7-1980

Microcrack Initiation and Growth

W. L. Morris
Rockwell International

M. R. James
Rockwell International

Follow this and additional works at: http://lib.dr.iastate.edu/cnde_yellowjackets_1979
Part of the Materials Science and Engineering Commons

Recommended Citation
http://lib.dr.iastate.edu/cnde_yellowjackets_1979/32

This document is brought to you for free and open access by the Interdisciplinary Program for Quantitative Flaw Definition Annual Reports at Iowa State University Digital Repository. It has been accepted for inclusion in Proceedings of the DARPA/AFML Review of Progress in Quantitative NDE, July 1978–September 1979 by an authorized administrator of Iowa State University Digital Repository. For more information, please contact digirep@iastate.edu.
A Monte Carlo simulation technique to predict the mean and scatter in fatigue lifetime of certain 2000 and 7000 series aluminum alloys is described. The method is based upon models of surface microcrack nucleation and early growth, which comprise the initiation phase of the fatigue failure process. The experimental basis for the models is discussed, and examples of predicted numbers and lengths of microcracks developed during fatigue of Al 2219-T851 are compared with experiments. Scatter in lifetime and mean lifetime predictions are also discussed. These are obtained from repeated running of the fatigue simulation, for which scatter in lifetime results from the effect of local microscopic variations in alloy microstructure on the time of crack initiation. The association of the method with a "retirement for cause" philosophy is noted.

INTRODUCTION

For many structural applications, the fatigue lifetime may be dominated by the crack initiation phase of the fatigue failure process. This is particularly true for components in "fail safe" usage, but applies to a significant number of "safe life" designs as well - for which the intent is to inspect components during service with sufficient frequency to remove damaged components before catastrophic failure. The large scatter in fatigue lifetime common to the crack initiation stage of fatigue failure presents a problem, both in the design and in the inspection of such components. In design, performance/cost considerations may dictate less conservative designs and a key design decision then hinges upon prediction of the effect of reduced design safety margin on early component failure. With respect to NDE, microcracking in the initiation stage is a challenge to detect, and the time of transition from initiation, to macroscopic cracking, to failure can be abrupt and sensitive to the statistics of the failure process. Conventional design practice is reasonably well equipped to specify component performance near the mean component lifetime. An approach is needed, however, to reliably predict the probability of infrequent early component failures.

Component fatigue lifetime may be sensitive to alloy surface condition, to residual stresses induced in manufacture and modified in use, and to alloy microstructure. In principle, at least, surface condition and residual stresses introduce scatter in lifetime only if they are uncharacterized. In contrast, the effect of the statistics of the microscopic failure mechanisms on lifetime is inherently statistical in nature and must be dealt with in that vein. Reported here are the initial results of a technique to predict the scatter in lifetimes which arises from the statistical nature of the microscopic fatigue failure process of certain aluminum alloys. It is hoped that, in the future, the method may serve as a model for similar prediction procedures for other structural alloys. In the longer term, our concept is to integrate the lifetime prediction procedure with a contemporary computer aided design facility so that a designer can obtain "real time" predictions of the fatigue performance of a component during the design process.
of approximately one grain size in length. It is
the nucleation, early growth and coalescence of
such cracks which comprise the initiation phase of
fatigue failure. Nucleation in this alloy commonly
occurs by fracture of intermetallic particles at or
near the surface as the case illustrated in
Fig. 1c. By definition, initiation ends with the
formation of a macrocrack of sufficient size that
its subsequent propagation rate can be estimated
with conventional fracture mechanics. During
initiation, rates of microcrack nucleation and
growth are highly sensitive to the alloy micro­
structure in the region of a microcrack site.
Scatter in the time of initiation reflects the
statistics of nucleation and growth as determined
by the possible distribution in microstructure
which can occur in regions in a component vulner­
able to crack initiation. Naturally, the initiation
events illustrated in Fig. 1b have occurred
much too late to have any appreciable effect on
lifetime. It was similar events early in the for­
mation of the 0.015 in. crack which were a prin­
cipal factor in determining fatigue life.

Fatigue lifetime for the alloys of interest may
be predicted using a Monte Carlo simulation of
microcrack nucleation, early growth and coalescence.
The computation scheme is illustrated in Fig. 2.

![Fig. 2 Schematic illustration of steps in com­
puter simulation of alloy fatigue failure.](image)

A finite element analysis is used to identify
regions in a tentative component design vulner­
able to fatigue, and to calculate the distribution of
surface stress in those areas. An additional input
to the calculation is the material orientation and
"average" microstructure in the vulnerable areas.
Microstructure includes the distribution in grain
and intermetallic particle sizes in the material,
relative to the surface stress axis. The fatigue
model combines a Monte Carlo simulation of initia­
tion with a conventional fracture mechanics calcu­
lation of lifetime spent in propagation of a micro­
crack to a critical size. By repeated running of
the simulation, we obtain the scatter and mean in
fatigue lifetime. At intermediate points in the
lifetime of an individual specimen, a prediction of
the numbers and lengths of the surface microcracks
can be obtained. This provides simulation data for
a sensitive comparison between model and experi­
ment. Such data also define the nature of cracking
distributions which must be detected by NDE methods
to determine the state of fatigue of a component at
that point.

**INITIATION MODELS AND MICROSCOPIC
COMPARISONS OF SIMULATION TO EXPERIMENT**

The models used to predict microscopic nuclea­
tion and early growth behavior are briefly des­
cribed. Independent tests of the models are also
shown, along with selected comparisons of micro­
cracking parameters obtained experimentally and by
simulation. With respect to crack coalescence, our
simulation employs a rather simple view of the pro­
cess. It considers only coalescence of cracks loc­
cated initially, either in neighboring grains or
separated by no more than one intervening grain.
Our rationale is that coalescence of more widely
spread cracks comes too late in the initiation pro­
cess to have any substantial effect on lifetime.
Additionally, we assume that the presence of one
microcrack does not affect the propagation of
another until their tips are in a common grain.
There is some experimental evidence to support such
a model, as is discussed below for propagation of
isolated cracks. The key factor is that the plastic
deformation at a microcrack tip is substantially
confined inside the grain containing the
microcrack for a wide range in grain sizes and
crack lengths.

Nucleation. A dislocation based model has been pro­
posed by Chang et al. which relates the rate of
cracking of intermetallic particles during fatigue
to the accumulated strain energy density at a po­
tential nucleation site. A two step process is
identified: I) fracture of the intermetallic; II)
propagation of the crack into the matrix. Each
process is described by its own rate equation.

With reference to Fig. 3 for microscopic nomen­
clature, the earliest number of cycles \( N_1 \) at which

\[
N_1 = \frac{C_0}{D^3W^3(t_{eff-o})^2} \tag{1}
\]

Subsequent cracking into the matrix can occur no
earlier than

\[
N_{11} = \frac{C_0'}{D^3W(t_{eff-o})^2} \tag{2}
\]
$C_0, C_0'$ and $t_0$ are material constants. $D$ is the maximum slip distance at $45^\circ$ to the stress axis in the grain containing the intermetallic. $W$ is the intermetallic width normal to the stress axis.

$T_{off}$ is an effective surface shear stress which is calculated from the grain's crystallographic orientation and external load, and accounts for the effect of grain orientation on the propensity for slip.

The predicted trend is, therefore, for nucleation to occur first in the largest grains (large $D$) at the largest intermetallics, and to proceed to smaller $D, W$ with increasing fatigue. This behavior is modified, however, by the crystallographic orientation of the grains which can lead to instances of early nucleation at small $D, W$ sites for favorable grain orientations.

To calculate an average nucleation rate, it is necessary to integrate over all possible $D$ and $W$ combinations which can occur as determined by the probability distribution of these quantities for the alloy microstructure. Results of such a calculation are shown in Fig. 4. Plotted are the number of broken intermetallics as a function of number of fatigue cycles. (Thus, the figure describes the step I nucleation process.) Data are shown for four cyclic stress amplitudes for an Al 2219-T851 alloy. Fatigue was terminated when the largest microcrack reached approximately 400µm. The solid curves are from the computer simulation, the dots are experimental points. Several features should be noticed. Cyclic amplitude effects crack density by several orders of magnitude. Nucleation begins early in the specimen lifetime even for the smallest stress amplitude. The decreasing rate of nucleation with fatigue for the small stress amplitudes is attributed to a cyclic hardening of the surface. This is strictly a near surface effect which disappears in dry air.

The derivative of the curves in Fig. 4 gives, naturally, the corresponding average rates at which intermetallics fracture with fatigue. By random selection, these are used to choose the cycles $N_I$ of each nucleation event. $D$ and $W$ values are then selected, consistent with constraints of the rate equation, and $N_{II}$ is calculated. Figure 5 shows a comparison between measured values of $D$ vs. $W$, and values obtained from simulation for the step II nucleation process for a selected set of fatigue cycles and cyclic stress amplitude. Each datum for both cases corresponds to an individual nucleation event. The dashed lower bound to $D$ vs. $W$ on the experimental data is obtained from the simulation. Scatter behind the lower bound in the simulation, and by inference in the experiment, results from the distribution in crystallographic orientations of the grains.

**Fig. 4** Comparison of measured (dots) to predicted (lines) number/cm² of intermetallics fractured by fatigue of Al 2219-T851 for four maximum cyclic stress amplitudes (fractions of the yield strength, $\sigma_{yield}$) for fully reversed loading.

**Fig. 5** Comparison of measured to simulated values of $D, W$ producing cracking into the matrix for Al 2219-T851 for 500 fatigue cycles with maximum stress amplitude of 90% of the yield strength. Dashed lower bound of experimental data (Top) is obtained from the simulation (Bottom).

**Early Propagation.** Microcrack closure stress is the dominate factor in controlling the rate of microcrack growth in the aluminum alloys of interest.

In the presence of crack tip closure, strain at the crack tip induced on a tensile loading cycle is not completely reversed on unloading and the resulting rate of propagation can be described by
A and m are material constants. The effective
stress intensity range, \( \Delta K_{\text{eff}} \), includes only that
portion of the fatigue cycle for which the crack is
open to the tip (i.e., for the maximum cyclic
stress, \( \sigma_{\text{max}} \), minus the crack closure stress,
\( \sigma_{\text{cc}} \)). The crack closure stress is found to be
highly sensitive to the location of the crack tip
relative to the grain boundaries. The plastic zone
size at a surface microcrack tip is not determined
by continuum constraints, but is instead princip­
ally determined by the distance of the crack tip to
the next grain boundary. The majority of plastic
slip is not continuous across a grain boundary and
hence, most deformation accumulates from within the
grain itself. If the slip distance within the
grain is large, the residual tensile strain devel­
oped at the crack tip is also large, and the crack
closure stress is correspondingly large. An empir­
ical expression which relates distance of a crack
tip to grain boundary to \( \sigma_{\text{cc}} \) allows one to estimate
\( \sigma_{\text{cc}} \) from the instantaneous location of a microcrack
relative to the alloy microstructure. Thus,
\[
\frac{\sigma_{\text{cc}}}{\sigma_{\text{max}}} = \frac{a_c}{2c} .
\]  
\( a \) is a material constant which is a function of
relative humidity, but is approximately unity.

With reference to Fig. 6, \( Z_0 \) is the distance of a
crack tip to the grain boundary and \( 2c \) is crack
length. Of course, if \( \frac{\sigma_{\text{cc}}}{\sigma_{\text{max}}} > 1 \), no propagation
of tensile opening (Mode I) cracks can occur. With
the initial condition \( 2c = W \), it therefore results
that only cracks located so that, at least for one
tip, \( Z_0 < 2c \) will propagation occur. Consequently,
most fractured intermetallics never lead to propa­
gation into the matrix as \( \sigma_{\text{max}}/\sigma_{\text{cc}} \) exceeds unity at
both surface tips. The rare exceptions which re­
sult in propagation are intermetallics located near
grain boundaries with a succession of small grains
beyond. Propagation begins and continues in the
direction of maximum constraint of the plastic zone
size (i.e., small \( Z_0 \)). An alternate mode of
propagation occasionally observed in the alloys of
interest is for propagation from an intermetallic
to proceed by a shear mode, which is unaffected by
tensile closure. Ultimately, this process must
also be modeled. Currently, it is deemed suffi­
ciently rare to be handled in an empirical way by
assigning smaller \( a \) values to a fraction of the
microcrack on a random basis.

To illustrate the success of the closure model­
ing of growth of cracks of approximately the grain
size, we compare measured microcrack propagation
rates to \( \Delta K_{\text{eff}} \) for Al 2219-T851. In Fig. 7, the
\( \Delta K_{\text{eff}} \) values have been determined experimentally by
measuring \( \sigma_{\text{cc}} \) directly at the surface crack tips.
The data are taken near the threshold of propaga­
tion for \( \Delta K \) and, hence, no correlation been \( dc/dN \)
and \( \Delta K \) is found. Conversely, the \( \Delta K_{\text{eff}} \) model pro­
vides a good description of microcrack growth rate.

The distribution of lengths of surface micro­
cracks at any point in the fatigue life is also
obtained by Monte Carlo methods. The nucleation
and early growth of each crack is simulated, and
the sizes of the grains in the region of each crack

\[
\frac{dc}{dn} = A(\Delta K_{\text{eff}})^m ,
\]  
\( A, m \) are material constants. The effective
stress intensity range, \( \Delta K_{\text{eff}} \), includes only that
portion of the fatigue cycle for which the crack is
open to the tip (i.e., for the maximum cyclic
stress, \( \sigma_{\text{max}} \), minus the crack closure stress,
\( \sigma_{\text{cc}} \)). The crack closure stress is found to be
highly sensitive to the location of the crack tip
relative to the grain boundaries. The plastic zone
size at a surface microcrack tip is not determined
by continuum constraints, but is instead princip­
ally determined by the distance of the crack tip to
the next grain boundary. The majority of plastic
slip is not continuous across a grain boundary and
hence, most deformation accumulates from within the
grain itself. If the slip distance within the
grain is large, the residual tensile strain devel­
oped at the crack tip is also large, and the crack
closure stress is correspondingly large. An empir­
ical expression which relates distance of a crack
tip to grain boundary to \( \sigma_{\text{cc}} \) allows one to estimate
\( \sigma_{\text{cc}} \) from the instantaneous location of a microcrack
relative to the alloy microstructure. Thus,
\[
\frac{\sigma_{\text{cc}}}{\sigma_{\text{max}}} = \frac{a_c}{2c} .
\]  
\( a \) is a material constant which is a function of
relative humidity, but is approximately unity.

With reference to Fig. 6, \( Z_0 \) is the distance of a
crack tip to the grain boundary and \( 2c \) is crack
length. Of course, if \( \frac{\sigma_{\text{cc}}}{\sigma_{\text{max}}} > 1 \), no propagation
of tensile opening (Mode I) cracks can occur. With
the initial condition \( 2c = W \), it therefore results
that only cracks located so that, at least for one
tip, \( Z_0 < 2c \) will propagation occur. Consequently,
most fractured intermetallics never lead to propa­
gation into the matrix as \( \sigma_{\text{max}}/\sigma_{\text{cc}} \) exceeds unity at
both surface tips. The rare exceptions which re­
sult in propagation are intermetallics located near
grain boundaries with a succession of small grains
beyond. Propagation begins and continues in the
direction of maximum constraint of the plastic zone
size (i.e., small \( Z_0 \)). An alternate mode of
propagation occasionally observed in the alloys of
interest is for propagation from an intermetallic
to proceed by a shear mode, which is unaffected by
tensile closure. Ultimately, this process must
also be modeled. Currently, it is deemed suffi­
ciently rare to be handled in an empirical way by
assigning smaller \( a \) values to a fraction of the
microcrack on a random basis.

To illustrate the success of the closure model­
ing of growth of cracks of approximately the grain
size, we compare measured microcrack propagation
rates to \( \Delta K_{\text{eff}} \) for Al 2219-T851. In Fig. 7, the
\( \Delta K_{\text{eff}} \) values have been determined experimentally by
measuring \( \sigma_{\text{cc}} \) directly at the surface crack tips.
The data are taken near the threshold of propaga­
tion for \( \Delta K \) and, hence, no correlation been \( dc/dN \)
and \( \Delta K \) is found. Conversely, the \( \Delta K_{\text{eff}} \) model pro­
vides a good description of microcrack growth rate.

The distribution of lengths of surface micro­
cracks at any point in the fatigue life is also
obtained by Monte Carlo methods. The nucleation
and early growth of each crack is simulated, and
the sizes of the grains in the region of each crack

\[
\frac{dc}{dn} = A(\Delta K_{\text{eff}})^m ,
\]  
\( A, m \) are material constants. The effective
stress intensity range, \( \Delta K_{\text{eff}} \), includes only that
portion of the fatigue cycle for which the crack is
open to the tip (i.e., for the maximum cyclic
stress, \( \sigma_{\text{max}} \), minus the crack closure stress,
\( \sigma_{\text{cc}} \)). The crack closure stress is found to be
highly sensitive to the location of the crack tip
relative to the grain boundaries. The plastic zone
size at a surface microcrack tip is not determined
by continuum constraints, but is instead princip­
ally determined by the distance of the crack tip to
the next grain boundary. The majority of plastic
slip is not continuous across a grain boundary and
hence, most deformation accumulates from within the
grain itself. If the slip distance within the
grain is large, the residual tensile strain devel­
oped at the crack tip is also large, and the crack
closure stress is correspondingly large. An empir­
ical expression which relates distance of a crack
tip to grain boundary to \( \sigma_{\text{cc}} \) allows one to estimate
\( \sigma_{\text{cc}} \) from the instantaneous location of a microcrack
relative to the alloy microstructure. Thus,
\[
\frac{\sigma_{\text{cc}}}{\sigma_{\text{max}}} = \frac{a_c}{2c} .
\]  
\( a \) is a material constant which is a function of
relative humidity, but is approximately unity.

With reference to Fig. 6, \( Z_0 \) is the distance of a
crack tip to the grain boundary and \( 2c \) is crack
length. Of course, if \( \frac{\sigma_{\text{cc}}}{\sigma_{\text{max}}} > 1 \), no propagation
of tensile opening (Mode I) cracks can occur. With
the initial condition \( 2c = W \), it therefore results
that only cracks located so that, at least for one
tip, \( Z_0 < 2c \) will propagation occur. Consequently,
most fractured intermetallics never lead to propa­
gation into the matrix as \( \sigma_{\text{max}}/\sigma_{\text{cc}} \) exceeds unity at
both surface tips. The rare exceptions which re­
sult in propagation are intermetallics located near
grain boundaries with a succession of small grains
beyond. Propagation begins and continues in the
direction of maximum constraint of the plastic zone
size (i.e., small \( Z_0 \)). An alternate mode of
propagation occasionally observed in the alloys of
interest is for propagation from an intermetallic
to proceed by a shear mode, which is unaffected by
tensile closure. Ultimately, this process must
also be modeled. Currently, it is deemed suffi­
ciently rare to be handled in an empirical way by
assigning smaller \( a \) values to a fraction of the
microcrack on a random basis.

To illustrate the success of the closure model­
ing of growth of cracks of approximately the grain
size, we compare measured microcrack propagation
rates to \( \Delta K_{\text{eff}} \) for Al 2219-T851. In Fig. 7, the
\( \Delta K_{\text{eff}} \) values have been determined experimentally by
measuring \( \sigma_{\text{cc}} \) directly at the surface crack tips.
The data are taken near the threshold of propaga­
tion for \( \Delta K \) and, hence, no correlation been \( dc/dN \)
and \( \Delta K \) is found. Conversely, the \( \Delta K_{\text{eff}} \) model pro­
vides a good description of microcrack growth rate.

The distribution of lengths of surface micro­
cracks at any point in the fatigue life is also
obtained by Monte Carlo methods. The nucleation
and early growth of each crack is simulated, and
the sizes of the grains in the region of each crack

Fig. 7 Relationship of microcrack propagation rate to $\Delta K_{\text{eff}}$.

Fig. 8 Comparison between predicted and measured microcrack lengths for Al 2219-T851 with maximum crack lengths approximately 200\,\mu m. Smaller surface area would have a longer fatigue lifetime on the average. A computer simulation illustrates this effect (Fig. 9). Plotted against fatigue cycles is the probability density of failure for the two bar geometries. The total area under each curve is 100\% failure. The cyclic stress amplitude has been chosen so that lifetime for the short bar is determined principally by nucleation statistics. For the longer bar, operating at the same cyclic stress amplitude, the resulting lifetime is dominated by crack growth statistics. The early dropout tail at short lifetimes for the long bar is the result of crack coalescence.

Mean lifetime is directly obtainable from such plots. Figure 10 illustrates a comparison between measured and predicted mean fatigue lifetimes for Al 7075 alloys having different mean grain sizes (75\,\mu m vs. 25\,\mu m). A large grain as compared to a fine grain Al 7075 alloy. Mean grain size differs by approximately a factor of 3 for the two materials. The improved lifetime in the simulation results from an increased time to first crack nucleation in the fine grain material.
SUMMARY

A Monte Carlo simulation of the initiation stage of fatigue failure of certain 2000 and 7000 series aluminum alloys has been described. The technique allows one to predict the scatter in lifetime of a material which results from the statistics of the microscopic failure processes, which are determined by the alloy microstructure. An ultimate application of the method may be to predict the probability density of early fatigue failure of components. It is unlikely that such predictions can be made with confidence with any other approach as the low probability events, which the procedure addresses, are extremely expensive to define experimentally. Furthermore, the technique, in principle, allows one to predict the effect of a change in alloy heat (microstructure) on the early failure statistics with a minimum of additional experimental characterization. Computer simulations which accurately reflect the microscopic nature of crack initiation may also be useful in defining inspection schedules in the event that NDE techniques are developed which can characterize the small cracks present during the initiation process.

ACKNOWLEDGEMENTS

This research was supported by Rockwell IR&D funds. The scatter in lifetime predictions were made using a one-dimensional computer simulation of the nucleation process developed under an Office of Naval Research Contract No. N00014-76-C-0952.

REFERENCES

Michael Resch (Stanford University): I would like to know what experimental techniques you use to measure your crack growth rates at the very low $K$ effective for the small surface cracks.

Fred Morris: All the measurements were done with a scanning electron microscope. The specimens were fatigued in air and then transferred to the microscope at 500 cycle increments. Crack length measurements were done with the specimen under load so we could see the cracks. It's tedious, but that's the only way I know to do it with any degree of accuracy.

Michael Resch: What size were the cracks?

Fred Morris: The crack lengths were in the range of five microns to 150 microns.

Phil Hodges (Rockwell, Los Angeles Division): It looks to me like you have very excellent data base for using acoustic emission for low cracking density to verify the number of fractures. Have you thought of that?

Fred Morris: Yes. Acoustic emission is, of course, one way to observe the brittle fracture of the individual intermetallic particles. Lloyd Graham, at the Science Center, has done some of that research several years ago. The other nice thing about being able to predict crack densities is that there is now available a surface technique using harmonic waves to count cracks. Otto Buck is going to be talking about that tomorrow.

Mark Weinberg (U.S. Army RADCOM): I'm kind of an outsider in this R&D community. I'm more concerned with applications. I would like a clarification of what you mean by crack closure. I can understand forces tending to keep cracks closed or from propagating. Do you mean by crack closure that there is a tendency to actually heal the material?

Fred Morris: I'm using the same definition used by Harris Marcus this morning. What happens with a crack that's very near the surface, that's in a plain stress condition, is that as you load it up to a maximum external surface stress, excess material is drawn into the region of the plastic zone near the crack tip. As you release the external stress the crack tip starts to close down on itself before the external stress reaches zero. The stress at which the first closure begins, we call the closure stress. The effect of having a positive closure stress is to reduce the plastic deformation per cycle during fatigue, and the altered crack propagation rate can be calculated. It's only an approximate technique because the crack closure stress and the crack opening stress are not exactly the same. But it's the best we can presently do.

Robb Thomson: Thank you very much.