ABSTRACT

A methodology is evaluated to predict the probability of specimen failure with subsequent fatigue, after a short surface crack has been detected in Al 2219-T851 alloy. Cracks are detected and tracked to failure using optical microscopy. Predictions of remaining lifetime distributions are made with a Monte Carlo procedure in conjunction with growth laws which model the effect of grains of differing size, shape and crystallographic orientation in the crack path on propagation rate. Because the surface of the alloy cyclically hardens, the average rate of crack growth is less for cracks formed later during fatigue. The predictive methodology successfully describes this phenomenon, as well as predicts the probability of early failure arising from the statistical nature of the growth process, for failure probabilities substantially smaller than conveniently measurable in the laboratory.

INTRODUCTION

The philosophy of Retirement-For-Cause (RFC) is to derive the optimum use from expensive high performance components by retaining them in service until failure by fatigue is potentially imminent. This is accomplished by inspection of each component at intervals to detect and size cracks which may have developed. A remaining lifetime prediction methodology is then employed to decide how much longer the component may safely be used. Of special concern in developing an RFC strategy is the statistical nature of the crack growth process, which may lead to failures in service substantially earlier than observable in testing small numbers of laboratory specimens. Unexpectedly early failures are most likely if the critical crack size is small, so that the growth of short cracks dominates the statistics of failure. In this case small variations in alloy microstructure from component to component can affect growth rates and hence remaining lifetime.

In low cycle fatigue early failures can also occur for a large critical crack size and arise from linking of multiply initiated cracks, some of which may have gone undetected during inspection.

Recent advances in modeling of the effect of alloy microstructure on the growth of short surface cracks in aluminum alloys makes it practical to consider calculating the probability distribution of remaining lifetime once a short crack has been detected. In this paper we evaluate this procedure by comparing experimentally measured failure probabilities to those predicted from the alloy microstructure. Surface cracks in fatigue specimens are detected and tracked to "failure" using optical microscopy. Crack growth laws which relate the rate of propagation to the size, shape and crystallographic orientation of grains in the crack path are used, in conjunction with a Monte Carlo technique, to obtain predicted probabilities of failure with additional fatigue after detection of a crack of a pre-determined size. Cycles to failure are calculated using measured rates as small as $10^{-4}$ and predicted and experimentally measured failure probabilities are compared for probabilities in the range of 0.03 to 0.5.

The surface of aluminum alloys cyclically hardens during fatigue, which reduces the rate at which short surface cracks propagate across grain boundaries. Consequently isolated cracks initiated early during high cycle fatigue, propagate more rapidly on an average than cracks initiated later. The hardening is an environmentally enhanced phenomenon which essentially disappears for fatigue in dry as opposed to in humid air. In low cycle fatigue, on the other hand, crack coalescence is more likely for cracks initiated later in lifetime because the density of cracks is larger. Thus, the probability of early fatigue failure for both high and low cyclic stress amplitudes is sensitive to the load history prior to detection of a crack, as well as to the number of cycles applied after a crack has been detected. These phenomena are investigated both experimentally and by computer simulation. Because the detected crack size and post detection loading amplitudes are well known, our predicted probability of early failure defines a lower bound to the scatter in remaining lifetime that would be expected in service for constant amplitude loading.

EXPERIMENTAL PROCEDURE

Flexural fatigue specimens (Fig. 1) of Al 2219-T851, with a yield strength $\sigma_{\text{y}} = 350$ MN/m$^2$, were prepared by machining with progressively decreasing cutting depths (ending with 20 µm) to minimize residual surface stresses. 80 µm of additional material were then removed from each surface using 600 grit emery paper, and a final polishing sequence with Al$_2$O$_3$ powder was used to give specimens a mirror-like finish for ease of crack length measurement. Repeated thickness measurements were made during polishing to maintain flatness and final specimen thickness was 0.1422 cm ± 0.0003 cm. The resulting residual surface stress, measured by X-ray diffraction, was less than 20 MN/m$^2$.

Specimens were fatigued in stroke control in laboratory air with fully reversed loading at a frequency of 5 Hz. Two maximum stress amplitudes ($\sigma_{\text{max}}$) were used with $\sigma_{\text{max}} = 0.63 \sigma_{\text{y}}$ or $\sigma_{\text{max}} = 0.9 \sigma_{\text{y}}$. Specimens were transferred at intervals to an optical microscope to measure...
Characterization of early failure requires that lifetime for the growth of large numbers of cracks be determined. At the lower stress amplitude the density of grain sized surface cracks initiated in Al 2219-T851 is extremely small, however. Experimental values are approximately 0.01/cm² after 1.5 x 10⁶ fatigue cycles and 0.2/cm² after 4. x 10⁶ fatigue cycles. We discovered that a small dot of ink from a felt tip pen placed on the specimen surface during fatigue greatly accelerated the rate of crack initiation, provided that the ink remained liquid during fatigue. The mode of crack initiation was unchanged by this procedure. Cracks could be seeded at any point during fatigue by delaying application of the ink until the specimen had been prefatigued a desired amount. Our procedure was to apply ink dots of approximately 50 µm in diameter along a line with separation between dots of several millimeters. After 5 x 10³ fatigue cycles the ink was removed using an acetone solvent and 5 x 10³ additional cycles were applied to grow the cracks to grain size. Growth rates after removal of the ink were apparently the same as on un-inked specimens, and we therefore believe that this procedure leaves subsequent growth behavior unaltered.

Using this technique we were able to initiate numerous microcracks within a desired range in fatigue cycles. Our goal was to characterize the growth of cracks of a specific length formed by a specific number of fatigue cycles. What happens in practice, of course, is that most of the early crack growth is arrested by grain boundaries; with the result that a distribution in crack lengths is observed, governed by the distribution in grain sizes. To obtain sufficient data for reasonable statistics, we analyzed the growth of cracks with initial lengths of 80±30 µm. The new length of each crack after each progressive fatigue increment was plotted against its initial length at the first fatigue increment (10⁴ cycles) (Fig. 2). A linear interpolation of these data was used to construct an effective distribution in lengths for each fatigue increment that approximates the distribution expected if all the cracks were initially 80 µm in length (Fig. 3). This procedure compensates for the tendency of the average growth rate to increase slightly with a larger initial crack length.

Morris has shown that the growth rate of short surface cracks in aluminum is influenced by non-continuum constraints of the plastic zone size at the surface, caused by the grain boundaries. As crack lengths reach approximately 500 µm, the boundaries can no longer constrain slip and the plastic zone size expands to a size determined by continuum constraints. Thus, the statistical effect of microstructure on growth rate is expected to be of major importance when the cracks are short. Growth to a critical crack size beyond 500 µm adds to the lifetime but does not substantially increase the scatter in remaining lifetime, unless crack coalescence is involved. Additionally, the numbers of cycles to failure after 500 µm is very sensitive to specimen geometry, and for
Fig. 3 Probability of occurrence of cracks of length smaller than specified for five increments in fatigue, after detection of cracks 80\(\mu\)m in length after 10\(\times\)10^3 fatigue cycles.

instance, would be a small fraction of the total lifetime for smooth bar specimens. We, therefore, define "failure" for our experiments to have occurred for a given crack when it reaches 500\(\mu\)m in length. The cycles required to reach this length for each crack is determined by a linear interpolation between fatigue increments as the length passes 500\(\mu\)m. Only the faster growing cracks can be characterized with our technique as shadowing of short cracks by neighboring long cracks further suppresses the rate of growth of the shorter cracks. Furthermore, the specimen ultimately fails before all the cracks being tracked can reach 500\(\mu\)m. The experimental method, therefore, is useful to determine fractional probabilities of failure slightly larger than the reciprocal of the total number of cracks initially seeded.

RESULTS

Remaining specimen fatigue lifetime is frequently determined by growth of the first and largest crack detected. This is not always the case, however, as statistical fluctuations in growth rate can lead to failure from propagation of a crack initiated at a site unrelated to the main detected crack. The probability of occurrence of such initially undetected independent initiations increases with both specimen surface area and with cyclic stress amplitude. Our experimental procedure determines probable remaining lifetime where growth of a detected single crack determines the lifetime. This is a zero surface area limiting case. For comparison, later we calculate the probability of remaining lifetime for a non-zero surface area case, in which cracks unrelated to the main crack may participate in specimen failure.

Probable remaining fatigue lifetime (zero surface area limit) results for the \(\sigma_{\text{max}} = 0.63\)\sigma_{\text{yield}} case are shown in Fig. 4. Three cases are illustrated corresponding to 80\(\mu\)m cracks initiated at 10K, 25K or 40K cycles. The probability of failure is plotted as a function of additional cycles to failure. The data points are experimental whereas the curves are theoretical predictions obtained using growth models discussed later. The trend is for cracks initiated earlier in fatigue lifetime to grow more rapidly on an average. This is entirely the result of a progressive hardening of the surface which makes it more difficult for the short surface cracks to propagate across grain boundaries later in the lifetime.

Fig. 4 Additional cycles to "failure" after an 80\(\mu\)m long crack has been detected at 10, 25, or 40\(\times\)10^3 cycles, for a peak stress amplitude \(\sigma_{\text{max}} = 0.9\)\sigma_{\text{yield}}. Increased mean in remaining lifetime for later initiated cracks is due to cyclic hardening of the surface. Symbols are experimental data, curves are computer predictions.

Probable remaining fatigue lifetime (zero surface area limit) results for the \(\sigma_{\text{max}} = 0.9\)\sigma_{\text{yield}} case are illustrated in Fig. 5. The additional cycles to failure are plotted for cracks initially of 200\(\mu\)m in length detected at 5K or 10K cycles. Again "failure" is defined to occur as the crack length reaches 500\(\mu\)m. In this case crack coalescence dominates the growth of cracks initiated at 5K cycles leading to average growth rates faster than cracks formed at 10K cycles.

MODELING AND DISCUSSION

The rate of growth of short surface cracks in Al 2219-T851 is controlled by the size, shape and crystallographic orientation of grains in the path of the crack. As a surface crack tip approaches a grain boundary, slip begins into the next grain. Growth stops at that point on the crack front for a period of incubation during which a mature plastic zone develops in the new grain. The duration of incubation is longer the smaller the grain and the shorter the crack. Progressive
cyclic hardening of the surface reduces the propensity for slip into the new grain, increasing the incubation period for cracks encountering grain boundaries later in the fatigue lifetime. At sufficiently low stress amplitudes the duration of incubation essentially determines the average growth rate. Transgranular propagation into the new grain begins when sufficient deformation has been accumulated to satisfy a critical strain energy density criterion for propagation.\(^4\) The rate of subsequent crack growth within the new grain is determined by closure stress developed at the surface crack tip.\(^4\) In the event that the subsequent growth is crystallographic, slip is principally in the plane of the crack, the closure stress is small and the growth rate is comparatively rapid.\(^4\) If subsequent growth is transgranular and non-crystallographic, the closure stress is proportional to the grain size, and hence the rate of crack growth is initially slower into large grains than into small grains. For both growth modes, crack growth rates can be predicted by an effective stress intensity range based growth law for which \(\Delta K_{\text{eff}}\) is calculated from crack length, crack depth, propagation mode and location of the crack tip with respect to the grain boundaries.\(^4\)

Transgranular propagation of short cracks becomes progressively less likely as fatigue progresses, because of the increasing duration of incubation (as the surface hardening during fatigue). A competing intergranular growth mode is observed later in life at low stress amplitude (i.e., when the surface hardening is most pronounced). Growth proceeds by fracture of grain boundaries ahead of but not directly connected to the crack tip, followed by linking of the secondary cracks with the main crack. We have found, by measuring the opening displacement of surface crack tips, that grain boundary cracking ahead of the main crack can occur even if hardening at a surface tip is sufficient to prevent slip from the crack tip into the next grain. In this case, we attribute the grain boundary cracking to an enhancement of the strain amplitude at the boundaries due to the elastic strain field associated with the crack.

Remaining Lifetime Prediction. A computer simulation of the growth process is used to determine the cycles to failure for a series of 80 \(\mu\)m long cracks. For each crack the size, shape and crystallographic orientation of grains in the path of the crack at the surface are selected at random for the growth simulation using distributions in size and shape measured for the alloy, and by assuming that the distribution of crystallographic orientation is random. A simplifying assumption is made in the initial conditions for propagation which makes the calculation substantially easier. We assume that each 80 \(\mu\)m crack is in an 80 \(\mu\)m grain with the surface tips at the grain boundaries. We further assume that the cracks arrive at the boundary at any time during an interval up to 5,000 fatigue cycles before they were detected. This is consistent with experimental observation. More accurate calculation would require modeling and simulation of the crack initiation as well as of the crack growth processes.

The mathematical details of the growth laws, outlined above and used to predict cycles necessary for crack lengths to reach 500 \(\mu\)m, have been discussed elsewhere.\(^4\) The predictions shown in Fig. 4 were obtained by simulation of growth for 10\(^3\) cracks for each of the three initial conditions of prefatigue prior to crack detection. The trend for cracks initiated earlier to propagate more rapidly, on an average, is the result of a progressive increase in the duration of the incubation period for transgranular crack propagation across grain boundaries, due to cyclic hardening of the surface. For the case in Fig. 4, in which 4 \(\times\) 10\(^3\) cycles is necessary for initiation, the mode of propagation is predominantly intergranular. Fig. 4 is replotted in Fig. 6 using a log scale to emphasize the very early failure characteristics. Some disagreement exists between prediction and experimental data for 2.5 \(\times\) 10\(^4\) cycles to initiation, which we attribute to the simplicity of our assumptions regarding the initial location of each crack with respect to the grain boundaries. In Fig. 7 remaining lifetime predictions for \(\sigma_{\text{max}} = 0.63\) \(\sigma_{\text{yield}}\) have been extended to probabilities of failure of one chance in ten million by calculating the effect which crack coalescence will have on remaining lifetime. In Fig. 8 we compare calculated remaining lifetime for a zero surface area to a 100 \(\text{cm}^2\) surface area case for \(\sigma_{\text{max}} = 0.63\) \(\sigma_{\text{yield}}\) and 10\(^4\) cycles to initiation. The shorter remaining lifetime for the larger surface area is due to initiation and propagation of cracks other than the main detected crack. Both the coalescence and non-zero surface area calculations were made using experimental values of the average numbers of cracks/cm\(^2\) developed during fatigue.
Fig. 6 Results of Fig. 4, shown using a log scale to emphasize the early failure behavior.

Fig. 7 Probability of failure with additional fatigue cycles for extremely low probabilities of failure is altered by crack coalescence for \( \sigma_{\text{max}} = 0.63 \sigma_{\text{yield}} \). Coalescence is more important for cracks initiated at \( 40 \times 10^3 \) cycles, because the cracking density is higher. Dashed line is approximately the smallest failure probability that can practically be evaluated in the laboratory.

Fig. 8 Probability of failure distribution shifts to smaller numbers of additional cycles for specimens with larger surface area, because of crack initiation and propagation at sites away from the main crack.

Comments on RFC Strategy - For Al 2219-T851 four separate growth mechanisms govern the statistics of early failure for propagation of short surface cracks. The relative importance of these are a function of the cyclic stress amplitude, of the cycles at which the main crack is formed, and of the failure probability. For the case of \( \sigma_{\text{max}} = 0.63 \sigma_{\text{yield}} \) with \( 10^3 \) cycles to initiation, specimens with the shortest remaining lifetimes fail by crack coalescence and at longer lifetimes fail by crystallographic crack propagation with growth rates controlled by incubation. Still longer lifetime specimens fail due to transgranular non-crystallographic propagation with rates controlled by crack closure stress and by incubation, and the long-life specimens fail predominantly by intergranular crack propagation. Each separate growth mechanism is sensitive to different features of the alloy microstructure and mechanical properties. Thus, the statistical probability of early failure is governed by different material features depending upon the probability of failure. The probability of very early failure cannot, therefore, be realistically obtained by extrapolation from probability of failure statistics determined in the laboratory and instead must be calculated from a knowledge of the crack growth mechanisms.

**SUMMARY**

Once a short surface crack of predetermined size has been detected by NDE, statistical variations in the number of cycles to failure, can be substantial. Exceptionally early failures can
Occur at a lifetime 1/10 of the mean remaining lifetime determined in the laboratory. These early failures result from rare combinations of the alloy microstructure in the crack path which lead to accelerated rates of crack propagation. Because of their rarity it is unlikely that such failures will be observed during laboratory testing, and even if a single anomalous event is recorded it is likely to be discounted. We have described a method to calculate the probable remaining fatigue lifetime from alloy microstructure, for small probabilities of failure. Such predictions are an essential element in the development of a quantitatively based RFC strategy. A further complexity of the early failure process, which can place additional demands upon NDE of components, is also discussed. For aluminum alloys, remaining fatigue lifetime after crack detection is a function of the prior loading history. If service records on the component are inadequate to extract loading history information an additional requirement is that NDE provide a measure of the accumulated fatigue damage of the surface, as well as detect and size surface cracks, before reliable calculations of probable remaining lifetime calculations can be made.

REFERENCES

SUMMARY DISCUSSION

Mike Buckley, Chairman (DARPA): Time for a couple quick questions.

Kamel Salama (University of Houston): Do you have any idea what the mechanism is for the hardening of the surface? Is the formation of a surface oxide important?

Fred Morris (Rockwell Science Center): We know it's environmentally dependent and John Wert has looked at the near dislocation structure at the surface. What we find in the presence of humidity is tangled dislocations at the surface where there are bands of persistent dry air. I think simply that an oxide forms on the surface in humid air and prevents a certain number of dislocations from escaping through the surface, producing a harder surface.

Ross Stone (IRT Corporation): Don't you notice an effect with the rate of cycling in your fatigue tests?

Fred Morris: We haven't looked very carefully at frequency effects. Our measurements have been done at 10 Hertz. We expect no problem with heating at this frequency, but there must be a frequency dependence of the environmental effect if you go to high enough frequencies.

Ross Stone: Apparently you did ignore residual surface stress in the specimens. Is that because they were negligible?

Fred Morris: They were for our specimens. Each specimen came from the machine shop having approximately 22 ksi residual surface stress. We then took 0.003 of an inch of material off both sides and I polished each one lovingly by hand to an accuracy of 0.00001 of an inch in thickness. We measured until we got down to quite low value of residual surface stress. But we also have a model which predicts the rate at which the residual stresses decay during fatigue. That's one of the things we plan to incorporate into our failure model in the future.

Steve Kellman (Rockwell International): Fred, you mentioned the effect of moisture in the environment. How pervasive are they?

Fred Morris: In aluminum alloys, they show up in the crack initiation rate because it affects the hardening of the surface. They show up in the crack propagation rate both in the hardening of the surface and in surface ductility. And they also show up in the rate at which individual residual surface stresses relax during fatigue. If the air is humid, the surface is harder, there is less relaxation through the surface and the residual stresses stay high. So it's really pervasive, and it's very dependent on alloy type.

Otto Buck (Rockwell Science Center [now Ames Laboratory]): There is a constant in the computer calculations, in the equations that reflect the moisture effect.

Fred Morris: It's all in the models. The results you saw on moisture effect is entirely determined from the computer calculation, and are derived from what we have learned so far about the effects of moisture on growth mechanisms.

Steve Kellman: If you have initial information on the humidity variation with time can you actually calculate life with it?

Fred Morris: Yes, we have done it.

John Rodgers (Acoustic Emission Technology): I'm rather curious, considering the kind of environment that aircraft might be flying in in the European area with what is acknowledged to be acid rain conditions, with pH's ranging down as low as 1.5 or 2 - is this not an important consideration in the moisture effects on the crack initiation and propagation?

Fred Morris: I'm sure it's important. I told you how we started the cracks in those materials. We used felt-tip pen ink to start the crack. So anything you put on that aluminum surface that modifies the near surface ductility can have an enormous effect on the life time.

Chris Burger (Ames Laboratory): What did you consider failure?

Fred Morris: The complete calculations were done to a crack length of 500 microns.

Chris Burger: You were talking then about the surface length of the crack?

Fred Morris: Yes.

Chris Burger: What concerns one from a failure reliability standpoint are the depth of those cracks.
Fred Morris: Of course. And they can be quite shallow, but this is really an exercise in how well our models work. We're also attempting to do the same types of predictions on smooth-bar specimens, carrying it right on up to the critical crack size, but we suffered from some experimental constraints in this measurement, which is the main reason we fixed the maximum length of interest at 500 microns.

Chris Burger: That is surface length?

Fred Morris: That's surface length, and the crack shape factors are typically about 0.1 when the cracks get that long. But if you would take that same crack and put it in a smooth-bar specimen, it would be about 250 microns deep.

Mike Resch (Stanford University): What do you think it is about the felt pen ink that accelerates crack initiation, and what effect might the ink have on the growth kinetics on these very small cracks?

Fred Morris: What we do is use it to start the crack and then take it off.

Mike Resch: How do you take it off? How do you know you don't have ink down in the crack tip that's going to, say, affect the kinetics.

Fred Morris: I wouldn't say it doesn't. We try to get the ink out by ultrasonically cleaning in an acetone bath. However, the growth rates are consistent with uninked specimens. Of course, we have to use that technique, because if we were to rely on naturally started cracks, the chance of finding a natural crack after 10 thousand cycles in that material is one per 10 square centimeters of surface area. And it will take you a week to find it.

Mike Buckley, Chairman: Thank you very much, Fred.