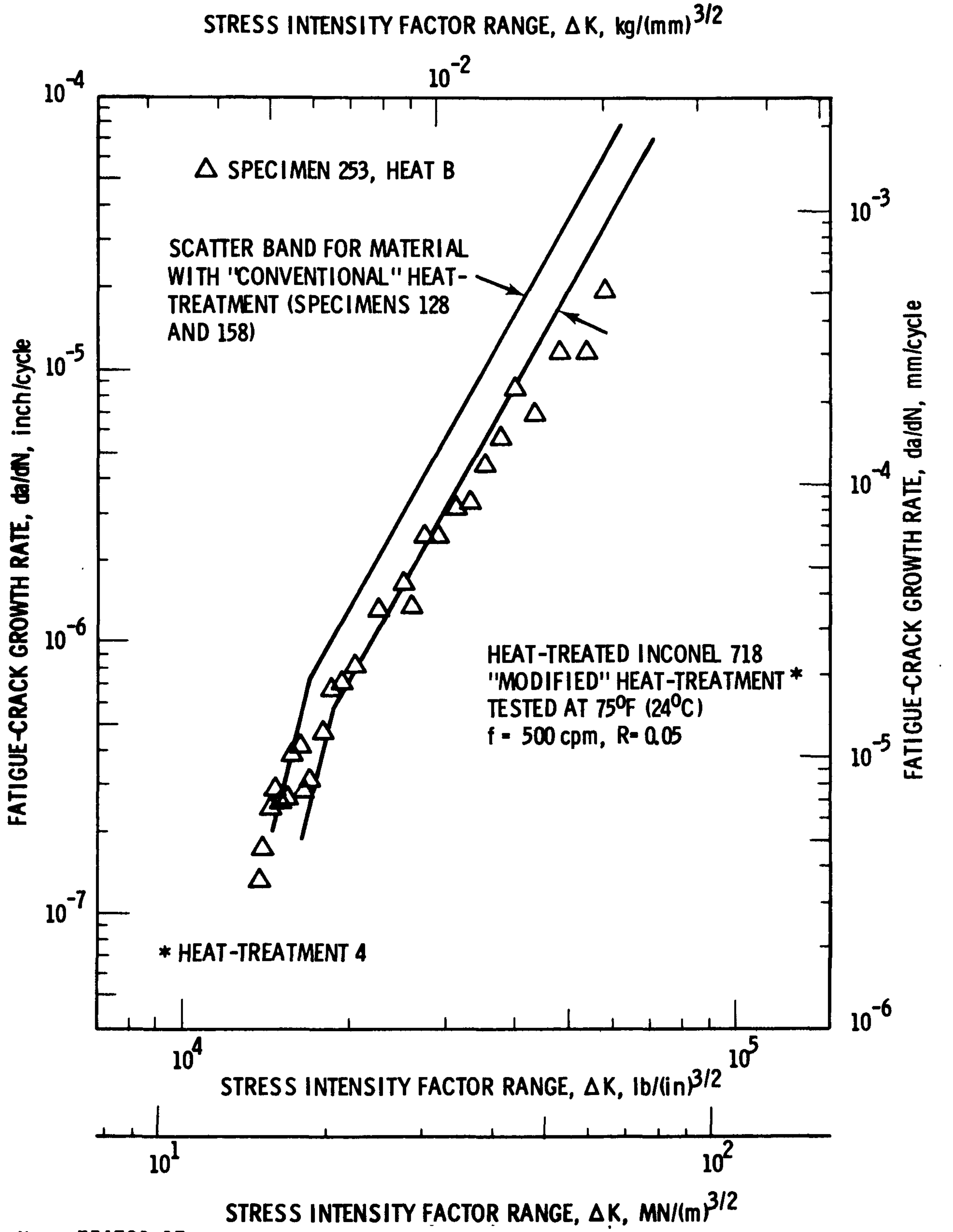
The fatigue-crack growth behavior of base-metal specimens receiving the "conventional" heat-treatment (i.e. Heat-Treatments 2 or 3) has already been shown in Figures 1-5. Heat-Treatments 2 and 3 vary only by the annealing temperature preceding precipitation heat-treatment, and no observable difference is noted in the crack growth behavior between these two treatments (see Figures 1, 2, and 4).

The fatigue-crack propagation behavior of base metal specimens receiving the "modified" heat-treatment was determined at test temperatures of 75°F



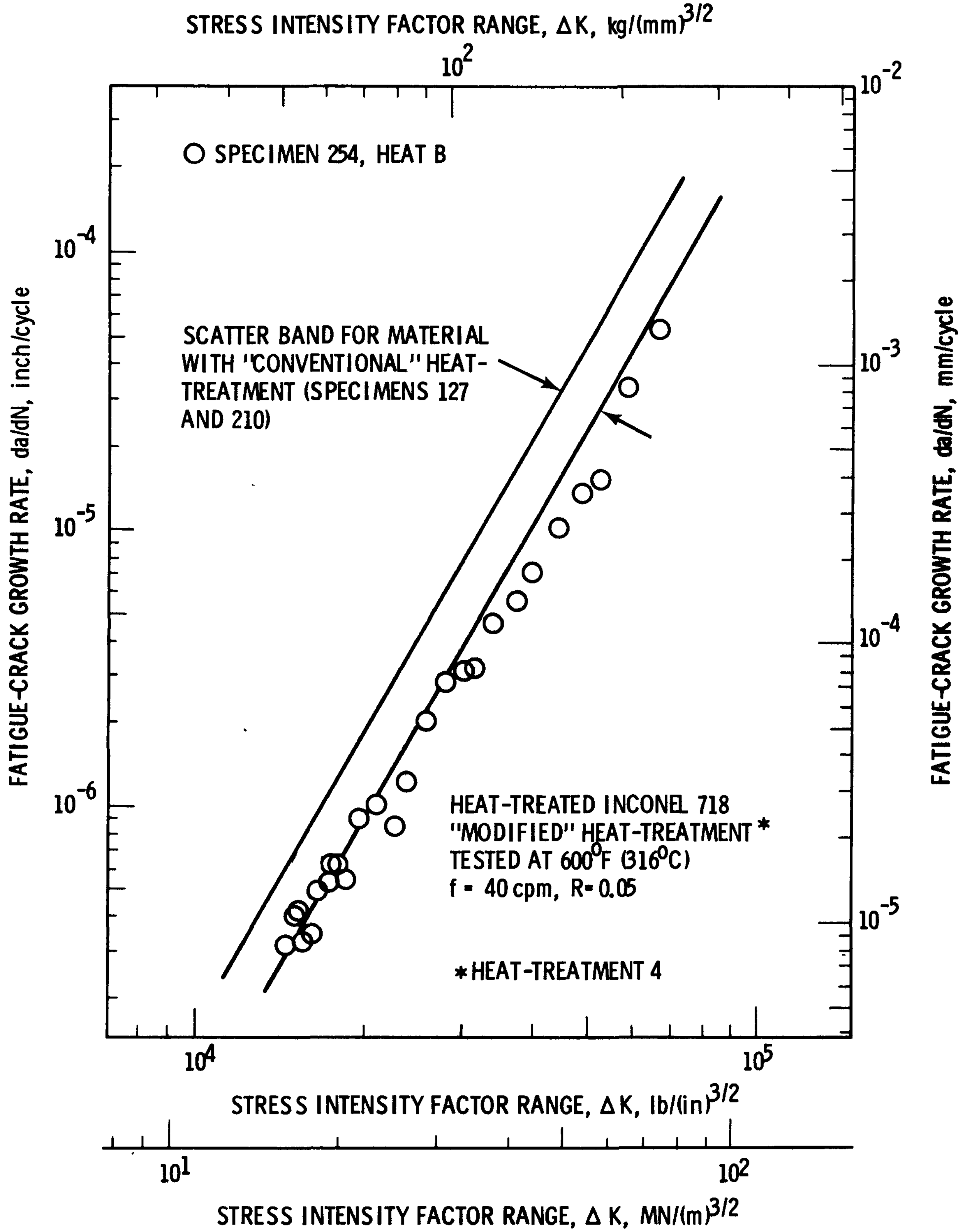
Neg. 754782-21

FIGURE 16. The Effect of Temperature Upon the Fatigue-Crack Growth Behavior of Inconel 718 Tested in a Hydrogen Environment. From Reference 7.



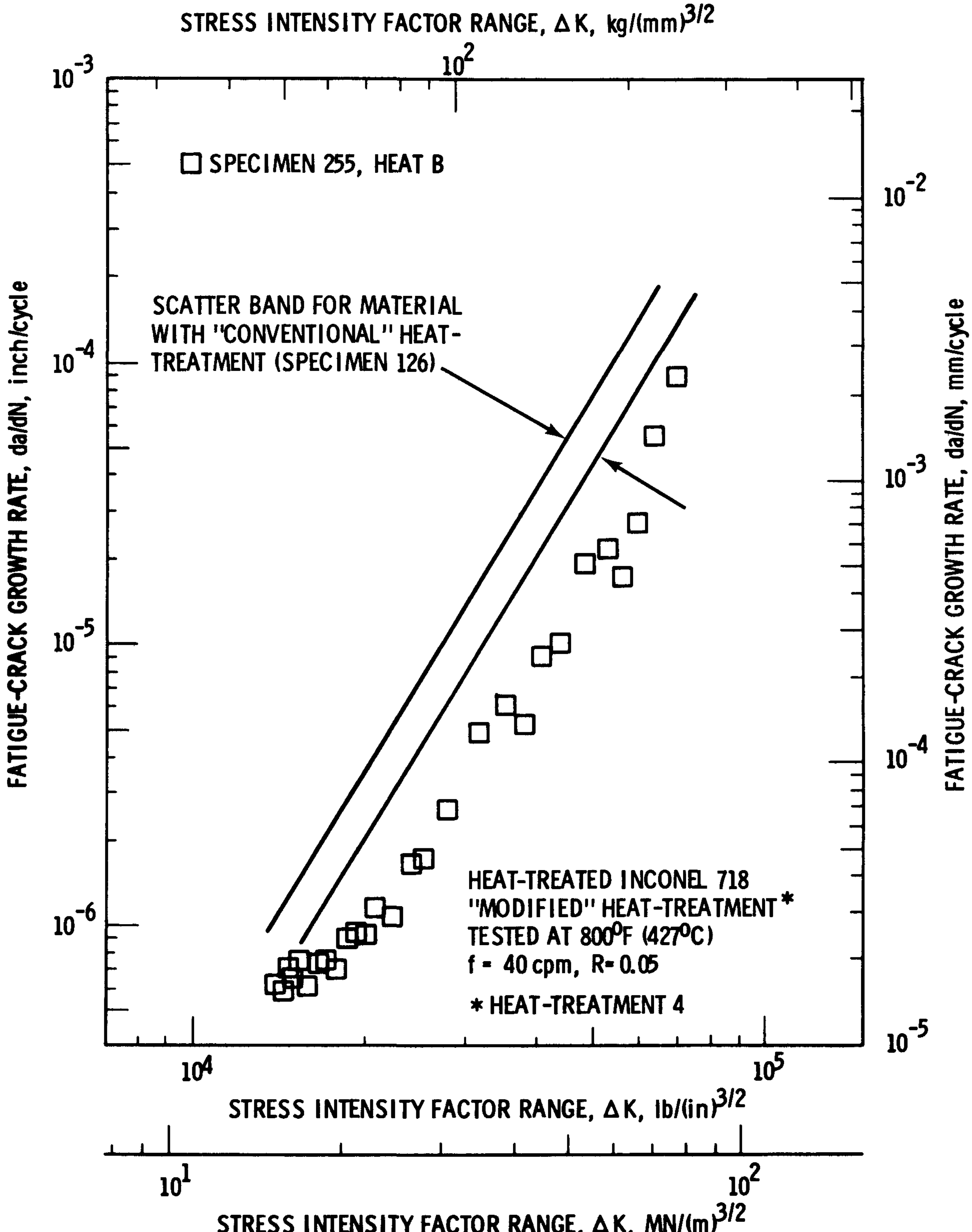
Neg. 754782-17

FIGURE 17. Fatigue-Crack Growth Behavior of "Modified" Heat-Treated Inconel 718 in an Air Environment at Room Temperature.



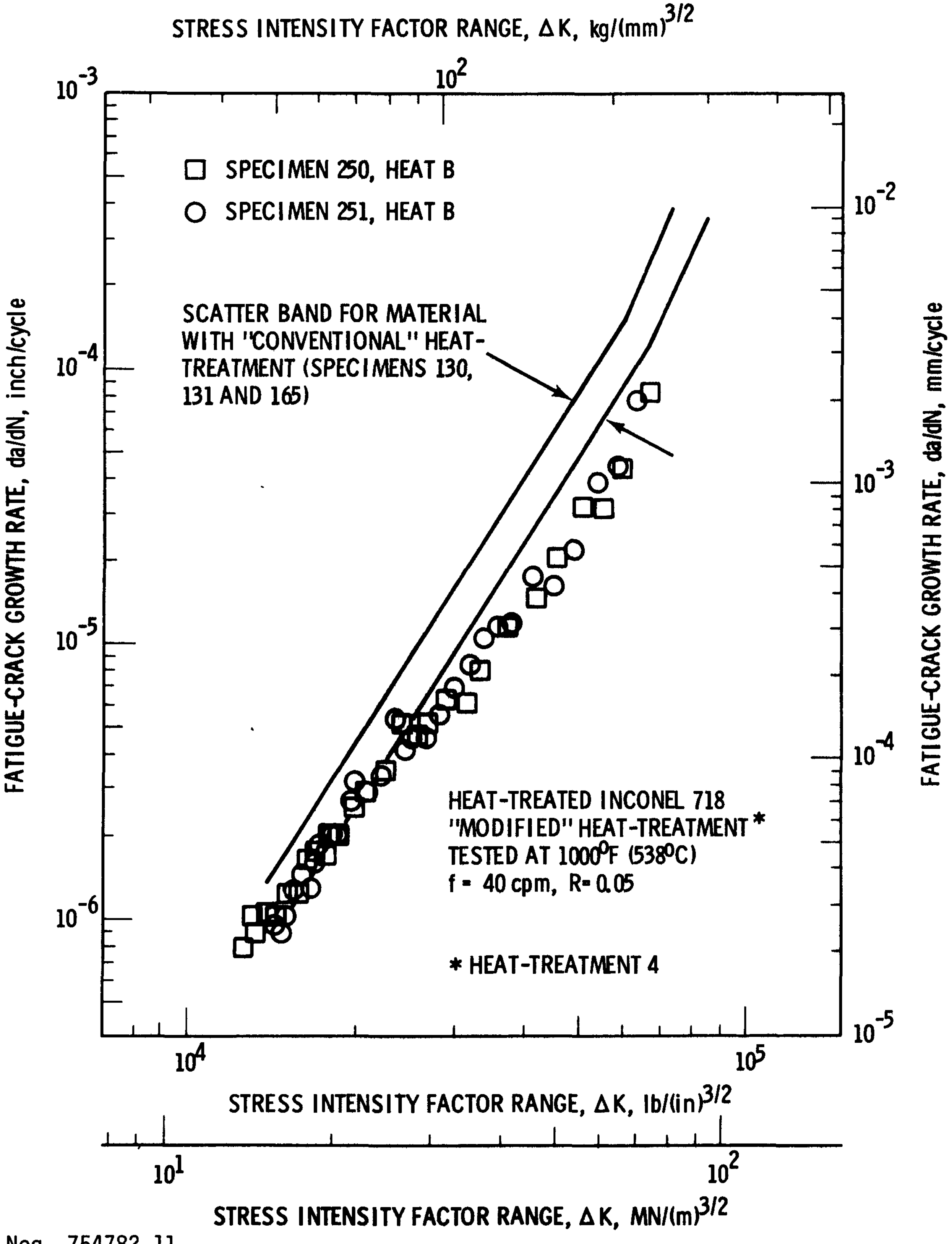
Neg. 754782-16

FIGURE 18. Fatigue-Crack Growth Behavior of "Modified" Heat-Treated Inconel 718 in an Air Environment at 600°F (316°C).



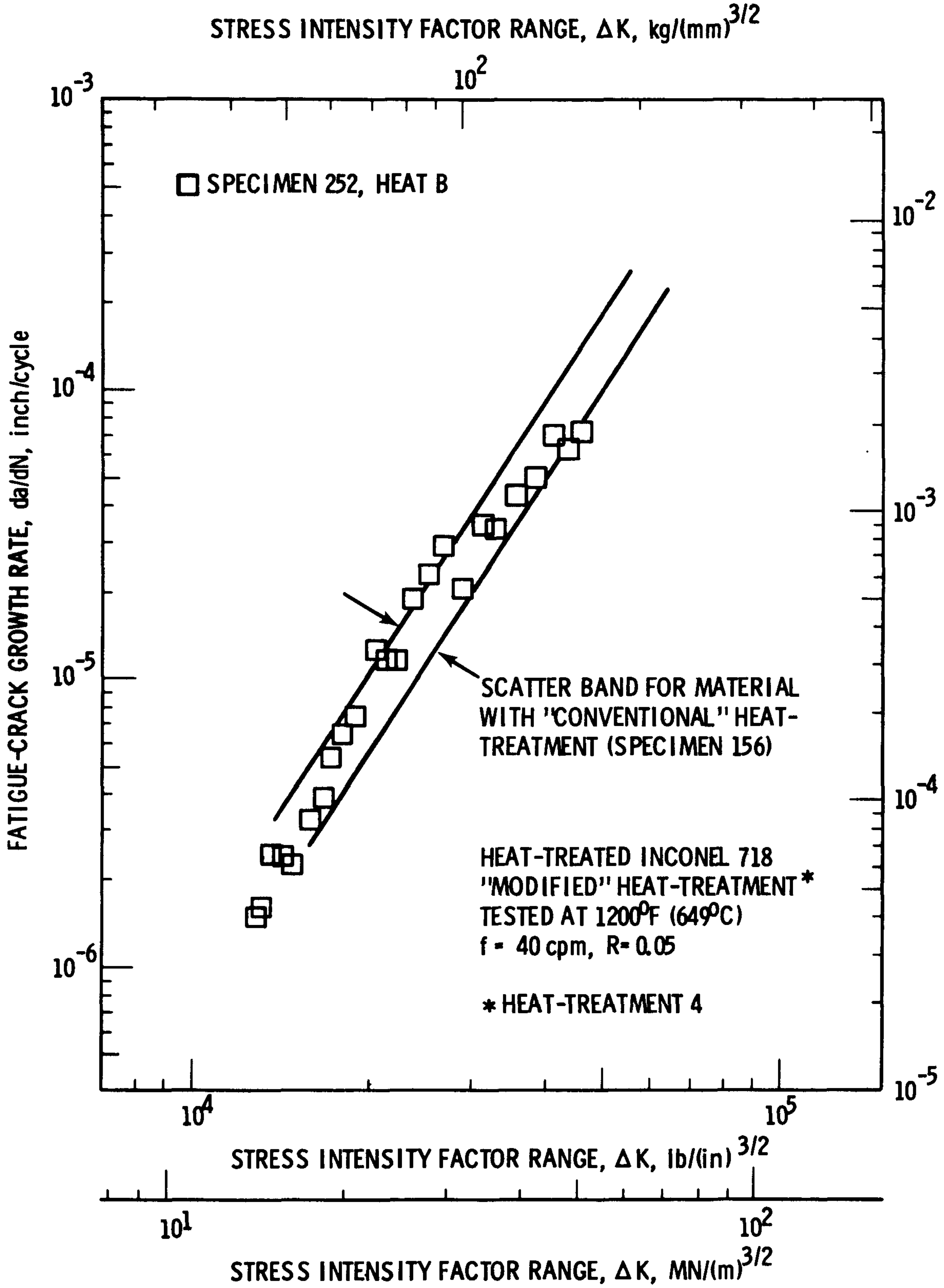
Neg. 754782-19

FIGURE 19. Fatigue-Crack Growth Behavior of "Modified" Heat-Treated Inconel 718 in an Air Environment at 800°F (427°C).



Neg. 754782-11

FIGURE 20. Fatigue-Crack Growth Behavior of "Modified" Heat-Treated Inconel 718 in an Air Environment at 1000°F (538°C).



Neg. 754782-8

FIGURE 21. Fatigue-Crack Growth Behavior of "Modified" Heat-Treated Inconel 718 in an Air Environment at 1200°F (649°C).

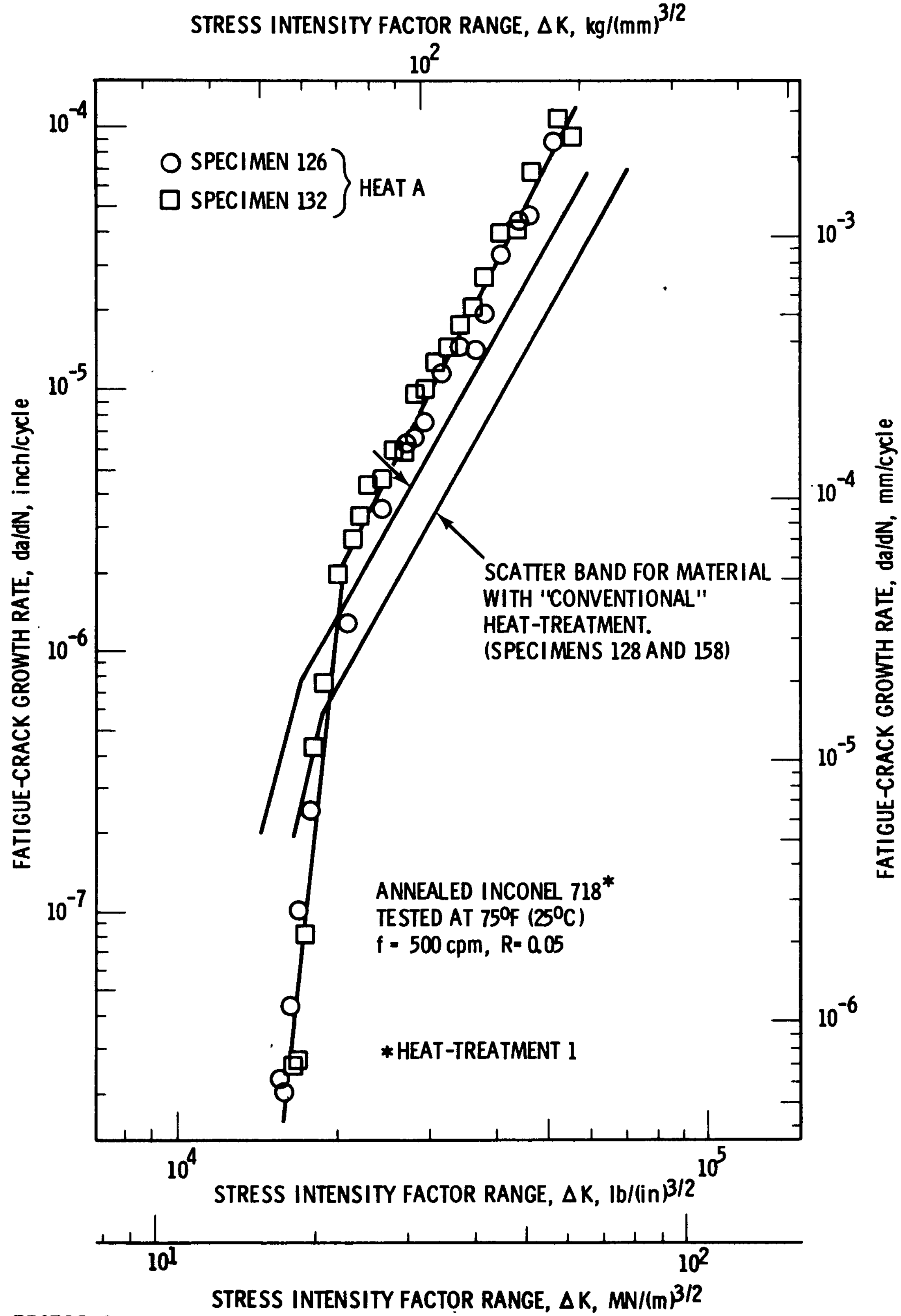
(24°C), 600°F (316°C), 800°F (427°C), 1000°F (538°C) and 1200°F (649°C), and the results are shown in Figures 17-21. Also shown, for comparison purposes, are scatter bands for the respective "conventionally" treated data from Figures 1-5. At test temperatures of 75°F (24°C) and 600°F (316°C), comparison of Figures 1 and 17, and 2 and 18, respectively, reveals that the "modified" heat-treatment appears to produce a slight improvement in crack growth behavior (i.e. lower growth rates) relative to the "conventional" treatment. However, Figures 19 and 20 show a greater improvement in "modified" treated material at test temperatures of 800°F (427°C) and 1000°F (538°C), respectively. At a test temperature of 1200°F (649°C), which is above the final precipitation aging temperature, the two heat-treatments produce essentially identical crack growth behavior (see Figure 21). Thus, the "modified" heat-treatment produces improvements (relative to the "conventional" treatment) in the crack growth behavior ranging from slight to moderate below the final precipitation aging temperature (1150°F or 621°C), and essentially identical behavior above this temperature.

A limited amount of testing was done on annealed (Heat-Treatment 1 in Table II) Inconel 718, and the results are shown in Figures 22-24. The scatter bands for "conventionally" treated material are also shown for comparison purposes, and it will be seen that at each of the test temperatures, the crack growth rates are higher in the annealed material than in the material receiving the "conventional" heat-treatment.

Three specimens which had originally received the "conventional" heat-treatment (Heat-Treatment 3) were then aged from 3000, 6000, and 12,000 hours at 1000°F (538°C) in an air environment to determine if additional long-time thermal aging had an influence on the fatigue-crack growth behavior. The results, shown in Figure 25, indicate that the additional aging had no effect upon the crack extension behavior.

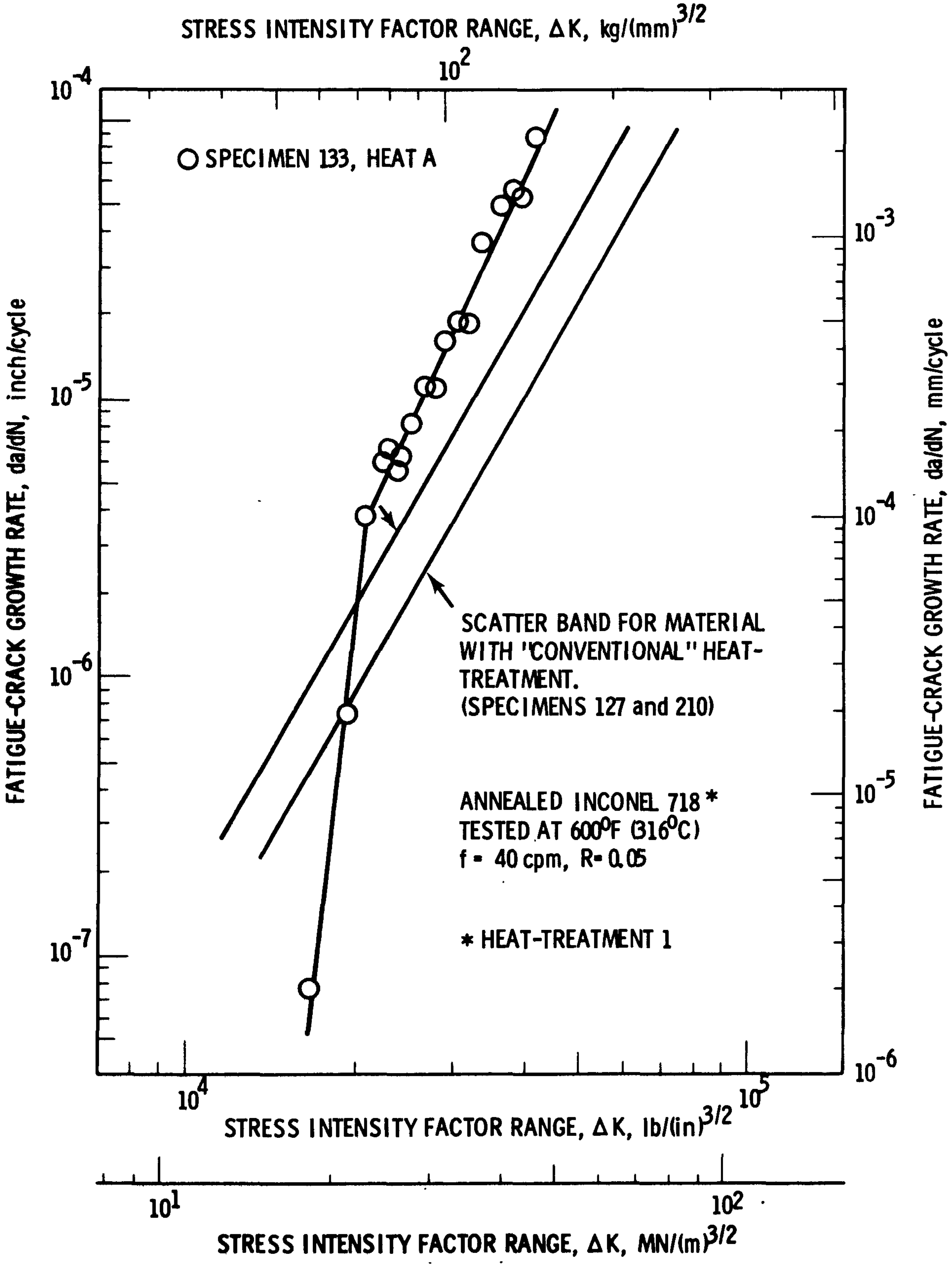
F. CRACK GROWTH IN WELDMENTS

As mentioned in the previous section, problems of inadequate ductility have been encountered when Inconel 718 weldments are given the "conventional" heat-treatment following welding[35,36]. The main difficulty appears to be



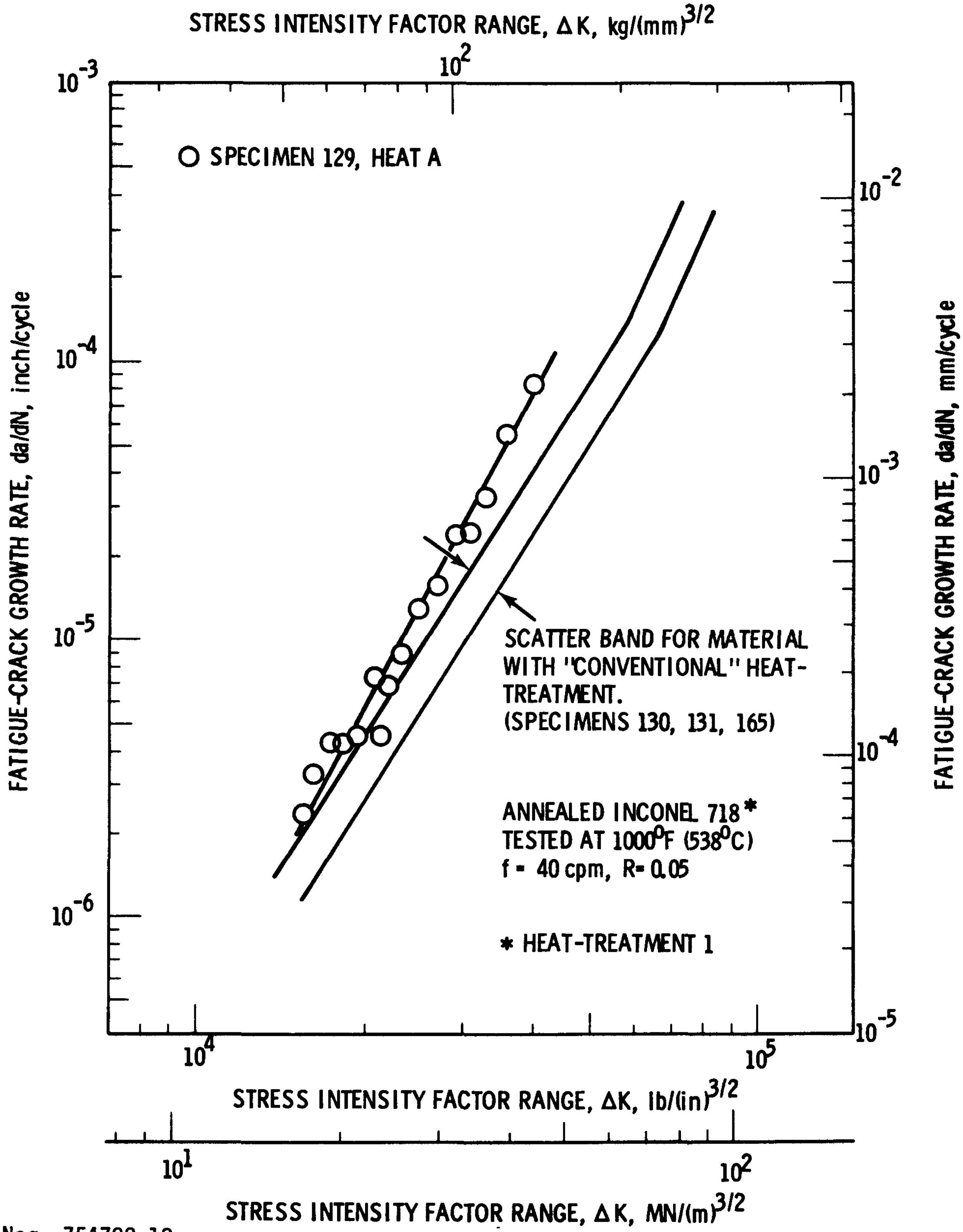
Neg. 754782-3

FIGURE 22. Fatigue-Crack Growth Behavior of Annealed Inconel 718 in an Air Environment at Room Temperature.



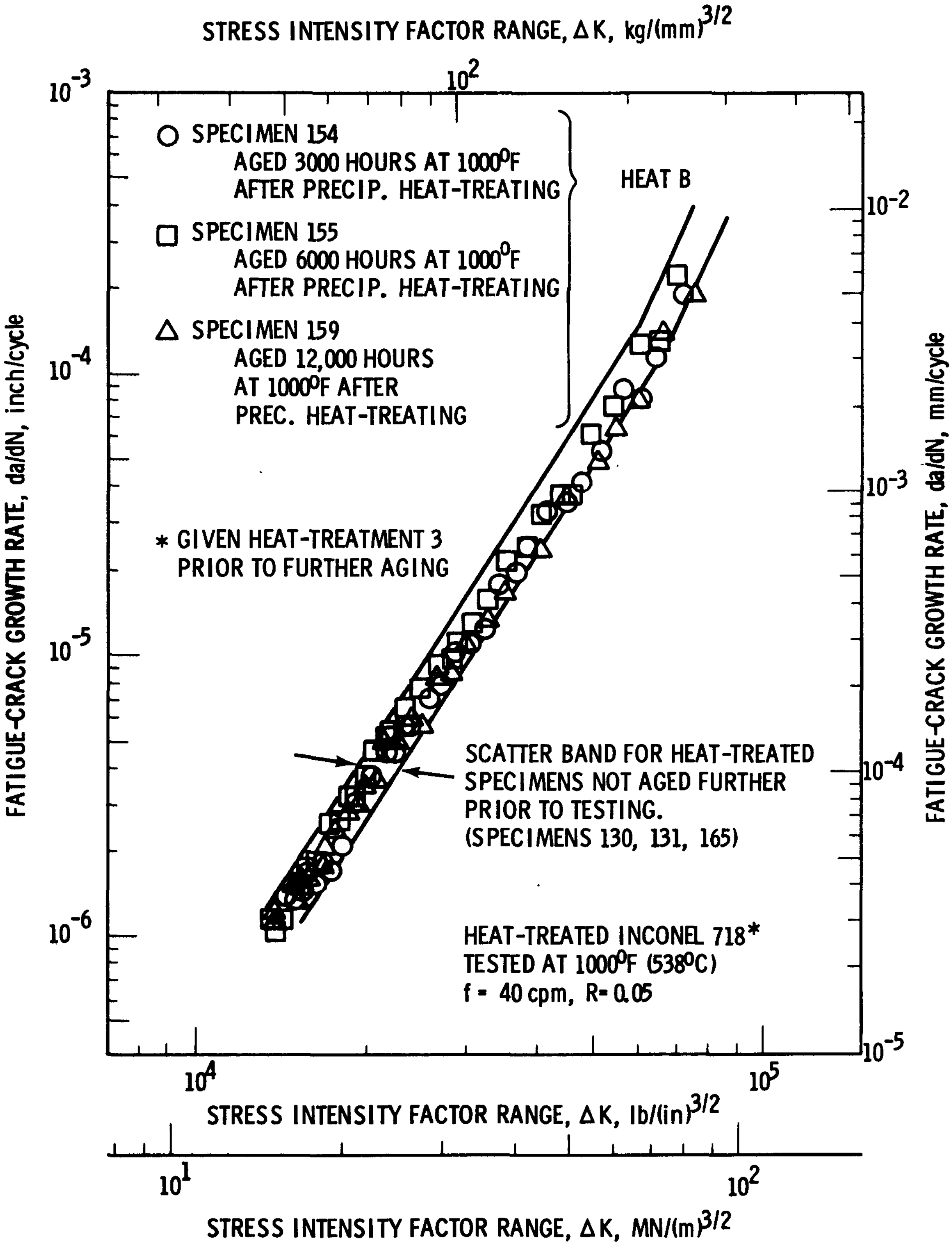
Neg. 754782-18

FIGURE 23. Fatigue-Crack Growth Behavior of Annealed Inconel 718 in an Air Environment at 600°F (316°C).



Neg. 754782-18

FIGURE 24. Fatigue-Crack Growth Behavior of Annealed Inconel 718 in an Air Environment at 1000°F (538°C).



Neg. 754782-13

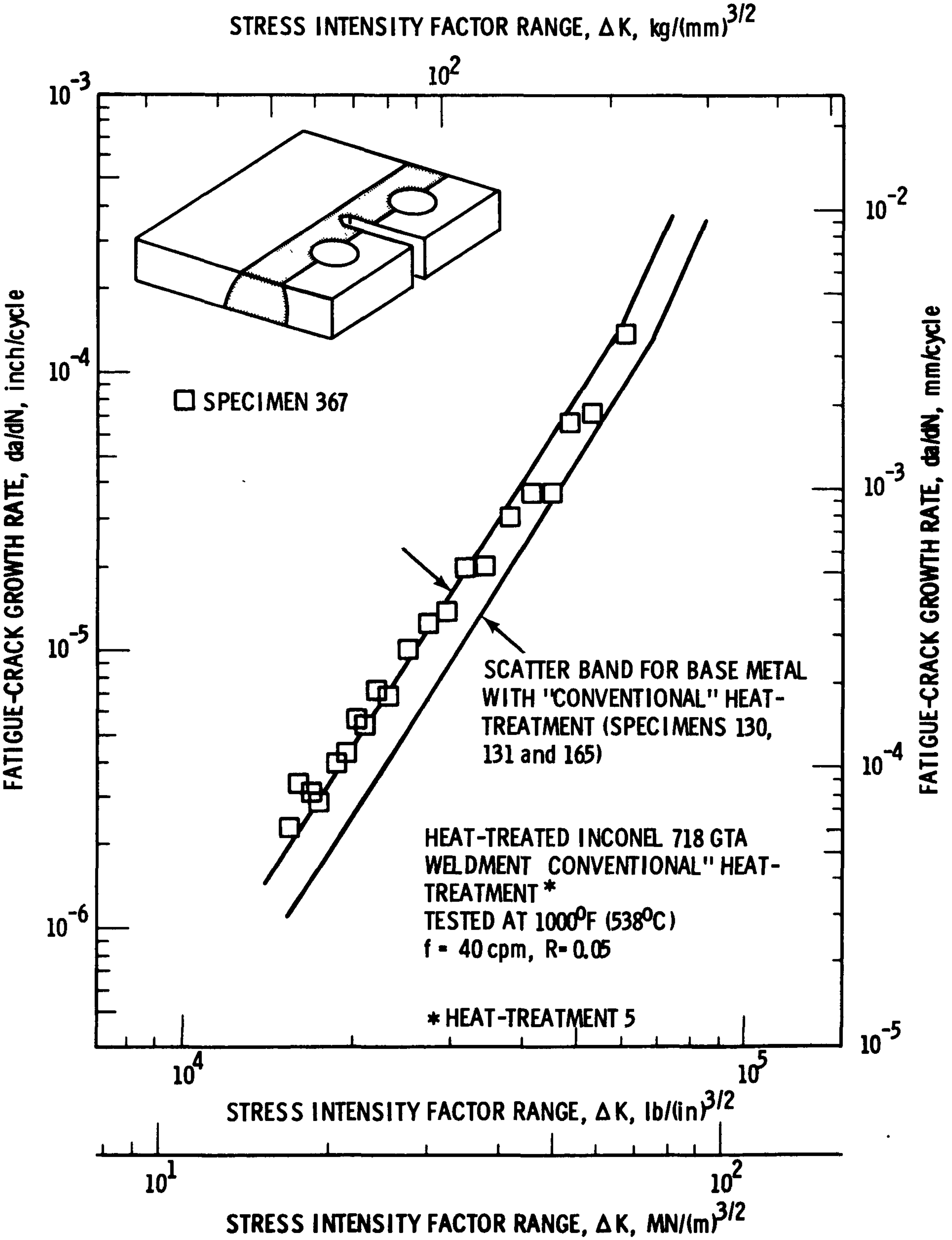
FIGURE 25. Fatigue-Crack Growth Behavior of "Conventionally" heat-Treated Inconel 718 Subjected to Additional Thermal Aging Prior to Fatigue Testing.

that the 1750°F (954°C) annealing temperature following welding (see Heat-Treatment 5 in Table II) is not sufficiently high to dissolve the "brittle" Laves phase that forms during the welding process, and annealing temperatures of about 2000°F (1093°C) appear to be required in order to achieve dissolution of the Laves phase[35,36]. However, the higher post-weld annealing temperature can result in a lower than optimum strength in the base metal following the precipitation age. The purpose, therefore, of the Aeroject Nuclear study[35] was to eliminate the "brittle" Laves phase without a significant weakening of the base metal.

GTA weldment specimens given a "conventional" heat-treatment following welding (Heat-Treatment 5) were tested at 1000°F (538°C) in two orientations of the crack relative to the weld, and these results are shown in Figures 26 and 27. The data of Figure 26 are for an "AB" orientation*, and it will be noted that at the lower values of K (i.e. shorter crack lengths) when the crack is propagating through the weld metal, the crack growth rates are slightly higher than those observed in similarly tested base metal. However, as the crack continues to extend and pass out of the weld metal and into the base metal, the crack growth rates become essentially identical to those observed for base metal specimens. The data of Figure 27 are for a "BA" orientation in which the crack propagates entirely within the weld metal. It will be noted in Figure 27 that the crack growth rates are higher in the weldment specimens than in similarly tested and heat-treated base metal specimens. The results of Figures 26 and 27, therefore, indicate that at 1000°F (538°C), "conventionally" treated weldment specimens exhibit greater crack growth rates than "conventionally" treated base metal. This is opposite to the trends noted for weldments in Type 304 stainless steel, which generally exhibited lower crack growth rates at 1000°F (538°C) than base metal[37,38].

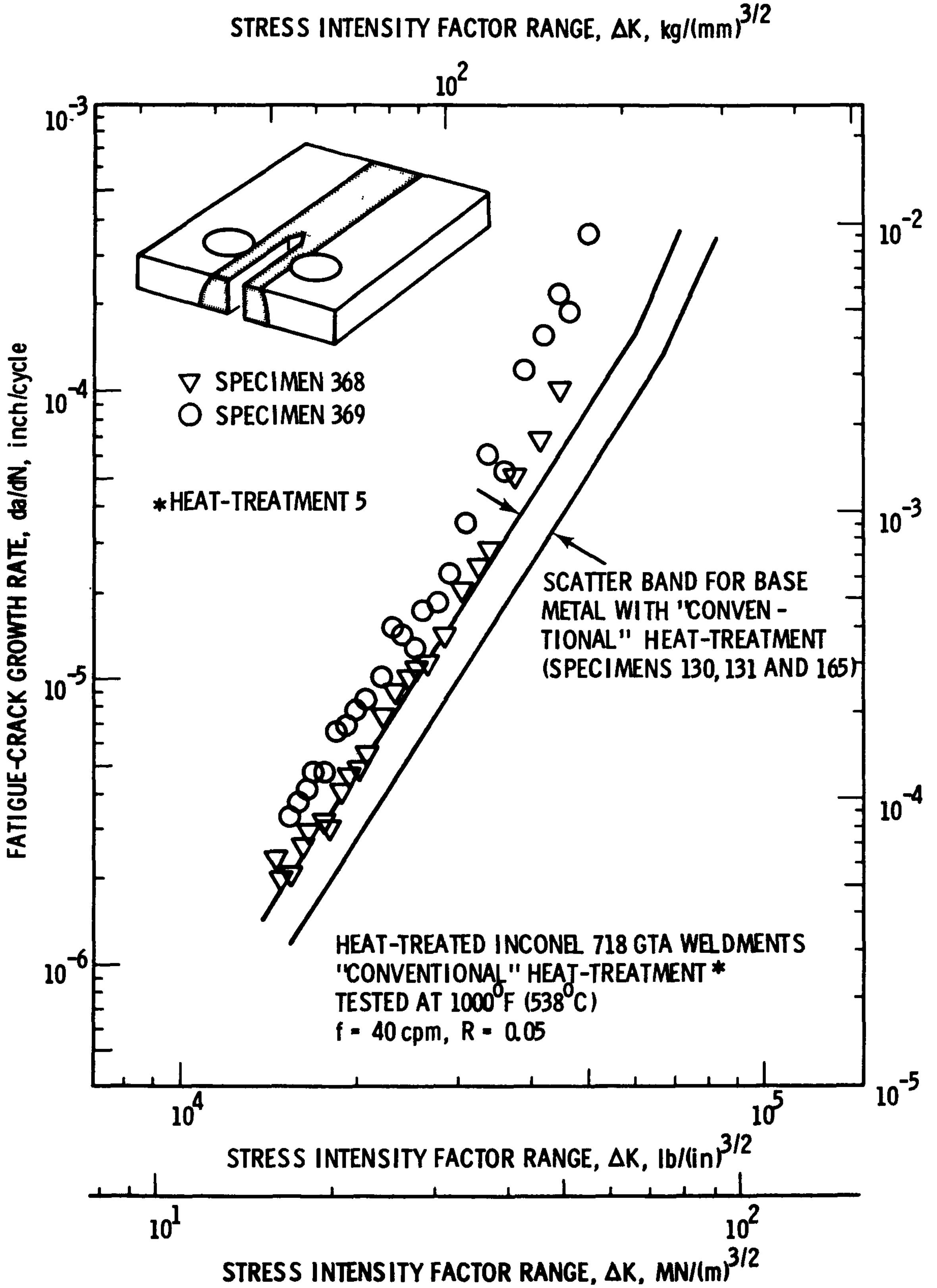
One of the "conventionally" treated specimens (Specimen 368 in Figure 27) failed during fatigue testing. The fracture surfaces of this specimen are shown in Figure 28 and it will be noted that the fracture surfaces are flat and devoid of shear lips indicating a low-energy fracture. Assuming that the failure occurred at the maximum fatigue load, and using the final crack

*See Reference 37 for the crack orientation system employed for weldment specimens.



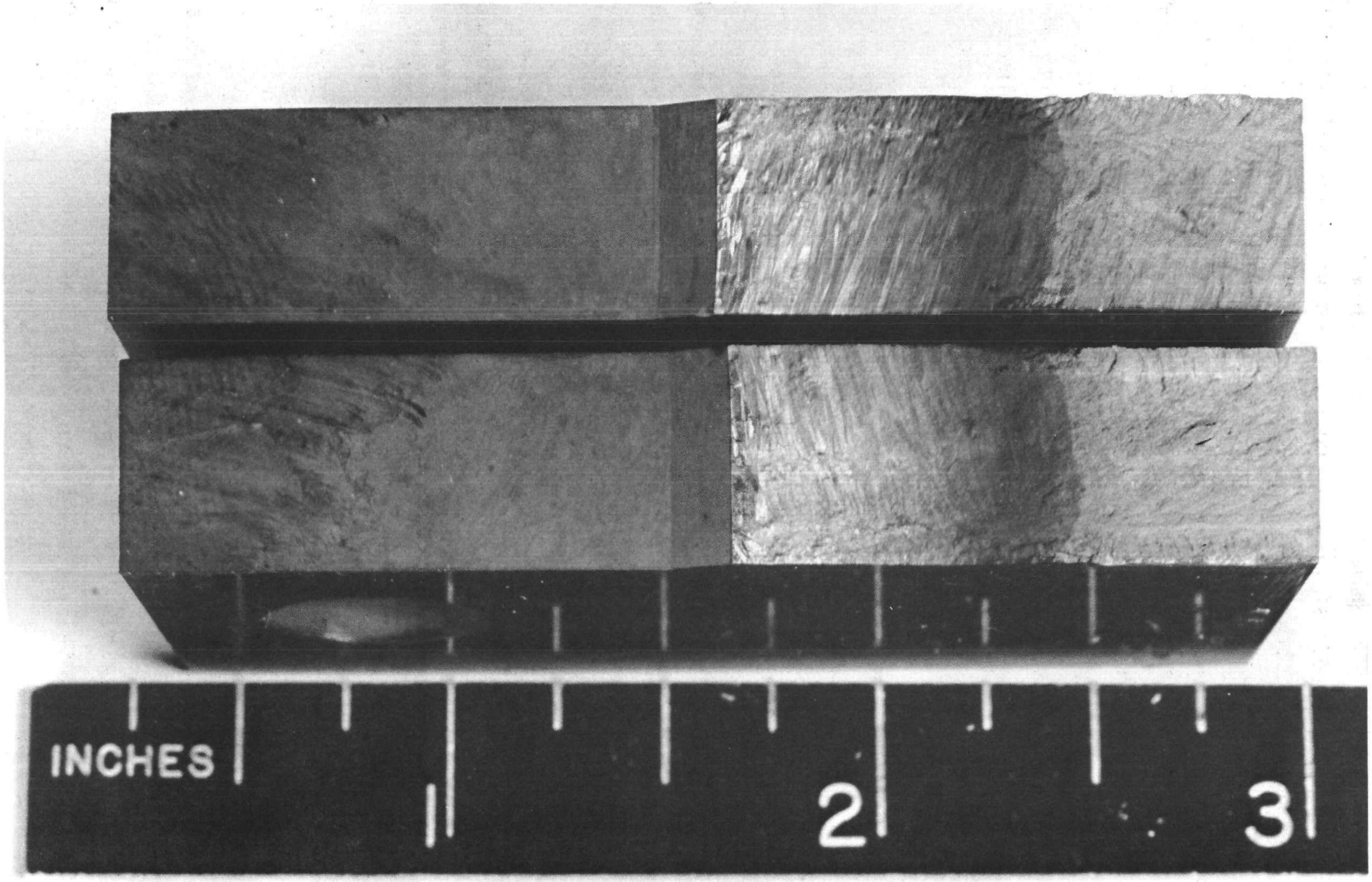
Neg. 754782-30

FIGURE 26. Fatigue-Crack Growth Behavior at 1000°F (538°C) of Inconel 718 GTA Weldment Given "Conventional" Heat-Treatment After Welding. (AB Orientation).



Neg. 754782-27

FIGURE 27. Fatigue-Crack Growth Behavior at 1000°F (538°C) of Inconel 718 GTA Weldment Given "Conventional" Heat-Treatment After Welding. (BA Orientation).



Neg. 736163-1
FIGURE 28. Fracture Surfaces of Specimen 368 Which Failed During Fatigue Testing.

length (taken from the fracture faces), an approximate final value of K may be calculated to be 56,900 psi $\sqrt{in}$. There are a number of reasons why this value should not be construed as K_{Ic} since many of the criteria of Reference 10 were not met: a) since this was a fatigue test, no crack opening deflection transducers were installed as would be necessary in a fracture toughness test, b) the final fatigue crack length was longer than that specified in Reference 10 and "precracking" occurred at approximately the same load as the failure load, rather than at the lower precracking loads required in Reference 10, and c) the load rise time (one-fourth the fatigue cyclic period for a sine wave) was faster than allowed by Reference 10. Nevertheless, it is interesting to make the calculation (per Ref. 10) to see if the specimen was sufficiently thick to produce a "plane-strain" fracture:

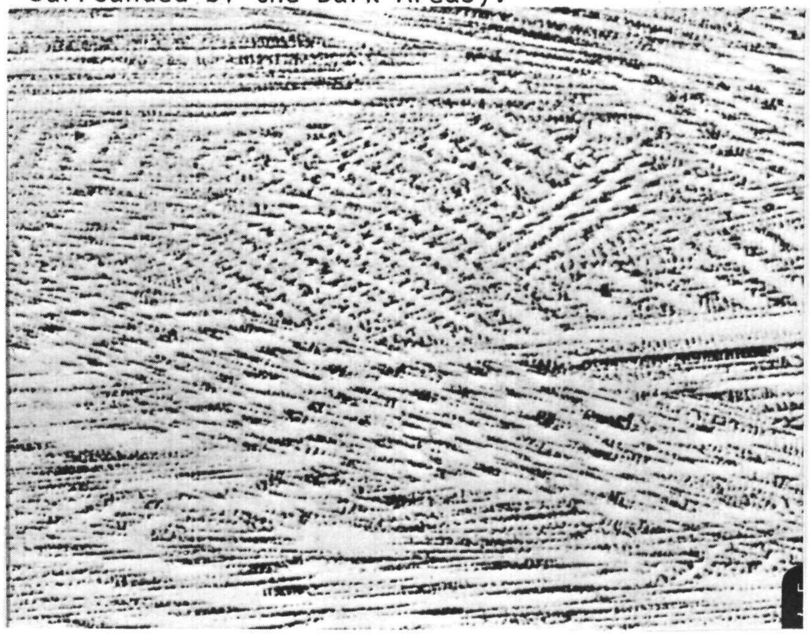
$$B > 2.5 \left(\frac{K}{\sigma_{ys}}\right)^2 \qquad \text{Eq. [5]}$$

Substituting K = 56,900 psi $\sqrt{in}$ and σ_{ys} = 134,700 psi (from Table IV), Equation [5] predicts that a thickness (B) of 0.446 inch would have been sufficient to produce a "valid" plane-strain fracture (providing, of course, all the other requirements of Reference 10 were met). From Table V it will be seen that this specimen had a thickness of approximately 0.454; slightly greater than that required by Equation [5].

The weld in Specimen 367 (which had the identical heat-treatment as Specimen 368, above) was examined metallographically. The photomicrographs of the fusion zone shown in Figure 29 indicate the presence of the Laves phase (the small white-etching areas surrounded by the dark areas in Figure 29b), which presumably contributed to the observed low-energy fracture. Thus we see that weldments given the "conventional" heat-treatments following welding exhibit higher crack growth rates than similarly treated base metal, and that these weldments may possess a relatively low fracture toughness.

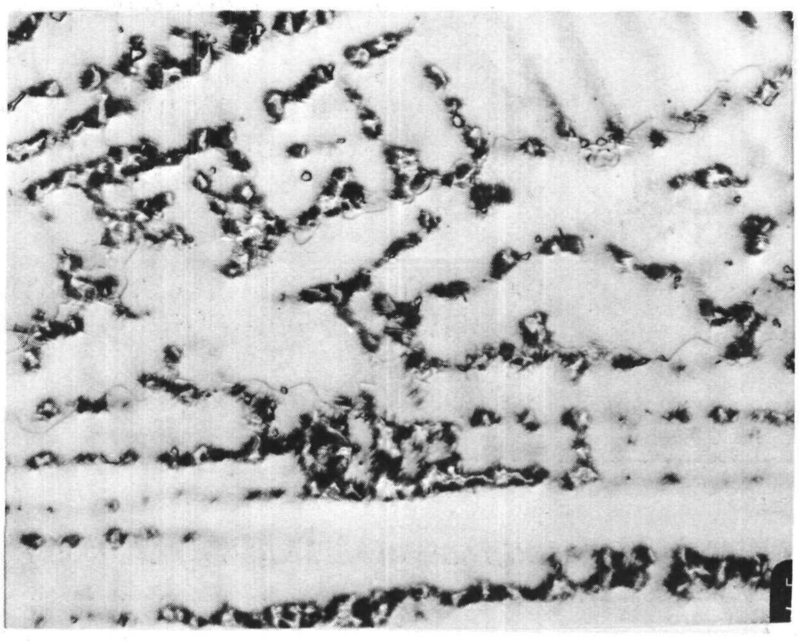
Weldment specimens receiving the "modified" heat-treatment (Heat-Treatment 7 in Table II) were tested at 1000°F (538°C) and 1200°F (649°C), and the results are shown in Figures 30 and 31, respectively. It will be noted in Figure 30 that the crack growth behavior in weldments with the "modified" heat-treatment is generally superior (i.e. lower growth rates) than that in

(a) Showing Laves Phase (the Small White-Etching Areas Surrounded bv the Dark Areas).



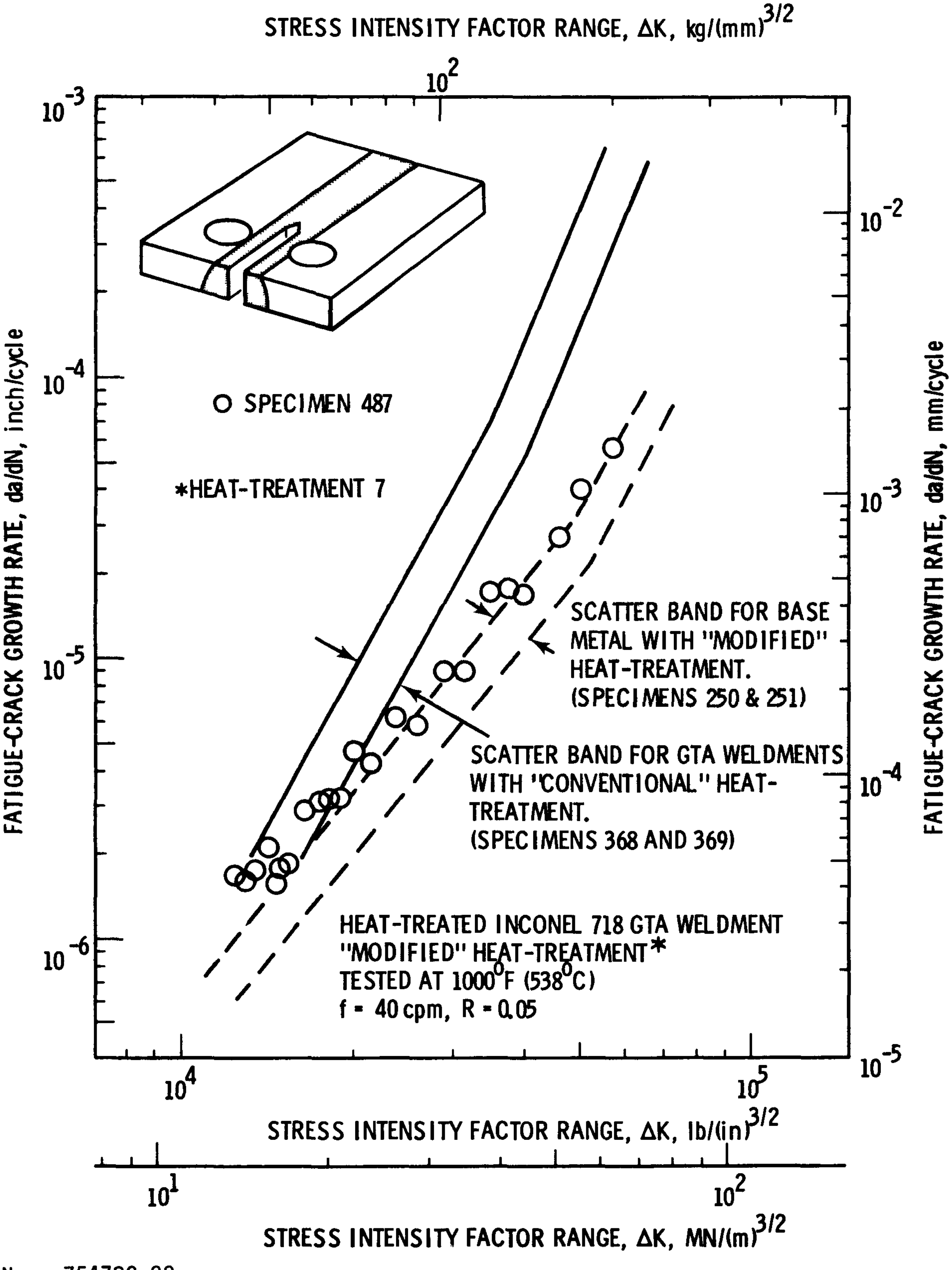
Neg. 473-1445D   10% Oxalic Acid 100x

(b) Enqlargment of Area of Fiqure 29.



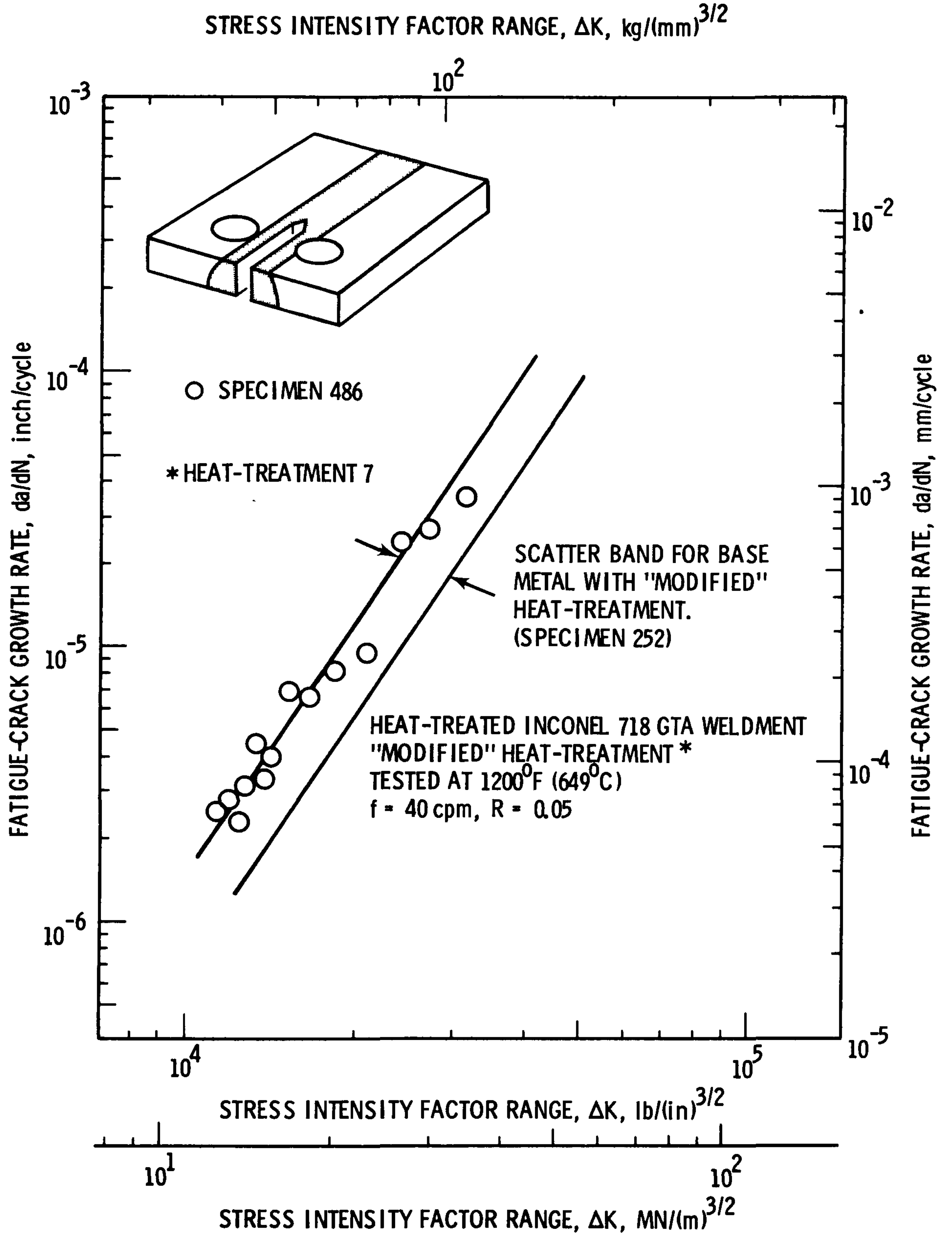
Neg. 473-1445F   10% Oxalic Acid 750x

FIGURE 29. Photomicrograph of Weld Deposit in Speciment 367.



Neg. 754782-22

FIGURE 30. Fatigue-Crack Growth Behavior at 1000°F (538°C) of Inconel 718 GTA Weldment Given "Modified" Heat-Treatment After Welding.



Neg. 754782-23

FIGURE 31. Fatigue-Crack Growth Behavior at 1200°F (649°C) of Inconel 718 GTA Weldment Given "Modified" Heat-Treatment After Welding.

"conventionally" treated weldments. This is especially true at the higher values of ΔK where the maximum cyclic value of K is starting to approach the fracture toughness. It will also be noted that the crack growth rates in the weldment with the "modified" treatment are slightly higher than in similarly treated base metal. The "modified" treatment weldment specimen tested at 1200°F (649°C) (see Figure 31) also exhibits crack growth rates slightly higher than in similarly treated base metal.

The weld in Specimen 487 which was given the "modified" heat-treatment (Heat-Treatment 7 in Table II) was examined metallographically. The photomicrographs of the fusion zone shown in Figure 32 show little or no trace of the Laves phase that was present in the "conventionally" treated weldments (e.g. see Figure 29). The differences between Figures 29 and 32 are striking, but almost identical photomicrographs may be seen in Reference 36. Therefore, it appears that the dissolution of the Laves phase due to the "modified" heat-treatment is primarily responsible for the improvements noted in the crack growth behavior relative to "conventionally" treated weldments.

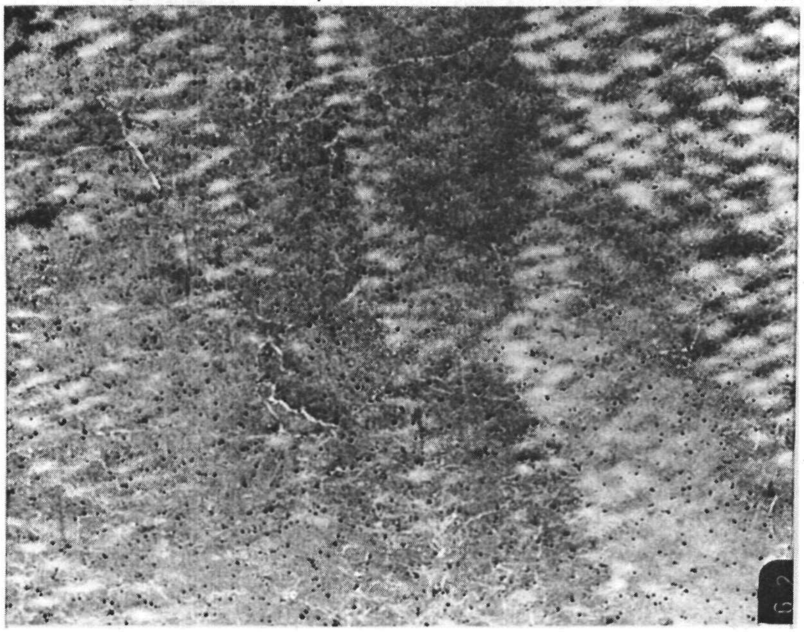
One weldment specimen was tested which had recieved what will be called the "irregular" heat-treatment (Heat-Treatment 6 in Table II). The higher annealing temperature could have produced some grain growth, and the higher precipitation aging temperatures could have over-aged the structure; nevertheless it is interesting to compre the crack growth results to those for "conventionally" treated weldments. This is done in Figure 33 and it will be noted that the "irregular" heat-treatment produced crack growth rates generally lower than those observed for the "conventional" treatment, and that the behavior is quite similar to that observed for the "modified" treatment specimen shown in Figure 30.

IV. SUMMARY AND CONCLUSIONS

The results of the present study may be summarized as follows:

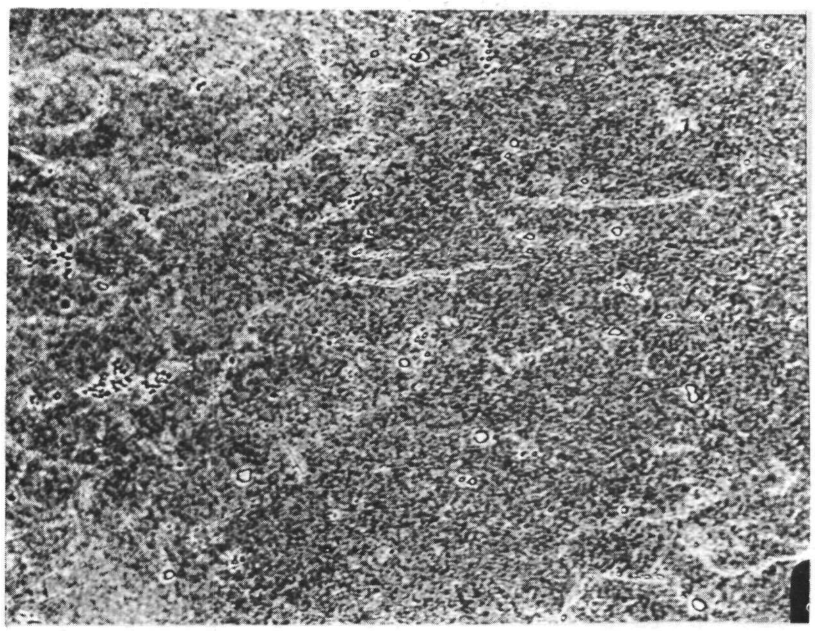
- When Inconel 718 is fatigue tested in an air environment, the fatigue-crack growth rate increases with increasing test temperature.

(a) Showing Almost Complete Dissolution of Laves Phase.



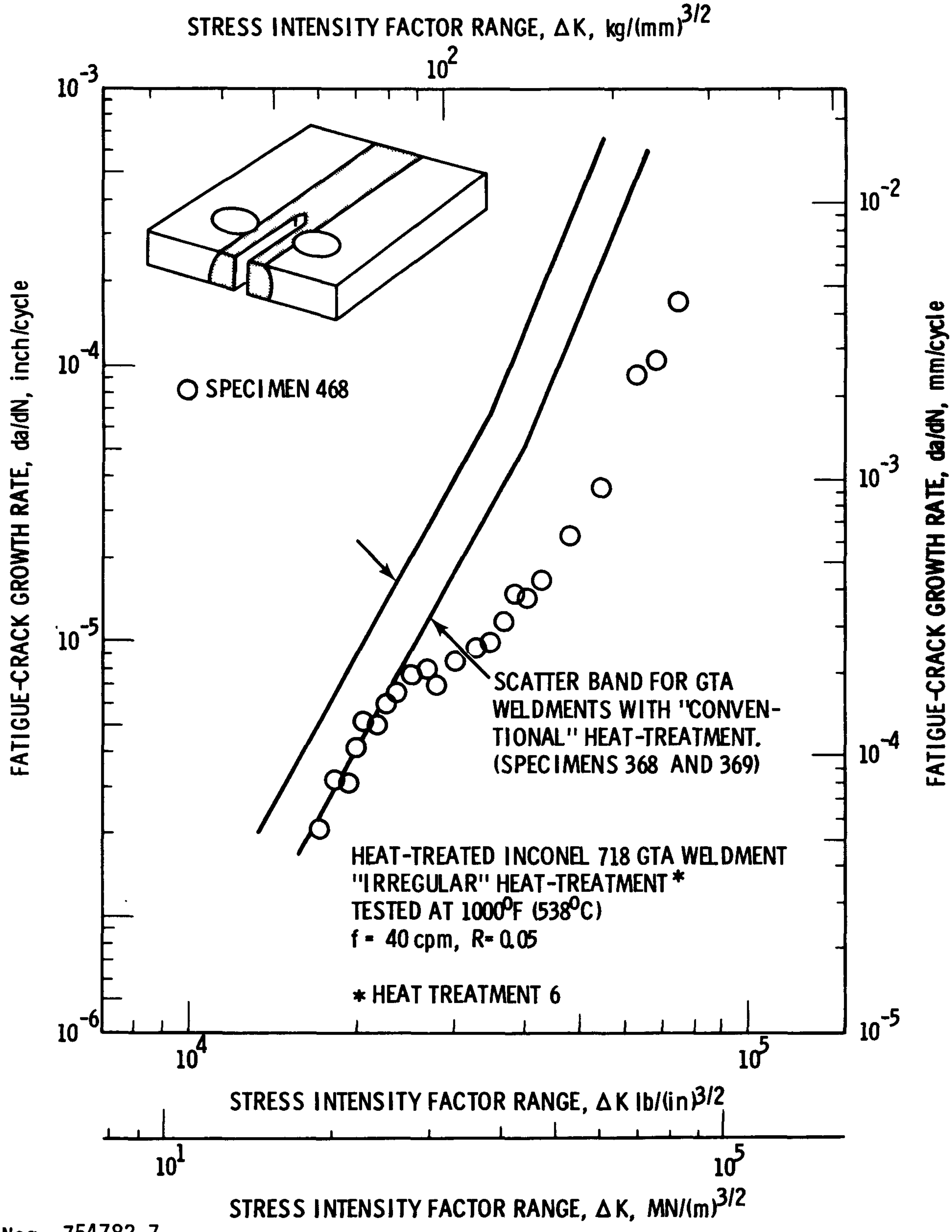
Neg. 475-2261A 10% Oxalic Acid 100x

(b) Enlargement of Area of Figure 32



Neg. 475-2261B 10% Oxalic Acid 750x

FIGURE 32. Photomicrograph of Weld Deposit in Specimen 487.



Neg. 754782-7

FIGURE 33. Fatigue-Crack Growth Behavior at 1000°F (538°C) of Inconel 718 GTA Weldment Given "Irregular" Heat-Treatment After Welding.

- When Inconel 718 is fatigue tested in an air environment at elevated temperatures, the fatigue-crack growth rate increases with decreasing cyclic frequency.

- At a given value of ΔK, the fatigue-crack growth rate increases with increasing values of the stress ratio R. However, the "effective" stress intensity factor can be used to normalize the effect of stress ratio.

- Limited data indicates that the fatigue-crack growth rate in a sodium environment at 800°F (427°C) is approximately an order of magnitude lower than in air under similar conditions.

- The "modified" heat-treatment developed by Aerojet Nuclear Company produces improvements in crack growth behavior (i.e. lower crack growth rates) relative to the "conventional" treatment ranging from slight to significant at test temperatures below the final aging temperature 1150°F (621°C), and essentially the same behavior above that temperature. This is the case for both base metal specimens and weldment specimens.

- Weldment specimens given the "conventional" heat-treatment exhibited higher crack growth rates and lower apparent ductility than did specimens receiving the "modified" treatment. Since the latter treatment effectively removed most of the Laves phase, it is assumed that the Laves phase was responsible for the inferior performance of the "conventionally" treated weldments.

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