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Henri Albert Sirot
(M.S. thesis)

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CRACK PROPAGATION RATES IN Fe-Ni ALLOYS

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CRACK PROPAGATION RATES IN Fe-Ni ALLOYS

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ABSTRACT

Fatigue crack propagation rates of iron nickel cryogenic alloys with 4, 8, 12 and 16 wt% Ni were measured as a function of the stress intensity range \( \Delta K \) at different temperatures. Room temperature tests showed that stage II crack propagation rates for cracks propagating faster than .2 \( \mu m \)/cycle are relatively insensitive to grain size, yield strength or nickel content, and the rates were comparable to those found in a 5 Mo .3 C steel. An equiaxed alpha structure, obtained by a very slow cool, was found slightly less resistant to fatigue crack propagation than a structure containing martensite, which was obtained by an iced brine quench. This slight effect could be due to either the change in substructure or the residual stresses associated with the production of martensite. The 12 Ni 0.5 Ti alloy which was iced brine quenched showed the best crack propagation resistance.

Crack growth rates at -77°C, -116°C and -196°C were measured as the interspace between the striations seen on the fatigue fracture surface using a scanning electron microscope. Crack growth rates faster than 0.2 \( \mu m \) per cycle appeared to be insensitive to the temperature of the test, even for the 8 Ni alloy fatigued at -116°C which is slightly below its charpy V-notch transition temperature. The longer
fatigue lives of the 8 Ni alloy specimens fatigued close to their transition temperature was attributed to slower crack growth rates during the initial stage of the propagation. Since \( K_{IC} \) varies with temperature, the above result showed that crack growth rates cannot be related to \( K_{IC} \).

It was impossible to initiate a fatigue crack at \(-196^\circ C\) in the 8 Ni alloy nor was it possible to propagate a crack previously initiated at room temperature. The specimen did not crack, or it failed at higher \( \Delta K \) in a brittle mode. The critical stress intensity values obtained from the final failure under fatigue conditions were lower than \( K_{IC} \) values measured by a conventional test. This effect may be attributed to the differences in strain rates or to the statistical distribution of strength.

Short periods of high stress cycling was shown to retard subsequent crack propagation under normal stress cycling at room temperature as well as at \(-196^\circ C\) in the 12 Ni alloy.

A good correlation between the standard optical crack tip observation technique and the striation measurement technique has been obtained for cracks growing faster than 0.2 \( \mu m/\text{cycle} \). The systematic difference observed in crack growth rates for high values of \( \Delta K \) has been explained by the curvature of the crack.
I. INTRODUCTION

A family of iron-nickel based cryogenic alloys has been developed recently in this laboratory. The thermomechanical processes designed to make these alloys both strong and tough at very low temperatures have been specified.

It is well known that a majority of service failures in structural components occur by fatigue. The characteristic mode of fatigue fracture consists of crack initiation at a local stress concentration and growth of the crack under cyclic loading until failure finally occurs. The final fracture mode is analogous to a static failure mechanism. It is interesting to note that if new smaller specimens are machined from the two broken pieces, they will show fatigue strengths comparable to new specimens. Thus, fatigue properties are much more dependent on local conditions than other mechanical properties such as yield strength and impact toughness. Therefore, it is imperative that the fatigue crack propagation characteristics of any newly developed cryogenic alloy be determined before the material can be considered for use in structural applications. Factors affecting the crack growth rates should be defined with the aim of improving the fatigue strength.

A majority of the fatigue properties found in the literature are based on tests to determine the fatigue life as a function of nominal stress. However, the present trend is to describe the crack growth rate as a function of the stress intensity factor $K^*$, since $K$ completely determines the stress distribution at the crack tip region. It has been

*NOTE: Stress intensity for an infinite plate, $K = \phi \sigma \sqrt{c}$ where $\phi$ is a geometric parameter, $\sigma$ is the nominal stress and $c$ is the instantaneous crack length.
shown that $K$, and not the nominal stress, governs the crack propagation rate. Most investigators express their results in the form:

$$\frac{da}{dN} = c(\Delta K)^m$$

where $da/dN$ is the crack growth rate per stress cycle, $\Delta K$ is the stress intensity range and $c$ and $m$ are experimental constants.

Various attempts have been made to relate $c$ and $m$ to the microstructure, stress level, environment and so on. In spite of intense efforts, as evident by the number of papers on this topic, no general agreement exists on what metallurgical or physical factors govern the rate of crack growth. For example, many investigators have tried to relate $c$ and $m$ to the mechanical properties (yield strength, tensile strength, $K_{IC}$, etc...) whereas others, (see Halm et al., Barson et al., Bathias and Pelloux, Johnson and Paris) claim that crack growth rates are insensitive to mechanical properties, composition or microstructure. Fatigue crack propagation is divided into stage I which occurs on slip planes inclined at 45° to the tensile axis (the highest shear stress planes) and stage II in which the crack grows normal to the stress axis. The extent of stage I crack growth increases as the load amplitude is decreased. The fraction of total life spent in stage II crack growth increases with increasing load amplitude. In the low cycle range (when fatigue fracture occurs within 50,000 cycles) most of the life consists of stage II cracking. Usually the distance traveled by a crack in the stage I mode is so small, i.e. about 0.2 mm or less, that for all practical purposes the fatigue fracture surface is formed by crack growth in the stage II mode.
Crack propagation rates of the cryogenic alloys were measured at low temperature: -77°C, -116°C and -196°C. The experimental difficulties involved in measuring the crack growth rates directly (using optical methods at very low temperatures) required the use of an indirect technique, one in which the crack growth rate was deduced from examination of the fracture surface. Therefore stage II crack growth only could be investigated in this study. In order to compare the results of this indirect technique with the standard optical measurements of the crack positions, several experiments were made at room temperature. In these room temperature experiments the influence of microstructure was also investigated.
II. EXPERIMENTAL PROCEDURE

A. Material Preparation

The alloys used in this investigation were made by vacuum induction melting; the compositions are given in Table I. To eliminate brittleness due to interstitials such as C,N,O and H, 0.15 Ti and 0.10 Al were added to the alloys.\textsuperscript{17,18} The as cast 20 lbs. ingots were homogenized for 72 hours at 1060°C in vacuum and then forged at 1000°C to a 2 in. by 2 5/8 in. cross section. The bars were then hot rolled at 1000°C to 1 5/8 in. thickness. Blanks cut from the rolled bars were heat-treated and then these blanks were machined to the desired dimensions for mechanical testing.

B. Heat Treatments and Microstructures

The transformation diagram of the iron nickel system obtained from continuous heating and cooling data are shown in Fig. 1. For the 4\%Ni to 16\%Ni alloys used in the study, the temperature above which austenite is the stable phase varies from 790°C to 650°C.

For the 4\%Ni alloy and 16\%Ni alloy the temperature range within which $\alpha$ and $\gamma$ phases coexist in equilibrium are 700°C-790°C and 350°C-650°C respectively. At room temperature, $\alpha$ is the only stable phase. There are two different ways by which $\gamma$ can transform to $\alpha$:

1. For alloys containing up to 5\%Ni, $\gamma$ transform to $\alpha$ with a nucleation and growth mechanism over a range of cooling rates ranging from about 2°C to 600°C per second. The structure obtained is called equiaxed alpha.
2. In alloys containing more than 10%Ni the austenite transforms into alpha with a athermal shear type mechanism even for very slow cooling rates. The resulting product is called massive (or lath) martensite.

Both types of reactions can occur in alloys containing between 5 and 10% nickel depending on the cooling rate. It has already been pointed out that the 12Ni, 0.5Ti balance Fe alloy was found very promising for cryogenic applications. The recommended heat treatment for this alloy is to austenitize at 900°C (2 hours for 1 in. section) air cool to room temperature followed by a reheat to 700°C (just above the two phase region) (1/2 hour to reach temperature and 1 hour at temperature for 1 in. section) and quench in iced brine. The resulting microstructure is lath martensite with a grain size of approximately 15 microns diameter. (Fig. 2).

An alloy containing 8Ni, .15Ti, .10Al wt% was also found promising in a preliminary study for cryogenic applications above -77°C. This alloy appeared interesting because the microstructure can be varied and thus most of the experiments to study the effect of microstructure on crack growth rate were done using this alloy.

Unlike steels, in which it is relatively easy to vary the microstructure by changing the cooling rate, cooling rates have very little effect on the microstructure of Fe-Ni alloys. The only practical way to get completely equiaxed α or completely lath martensite is to change the composition. Therefore fatigue crack growth rates on 4%Ni and 16%Ni alloys were also studied.
The 4%Ni alloy was austenitized at 900°C and air cooled to get pure equiaxed alpha substructure and a grain size of 50 μm, Fig. 3.

The 16%Ni alloy was austenitized at 950°C and iced brine quenched to get a lath martensite substructure and a grain size of 50 μm, Fig. 3.

In the 8%Ni alloy the grain size and the substructure, that is the relative proportion of equixed alpha and lath martensite within the grains can be varied in the following way:

A 50 μm grain size can be obtained after a 2 hour austenitizing treatment at 900°C. This grain size can then be refined to 15 microns by air cooling or ice brine quenching from 900°C and by reheating the alloy slightly above the two phase region. It can be seen on Fig. 4 that a duplex structure is obtained after reheating at 700°C, and that refinement of the prior austenitic grain is obtained after heat treatment at 735°C. A reheating temperature of 745°C-750°C was chosen to make sure that grain refinement was obtained without segregation (Fig. 5). Further grain refinement by a cycling treatment proved to be difficult. It has been found that the heat up rate is important. In the present case, the heat up time was about 30 minutes. A specimen quenched in iced brine from 900°C has a mixed substructure of lath martensite and equiaxed alpha (Fig. 6). A tentative TTT diagram (see Figs. 7 and 8) was developed using a magnetic permeability technique described elsewhere. Because of experimental difficulties to determine the precise location of the $M_s$ temperature, the amount of equiaxed $\alpha$ was increased by the following method: The specimens were furnace cooled from 900°C or 750°C to 560°C, held at 560°C for 18 hours, furnace cooled to 450°C and then air cooled to room temperature. The difference between the microstructures obtained with the different cooling rates
is difficult to observe optically (Fig. 9). Table II gives the different structures obtained in the alloys used in this study.

C. Mechanical Testing

1. Tensile Tests

Tensile properties were determined using 0.15 inch thick, 0.125 inch wide flat specimens with 1 inch gage length (Fig. 10). An Instron Testing Machine with 11000 lbs. capacity was used. Tests were conducted at room temperature, -77°C and -196°C.

2. Fatigue Tests

Fatigue crack propagation properties were measured using WOL standard specimens (Fig. 11). Specimens with 0.25 in., 0.5 in. and 1.0 in. thicknesses were tested at room temperature to determine the optimal thickness to insure plane strain conditions during crack growth. From Fig. 12, it can be seen that the results from the different samples are similar, but the experiments run on the thinner specimens had to be stopped earlier since general plastic deformation occurred at smaller crack lengths in these samples. It can thus be inferred that the stress intensity range was limited to values less than 66 MN m\(^{-3/2}\) (60 ksi \(\sqrt{\text{in.}}\)) for the thin samples. Therefore it was decided to run most of the tests using 1 in. thick specimens. The tests were carried out on a 300,000 lbs. MTS machine in load control and under sinusoidal tension-tension cycling. The ratio of the maximum load to the minimum load was kept constant at 0.1.
The MTS testing machine, as is characteristic of all electro-hydraulic equipment, has a frequency response limited by the sizes of the servo valves and the hydraulic power supply. At low frequencies the machine can respond faithfully to the desired load, displacement or strain commands as determined by the set point. However at higher frequencies, the response will be less than the set point values. This is normally overcome by using a strip chart to empirically determine a set point which will lead to the desired response from the machine. Frequently, this set point might command the machine to deliver a load of, for example 7000 lbs. whereas the load delivered is only 5000 lbs.

This discrepancy between command and response will increase as the frequency increases. Thus, for a desired minimum load to 600 lbs, if the frequency set is about 10 cps, the set point could be -100 lbs, i.e. compressive load. On stopping the machine to measure crack length, the frequency will drop to 0 cycle per second, and the machine will apply the set point load i.e. 100 lbs compression. This is highly undesirable since it can destroy the surface features on the fracture surface. There is also the danger of the ram going out of control in its effort to find resistance to generate a 100 lbs compression loading, especially since the grips are loose in compression. However safety limits can be set on ram displacement. After some trial and error a cyclic frequency of 5 cycles/second and high gain on feed back was used in all the tests. This insured that for all the minimum loads used in this study, the lower set point never commanded compression and thus the tests could be stopped anytime for crack length measurements without specimen damage.
The stress intensity for this specimen geometry as determined from the boundary collocation solution, is given by:

\[ K_I = \frac{P}{W^{1/2}B} y(a/W) \]

\[ y(a/W) = \left( \frac{a}{W} \right)^{1/2} \left[ 29.60 - 185.5\left( \frac{a}{W} \right) + 655.7\left( \frac{a}{W} \right)^2 - 1017.0\left( \frac{a}{W} \right)^3 + 638.9\left( \frac{a}{W} \right)^4 \right] \]

for \( 0.3 > a/W < 0.7 \)

where \( P \) is the load, \( B \) the thickness, \( W \) the specimen width and \( a \) the crack length measured from the center line of pin holes.

The function \( y(a/W) \) increases with \( a/W \) (see Fig. 13). For a given crack length, \( a \),

\[ \Delta K = K_{\text{max}} - K_{\text{min}} = \frac{P_{\text{max}}}{W^{1/2}B} y(a/W) - \frac{P_{\text{min}} y(a/W)}{W^{1/2}B} = \frac{\Delta P}{W^{1/2}B} y(a/W) \]

The test was run under load control, so \( \Delta P \) was constant but the stress intensity range increases as the crack length increases.

**Crack Growth Rates: Room Temperature Tests**

At this temperature the crack lengths on both sides of the specimen were measured with a travelling microscope. Frequently these lengths were different, so an averaging step is necessary. Assume that in \( N \) cycles the crack length increases from \( a_1 \) to \( a'_1 \) on one side and from \( a_2 \) to \( a'_2 \) on the other side (Fig. 14). The crack advance per cycle was calculated as:

\[ \frac{da}{dN} = \frac{1}{N} \cdot \left( \frac{(a'_1-a_1) + (a'_2-a_2)}{2} \right) = \frac{\Delta a}{N} \]
and the stress intensity range was calculated for the crack length, \( a \), corresponding to the medium point \( M \),

\[
a = \frac{(a_1 + a_2)}{2} - \frac{\Delta a}{2}
\]

This procedure is reasonable when the crack is straight and if the change in crack length is small during that \( N \) cycles, so that \( da/dN \) can then be considered as a linear function of the distance. In this case, the average value \( \Delta a/N \) corresponds to the value of \( da/dN \) at the middle of the interval \([a,a']\).

When the crack front is not advancing as a straight line there is considerable uncertainty regarding the value of \( a \) to be used in the calculation of \( \Delta K \) since the value of \( \Delta K \) and thus the stress around the tip of the crack will vary from one point to the other on the crack front. A number of attempts were made to insure a straight crack front to minimize errors in \( \Delta K \):

a) The holes for the loading pins were cleaned and the specimens were carefully aligned.

b) Thin samples were tried initially, but it soon became apparent that the difference in crack lengths on the two sides were about the same or greater than those seen on thicker samples.

c) Large values of \( \Delta K \) can be obtained either by keeping the load at a relatively low value and letting the crack grow or by increasing the load. It can be seen from Fig. 13 that a given error in a while \( a \) is small, leads to smaller errors in \( \Delta K \) than when \( a \) is large. Therefore, in some of the tests, the load was increased in steps
relatively quickly so that crack growth rates could be measured in the same specimen at values of $\Delta K$ close to the crack propagation threshold value and also at large values of $\Delta K$ while the crack length was fairly small.

**Low Temperature Tests**

The crack growth rates at $-77^\circ$C, $-116^\circ$C and $-196^\circ$C were measured from the striations seen on the fracture surface using a Scanning Electron Microscope (S.E.M.).

The length of the crack i.e. the distance from the slot of the specimen to the area of interest plus the distance between the slot and the load axis was measured using the vernier scale of the SEM. The crack growth rate per cycle was determined as an average of the distances between 2 striations. In the above, care was taken to get the micrographs close to the mid-thickness of the specimen.

A better resolution of the striations can be seen on carbon replicas of the fracture surface using a transmission electron microscope as compared to the resolution obtained through direct examination of the fracture surface using a scanning electron microscope. However, it was impossible to measure the crack length precisely using the replica method, thus almost all the crack lengths and crack growth per cycle measurements were made using the scanning electron microscope.

This technique is evaluated by comparing the results to those obtained from the standard method of measuring the crack length with a telescope. This subject will be discussed in a later section.
3. Fracture Toughness Testing

Plane strain fracture toughness tests were performed using the W.O.L standard specimens already described in the previous sections (Fig. 11). These specimens were first machined oversize and after heat treatment (austenitizing at 900°C followed by an ice brine quench) were ground to the final thickness. The fracture toughness specimens were tested on the M.T.S. machine. Tests were conducted at -77°C (dry ice) -116°C (freezing point of ethanol) and -196°C (LN).

A crack opening displacement gage was used and load and crack opening displacements were recorded on an X-Y recorder. The specimens were first fatigued to a final crack length of about 1.00 inch with a $K_{max}$ of about 30 ksi $\sqrt{\text{in.}}$. This value of $K_{max}$ was very close to the crack propagation threshold value and almost 100,000 cycles were required for 0.05 inch crack growth. However this value of $K_{max}$ did not meet all the ASTM recommendations. The recommendations of the ASTM are:

\[
\frac{K_{max}}{E} < \approx 0.0012 \text{ in.}^{1/2}
\]

\[
\frac{K_{max}}{\sigma_y} < \approx 0.02
\]

\[
K_{max} < \frac{K_Q}{2} \frac{\sigma_y \text{ temperature of test}}{\sigma_y \text{ temperature of fatiguing}}
\]

$K_{max}$ is the maximum stress intensity during fatiguing, $E$ is the Young modulus, and $\sigma_y$ is the yield strength. $K_Q$ is the conditional stress intensity determined in the subsequent fracture test. In particular the condition $(K_{max}/\sigma_y)^2 < 0.02$ inch was not fulfilled.
D. Microscopy

1. Optical Metallography

Specimens for optical metallography were ground on silicon carbide papers to 600 grit, polished on a 1 \( \mu \)m diamond abrasive wheel and given a final electro-polish with a solution of chromium oxide at about 30 volts for 2 mins.

They were then etched with a solution of 50\% Nital 5\% and 50\% Kallings Reagent, for one or two minutes.

2. Scanning Electron Microscopy

The fracture surface of each broken WOL specimen was examined using a JEOLCO JSM-U3 scanning electron microscope set at 25 kv. The specimens were sectioned to dimensions suitable for insertion into the SEM. The fracture surfaces were covered with acetate tape during the cutting operation. This tape was then dry stripped and the fracture surfaces ultrasonically cleaned with acetone.

3. Transmission Electron Microscopy of Carbon Replicas

After the dry stripping, some replica tapes were shadowed with carbon, cut to the dimensions of a electron microscope grid and placed on the grids. The plastic was then dissolved in acetone vapour, and the carbon replicas were examined with the Siemens 1A electron microscope operated at 60 kV.
III. RESULTS AND DISCUSSION

Tensile Data.

From the results of the tensile tests given in Table 3, several conclusions can be drawn.

1. As the Ni content was increased from 4\%Ni to 16\%Ni both the yield strength and tensile strength increased e.g. the Y.S and T.S are 29.3 ksi and 41.2 ksi for 4\%Ni, and 92.0 ksi and 97.5 ksi for 16\%Ni, respectively.

2. The ductility decreased as Ni content increased

3. Ductility also decreased as the test temperature was decreased.

4. In the 8\%Ni alloy slowly cooled samples showed slightly higher strengths than those quenched into iced brine.

5. There was no significant difference in either the yield strength or the tensile strength of the 8\%Ni alloy heat treated to give 50 gm and 15 gm grain sizes.

Fatigue Tests

In the literature, fatigue data such as crack propagation rates \( \frac{da}{dN} \) versus \( \Delta K \) (i.e. \( K_{\text{max}} - K_{\text{min}} \)) are usually presented on log-log plots in which the data may be fitted with a least square line giving the relationship \( \frac{da}{dN} = c(\Delta K)^m \). This is the well known Paris equation indicating that \( \frac{da}{dN} \) is a continuously increasing function of \( \Delta K \), since \( m \) is always larger than one.

A disadvantage of the least square fitting method applied on a log-log plot is that it is very sensitive to the range of growth rate measurements over which the test is run.\(^{13}\) It has been pointed out
that m is not an invariant integer, as assumed in many of the crack growth models. Also, C and m are more or less interdependent and thus a comparison of crack growth rates should always take both parameters into account. Therefore the expression da/dN = c(ΔK)^m is only a useful tool to fit the relationship between da/dN and ΔK within a limited range. In particular this expression without range restriction implies that there is not threshold value for ΔK. The existence of a fatigue limit in steels indicates otherwise.

The advantage of the log scale is that it emphasises low values of da/dN for ΔK values close to the threshold value. These values are most significant for engineers concerned with conditions to achieve long fatigue lives in materials. In this work, however, the range of data studied was fairly limited, the data could therefore be plotted on linear scales and the comparisons were made by superposition of the different graphs. The international system of units has been used:

\[ da/dN: \text{micron/cycle} \quad 1 \text{ml} = 25.4 \text{μm} \quad (10^{-5} \text{ inch} = 0.254 \text{μm}) \]
\[ \Delta K: \text{ mega Nm}^{-3/2} \quad 1 \text{ksi} \sqrt{\text{in}} \sim 1.1 \text{ MNm}^{-3/2} \]

A. Room Temperature Tests

Influence of Grain Size

The fatigue crack growth rates of specimens heat treated to get 50 μm and 15 μm grain sizes and subsequently quenched into iced brine are shown in Fig. 15. In Fig. 16 similar data on specimens heat treated to give 50 μm and 15 μm grain sizes and then cooled very slowly to room temperature are given.
It can be noted that in the slowly cooled alloys, grain sizes in the range studied have little or no effect. But the results of iced brine quenched alloys show considerable scatter. The sample with a 15 µm grain size shows propagation rates at the high and of the scatter band for high ΔK values. This sample appeared to exhibit faster crack growth rates. This was very surprising since a grain size effect, if found, would normally be expected to lead to slower crack growth rates. The discrepancy will be discussed later.

The reason why a variation in grain size largely showed a negligible effect on crack propagation rates can be rationalized as follows. The Fe-8Ni alloy has a wavy slip mode i.e. the stacking fault energy is high thus making cross slip easy. It is well established that only materials with planar slip show an improvement in fatigue life if the grain size is reduced. On the other hand, if the alloy has a planar slip mode the lack of any visible grain size effect seen may be because the grain diameters in this study are on the average much larger than the crack advance per cycle. In a Ti-6Al-4V alloy, when the grain size was reduced from 16 µm to 6 µm, the fatigue strength at 10^6 cycles increased from 23 ksi to 28 ksi. Thus it is reasonable to assume that the variations in the substructure within the grain may cause significant changes in crack propagation (a martensite lath is only about 1 µm wide).
Influence of Substructure

In the 8 Ni alloys, substructure (equiaxed alpha or lath martensite or varying proportions of these two) can be varied independently of grain size simply by varying the cooling rate.

The crack growth results of slowly cooled samples of the alloy are given in Fig. 10. The stress intensity range can vary as much as $10 \text{ MNm}^{-3/2}$ for a given $da/dN$. For example, for a $da/dN$ of 1 μ the stress intensity range may vary from 50 to 60 $\text{MNm}^{-3/2}$ (Fig. 16). Part of this scatter arises from errors in loads and in crack length measurements both of which affect $\Delta K$. Alternatively, for a given value of $\Delta K$ the values of $da/dN$ may differ by a factor of 2. This is much larger than the error in $da/dN$ which arises from the differences in crack growth advances on either side of the specimen (estimated at about 15%). Wider scatter occurs at lower values of $\Delta K$.

Crack growth data for the austenitized slowly cooled and iced brine quenched specimens are compared in Fig. 17. Although the scatter band of the iced brine quenched samples is wider than that for the slowly cooled samples, the resistance to fatigue crack propagation is definitely better in the iced brine quenched samples. Also the crack propagation threshold value of $\Delta K$ is higher in the iced brine quenched samples. The superior properties obtained by quenching is also seen from the fatigue lives in Table IV.

Additional data supporting the above conclusions was the nature of the crack front seen in the fracture toughness samples, Fig. 18a. Note that the shape of crack point implies that crack propagation was definitely slower near the surface. This fracture toughness specimens,
unlike the fatigue specimens, were heat treated after machining. Thus, the surfaces were exposed to very high cooling rates. The surfaces also showed a slightly higher hardness. This suggests that the slower crack propagation rate at the surface was caused either by the greater amount of martensite in the rapidly cooled parts of the specimen or by the residual stresses created by the fast cool.

From the T.T.T. diagram, Fig. 7, it is seen, that minor variations in cooling rates, especially at short times can lead to large differences in the proportion of equiaxed alpha and lath martensite. For example, in Fig. 7, for isothermal transformation at 520°C the amount of γ transformed is about 50% and 90% in 2 sec and 5 sec, respectively. This could explain the wide scatter in the data obtained from the ice brine quenched alloys. This could also explain why the small grain size samples show faster crack growth rates. The hardenability is poorer if the grain size is smaller, and thus this sample would contain more equiaxed alpha and less lath martensite than a larger grain sample for the same cooling rate. This means that the beneficial effect, if any, of small grain size is offset by the fact that less lath martensite occurred in the structure.

The scatter in the data obtained from the iced brine quenched alloy can also be easily explained by variations in the residual stresses created by the quench and left after machining.

The results obtained when two specimens of 12Ni .5Ti alloy with a 15 μ grain size were tested are shown in Fig. 19. In one specimen the crack propagated rapidly in the final stages so that measurements at
large \( \Delta K \) could not be made. The second specimen was given an unintentional initial overload. It will be shown later that any overload merely affects the immediate subsequent growth rate of the crack which then will grow at a rate characteristic of that \( \Delta K \). This alloy showed the highest propagation resistance among all iced brine quenched samples.

The substructure which is usually obtained in the 8%Ni alloy is a mixture of equiaxed \( \alpha \) and lath martensite. It is not possible to get either a completely equiaxed structure or lath martensite structure by varying the cooling rate because changes in cooling rates affect only a limited ratio of the amounts of lath martensite and equiaxed alpha. Therefore a slowly cooled 4%Ni alloy and a iced brine quenched 16%Ni alloy were also tested with the former containing 100% equiaxed structure and the latter structure being completely martensitic. The fatigue tests were run in dry argon at room temperature. It can be seen that the data of the 4%Ni and 16%Ni alloy fall within the scatter band obtained from slowly cooled and ice brine quenched 8Ni alloys. (Fig. 20) The 16%Ni alloy showed propagation rates generally slower than the 4%Ni alloy, but appeared worse when compared to the best values of growth resistance obtained for the 8Ni and 12Ni alloys. This was surprising because a test in a neutral environment was expected to show that a completely martensitic structure has slower propagation rates. The difference between the crack propagation rates of the 4Ni and 16Ni alloys was slight. This was also surprising, even though the 4%Ni alloy was much softer than the 16%Ni alloys. In the 4%Ni alloy, the fatigue crack propagation is arrested when plane strain conditions
are lost. But the crack opening continues because of plastic deformation in the remaining ligament to the point of final failure through general plastic deformation at a $K_{\text{max}}$ of about $56 \text{ MNm}^{-3/2}$. The crack growth rate measurements were limited to the crack length existing prior to the onset of plane stress conditions.

It should be noted that all these alloys showed crack growth resistances comparable to those found in Fe, 5 Mo, .5 C$^{24}$ (Fig. 21).

From these last results it may be concluded that the rate of crack propagation for rates faster than 0.2 $\mu$m/cycle is relatively insensitive to grain size, composition and yield strength in the alloys studied. These results are consistent with the conclusions found in the literature,$^{14,25-32}$ in which it was pointed out that stage II fatigue is insensitive to grain size and yield strength of the alloy, especially in materials showing a wavy slip mode. However, fine grain size can lead to long fatigue lives$^{26,33}$ if crack growth is by mode I probably because a greater amount of plastic deformation is involved.

In the 8 Ni alloy a mixture of lath martensite and equiaxed alpha obtained by an iced brine quench definitely shows a slightly better fatigue crack resistance than equiaxed structures obtained by a slow cool. This result supports Grosskreutz's suggestion$^{26}$ that by homogenizing slip through multiplication of dislocation sources, e.g. by cold working and quenching, the fatigue resistance can be improved, but could also be simply explained by the fact that a fast quench creates residual stresses which can reduce the effective $\Delta K$.

The best fatigue crack resistance among all the iced brine quenched alloys [8 Ni, 12 Ni, 16 Ni] was shown by the grain refined 12 Ni, .5 Ti alloy.
From the fact that the 16 Ni and 4 Ni alloys fatigued in dry argon showed crack growth rates falling in the middle of the scatter band of the 8 Ni alloys fatigued in air, it can be concluded that the effect of environment (air or dry argon) on propagation rates above 0.2 μm/cycle is negligible in these alloys and for these test conditions.

B. Crack Growth Rates Measured from Striations: Low Temperature Tests

It was pointed out earlier that the Fe-Ni-Ti alloys were developed for cryogenic applications and the objective of this study is to determine the fatigue behavior of these alloys at cryogenic temperatures; in particular to determine whether there is any difference in fatigue properties above and below the transition temperature. The transition temperature lies between -77°C and -110°C in the 9Ni, .15Ti alloy\(^{20}\) and below -196°C in the 12Ni, .5Ti alloy\(^{2,7}\).

A practical way to measure crack growth rates at low temperatures is by analyzing the striations seen on the fracture surface. Some examples of the micrographs of striations obtained in this study are shown in Figs. 22-24. Within each grain the striations were curved with their center of curvature towards the point where the crack first entered the grain. The crack directions were slightly different in adjacent grains. A preliminary study showed that there was a slight discrepancy between the crack growth rates measured using the optical method described previously and that obtained by measuring the distance between striations seen on the fracture surface (Fig. 25). This will be discussed in more detail later.
Fatigue Behavior above the Transition Temperature

Crack propagation rates in 1/4 in. thick specimens of 8Ni, 0.15 Ti alloys heat treated to get 50 μm grain size and then quenched in iced brine and fatigued at room temperature and -77°C are given in Fig. 26. Results obtained from the striation interspace measurements are very similar, indicating that the temperature does not seem to have a discernable influence on crack propagation in the 8Ni alloy at temperatures above transition temperature. Further tests of 8Ni alloy were run at -77°C using 1/2 in. and 1 in. thick iced brine quenched specimens. The 1 in. thick specimen was heat treated to get a 15 μm grain size and the grain size obtained was verified by microscopic examination. The 1/2 in. thick specimen contained 50 μm grains and failed in an almost brittle manner as seen from the sharp transition from striations to quasi-cleavage facets (Fig. 31a). The specimen with a 15 μm grain size failed by general plastic deformation as can be seen in Fig. 18b, indicating that the transition temperature is lower for the smaller grain size alloy. From Fig. 27, it can be noted that the fatigue crack data of the 1 in. thick sample (with 15 μm grains) appeared to have the slowest crack growth rate. This is contrary to expectations since previous results had indicated virtually no effect of grain size on crack propagation rates. The slight anomaly can be explained by the effect of the curvature of the crack on the calculated value of ΔK (effect probably more sensitive in thicker specimens) as will be discussed in a later section.
In Fig. 28, crack growth rates (using the striation method) in the 12Ni 0.5Ti alloy heat treated to get 15 μm grain size, and measured at room temperature and liquid nitrogen temperatures are compared. The results were very similar.

The results of all the tests run on the 8Ni and 12Ni alloys above their transition temperatures are given in Fig. 29. The scatter band is fairly wide and no particular trend can be found. Thus the conclusion is that at the high crack growth rates used in this study, the crack propagation rates above 0.2 μm/cycle are independent of nickel content and temperature for test temperatures between room temperature and the transition temperature. These conclusions are consistent with those of Bucci et al., who studied fatigue crack propagation in 5 Ni and 9 Ni steels at cryogenic temperatures, using a compliance technique for converting displacement measurements to crack length.

**Fatigue Behavior Below the Charpy V-notch Transition Temperature**

The 8Ni alloys specimens with two different grain sizes were precracked at room temperature and then fatigued at -116°C (freezing point of ethanol). The results, Fig. 30, show that the crack propagation rates at -116°C are similar to the rates observed at temperatures above the transition temperature. Also variation in grain size did not affect the crack growth rate but the sample with the smaller grain size fractured at larger K values. These, K values for the 8Ni alloy with 50 μm and 15 μm grain sizes are 55±2 MNm$^{-3/2}$ and 61±3 MNm$^{-3/2}$, respectively. Although growth rates were apparently unaffected by testing at -116°C, the fatigue life increased compared to the fatigue lives obtained at
room temperature (Table V). The longer fatigue life is probably due to the longer time required before the precrack starts to move and to slower crack growth rate in the very early stage of growth. An interesting feature observed in this test was that, after the crack had grown to a certain length under fatigue, the specimen fractured in a catastrophic manner by a cleavage mode. Figure 31B shows how the striations were replaced by flat cleavage facets. This explains why the smaller grain size sample fractured at a higher $K_\text{I}$ value since brittle cleavage strength increased if grain size is smaller.

Tests run at liquid nitrogen temperature showed that it was very difficult either to initiate or propagate a crack by fatigue at this temperature. For example for $\Delta K$ values up to $41 \text{ MNm}^{-3/2}$ ($K_{\text{max}} = 45.5 \text{ MNm}^{-3/2}$) the crack would not initiate and raising the load to obtain a $\Delta K$ of $57.6 \text{ MNm}^{-3/2}$ (and a $K_{\text{max}}$ of $64 \text{ MNm}^{-3/2}$) produced a catastrophic brittle fracture. (Fig. 32)

A specimen which was precracked at room temperature, failed on a load application corresponding to a $\Delta K$ of $36.8 \text{ MNm}^{-3/2}$ ($K_{\text{max}} = 41 \text{ MNm}^{-3/2}$) at $-196^\circ\text{C}$ without any further fatigue crack growth. Another precracked specimen fatigued at $\Delta K$ of $32.5 \text{ MNm}^{-3/2}$ ($K_{\text{max}} = 36 \text{ MNm}^{-3/2}$) showed no crack propagation even after 30000 cycles. When this specimen was warmed to room temperature to examine the crack length and recooled to liquid nitrogen temperature, it cracked catastrophically after a few cycles at the same load. The room temperature crack length measurement indicated no crack growth. Thus it is clear that the applied load was very close to the critical stress intensity required for brittle fracture in that specimen. The values of $K_{\text{max}}$ at which the crack propagates
catastrophically in fatigue are compared to the results obtained from standard fracture toughness tests in Table VI. The two ranges of $K_{IC}$ values given at each temperature for the fracture toughness tests were obtained by using the maximum load and the $P_Q$ load which is defined by the intersection of the load deflection curve with the 95% slope line according to the ASTM standards.

The ranges of $K_{IC}$ values rather than a single unique value are given for both fatigue and standard tests at -116°C and for the standard test at -196°C because the crack front was curved. Thus the low value was obtained by taking the crack length value seen on the sides and the high value was for the crack length found in the middle. Because the curvature of the crack front is larger than that allowed in the ASTM specifications, the tests cannot be considered to be valid. Although $K_{IC}$ (fatigue) is closer to $K_{IC}$ (standard test) using $P_Q$, $K_{IC}$ (fatigue) should really be compared to $K_{IC}$ (standard test) using $P_{\text{max}}$ since the fatigue test specimen with a certain crack length, fractured on the next maximum load. The discrepancy between the $K_{IC}$ obtained from the fatigue tests and from tensile tests at -116°C and at -196°C may be due to differences in strain rates, since the fatigue tests were run at 5 cps. This discrepancy can also be due to the statistical distribution of strength. In the specimens tested in a standard test, the precrack is stopped in an arbitrary position, whereas in a fatigue test the crack keeps advancing until it reaches a low strength region with a corresponding low value of $K_{IC}$.

From Table IV it can also be noted that there is a larger difference between $K_{IC}$ (fatigue) and $K_{IC}$ (standard test) at -116°C than at -196°C.
C. Low Crack Growth Rates

Very low crack growth rates were not investigated in this study. Since the striations could not be seen in the S.E.M., they were not measured in the low temperature experiments. Also the above growth rates show considerable scatter in the room temperature experiments. In particular the threshold values seem to vary from 15 to 25 MNm\(^{-3/2}\). The reason is not known. Some uncontrolled factors, for example, residual stressed or variations in the environment might explain the very large scatter in these low crack growth rates at rates less than 0.2 \(\mu m/cycle\). It seems also that at the lowest \(\Delta K\) the crack front becomes curved as shown in Fig. 18a. The crack appears to nucleate at the mid-thickness of the specimen where it propagates more rapidly than at the edges. In consequence the crack is really longer than it looks when seen on both sides of the specimen. This discrepancy between crack growth rates in the center and on the sides of the specimen does not seem to occur so drastically for higher \(\Delta K\).

Very low crack growth rates are also extremely sensitive to any preload. A 8\%Ni alloy 1/4 inch thick specimen was accidently overloaded and subsequently the crack propagation was reduced considerably, and the apparent threshold value of \(\Delta K\) was increased to 35 MNm\(^{-3/2}\). After the crack had advanced 1 or 2 mm, it resumed normal growth behavior. A 12Ni alloy specimen was also accidently overloaded. Subsequently it showed no propagation at all for a \(\Delta K = 35\) MNm\(^{-3/2}\) and the propagation started very slowly at \(\Delta K = 45\) MNm\(^{-3/2}\).
The effect of short periods of high stress cycling on subsequent crack growth under normal stress cycling was studied in detail in a 8Ni 1.0Ti 0.1Al alloy at room temperature and a 12Ni .5Ti alloy at liquid nitrogen temperature. A high stress cycling defined by a $\Delta K$ of 70 MNm$^{-3/2}$ and a $K_{\text{max}}$ of 77 MNm$^{-3/2}$ arrested the subsequent crack advance for $\Delta K$ values up to 35 MNm$^{-3/2}$ in the 8Ni alloy, Fig. 33. After the crack advanced 1 or 2 mm the crack propagation resumed at rates similar to the rates obtained with non overloaded specimens.

The fracture surface of a 12Ni specimen fatigued at liquid nitrogen temperature is shown in Fig. 34. This specimen was alternately fatigued under a $\Delta K$ of 77±3 MNm$^{-3/2}$ (and a $K_{\text{max}}$ of 79±3 MNm$^{-3/2}$) for 10 cycles and then under a $\Delta K$ varying from 36 MNm$^{-3/2}$ to 39 MNm$^{-3/2}$ for 10 000 cycles. It can be seen that the advance of the crack at low $\Delta K$ after high $\Delta K$ cycling is extremely slow. Crack growth rates vary from 0.005 $\mu$m to 0.015 $\mu$m per cycle during low $\Delta K$ cycling. These values are about ten times lower than what is normally obtained for the same $\Delta K$ range prior to high $\Delta K$ cycling.

This delaying effect in crack propagation caused by intermittent high loads has been frequently observed. The high loads produce residual compressive stresses at the crack tip which reduce the subsequent crack growth at lower amplitudes. This effect persists until the crack grows beyond the residual stressed region.
D. Comparison of the Crack Growth Rates Measured by the Travelling Telescope and that Obtained from Striations

Previous studies of programmed loading have clearly proved that there is a one to one correspondence between striations and load cycles.\textsuperscript{37,39} McMillan and Pelloux\textsuperscript{38} also showed, using the same method, that fatigue cracking occurs only during a load rise when the crack opens. The mechanism of stage II fatigue has been extensively studied, especially on aluminum alloys, and there is good agreement that a stage II crack has a blunt shape during the tensile part of a fatigue cycle and is less blunt in the compression part, thereby forming a striation. However, although some workers\textsuperscript{40,41} found good correlation between the striation spacings and microscopic growth rates, others\textsuperscript{13} have found discrepancies. These will be discussed for low $\Delta K$ and for high $\Delta K$ conditions.

For tests run at $\Delta K$ values less than 34 $\text{MNm}^{-3/2}$, the striation spacing measurements gave crack growth rates up to two times larger than those measured optically.\textsuperscript{42,43} It was concluded that the above discrepancy existed because the crack front advance was not continuous. The present study also showed that for $\Delta K$ of 30-35 $\text{MNm}^{-3/2}$ the interstriation spacing indicated faster growth rates. Part of the problem may be that interstriation measurements are made on very small areas and thus are sensitive to slight differences in crack growth rates along the crack front.
However, at high growth rates or with large stress intensity factors, striation spacings indicate slower crack growth rates than those obtained from optical measurements, (Fig. 25 and Fig. 35). This has been pointed out by previous investigators. Bathias and Pelloux explained the above result by stating that fatigue crack growth at high growth rates involves void formation and tearing by ductile fracture. This proposal is not very satisfactory because for large growth rates especially, the striations are continuously visible across the whole width of the specimen. An alternative explanation can be offered.

A specimen in which the precracked crack length can be definitely established as, for example, in a fatigue specimen marked by oxidation or in a fracture toughness specimen, shows that during the early stage of fatigue, the crack front assumes a significantly curved shape with the center of curvature on the crack nucleation side of the crack front (Fig. 18). In one 25.4 mm thick (1 in. thick) specimen the crack can propagate up to 2 mm in the middle without any crack advance close to the surface. This indicates that during the early stage of fatigue where $\Delta K$ is close to the threshold $\Delta K$, the crack propagates much faster at the mid-thickness of the specimen where the pictures were taken with the SEM than close to the surface. Thus for low crack propagation rates the striation spacings measured close to the middle of the specimen indicate crack growth rates which are systematically higher than the rates measured along the surface. After this early stage, i.e. for $\Delta K = 35 - 45 \text{ MNm}^{-3/2}$, the crack front appears to retain a constant shape thus the propagation rates both within and on the specimen surface are in good agreement.
The discrepancy at high propagation rates can be explained by the fact that a large $\Delta K$ often corresponds to a long crack length and thus the calculated value of $\Delta K$ becomes sensitive to the value of the crack length $a$ as shown in Fig. 13. It is reasonable to assume that the crack front shape has become constant and can be represented by the sketch shown in Fig. 36. The two curved lines represent the position of the crack before and after a load cycle.

For a given propagation rate, $da/dN$, a measured crack length will be different at different positions within the specimen thickness. Let $a_1$ be the average length of the crack determined by optical method of the crack position seen on the surfaces, and let $a_2$ be the crack length measured at mid-thickness on the fracture surface using the SEM. In general $a_2 > a_1$ (Fig. 36). Therefore, for a given value of $da/dN$ the calculated value of $\Delta K$ will be larger with the striation method than the optical method.

When striations were measured close to the side of the specimen the crack growth rates obtained were very close to those obtained by optical measurements with a travelling telescope (Fig. 25). Also the discrepancy in growth rates obtained by two different methods are less pronounced with thinner specimens.

It may be concluded that because the crack front is not a straight line perpendicular to the edges of the specimen, systematical errors are made in the estimation of $\Delta K$ values.

When the fracture surface is rough at large $\Delta K$ values, care should be taken when measuring the interstriation distance since the angle at which the SEM photomicrograph was taken can affect the results.
It is obvious then that, by judicious choice of specimen position for the analysis of striation interspaces, the striation method is a valid means of measuring crack growth rates. The method can be expected to give consistent results for crack growth in the bulk of the specimen, since the value of the crack length used for the calculation of $\Delta K$ is no more arbitrary than the value used in the optical method.
CONCLUSIONS

1. The tensile properties of 4Ni, 8Ni, 12Ni, 16Ni alloys and 5Mo, .3C steels are different, but the fatigue crack growth rates in those alloys fall within the same scatter band, indicating that the crack growth rates are relatively insensitive to tensile properties and composition.

2. In the 8Ni alloy, variation of grain size from 50μ to 15μ has no significant effect on crack propagation rates.

3. An equiaxed alpha structure, obtained by a very slow cool, was found slightly less resistant to fatigue crack propagation than a structure containing martensite, which was obtained by an iced brine quench. This slight effect could be due to either the change in substructure or the residual stress associated with the production of martensite. Among all the iced brine quenched samples, the 12Ni 0.5Ti alloy showed the best fatigue crack resistance.

4. The crack growth data for alloys containing from 4 to 16Ni fatigued in air or argon fall inside the same large scatter band. Thus, at fast crack growth rates (0.2 μm/cycle) the air or argon environment does not have a large effect on crack growth.

5. Crack growth rates at room temperature, -77°C, -116°C are similar in the 8Ni alloys containing 15μ and 50μ grain sizes. The transition temperature of the 8Ni alloy is about -100°C. Thus the crack growth rates are not affected by temperature even at temperatures
somewhat below the transition temperature. In the 12Ni alloy crack
growth rates at room temperature and -196°C are similar. The transition
temperature of the 12Ni alloy is below -196°C.

6. Although temperature has no significant effect on crack growth
rates at rates above 0.2 μm, the fatigue life increased significantly
with decreasing temperature. This is probably due to the long times
required before the precracked cracks start to move and also to the
slower initial propagation at low temperatures.

7. Although the fatigue crack growth rates at different temperatures
remained about the same fracture toughness decreased with decrease in
temperature. Thus, crack growth rates cannot be related to \( K_{IC} \) values.

8. It proved impossible to initiate or to propagate a crack by
fatigue in the 8 Ni alloy at liquid nitrogen temperature. Brittle
fracture occurred when the cyclic load increased above certain values.

9. The critical values of fracture toughness calculated from the
fatigue test data were lower than those obtained from the standard tests.
This difference may be attributed to the effect of strain rates or to
the statistical distribution of strength. In the specimens tested in
a standard test, the precrack is stopped in an arbitrary position,
whereas in a fatigue test the crack keeps advancing until it reaches a
low strength region with a corresponding low value of \( K_{IC} \).

10. A high overload cycle retards crack propagation at subsequent
lower ΔK cycles at room temperature for the 8Ni alloys. Similar behavior
was observed in the 12Ni alloy tested at -196°C. This effect is probably
due to the residual stresses left by the high loads. After the crack
had slowly grown beyond the residual stress region, crack propagation was
observed to resume normal behavior.
11. A good correlation between the standard optical technique and the striation technique has been obtained especially for cracks growing faster than 0.2 μm/cycle. The systematic difference in the two methods has been explained by the fact that the crack front is curved.
ACKNOWLEDGEMENTS

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The assistance of the support staff of the Inorganic Materials Research Division was also greatly appreciated. The author wishes to thank all those who, by their efficient help and constant readiness, contributed to make his work in the laboratory most enjoyable.

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Data obtained from slowly cooled 8Ni alloys.

**SPECIMEN 123**

8Ni, 0.15Ti, 0.07Al, Bal. Fe
50 MICRON GRAINS
TEST RUN AT 25 DEGREES C

**THICKNESS 1.000 INCH**

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**SPECIMEN 124**

8Ni, 0.15Ti, 0.07Al, Bal. Fe
50 MICRON GRAINS
TEST RUN AT 25 DEGREES C

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Data obtained from slowly cooled 8M1 alloys.

**SPECIMEN: 121**

8M1, 0.15Ti, 0.07Al Bal. Fe

15 MICRON GRAINS

TEST RUN AT 25 DEGREES C

THICKNESS 1.000 INCH

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</table>

**SPECIMEN: 122**

8M1, 0.15Ti, 0.07Al Bal. Fe

15 MICRON GRAINS

TEST RUN AT 25 DEGREES C

THICKNESS 1.000 INCH

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK MM^N-3/2</th>
<th>DADN MICRON</th>
</tr>
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<tbody>
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<td>37.357</td>
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<td>5400.0</td>
<td>35.40</td>
<td>79.616</td>
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</tbody>
</table>
Data obtained from slowly cooled 8Ni alloys.

**Linear Regression of \( \frac{da}{dN} \) against \( \Delta K \)**

<table>
<thead>
<tr>
<th>IND. VAR(I)</th>
<th>COEF.(I)</th>
<th>S.E. COEF.</th>
<th>T-VALUE</th>
<th>R(I)SQRD</th>
<th>MIN X(I)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>-1.0580</td>
<td>0.056</td>
<td>12.23</td>
<td>0.9988</td>
<td>15.994</td>
</tr>
<tr>
<td>1</td>
<td>0.0403</td>
<td>0.017</td>
<td>23.08</td>
<td>0.9988</td>
<td>2.832</td>
</tr>
</tbody>
</table>

**ESTIMATED COVARIANCE MATRIX OF REGRESSION COEFFICIENTS**

\[
\begin{pmatrix}
0 & 7.4389 \times 10^{-3} \\
1 & -1.4326 \times 10^{-4} & 3.0487 \times 10^{-6}
\end{pmatrix}
\]

**SUM OF Observations**

- 58 observations
- \( \frac{da}{dN} = 0.0403 \Delta K - 1.0580 \)

**SUM OF RESIDUAL VARIABLES**

- 1 variable
- 56 degrees of freedom
- 532.5506

**SUM OF SQUARES**

- Residual root mean square: 0.2305
- Residual mean square: 0.0439
- Residual sum of squares: 2.4574
- Sum of squares about mean: 25.4268
- Mult. Correl. Coef. squared: 0.9049
- Mult. Correl. Coef.: 0.9512

**Linear Regression of Loge(\( \frac{da}{dN} \)) against Loge(\( \Delta K \))**

<table>
<thead>
<tr>
<th>IND. VAR(I)</th>
<th>COEF.(I)</th>
<th>S.E. COEF.</th>
<th>T-VALUE</th>
<th>R(I)SQRD</th>
<th>MIN X(I)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>-3.6465</td>
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<td>11.92</td>
<td>0.9988</td>
<td>15.994</td>
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<tr>
<td>1</td>
<td>1.5628</td>
<td>1.121</td>
<td>13.95</td>
<td>0.9988</td>
<td>2.832</td>
</tr>
</tbody>
</table>

**ESTIMATED COVARIANCE MATRIX OF REGRESSION COEFFICIENTS**

\[
\begin{pmatrix}
0 & 1.3174 \times 10^{-6} \\
1 & -4.73 \times 10^{-6} & 1.2559 \times 10^{-6}
\end{pmatrix}
\]

**SUM OF OBSERVATIONS**

- 58 observations
- \( \frac{da}{dN} = (\Delta K)^{1.5628} \)

**SUM OF RESIDUAL VARIABLES**

- 1 variable
- 56 degrees of freedom
- 194.4702

**SUM OF SQUARES**

- Residual root mean square: 0.3211
- Residual mean square: 0.1031
- Residual sum of squares: 5.7743
- Sum of squares about mean: 25.8268
- Mult. Correl. Coef. squared: 0.7764
- Mult. Correl. Coef.: 0.8811
Data obtained from iced brine quenched 8Ni alloys.

**SPECIMEN 112**

8Ni, 0.15Ti, 0.06Al Bal. Fe

15 MICRON GRAINS
TEST RUN AT 25 DEGREES C

THICKNESS .500 INCH

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK MM*MM-3/2</th>
<th>DADN MICRON</th>
</tr>
</thead>
<tbody>
<tr>
<td>1 3000.0</td>
<td>22.90</td>
<td>35.057</td>
<td>.130</td>
</tr>
<tr>
<td>2 3000.0</td>
<td>24.40</td>
<td>38.047</td>
<td>.160</td>
</tr>
<tr>
<td>3 3000.0</td>
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<td>43.557</td>
<td>.300</td>
</tr>
<tr>
<td>4 3000.0</td>
<td>28.60</td>
<td>49.435</td>
<td>.730</td>
</tr>
<tr>
<td>5 3000.0</td>
<td>29.90</td>
<td>51.967</td>
<td>.590</td>
</tr>
<tr>
<td>6 3000.0</td>
<td>30.80</td>
<td>54.335</td>
<td>.650</td>
</tr>
<tr>
<td>7 3000.0</td>
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</tr>
<tr>
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<td>34.00</td>
<td>66.042</td>
<td>1.640</td>
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<tr>
<td>9 3000.0</td>
<td></td>
<td>77.107</td>
<td>1.770</td>
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</table>

**SPECIMEN 126**

8Ni, 0.15Ti, 0.06Al Bal. Fe

50 MICRON GRAINS
TEST RUN AT 25 DEGREES C

THICKNESS 1.000 INCH

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK MM*MM-3/2</th>
<th>DADN MICRON</th>
</tr>
</thead>
<tbody>
<tr>
<td>1 6000.0</td>
<td>22.30</td>
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<td>.060</td>
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<tr>
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<td>23.80</td>
<td>36.802</td>
<td>.160</td>
</tr>
<tr>
<td>4 6000.0</td>
<td>24.80</td>
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</table>
Data obtained from iced brine quenched 8Ni Alloys.

**SPECIMEN 221**

8Ni, 1.0Ti Bal. Fe

50 MICRON GRAINS

TEST RUN AT 25 DEGREES C

THICKNESS 1.000 INCH

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK MN*MM-3/2</th>
<th>DADN MICRON</th>
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**SPECIMEN 22200**

8Ni, 1.0Ti Bal. Fe

50 MICRON GRAINS

TEST RUN AT 25 DEGREES C

THICKNESS 1.000 INCH

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK MN*MM-3/2</th>
<th>DADN MICRON</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
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<td>29.00</td>
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</table>
Data obtained from iced brine quenched 8Ni alloys.

**SPECIMEN 136**

8Ni, 0.15Ti, 0.05Al Bal Fe  
50 MICRON GRAINS  
TEST RUN AT 25 DEGREES C  
THICKNESS 1.005 INCH

<table>
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<th>DELTAK MN*M-3/2</th>
<th>DADN MICRON</th>
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</table>
Data obtained from iced brine quenched large grains 8Ni alloys.

Linear Regression of da/dN against ΔK

<table>
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<tr>
<th>IND. VAR(I)</th>
<th>COEF.B(I)</th>
<th>S.E. COEF.</th>
<th>T-VALUE</th>
<th>P(I)</th>
<th>MIN X (I)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
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<td>.0025</td>
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ESTIMATED COVARIANCE MATRIX OF REGRESSION COEFFICIENTS

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</thead>
<tbody>
<tr>
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</tr>
<tr>
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</table>

NC. OF OBSERVATIONS: 60
NC. OF IND. VARIABLES: 1
RESIDUAL DEGREES OF FREEDOM: 58
F-VALUE: 283.0043  da/dN = 0.418 ΔK-1.3276

RESIDUAL ROOT MEAN SQUARE: .3296
RESIDUAL MEAN SQUARE: 1.086
RESIDUAL SUM OF SQUARES: 6.3009
SUM OF SQ ABOUT MEAN: 37.0453
MULT. CORREL. COEF. Squared: .8299
MULT. CORREL. COEF.: .9110

Linear Regression of Log_e(da/dN) against Log_e(ΔK)

<table>
<thead>
<tr>
<th>IND. VAR(I)</th>
<th>COEF.B(I)</th>
<th>S.E. COEF.</th>
<th>T-VALUE</th>
<th>P(I)</th>
<th>MIN X (I)</th>
</tr>
</thead>
<tbody>
<tr>
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</table>

ESTIMATED COVARIANCE MATRIX OF REGRESSION COEFFICIENTS

<table>
<thead>
<tr>
<th></th>
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<th>1</th>
</tr>
</thead>
<tbody>
<tr>
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<td>2.39431E-02</td>
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</table>

NC. OF OBSERVATIONS: 60
NC. OF IND. VARIABLES: 1
RESIDUAL DEGREES OF FREEDOM: 58
F-VALUE: 137.5599  da/dN = 0.001976(ΔK)1.8148

RESIDUAL ROOT MEAN SQUARE: .4352
RESIDUAL MEAN SQUARE: .1894
RESIDUAL SUM OF SQUARES: 10.9871
SUM OF SQ ABOUT MEAN: 37.0453
MULT. CORREL. COEF. Squared: .7034
MULT. CORREL. COEF.: .8397
Data obtained from iced brine quenched alloys.

SPECIMEN 141
8Ni, 0.15Ti, 0.06Al Bal. Fe
15 MICRON GRAINS
TEST RUN AT 25 DEGREES C
THICKNESS 1.000 INCH

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK MM**3/2</th>
<th>DAHN MICRON</th>
</tr>
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<td>33.60</td>
<td>74.257</td>
</tr>
<tr>
<td>13</td>
<td>6000.0</td>
<td>34.00</td>
<td>77.107</td>
</tr>
</tbody>
</table>
Data obtained from iced brine quenched 15 μm grain alloy.

### Linear regression of $\frac{da}{dN}$ against $\Delta K$

<table>
<thead>
<tr>
<th>IND. VAR(1)</th>
<th>CCEF. B(1)</th>
<th>S.E. CCEF.</th>
<th>T-VALUE</th>
<th>F(1)SQ.</th>
<th>N: X(I)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>-1.5236</td>
<td>.1752</td>
<td>8.70</td>
<td>.0032</td>
<td>15.99</td>
</tr>
</tbody>
</table>

**ESTIMATED COVARIANCE MATRIX OF REGRESSION COEFFICIENTS**

\[
\begin{bmatrix}
C & 0 \\
0 & 1
\end{bmatrix}
\]

\[
\begin{bmatrix}
1 & -3.0339E-02 \\
-3.0339E-02 & 1.0403E-05
\end{bmatrix}
\]

### Linear regression of $\log_2(\frac{da}{dN})$ against $\log_2(\Delta K)$

<table>
<thead>
<tr>
<th>IND. VAR(1)</th>
<th>CCEF. B(1)</th>
<th>S.E. CCEF.</th>
<th>T-VALUE</th>
<th>F(1)SQ.</th>
<th>N: X(I)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>-7.4020</td>
<td>.5341</td>
<td>7.54</td>
<td>.2533</td>
<td>9.13</td>
</tr>
</tbody>
</table>

**ESTIMATED COVARIANCE MATRIX OF REGRESSION COEFFICIENTS**

\[
\begin{bmatrix}
C & 0 \\
0 & 1
\end{bmatrix}
\]

\[
\begin{bmatrix}
1 & -2.4493E-01 \\
-2.4493E-01 & 6.4268E-02
\end{bmatrix}
\]

### Linear regression of $\log_2(\frac{da}{dN})$ against $\log_2(\Delta K)$

\[
\text{da/dN} = 0.0516(\Delta K) - 1.5236
\]

### Linear regression of $\log_2(\frac{da}{dN})$ against $\log_2(\Delta K)$

\[
da/dN = 0.000370(\Delta K)^2 + 2.3262
\]
Data obtained from the 42Ni Alloy

SPECIMEN 4
4Ni, 0.15Ti, 0.1Al Bal. Fe
50 MICRON GRAINS Air Cooled
TEST RUN AT 25 DEGREES C in a flow of dry argon.

THICKNESS .550 INCH

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK MNEM-3/2</th>
<th>DADN MICRON</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>2700.0</td>
<td>22.20</td>
<td>27.648</td>
</tr>
<tr>
<td>2</td>
<td>2700.0</td>
<td>23.00</td>
<td>28.836</td>
</tr>
<tr>
<td>3</td>
<td>2700.0</td>
<td>23.90</td>
<td>30.276</td>
</tr>
<tr>
<td>4</td>
<td>2700.0</td>
<td>24.80</td>
<td>31.843</td>
</tr>
<tr>
<td>5</td>
<td>3600.0</td>
<td>26.00</td>
<td>45.555</td>
</tr>
<tr>
<td>6</td>
<td>3600.0</td>
<td>27.20</td>
<td>49.095</td>
</tr>
</tbody>
</table>
Data obtained from the 4Ni slowly cooled alloy.

Linear regression of \( \frac{da}{dN} \) against \( \Delta K \)

<table>
<thead>
<tr>
<th>IND. VAR(I)</th>
<th>COEF.R(I)</th>
<th>S.E. COEF.</th>
<th>T-VALUE</th>
<th>R(I)SQRT</th>
<th>MIN X(I)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>-0.8529</td>
<td>0.0952</td>
<td>8.96</td>
<td>0.92</td>
<td>27.648</td>
</tr>
</tbody>
</table>

Estimated covariance matrix of regression coefficients

\[
\begin{bmatrix}
0 & 1 \\
1 & -7.4169E-04 & 6.7854E-06
\end{bmatrix}
\]

- Number of observations: 6
- Number of independent variables: 1
- Residual degrees of freedom: 4
- F-value: 148.8553
- Residual root mean square: 0.0542
- Residual mean square: 0.0029
- Residual sum of squares: 0.0117
- Sum of sq. about mean: 4487
- Multi. correl. coeff. squared: 0.9738
- Multi. correl. coeff.: 0.9868

Linear regression of \( \log_e(\frac{da}{dN}) \) against \( \log_e(\Delta K) \)

<table>
<thead>
<tr>
<th>IND. VAR(I)</th>
<th>COEF.R(I)</th>
<th>S.E. COEF.</th>
<th>T-VALUE</th>
<th>R(I)SQRT</th>
<th>MIN X(I)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>-3.9222</td>
<td>0.4397</td>
<td>8.92</td>
<td>0.957</td>
<td>3.320</td>
</tr>
</tbody>
</table>

Estimated covariance matrix of regression coefficients

\[
\begin{bmatrix}
0 & 1 \\
1 & -5.43347E-02 & 1.53307E-02
\end{bmatrix}
\]

- Number of observations: 6
- Number of independent variables: 1
- Residual degrees of freedom: 4
- F-value: 91.5517
- Residual root mean square: 0.0685
- Residual mean square: 0.0047
- Residual sum of squares: 0.0188
- Sum of sq. about mean: 4487
- Multi. correl. coeff. squared: 0.9581
- Multi. correl. coeff.: 0.9788
Data obtained from the 16%wt Ni Alloy

SPECIMEN 16000

16Ni, 0.15Ti, 0.10Al Bal. Fe
50 MICRON GRAINS
TEST RUN AT 25 DEGREES C
THICKNESS .550 INCH

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK MM/M-3/2</th>
<th>DADN MICRON</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>3300.0</td>
<td>22.10</td>
<td>33.617</td>
</tr>
<tr>
<td>2</td>
<td>3300.0</td>
<td>23.10</td>
<td>35.433</td>
</tr>
<tr>
<td>3</td>
<td>3300.0</td>
<td>23.70</td>
<td>36.601</td>
</tr>
<tr>
<td>4</td>
<td>3300.0</td>
<td>25.20</td>
<td>39.826</td>
</tr>
<tr>
<td>5</td>
<td>3300.0</td>
<td>26.50</td>
<td>43.057</td>
</tr>
<tr>
<td>6</td>
<td>3300.0</td>
<td>27.20</td>
<td>45.004</td>
</tr>
<tr>
<td>7</td>
<td>3300.0</td>
<td>27.60</td>
<td>46.192</td>
</tr>
<tr>
<td>8</td>
<td>3900.0</td>
<td>28.20</td>
<td>56.831</td>
</tr>
<tr>
<td>9</td>
<td>3900.0</td>
<td>29.10</td>
<td>62.532</td>
</tr>
<tr>
<td>10</td>
<td>3900.0</td>
<td>30.20</td>
<td>65.701</td>
</tr>
<tr>
<td>11</td>
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<td>31.20</td>
<td>71.129</td>
</tr>
<tr>
<td>12</td>
<td>3900.0</td>
<td>32.10</td>
<td>76.710</td>
</tr>
<tr>
<td>13</td>
<td>3900.0</td>
<td>33.10</td>
<td>83.809</td>
</tr>
<tr>
<td>14</td>
<td>3900.0</td>
<td>34.10</td>
<td>91.999</td>
</tr>
</tbody>
</table>
Data obtained from the 16Ni iced brine quenched alloy.

Linear regression of $\frac{da}{dn}$ against $\Delta K$

<table>
<thead>
<tr>
<th>IND. VAR(I)</th>
<th>COEF. R(I)</th>
<th>S.E. COEF.</th>
<th>T-VALUE</th>
<th>R(I) SQRT</th>
<th>MIN X(I)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$c$</td>
<td>-1.9570</td>
<td>0.047</td>
<td>4.84</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

ESTIMATED COVARIANCE MATRIX OF REGRESSION COEFFICIENTS

<table>
<thead>
<tr>
<th></th>
<th>0</th>
<th>1</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>1.6379E-01</td>
<td></td>
</tr>
<tr>
<td>1</td>
<td>-2.6307E-03</td>
<td>4.6831E-05</td>
</tr>
</tbody>
</table>

NO. OF OBSERVATIONS: 14
NO. OF IND. VARIABLES: 1
RESIDUAL DEGREES OF FREEDOM: 12

F-VALUE: 69.3458 $da/dn = 0.0532\Delta K - 1.9570$
RESIDUAL ROOT MEAN SQUARE: 0.4736
RESIDUAL MEAN SQUARE: 0.2243
RESIDUAL SUM OF SQUARES: 2.6910

SUM OF SQ ABOUT MEAN: 16.2237
MULT. CORREL. COEF. SQUARED: 0.8341
MULT. CORREL. COEF.: 0.5133

Linear regression of $\log_e(\frac{da}{dn})$ against $\log_e(\Delta K)$

<table>
<thead>
<tr>
<th>IND. VAR(I)</th>
<th>COEF. R(I)</th>
<th>S.E. COEF.</th>
<th>T-VALUE</th>
<th>R(I) SQRT</th>
<th>MIN X(I)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$c$</td>
<td>-10.6121</td>
<td>1.8772</td>
<td>5.55</td>
<td></td>
<td></td>
</tr>
<tr>
<td>$x$</td>
<td>2.8782</td>
<td>0.1707</td>
<td>6.12</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

ESTIMATED COVARIANCE MATRIX OF REGRESSION COEFFICIENTS

<table>
<thead>
<tr>
<th></th>
<th>0</th>
<th>1</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>3.5238E+00</td>
<td></td>
</tr>
<tr>
<td>1</td>
<td>-2.8053E+01</td>
<td>2.2151E+01</td>
</tr>
</tbody>
</table>

NO. OF OBSERVATIONS: 14
NO. OF IND. VARIABLES: 1
RESIDUAL DEGREES OF FREEDOM: 12

F-VALUE: 37.3971 $da/dn = 0.000030(\Delta K)2.8782$
RESIDUAL ROOT MEAN SQUARE: 0.5731
RESIDUAL MEAN SQUARE: 0.3284
RESIDUAL SUM OF SQUARES: 3.9412

SUM OF SQ ABOUT MEAN: 16.2237
MULT. CORREL. COEF. SQUARED: 0.7571
MULT. CORREL. COEF.: 0.8701
Data obtained from the 12Ni 0.5Ti Alloy

SPECIMENS 1005  Two specimens.

12Ni 0.5Ti  bal. Fe

15 MICRON GRAINS  IB Quenched

TEST RUN AT  25 DEGREES C

THICKNESS 1.000 INCH

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK MN*M-3/2</th>
<th>DADN MICRON</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>4500.0</td>
<td>22.30</td>
<td>25.476</td>
</tr>
<tr>
<td>2</td>
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</tr>
<tr>
<td>5</td>
<td>4500.0</td>
<td>24.80</td>
<td>29.189</td>
</tr>
<tr>
<td>6</td>
<td>4500.0</td>
<td>25.40</td>
<td>30.220</td>
</tr>
<tr>
<td>7</td>
<td>4500.0</td>
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<td>31.319</td>
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<td>8</td>
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<td>26.50</td>
<td>32.292</td>
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<td>27.00</td>
<td>33.323</td>
</tr>
<tr>
<td>10</td>
<td>4500.0</td>
<td>27.60</td>
<td>34.644</td>
</tr>
<tr>
<td>11</td>
<td>4500.0</td>
<td>28.10</td>
<td>35.821</td>
</tr>
<tr>
<td>12</td>
<td>4500.0</td>
<td>28.60</td>
<td>37.076</td>
</tr>
<tr>
<td>13</td>
<td>4500.0</td>
<td>29.20</td>
<td>38.693</td>
</tr>
<tr>
<td>14</td>
<td>4500.0</td>
<td>29.70</td>
<td>40.143</td>
</tr>
<tr>
<td>15</td>
<td>4500.0</td>
<td>30.40</td>
<td>42.346</td>
</tr>
<tr>
<td>16</td>
<td>4500.0</td>
<td>31.40</td>
<td>45.888</td>
</tr>
<tr>
<td>17</td>
<td>4500.0</td>
<td>32.40</td>
<td>49.966</td>
</tr>
<tr>
<td>18</td>
<td>4500.0</td>
<td>33.40</td>
<td>54.669</td>
</tr>
<tr>
<td>19</td>
<td>7500.0</td>
<td>23.40</td>
<td>45.011</td>
</tr>
<tr>
<td>20</td>
<td>7500.0</td>
<td>24.10</td>
<td>46.769</td>
</tr>
<tr>
<td>21</td>
<td>7500.0</td>
<td>25.20</td>
<td>49.782</td>
</tr>
<tr>
<td>22</td>
<td>7500.0</td>
<td>26.60</td>
<td>54.156</td>
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<td>57.740</td>
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<tr>
<td>24</td>
<td>7500.0</td>
<td>28.10</td>
<td>59.702</td>
</tr>
<tr>
<td>25</td>
<td>7500.0</td>
<td>28.60</td>
<td>61.793</td>
</tr>
<tr>
<td>26</td>
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</tr>
<tr>
<td>27</td>
<td>7500.0</td>
<td>30.10</td>
<td>68.960</td>
</tr>
<tr>
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<td>7500.0</td>
<td>31.50</td>
<td>77.117</td>
</tr>
<tr>
<td>29</td>
<td>7500.0</td>
<td>32.50</td>
<td>84.011</td>
</tr>
</tbody>
</table>
Data obtained using the striation method

**SPECIMEN 131CJ**

8Ni 0.15Ti 0.07Al Bal.
15 MICRON GRAINS IB Quenched
TEST RUN AT -116 DEGREES C

THICKNESS 1.005 INCH

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTA K MN*M-3/2</th>
<th>DADN MICRON</th>
<th>Negative Number</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>6000.0</td>
<td>26.00</td>
<td>41.551</td>
<td>.335</td>
</tr>
<tr>
<td>2</td>
<td>6000.0</td>
<td>29.10</td>
<td>47.524</td>
<td>.290</td>
</tr>
<tr>
<td>3</td>
<td>6000.0</td>
<td>28.40</td>
<td>48.510</td>
<td>.310</td>
</tr>
</tbody>
</table>

**SPECIMEN 135CJ**

8Ni .15Ti 007Al
50 MICRON GRAINS IB Quenched
TEST RUN AT -116 DEGREES C

THICKNESS 1.005 INCH

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTA K MN*M-3/2</th>
<th>DADN MICRON</th>
<th>Negative Number</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>4800.0</td>
<td>22.60</td>
<td>27.467</td>
<td>.013 (optical method)</td>
</tr>
<tr>
<td>2</td>
<td>5400.0</td>
<td>23.80</td>
<td>32.957</td>
<td>.068</td>
</tr>
<tr>
<td>3</td>
<td>6000.0</td>
<td>27.10</td>
<td>44.493</td>
<td>.400</td>
</tr>
<tr>
<td>4</td>
<td>6000.0</td>
<td>28.30</td>
<td>48.177</td>
<td>.300</td>
</tr>
<tr>
<td>5</td>
<td>6000.0</td>
<td>29.20</td>
<td>51.334</td>
<td>.450</td>
</tr>
</tbody>
</table>
Data obtained using the striation method

**SPECIMEN 11402**

8Ni 0.15Ti 0.06Al Bal. Fe  
50 MICRON GRAINS IB Quenched  
TEST RUN AT -77 DEGREES C

**THICKNESS .500 INCH**

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK NN*M-3/2</th>
<th>DADN MICRON</th>
<th>Negative Number</th>
</tr>
</thead>
<tbody>
<tr>
<td>1 3000.0</td>
<td>24.40</td>
<td>38.047</td>
<td>.300</td>
<td>114-2=2,3</td>
</tr>
<tr>
<td>2 3000.0</td>
<td>25.40</td>
<td>40.294</td>
<td>.400</td>
<td>114-2=5,6</td>
</tr>
<tr>
<td>3 3000.0</td>
<td>26.60</td>
<td>43.325</td>
<td>.270</td>
<td>114-2=8,9</td>
</tr>
<tr>
<td>4 3000.0</td>
<td>27.30</td>
<td>45.295</td>
<td>.320</td>
<td>114-2=11,12</td>
</tr>
<tr>
<td>5 3000.0</td>
<td>28.60</td>
<td>49.435</td>
<td>.400</td>
<td>114-2=13,14</td>
</tr>
<tr>
<td>6 3000.0</td>
<td>30.30</td>
<td>56.024</td>
<td>.500</td>
<td>114-2=15,16</td>
</tr>
<tr>
<td>7 3000.0</td>
<td>31.10</td>
<td>59.697</td>
<td>.770</td>
<td>114-2=17,18</td>
</tr>
<tr>
<td>8 3000.0</td>
<td>32.60</td>
<td>67.805</td>
<td>1.350</td>
<td>114-2=19,20</td>
</tr>
<tr>
<td>9 3000.0</td>
<td>33.80</td>
<td>75.661</td>
<td>1.380</td>
<td>114-2=21,22</td>
</tr>
<tr>
<td>10 3000.0</td>
<td>35.00</td>
<td>84.585</td>
<td>1.500</td>
<td>114-2=23,24</td>
</tr>
</tbody>
</table>

**SPECIMEN 14200**

8Ni .15Ti 0.06Al Bal. Fe  
15 MICRON GRAINS IB Quenched  
TEST RUN AT -77 DEGREES C

**THICKNESS 1.000 INCH**

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK NN*M-3/2</th>
<th>DADN MICRON</th>
<th>Negative Number</th>
</tr>
</thead>
<tbody>
<tr>
<td>1 6000.0</td>
<td>23.80</td>
<td>36.802</td>
<td>.300</td>
<td>142-1</td>
</tr>
<tr>
<td>2 6000.0</td>
<td>24.80</td>
<td>38.519</td>
<td>.230</td>
<td>142-2</td>
</tr>
<tr>
<td>3 6000.0</td>
<td>27.30</td>
<td>46.808</td>
<td>.600</td>
<td>142-3</td>
</tr>
<tr>
<td>4 6000.0</td>
<td>29.50</td>
<td>52.735</td>
<td>.330</td>
<td>142-4</td>
</tr>
<tr>
<td>5 6000.0</td>
<td>30.70</td>
<td>57.810</td>
<td>.600</td>
<td>142-5</td>
</tr>
<tr>
<td>6 6000.0</td>
<td>32.50</td>
<td>67.209</td>
<td>.600</td>
<td>142-6</td>
</tr>
<tr>
<td>7 6000.0</td>
<td>34.00</td>
<td>77.107</td>
<td>.900</td>
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<td>8 6000.0</td>
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</table>
Data obtained using the striation method

**SPECIMEN 5**

12Ni, 0.5Ti Bal. Fe

15 MICRON GRAINS

TEST RUN AT 25 DEGREES C

THICKNESS 1.000 INCH

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK MN*^3/2</th>
<th>DAMN MICRON</th>
<th>Negative Number</th>
</tr>
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<tbody>
<tr>
<td>1</td>
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<td>30.044</td>
<td>0.140</td>
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<tr>
<td>2</td>
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<td>27.80</td>
<td>35.106</td>
<td>0.150</td>
</tr>
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**SPECIMEN 4**

12Ni, 0.5Ti Bal Fe

15 MICRON GRAINS

TEST RUN AT -198 DEGREES C

THICKNESS 1.000 INCH

<table>
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<th>CRACK LENGTH MM</th>
<th>DELTAK MN*^3/2</th>
<th>DAMN MICRON</th>
<th>Negative Number</th>
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</table>
Data obtained using the striation method

**SPECIMEN 11201**

Ni 0.15Ti 0.06Al Bal Fe

50 MICRON GRAINS IB Quenched

TEST RUN AT 25 DEGREES C

THICKNESS .500 INCH

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK MM=M-3/2</th>
<th>DADN MICRON</th>
<th>Number</th>
</tr>
</thead>
<tbody>
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**SPECIMEN 11301**

Ni 0.15Ti 0.06Al Bal Fe

50 MICRON GRAINS IB Quenched

TEST RUN AT 25 DEGREES C

THICKNESS .250 INCH

<table>
<thead>
<tr>
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<th>DELTAK MM=M-3/2</th>
<th>DADN MICRON</th>
<th>Number</th>
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</thead>
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<tr>
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<td>0.700</td>
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<td>32.10</td>
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**SPECIMEN 11502**

Ni 0.15Ti 0.06Al Bal Fe

50 MICRON GRAINS IB Quenched

TEST RUN AT -77 DEGREES C

THICKNESS .250 INCH

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<th>Number</th>
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<tr>
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<td>32.90</td>
<td>69.645</td>
<td>0.850</td>
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</table>
Data obtained using the striation method

**SPECIMEN 30**

12Ni 0.5Ti Bal. Fe

15 MICRON GRAINS
TEST RUN AT 196 DEGREES C

THICKNESS 1.000 INCH

<table>
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<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK MN*M-3/2</th>
<th>DADN MICRON</th>
<th>Negative Number</th>
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</thead>
<tbody>
<tr>
<td>1</td>
<td>22.50</td>
<td>74.371</td>
<td>1.500</td>
<td>003-7</td>
</tr>
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<td>1.500</td>
<td>003-3</td>
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<td>23.77</td>
<td>79.606</td>
<td>1.200</td>
<td>003-8</td>
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<td>25.30</td>
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**SPECIMEN 1**

12Ni 0.5Ti Bal. Fe

15 MICRON GRAINS
TEST RUN AT 25 DEGREES C

THICKNESS 1.000 INCH

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
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<th>DELTAK MN*M-3/2</th>
<th>DADN MICRON</th>
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</thead>
<tbody>
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</tr>
<tr>
<td>5</td>
<td>7500.0</td>
<td>30.10</td>
<td>68.960</td>
</tr>
<tr>
<td>6</td>
<td>7500.0</td>
<td>31.10</td>
<td>74.622</td>
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<td>7500.0</td>
<td>32.10</td>
<td>81.136</td>
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</tbody>
</table>
Data obtained using the striation method

Linear regression of $\frac{da}{dn}$ against $\Delta K$

<table>
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<tr>
<th>$t$</th>
<th>COEF. $b(1)$</th>
<th>S.E. COEF.</th>
<th>T-VALUE</th>
<th>F(1) SQED</th>
<th>MIN $X(1)$</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
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<td>2.6919</td>
<td>10.45</td>
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</table>

**ESTIMATED COVARIANCE MATRIX OF REGRESSION COEFFICIENTS**

<table>
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<th>1</th>
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</thead>
<tbody>
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</tr>
<tr>
<td>1</td>
<td>-1.2637E-04</td>
<td>2.27526E-06</td>
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**NO. OF OBSERVATIONS**: 70

**NO. OF INCL. VARIABLES**: 1

**RESIDUAL DEGREES OF FREEDOM**: 68

**F-VALUE**: 356.4762

**RESIDUAL TOTAL MEAN SQUARE**: 1.954

**RESIDUAL MEAN SQUARE**: 0.0386

**RESIDUAL SUM OF SQUARES**: 2.6228

**SUM OF SUM ABOUT MEAN**: 16.3721

**MULT. CORREL. COEF. SQUARED**: 0.8398

**MULT. CORREL. COEF.**: 0.9164

---

Linear Regression of $\log_e(\frac{da}{dn})$ against $\log_e(\Delta K)$

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<thead>
<tr>
<th>$t$</th>
<th>COEF. $b(1)$</th>
<th>S.E. COEF.</th>
<th>T-VALUE</th>
<th>F(1) SQED</th>
<th>MIN $X(1)$</th>
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</thead>
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<td>0.0009</td>
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**ESTIMATED COVARIANCE MATRIX OF REGRESSION COEFFICIENTS**

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**NO. OF OBSERVATIONS**: 70

**NO. OF INCL. VARIABLES**: 1

**RESIDUAL DEGREES OF FREEDOM**: 68

**F-VALUE**: 237.7232

**RESIDUAL TOTAL MEAN SQUARE**: 2314

**RESIDUAL MEAN SQUARE**: 0.0386

**RESIDUAL SUM OF SQUARES**: 3.6415

**SUM OF SUM ABOUT MEAN**: 16.3721

**MULT. CORREL. COEF. SQUARED**: 0.7776

**MULT. CORREL. COEF.**: 0.9818
Data obtained from overloaded specimens

**SPECIMEN 223**

8Ni, 1.0Ti Bal. Fe Alloy
50 MICRON GRAINS iced brine quenched
TEST RUN AT 25 DEGREES C

**THICKNESS 1.000 INCH**

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK MN*M-3/2</th>
<th>DADN MICRON</th>
</tr>
</thead>
<tbody>
<tr>
<td>1 4500.0</td>
<td>22.40</td>
<td>25.609</td>
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</tr>
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<td>23.00</td>
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<td>.010</td>
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<td>4 7500.0</td>
<td>23.40</td>
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<td>.150</td>
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<td>14 7500.0</td>
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**SPECIMEN 001**

12Ni, .5Ti Bal. Fe Alloy
15 MICRON GRAINS iced brine quenched
TEST RUN AT 25 DEGREES C

**THICKNESS 1.000 INCH**

<table>
<thead>
<tr>
<th>LOAD POUNDS</th>
<th>CRACK LENGTH MM</th>
<th>DELTAK MN*M-3/2</th>
<th>DADN MICRON</th>
</tr>
</thead>
<tbody>
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<td>.050</td>
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Data obtained from overloaded specimens (Cont.)

**SPECIMEN 11501**

8Ni 0.15Ti 0.06Al Bal. Fe Alloy

50 MICRON GRAINS

TEST RUN AT 25 DEGREES C

THICKNESS .250 INCH

<table>
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<td>15</td>
<td>1500.0</td>
<td>30.00</td>
<td>54.749</td>
</tr>
<tr>
<td>16</td>
<td>1500.0</td>
<td>32.60</td>
<td>67.805</td>
</tr>
</tbody>
</table>
Data obtained from overloaded specimen.

Linear regression of \( \frac{da}{dn} \) against \( \Delta K \)

<table>
<thead>
<tr>
<th>IND. VAR(I)</th>
<th>COEF. R(I)</th>
<th>S.E. COEF.</th>
<th>T-VALUE</th>
<th>R(I) SQRN</th>
<th>MIN X(I)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>-2.2649</td>
<td>.0592</td>
<td>16.21</td>
<td>0</td>
<td>-34.872</td>
</tr>
</tbody>
</table>

Estimated Covariance Matrix of Regression Coefficients

\[
\begin{bmatrix}
0 & 1.95129 \times 10^{-2} \\
1 & -7.32515 \times 10^{-4} & 6.3464 \times 10^{-6}
\end{bmatrix}
\]

No. of Observations: 30
No. of Ind. Variables: 1
Residual Degrees of Freedom: 28
F-Value: 551.7912
Residual Root Mean Square: .2505
Residual Mean Square: 0.0627
Residual Sum of Squares: 1.7565
Sum of Sq About Mean: 36.3709
Multi. Correl. Coef.: .9756

Linear regression of \( \log_e (\frac{da}{dn}) \) against \( \log_e (\Delta K) \)

<table>
<thead>
<tr>
<th>IND. VAR(I)</th>
<th>COEF. R(I)</th>
<th>S.E. COEF.</th>
<th>T-VALUE</th>
<th>R(I) SQRN</th>
<th>MIN X(I)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>-11.6296</td>
<td>.6604</td>
<td>13.55</td>
<td>0</td>
<td>3.552</td>
</tr>
</tbody>
</table>

Estimated Covariance Matrix of Regression Coefficients

\[
\begin{bmatrix}
0 & 7.40241 \times 10^{-1} \\
1 & -7.88341 \times 10^{-1} & 4.82336 \times 10^{-2}
\end{bmatrix}
\]

No. Observations: 30
No. of Ind. Variables: 1
Residual Degrees of Freedom: 28
F-Value: 212.3683
Residual Root Mean Square: .3490
Residual Mean Square: .1513
Residual Sum of Squares: 4.2368
Sum of Sq About Mean: 36.3709
Multi. Correl. Coef.: .9406
Table I. Composition of Alloys.

<table>
<thead>
<tr>
<th>Ni</th>
<th>Ti</th>
<th>Al</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>4</td>
<td>.15</td>
<td>.6</td>
<td>bal</td>
</tr>
<tr>
<td>8</td>
<td>.15</td>
<td>.6</td>
<td>bal</td>
</tr>
<tr>
<td>8</td>
<td>1.0</td>
<td>.6</td>
<td>bal</td>
</tr>
<tr>
<td>12</td>
<td>.5</td>
<td>bal</td>
<td></td>
</tr>
<tr>
<td>16</td>
<td>.15</td>
<td>.6</td>
<td>bal</td>
</tr>
</tbody>
</table>

Usual figures obtained with the pure metals used in this laboratory are:
- C: 0.001 wt%
- N₂: 0.004 wt%
- O₂: 0.002 wt%
Table II. Heat treatments and substructures.

<table>
<thead>
<tr>
<th>Ni Content wt%</th>
<th>Ti Content wt%</th>
<th>Grain Size</th>
<th>Cooling Rate (Substructure)</th>
</tr>
</thead>
<tbody>
<tr>
<td>4</td>
<td>0.15</td>
<td>50μ</td>
<td>slowly cooled (equiaxed)</td>
</tr>
<tr>
<td>8</td>
<td>0.15</td>
<td>50μ</td>
<td>slowly cooled</td>
</tr>
<tr>
<td>8</td>
<td>0.15 or 1.0</td>
<td>50μ 15μ</td>
<td>ice brine quenched</td>
</tr>
<tr>
<td>12</td>
<td>0.5</td>
<td>15μ</td>
<td>ice brine quench (lath martensite)</td>
</tr>
<tr>
<td>16</td>
<td>0.15</td>
<td>50μ</td>
<td>ice brine quench (lath martensite)</td>
</tr>
</tbody>
</table>
Table III. Tensile tests
Cross head speed 0.1 cm/mm
Yield strength measured at 0.2% strain

<table>
<thead>
<tr>
<th></th>
<th>+25°C</th>
<th></th>
<th></th>
<th>-77°C</th>
<th></th>
<th></th>
<th>-196°C</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Y.S.</td>
<td>T.S.</td>
<td>Elong-</td>
<td>Y.S.</td>
<td>T.S.</td>
<td>Elong-</td>
<td>Y.S.</td>
</tr>
<tr>
<td></td>
<td>ksi</td>
<td>ksi</td>
<td>(%)</td>
<td>ksi</td>
<td>ksi</td>
<td>(%)</td>
<td>ksi</td>
</tr>
<tr>
<td>4% Ni</td>
<td>29.3</td>
<td>41.2</td>
<td>31</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>8% Ni</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>slowly cooled</td>
<td>68.4</td>
<td>78.1</td>
<td>24</td>
<td>77.5</td>
<td>89.5</td>
<td>26</td>
<td>114.7</td>
</tr>
<tr>
<td>8% Ni</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>brine quenched</td>
<td>63.1</td>
<td>72.9</td>
<td>23</td>
<td>74.8</td>
<td>85</td>
<td>26</td>
<td>116.4</td>
</tr>
<tr>
<td>8% Ni</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>IBQ smaller grain</td>
<td>63.3</td>
<td>71.6</td>
<td>20</td>
<td>74.4</td>
<td>85.9</td>
<td>22.4</td>
<td>108.5</td>
</tr>
<tr>
<td>12% Ni</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>.5%</td>
<td>86.5</td>
<td>92.6</td>
<td>21</td>
<td>98.2</td>
<td>106.9</td>
<td>--</td>
<td>134.9</td>
</tr>
<tr>
<td>16% Ni</td>
<td>92.0</td>
<td>97.5</td>
<td>10.8</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
Table IV. Fatigue lives of the different specimens (room temperature)

<table>
<thead>
<tr>
<th>Specimen #</th>
<th>Thickness</th>
<th>Grain Size</th>
<th>Cooling rate</th>
<th>Maximum load (lbs)</th>
<th>Equivalent max. load for 1 in. thickness</th>
<th>Life</th>
<th>Initial crack length (mm)</th>
<th>Comments</th>
</tr>
</thead>
<tbody>
<tr>
<td>8Ni 112-1</td>
<td>1/2</td>
<td>50</td>
<td>IBQ</td>
<td>3000</td>
<td>6000</td>
<td>36536</td>
<td>22.3</td>
<td></td>
</tr>
<tr>
<td>8Ni 126</td>
<td>1</td>
<td>50</td>
<td>IBQ</td>
<td>5400</td>
<td></td>
<td>55000</td>
<td>22.7</td>
<td>precracking with p(_{\text{max}}) = 6000</td>
</tr>
<tr>
<td>8Ni 136</td>
<td>1</td>
<td>50</td>
<td>IBQ</td>
<td>4500</td>
<td></td>
<td>65000</td>
<td>21.8</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>6000</td>
<td></td>
<td>10835</td>
<td>21.8 (26.0)</td>
<td></td>
</tr>
<tr>
<td>8Ni 141</td>
<td>1</td>
<td>15</td>
<td>IBQ</td>
<td>4500</td>
<td></td>
<td>30000</td>
<td>21.7</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>6000</td>
<td></td>
<td>18100</td>
<td>21.7 (23.8)</td>
<td></td>
</tr>
<tr>
<td>8Ni 123</td>
<td>1</td>
<td>50</td>
<td>slowly cooled</td>
<td>4200</td>
<td></td>
<td>66774</td>
<td>21.7</td>
<td></td>
</tr>
<tr>
<td>8Ni 124</td>
<td>1</td>
<td>50</td>
<td>slowly cooled</td>
<td>6000</td>
<td></td>
<td>22500</td>
<td>22.1</td>
<td></td>
</tr>
<tr>
<td>8Ni 121</td>
<td>1</td>
<td>15</td>
<td>slowly cooled</td>
<td>3000</td>
<td></td>
<td>20000</td>
<td>21.8</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>4000</td>
<td></td>
<td>10000</td>
<td>21.8 (21.8)</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>3000</td>
<td></td>
<td>70000</td>
<td>22.0 (22.0)</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>4500</td>
<td></td>
<td>15000</td>
<td>23.8 (23.8)</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>6000</td>
<td></td>
<td>10000</td>
<td>28.1 (28.1)</td>
<td></td>
</tr>
<tr>
<td>8Ni 122</td>
<td>1</td>
<td>15</td>
<td>slowly cooled</td>
<td>5400</td>
<td></td>
<td>35000</td>
<td>22.9</td>
<td>precracking with p(_{\text{max}}) = 6000</td>
</tr>
<tr>
<td>8Ni 221</td>
<td>1</td>
<td>50</td>
<td>IBQ</td>
<td>4000</td>
<td></td>
<td>50000</td>
<td>21.7</td>
<td></td>
</tr>
<tr>
<td>(1 Ti)</td>
<td></td>
<td></td>
<td></td>
<td>5000</td>
<td></td>
<td>10000</td>
<td>24.2</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>6000</td>
<td></td>
<td>17300</td>
<td>25.5</td>
<td></td>
</tr>
<tr>
<td>Specimen #</td>
<td>Thickness</td>
<td>Grain Size</td>
<td>Cooling rate</td>
<td>Maximum load (lbs)</td>
<td>Equivalent max. load for 1 in. thickness</td>
<td>Life</td>
<td>Initial crack length</td>
<td>Comments</td>
</tr>
<tr>
<td>------------</td>
<td>------------</td>
<td>------------</td>
<td>--------------</td>
<td>--------------------</td>
<td>----------------------------------------</td>
<td>------</td>
<td>----------------------</td>
<td>----------</td>
</tr>
<tr>
<td>8Ni 222 (1 Ti)</td>
<td>1</td>
<td>50</td>
<td>IBQ</td>
<td>6000</td>
<td>39616</td>
<td>21.7</td>
<td></td>
<td></td>
</tr>
<tr>
<td>4Ni</td>
<td>.55</td>
<td>50</td>
<td>Slowly cooled</td>
<td>2700, 3600</td>
<td>4920, 6550</td>
<td>37000</td>
<td>5000</td>
<td>(25.2)</td>
</tr>
<tr>
<td>16Ni</td>
<td>.55</td>
<td>50</td>
<td>IBQ</td>
<td>3300, 3900</td>
<td>6000, 7100</td>
<td>34000</td>
<td>5200</td>
<td>(27.4)</td>
</tr>
<tr>
<td>12Ni 005</td>
<td>1</td>
<td>15</td>
<td>IBQ</td>
<td>4500</td>
<td></td>
<td>94278</td>
<td></td>
<td>21.6</td>
</tr>
</tbody>
</table>
Table V. Fatigue lives of the specimens (low temperatures)

<table>
<thead>
<tr>
<th>Specimen #</th>
<th>Thickness</th>
<th>Grain size</th>
<th>Cooling rate</th>
<th>Temp. °C</th>
<th>Maximum load (lbs)</th>
<th>Equivalent max. load for 1 in. thickness (lbs)</th>
<th>Life # cycle</th>
<th>Initial crack length (mm)</th>
<th>Comments</th>
</tr>
</thead>
<tbody>
<tr>
<td>8Ni 114-2</td>
<td>1/2</td>
<td>50</td>
<td>IBQ</td>
<td>-77</td>
<td>3000</td>
<td>6000</td>
<td>28100</td>
<td>~ 22.7</td>
<td>Precracked at RT with $\rho_{max} = 3000$ lbs.</td>
</tr>
<tr>
<td>8Ni 115-2</td>
<td>1/4</td>
<td>50</td>
<td>IBQ</td>
<td>-77</td>
<td>1500</td>
<td>6000</td>
<td>34000</td>
<td>~ 22.7</td>
<td>Precracked at RT with $\rho_{max} = 1500$ lbs.</td>
</tr>
<tr>
<td>8Ni 142</td>
<td>1.</td>
<td>15</td>
<td>IBQ</td>
<td>-77</td>
<td>6000</td>
<td></td>
<td>50300</td>
<td>21.8</td>
<td>Not precracked</td>
</tr>
<tr>
<td>8Ni 131</td>
<td>1.005</td>
<td>15</td>
<td>IBQ</td>
<td>-116</td>
<td>6000</td>
<td></td>
<td>41200</td>
<td>22.7</td>
<td>Precracked at RT with $\rho_{max} = 5400$ lbs.</td>
</tr>
<tr>
<td>8Ni 135</td>
<td>1.005</td>
<td>50</td>
<td>IBQ</td>
<td>-116</td>
<td>4800</td>
<td></td>
<td>25000</td>
<td>22.5</td>
<td>Precracked at RT with $\rho_{max} = 5400$ lbs.</td>
</tr>
<tr>
<td>12Ni 004</td>
<td>1.</td>
<td>15</td>
<td>IBQ</td>
<td>-196</td>
<td>7500</td>
<td></td>
<td>22000</td>
<td>21.7</td>
<td>Not precracked</td>
</tr>
</tbody>
</table>
Table VI  
Fracture toughness data  
*(do not meet the ASTM requirements)*  

8 Ni 50 jm grain size ice brine quenched specimens  

<table>
<thead>
<tr>
<th>Temperature</th>
<th>$K_I$ failure for fatigue tests 5%</th>
<th>$K_I$ failure for standard tests</th>
</tr>
</thead>
<tbody>
<tr>
<td>-116°C</td>
<td>53 to 57</td>
<td>71 to 97 (using $P_Q$)</td>
</tr>
<tr>
<td></td>
<td></td>
<td>88.5 to 121 (using $P_{max}$)</td>
</tr>
<tr>
<td>-196°C</td>
<td>36</td>
<td>40 to 50 (using $P_Q$)</td>
</tr>
<tr>
<td></td>
<td></td>
<td>43 to 54 (using $P_{max}$)</td>
</tr>
</tbody>
</table>

$\text{MNm}^{-3/2}$  

12 Ni 15µm grain $K_I$ for fatigue 6% 100 - 110 $\text{MNm}^{-3/2}$  

$1 \text{ ksi} \sqrt{\text{in.}} = 1.1 \text{ MNm}^{-3/2}$
REFERENCES


37. P. Forsyth and D. Ryder, Fatigue Fracture Aircraft, Engineering, 32, 96 (1960).


Fig. 1. Iron nickel transformation diagram obtained from continuous heating and cooling data (From Hansel: Constitution of in binary alloys McGraw Hill, 1958)
Fig. 2. Photo-micrograph a grain refined 12Ni .5Ti alloy. This alloy was austenitized at 900°C cooled in air to room temperature, then reheated to 700°C and iced brine quenched.
Fig. 3. Photo-micrographs of the 16Ni(A) and 4Ni(B) alloys. The 16Ni alloy contains a martensitic substructure, whereas the 4Ni alloy shows an equiaxed substructure. (Courtesy of Dr. M. Yokota)
Fig. 4. Mixed martensitic and equiaxed substructure (A) obtained in the 8Ni alloy which was austenitized at 900°C iced brine quenched, reheated to 700°C and iced brine quenched. In B, the specimen was reheated to 735°C and iced brine quenched. Note the refined prior austenitic grains in B. This indicates that the two phase region is below 735°C.
Fig. 5. Optical micrograph of the grain refined 8Ni alloy. The specimen was austenitized at 900°C for 2 hours, iced brine quenched and reheated to 750°C for one hour and again quenched in iced brine.
Fig. 6. Optical micrographs of the 8Ni 1.0Ti alloy. The specimen was austenitized at 900°C for 2 hours and quenched into iced brine. The grain size is about 50 μm. The structure contains martensite and equiaxed alpha.
Fig. 7. The curves show, approximately, the amount of austenite transformed versus time when the specimens of the 8Ni .15Ti alloy were held at different temperatures in a salt bath.
Fig. 8. An approximate TTT diagram of the 8Ni .15Ti alloy, obtained using a magnetic permeability technique.
Fig. 9. Optical micrograph of the 8Ni alloy.

a) 900°C × 2 hours IBQ + 750°C × 1 hour IBQ.

b) 900°C × 2 hours slow cool to room temperature + 750°C × 1 hour slow cool to 560°C + 18 hours at 560°C + slow cool.
TENSILE SPECIMEN

SCALE: 2:1

Fig. 10. The tensile specimen used in this study.
Fig. 11. The standard W.O.L specimen used in this study.
Fig. 12. Crack growth data obtained using 1/4 in., 1/2 in. and 1 in. thick specimens.
Fig. 13. The figure shows the variation in $Y(a/W)$ because of change in $a/W$. 
Fig. 14. A schematic diagram showing how the crack length was averaged from the different lengths measured on the two sides of the specimen.
Fig. 15. Fatigue crack growth data obtained from six 8Ni specimens which were quenched in iced brine. Five specimens contained 50 μm and one contained 15 μm grains.
Fig. 16. Fatigue crack growth data obtained from four 8Ni alloys which were slowly cooled. Two specimens contained 50 μm grains and the two other specimens contained 15 μm grains.
Fig. 17. Fatigue crack growth data obtained from all the 8Ni specimens are compared.
Fig. 18.  

a) Fracture surfaces seen on $K_{IC}$ specimens at $-77^\circ C$, $-116^\circ C$, and $-196^\circ C$.  

b) Fracture surfaces seen on specimens which failed under fatigue at $-77^\circ C$, $-116^\circ C$, and $-196^\circ C$.  

XBB 748-5376
Fig. 19. Fatigue crack growth rates seen in the 12Ni .5Ti alloy which was grain refined and then quenched in iced brine. The lowest dark point is due to an accidental overload and can be ignored. The dashed lines, and the full lines represent the limit of the scatter bands obtained when the 8Ni alloys were tested in the slowly cooled and iced brine quenched conditions. Note that the 12Ni alloy shows slightly greater fatigue resistance than the 8Ni alloys.
Fig. 20. Crack growth data obtained from the 16Ni and 4Ni alloy fatigued in a flow of dry argon. The two sets of lines represent the limit of the scatter bands obtained with the 8Ni alloys as described for Fig. 19.
Fig. 21. The data obtained by Horn\textsuperscript{24} using a 5Mo 0.3C steel are compared to the results obtained for the Fe-Ni alloys. The dashed and full lines show the limits of the scatter bands of the 8Ni as described for Fig. 19.
Fig. 22. Photomicrographs of the fatigue fracture surface of the 12Ni-5Ti fatigued at liquid nitrogen temperature taken using a scanning electron microscope.

(A. $\Delta K = 50.1 \text{ MNm}^{-3/2}$)  
(B. $\Delta K = 51.6 \text{ MNm}^{-3/2}$)  
(C. $\Delta K = 55.2 \text{ MNm}^{-3/2}$)  
(D. $\Delta K = 58.9 \text{ MNm}^{-3/2}$)

and photomicrographs taken from carbon replicas using a transmission electron microscope (a. $\Delta K = 44 \text{ MNm}^{-3/2}$)  
b. $\Delta K = 50 \text{ MNm}^{-3/2}$, c. $\Delta K = 71 \text{ MNm}^{-3/2}$, d. $\Delta K = 87 \text{ MNm}^{-3/2}$)
Fig. 22.
Fig. 22 (Cont.)
Fig. 22 (Cont.)
Fig. 23. Fatigue fracture surfaces of a 8Ni quenched specimen with a 50 μm grain size fatigued at -116°C as seen through a scanning electron microscope.
A. ΔK=44.5 MNm$^{-3/2}$  B. ΔK=48.2 MNm$^{-3/2}$  C. ΔK=51.3 MNm$^{-3/2}$
Fig. 24. Fatigue fracture surface of a 8Ni quenched specimen with a 50 μm grain size fatigued at room temperature examined using a scanning electron microscope.

A. $\Delta K = 44.1 \text{ MNm}^{-3/2}$  
B. $\Delta K = 42.7 \text{ MNm}^{-3/2}$  
C. $\Delta K = 66.0 \text{ MNm}^{-3/2}$  
D. $\Delta K = 70.9 \text{ MNm}^{-3/2}$
Fig. 25. Fatigue growth data obtained using the striation method and the optical method are compared. An 1/2 in. thick specimen of the quenched 8Ni alloy with a 50 μm grain size was used.
Fig. 26. Fatigue data obtained using the striation method on 1/4 in. thick specimens of the 8Ni alloy tested at room temperature and -77°C.
Fig. 27. Comparison of the results obtained using 1/4 in., 1/2 in. and 1 in. thick specimens of the 8Ni alloy tested at -77°C.
Fig. 28. Fatigue data obtained on testing the 12Ni alloy at 25°C and -196°C. Crack growth rates were measured using the striation method.
Fig. 29. The results of all tests run at 25°C, -77°C and -196°C in which the striation method was used to derive the fatigue crack growth rate are compared.
Fig. 30. Fatigue data obtained on testing the 8Ni alloy at -116°C.
Fig. 31. (A) shows the transition from fatigue striations to quasi cleavage in an 8Ni alloy tested at -77°C. (B) shows the transition from fatigue striations to cleavage facets in an 8Ni alloy tested at -116°C.
Fig. 32. Brittle fracture surface of the 8Ni alloy tested at -196°C.
Fig. 33. Effect of a high ΔK cycling on subsequent cycling at lower ΔK in the 8Ni 1.0Ti alloy fatigued at room temperature. The large dark square represents the crack growth rate corresponding to the high ΔK cycling at 0.5 cycles/cycle.
Fig. 34. Micrographs taken through the SEM of the 12Ni alloy fatigued at -196°C. The specimen had been precracked at room temperature. At -196°C the specimen was fatigued under a high load for about 10 cycles and then for 10 000 cycles under half the original load, and the above sequence was repeated.
Fig. 35. The scatter band of the optical method shows the results obtained using quenched 8Ni alloys whereas the scatter band of the striation method shows the results obtained on the same alloy and also includes data obtained on quenched 12Ni alloy.
Fig. 36. Schematic diagram illustrating a curved crack front.
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