Fatigue and Crack-Growth Analyses under Giga-Cycle Loading on Aluminum Alloys

J. C. Newman, Jr.*

Department of Aerospace Engineering, Mississippi State University, Mississippi State, MS 39762 USA

Abstract

Stress-life (S-N) data from the literature on uniaxial-loaded specimens made of 2024-T351 aluminum alloy and subject to constant-amplitude loading are used to support the crack-growth concept to calculate the fatigue behavior into the giga-cycle fatigue regime using initial flaw sizes consistent with micro-structural discontinuities in the material. Fatigue test data from the literature on uniaxial-loaded specimens made 7075-T6 under superimposed low- (5 Hz) and ultra-high- (20 kHz) cycle (sine-on-sine) fatigue loading are used to study the crack-growth concept for predicting fatigue behavior under giga-cycle fatigue conditions. Fatigue behavior under giga-cycle loading conditions was modelled fairly well with the crack-growth concept.

Keywords: cracks; crack closure; fatigue; fatigue-crack growth; aluminum alloy

1. Introduction

The fatigue endurance limit for many metallic materials has been questioned since the development of fatigue test machines that are capable of very high frequencies and fatigue failures have occurred after $10^7$ cycles at stress levels before the traditional fatigue limits [1-10]. Some typical fatigue stress-life (S-N) curves are shown in Fig. 1. The solid curve is the traditional behavior that has been observed on many materials, which had what appeared to be well-defined endurance limit from external (surface crack) nucleation sites, as shown in Fig. 2(a). In the past 20 years, however, fatigue testing in the very-high-cycle fatigue (VHCF) and giga-cycle fatigue (GCF) regime has indicated

* Corresponding author. Tel.: +1-901-734-6642; fax: +1-662-325-7730.
E-mail address: newmanjr@ae.msstate.edu
that failures are occurring below the traditional endurance limits and this behavior may be due to either external or internal (fish-eye) nucleation sites, as shown in Fig. 2(b), producing a higher slope on the S-N curve (dashed curve) or a mode change (dash-dot curve).

Fig. 1. Stress-life (S-N) curves for low-cycle fatigue (LCF) to giga-cycle fatigue (GCF) behavior.

During the past two decades, several international research programs have been conducted on small-crack behavior. One was by the AGARD Structures and Materials Panel [11-13] and the other was by the National Aeronautics and Space Administration (NASA) [14] and the Chinese Aeronautical Establishment (CAE) [15]. These programs conducted fatigue tests to generate small-crack data on a wide variety of materials and loading conditions. Most of these studies dealt with naturally-initiated cracks at notches, but some studies [12, 13] had also used small electrical-discharged-machined notches to initiate cracks. One of the main conclusions from these studies was that a large portion of the fatigue (S-N) lives were consumed by fatigue-crack growth (FCG) from micro-structural features.

Newman and many of his coworkers [11,14,16,17] have used continuum-mechanics concepts with initial discontinuity sizes, like those that initiated cracks at inclusion particles or voids, and the effective stress-intensity factor range (corrected for plasticity and crack closure), to predict the fatigue lives on a wide variety of materials and loading conditions. The baseline crack-growth-rate data for these materials were obtained from large-crack data, ignoring the large-crack threshold (generated with the load-shedding test method), and using small-crack data in the low crack-growth-rate regime. Small-crack thresholds were estimated from the endurance limits on the various materials. However, the estimated small-crack thresholds are revisited in this paper, since endurance limits may not exist on the basis of VHCF or GCF testing [5, 18, 19].

A typical ΔK_{eff}-rate relation for a material and test conditions is shown in Fig. 3(a), as the solid curve. The relation in the near-threshold regime was obtained from small-crack data with linear extrapolation into the ultra-low-rate regime (dashed line). (The ultra-low-rate regime is characterized by a very steep slope on the ΔK_{eff}-rate curve below about 10^{-10} m/cycle and exhibits discontinuous crack growth.) Stanzl and co-workers [2, 4, 5], Bathias [18] and Miller [19] concluded that “there is no fatigue limit in metallic materials”. One may also say that “there are no fatigue-crack-growth thresholds for small cracks in metallic materials”. The linear extrapolation of small-crack data into the ultra-low-rate regime is consistent with Tanaka et.al [20] assessment that small cracks may have very low stress-intensity-factor-range thresholds. Herein, no finite threshold stress-intensity-factor ranges were used for small-crack behavior.
The lower $\Delta K_{\text{eff}}$-rate relation was assumed to have a finite, but steep slope that was estimated because FCG data do not exist in this regime. Fig. 3(b) shows a corresponding calculated S-N curve from LCF to GCF loading using an initial discontinuity size typical of those that initiate failure in aluminum alloys. The finite slope in the VHCF and GCF region is directly related to the slope of the lower region on the $\Delta K_{\text{eff}}$-rate curve.

![Crack-growth-rate curve](image1.png) ![Stress-life curve](image2.png)  
(a) Crack-growth-rate curve  
(b) Stress-life curve from LCF to GCF  
Fig. 3. Characteristic fatigue-crack-growth and stress-life behavior from LCF to giga-cycle fatigue (GCF) behavior.

In this paper, stress-life (S-N) behavior from the literature on uniaxial-loaded specimens made of 2024-T351 aluminum alloys and subject to constant-amplitude loading is used to support the crack-growth concept to calculate the fatigue behavior into the giga-cycle fatigue regime using initial flaw sizes consistent with micro-structural discontinuities in the material. Fatigue test data from the literature on uniaxial-loaded specimens made 7075-T6 under superimposed low- (5 Hz) and ultra-high- (20 kHz) cycle (sine-on-sine) fatigue loading are used to study the crack-growth concept for predicting fatigue behavior under variable-amplitude giga-cycle fatigue conditions.

**Nomenclature**

- $a$: surface-crack depth, mm
- $B$: thickness, mm
- $c$: surface-crack length, mm
- $D$: specimen diameter, mm
- $K_T$: stress-concentration factor
- $N$: cycles
- $N_f$: cycles to failure
- $R$: stress ratio ($S_{\text{min}}/S_{\text{max}}$)
- $S$: applied stress, MPa
- $S_{a2}$: alternating stress, MPa
- $S_{\text{max}}$: maximum applied stress, MPa
- $S_{\text{min}}$: minimum applied stress, MPa
- $S_o$: crack-opening stress, MPa
- $\alpha$: constraint factor
- $\Delta K_{\text{eff}}$: effective stress-intensity factor, MPa$\sqrt{\text{m}}$
2. Materials

Fatigue-crack-growth-rate data on two aluminum alloys, 2024-T351 [21] and 7075-T6 [14], had previously been analyzed to determine the effective stress-intensity factor range ($\Delta K_{eff}$) against rate relation for large-crack data on either compact, C(T), and/or middle-crack-tension, M(T), specimens. These data were over a very wide range in rates from threshold to fracture. For 2024-T3 [11] and 7075-T6 [14] alloys, small-crack data had also been determined from the literature, so modifications to the crack-growth curve were made in the low-rate regime. The small-crack data for the 2024-T3 alloy were assumed to be the same as those for the T351 alloy. No crack-growth thresholds were modeled or assumed. The crack-growth-rate curve was modeled with a finite, but very steep slope in the low- to ultra-low rate regime. Of course, the slope was determined by trial-and-error to fit fatigue failures in the giga-cycle fatigue regime. The key development was the assumed linear extrapolation into the ultra-low-rate regime. This behavior needs to be validated or verified that finite thresholds do or do not exist. Unfortunately, the "load-shedding" threshold test method will always produce a finite threshold; thus, a new method is needed to determine the ultra-low rates. The flow stress, $\sigma_{o}$, was average between yield stress and ultimate tensile strength.

3. Loadings

Two types of loading were considered for fatigue and fatigue-crack-growth analyses: (1) constant-amplitude loading and (2) sine-on-sine loading, as shown in Fig. 4. The sine-on-sine loading was used by Arcari et. al [22] on uniaxial-loaded specimens with the LCF loading applied at 5 Hz and the VHCF loading applied at 19.4 kHz. The maximum stress was held constant at 380 MPa and the minimum stress was 20 MPa for all tests, but the alternating stress ($S_{a2}$) was varied from 0 to 90 MPa. The sine-on-sine loading had 3,880 cycles in one block (one cycle at 5 Hz). Note that $S_{a2} = 0$ was constant-amplitude loading.

![Fig. 4. Sine-on-sine loading with VHCF loading superimposed on LCF loading.](image-url)

The sine-on-sine loading condition is a special variable-amplitude loading sequences that will severely test fatigue-crack-growth algorithms. These spectra would normally have to be rainflow counted to extract the major stress amplitudes to compute the correct damage. But the FASTRAN life-prediction code [23] has what is called "rainflow-on-the-fly" methodology, so the load sequence is applied and the code computes the major damage plus damage due to the smaller excursions. If the crack closes during any unloading excursion, then the rainflow-on-the-fly method is reset. Fatigue-life analyses with FASTRAN used two options: (1) constant crack-opening stress option (fast model), and (2) the cycle-by-cycle (NMAX = 1) option (full model) [23].
4. Fatigue-crack growth

Pearson [24] presented the first comparisons of small and large cracks in an aluminum alloy. His results initiated a world-wide study on small-crack behavior and his results are shown in Fig. 5, as the dashed curve, along with additional small-crack data (light solid curves) from Lankford [25] on 7075-T6 aluminum alloy using un-notched (K_T = 1) specimens. Lankford’s data went down to ΔK values as low as 1.5 MPa·m. They both concluded that cracks of about the average grain size grew several times faster than large cracks at nominally identical ΔK values. The symbols and dash-dot curve show the large-crack data and the development of the large-crack threshold at about 3 to 4 MPa·m. Some general observations from Lankford were that the minimum in dcdN occurred when the crack length, c, was about the minimum dimension of the grain size and that the magnitude of the lower rates was controlled by the degree of micro-plasticity in the next grain penetrated by the crack. If the next grain is oriented like the first, then no deceleration will occur, as indicated by the uppermost small-crack curves. Nearly two decades later, Phillips as part of a NASA/CAE cooperative program [26] generated large-crack data on middle-crack tension specimens that agreed very well with small-crack data in the lower plateau region (10^-6 to 10^-5 mm/cycle). There are small-crack effects, as shown by Langford's data, but the small-crack data by Pearson agreed with the newer large-crack data. Thus, the issue appears to be more related to the early large-crack data, which used a load-shedding method. It has been shown that the "load-shedding" method previously used to generate large-crack data at low rates had a load-history effect with elevated thresholds and slower rates than steady-state behavior [27, 28].

![Fig. 5. Comparison of small- and large-crack data on 7075-T6 aluminum alloy.](image)

The ΔK_eff-rate data for the 7075-T6 alloy is shown in Fig. 6 for R = -1, 0, and 0.5 [26]. The data correlated quite well with several transitions in slope at about the same rate for all stress ratios. However, some differences were observed in the threshold regime. For these calculations, α = 1.2 was used for rates greater than 7E-06 m/cycle (end of transition from flat-to-slant crack growth), and α = 1.8 (nearly equivalent to Irwin's plane-strain condition) was used for rates lower than 7E-07 m/cycle (beginning of transition from flat-to-slant crack growth). For intermediate rates, α was varied linearly with the logarithm of crack-growth rate. Newman et.al [16] have proposed that the flat-to-slant crack-growth transition region may be used to indicate a change from nearly plane-strain to plane-stress behavior and, consequently, to indicate a change in constraint, where (ΔK_eff)_T = 0.5 σ_0 \√B [23].

Fig. 6 also shows the small-crack data [26] generated on the 7075-T6 alloy (solid circular symbols). Since the crack-opening stresses stabilize very quickly for R = 0 loading, the steady-state crack-opening equations [23] with α = 1.8 were used to calculate ΔK_eff values. The small-crack data had more scatter in the low-rate regime due to the interactions with the micro-structure. The small-crack data agreed well with the large-crack data in the lower plateau.
region, but the data did not show the transitions (slope changes) above 1E-08 m/cycle. Instead of having crack-growth expressions for both \( \frac{da}{dN} \) and \( \frac{dc}{dN} \), a single curve was selected. The straight lines with symbols show the multi-linear \( \Delta K_{\text{eff}} \)-rate relation selected to make fatigue-life predictions.

Fig. 6. Effective stress-intensity factor range against rate for 7075-T6 aluminum alloy.

5. Fatigue

Test data from the literature on the fatigue of aluminum alloys under giga-cycle loading conditions are used to illustrate the predictive capability of small-crack theory using initial flaw sizes consistent with inclusion-particle distributions in these alloys. The first application is under constant-amplitude loading, while the last application is under a variable-amplitude (sine-on-sine) loading.

5.1. Constant-amplitude loading

Stanzl and co-workers [2] have studied giga-cycle fatigue and fatigue-crack growth behavior on 2024 aluminum alloys. Mayer and Stanzl-Tschegg [29] conducted fatigue test from \( 10^5 \) to \( 10^{10} \) cycles on 2024-T351. These tests were conducted on smooth \( (K_T = 1) \) dog-bone specimens at \( R = -1 \) and 20 kHz. The test results are shown as solid symbols in Fig. 7. Again, using the \( \Delta K_{\text{eff}} \)-rate curve [21] and assumed initial discontinuity sizes of 10 to 30 \( \mu \text{m} \) radius external surface cracks in the dog-bone specimens, calculations of fatigue lives have been made from \( 10^4 \) to \( 10^{10} \) cycles. The 10 to 30 \( \mu \text{m} \) initial discontinuity sizes bound the HCF to GCF data quite well, but under predicted the LCF results. In the past, a \( (\Delta K_{\text{eff}})_{th} = 0.8 \) MPa-m\(^{1/2}\) has been used, but failures are still occurring at lower values.

5.2. Sine-on-sine (LCF/VHCF) loading

These tests [22] were conducted on smooth \( (K_T = 1) \) round specimens (\( D = 4 \text{ mm} \)). The loading used for these fatigue tests was variable amplitude loading, composed by a carrier sine wave and a superimposed 19.4 kHz sine wave of lower amplitude, called a sine-on-sine wave (Fig. 4). The amplitude of the superimposed sine wave was varied, from 20 to 90 MPa, and the amplitude of the carrier sine wave reduced accordingly to have the same value of \( S_{\text{max}} \) and \( S_{\text{min}} \) for all tests. Fatigue lives from these tests are shown as solid diamond symbols in Fig. 8.

Fatigue-life predictions [22] were performed with the Navy’s fatigue-life-prediction tool, FAMS (Fatigue Analysis of Metallic Structures) with material constants for 7075-T6; a strain-life analysis was perform to calculate the damage using a Smith-Watson-Topper (SWT) correction method. Also, fatigue-life predictions [22] were performed with NASGRO’s tool for fatigue NASFORM; using material constants from NASGRO database for 7075-T6, a stress-life
analysis was performed to calculate the damage for each of the loading conditions. A photograph of one nucleation sites is shown in Fig. 8(a); and the fatigue lives are shown in Fig. 8(b).

Photographic of nucleation sites on four (P1, P4, P6, P7; S_{a2} = 0, 60, 80, 90 MPa, respectively) of the 8 specimens showed an inclusion particle(s) as the source of fatigue failures. The particle areas were measured and used to compute an equivalent semi-circular flaw size. They ranged from 10 to 14 $\mu$m in radius. Specimen P1 (constant amplitude) had the smallest flaw size, whereas, P6 and P7 had the largest. But the inclusions in P6 and P7 were oblong shaped and one inclusion had very little penetrating the free surface; thus, crack growth could have been in a vacuum environment. The circular symbols with dashed curves show the fatigue-life predictions using the upper and lower flaw-size bounds with $S_{d}/S_{\text{max}} = 0.31$ (fast model); and the solid circular symbols show the predicted results using the average flaw size (12 $\mu$m) under the cycle-by-cycle (NMAX = 1) option (full model).

![Fig. 7. Comparison of measured and calculated fatigue life for 2024-T351 aluminum alloy.](image)

![Fig. 8. Comparison of measured [22] and predicted fatigue lives for 7075-T6 under sine-on-sine loading.](image)
6. Concluding remarks

The fatigue process in aluminum alloys have been successfully modeled with fatigue-crack-growth concepts using a crack-closure model, small-crack data, and an initial flaw size related to the inclusion-particle distribution in the materials under low- and high-cycle fatigue conditions. Studies on very high cycle fatigue have also revealed that failures still occur well below the traditional endurance limit at $10^8$ to $10^{10}$ cycles. During the past decade, the load-shedding method used to generate the very low fatigue-crack-growth-rate data in the threshold regime has been shown to be flawed. The load-shedding method generates a load-history effect, inadvertently high thresholds, and slower crack-growth rates than steady-state behavior, due to remote crack closure. Fatigue behavior modeled with small-crack growth involves ultra-slow rates well below the traditional threshold regime ($10^{-10}$ m/cycle), and crack growth is intermittent (average crack-growth rates modeled with the effective stress-intensity factor range). Fatigue behavior under giga-cycle loading conditions was modeled fairly well with the crack-growth concept.

References

[1] M. Kikukawa, K. Ohji, K. Ogura, K., The push-pull fatigue strength of mild steel at very high frequencies of stress up to 100 kHz, J. Basic Engng. (Trans. ASME), 87 (1965) 855.
[2] S. Stanzl, E. Tschegg, H. Mayer, Lifetime measurements for random loading in the very high cycle fatigue range, Int. J. Fatigue, 8(4) (1986) 195-200.
[3] R. Ebara, Y. Yamada, Ultrasonic corrosion fatigue testing of 13Cr stainless steel and Ti-6Al-4V alloys, Ultrasonic Technology, K. Toda (ed.), MY Research, 1987, pp. 329-342.
[4] S. Stanzl, H. Ebenberger, Concepts of fatigue crack growth thresholds gained by the ultrasound method, Fatigue Crack Growth Concepts, D. Davidson and S. Suresh, eds., The Met.Soc. AIME, 1984, 339-416.
[5] S. Stanzl-Tschegg, B. Schönauer, Mechanisms of strain localization, crack initiation and fracture of polycrystalline copper in the VHCF Regime, 17th European Conference on Fracture, Brno, Czech Republic, 2-5 September 2008, pp. 890-897.
[6] Y. Murakami, Effects of small defects and non-metallic inclusions, Int. J. Fatigue, 16 (1994) 163-182.
[7] J. Lankford, The growth of small fatigue cracks in 7075-T6 aluminum, Fatigue Engng. Mater. Struct., 5 (1982) 233-248.
[8] A. Arcari, N. Apetre, S. Sarkar, N. Iyyer, N. Dowling, S. Stanzl-Tschegg, A. Vasudevan, N. Phan, P. Kang, Influence of superimposed VHCF loadings in cyclic fatigue of 7075-T6 aluminum alloy, 53rd AIAA/ASME/ASH/ASC Structures, Structural Dynamics and Materials Conference, April 2012, Honolulu, Hawaii.
[9] J.C. Newman, Jr., P.R. Edwards, Short-crack growth behaviour in various aircraft materials, AGARD Report No. 732, 1988.
[10] J.C. Newman, Jr., P.R. Edwards, Short-crack growth behaviour in various aircraft materials, AGARD Report No. 767, P.R. Edwards and J.C. Newman, Jr., eds., 1990.
[11] J.C. Newman, Jr., E.P. Phillips, An assessment of the small-crack effect for 2024-T3 aluminum alloy, Small Fatigue Cracks, The Metallurgical Society, Inc., Warrendale, PA, 1986, pp. 427-452.
[12] J.C. Newman, Jr., E.P. Phillips, M.H. Swain, R.A. Everett, Fatigue mechanics: an assessment of a unified approach to life prediction,” ASTM STP-1122, 1992, pp. 5-27.
[13] C. Bathias, There is no infinite fatigue life in metallic materials, Fatigue Fract. Engng. Mater. Struct., 22 (1999) 559-565.
[14] K.J. Miller, The fatigue limit and its elimination, Fatigue Fract. Engng. Mater. Struct., 22 (1999) 545-557.
[15] K. Tanaka, Y. Nakai, M. Yamashita, Fatigue growth threshold of small cracks, Int. J. Fracture, 17(5) (1981) 519-533.
[16] J.C. Newman, Jr., X.R. Wu, S.L. Venneri, C.G. Li, Small-crack effects in high-strength aluminum alloys - a NASA/CAE Cooperative Program, NASA RP-1309, 1994.
[17] J.C. Newman, Jr., P.R. Edwards, Short-crack growth behaviour in various aircraft materials, AGARD Report No. 732, 1988.
[18] J.C. Newman, Jr., X.R. Wu, S.L. Venneri, C.G. Li, Small-crack effects in high-strength aluminum alloys - a NASA/CAE Cooperative Program, NASA RP-1309, 1994.
[19] J.C. Newman, Jr., P.R. Edwards, Short-crack growth behaviour in various aircraft materials, AGARD Report No. 767, P.R. Edwards and J.C. Newman, Jr., eds., 1990.
[20] J.C. Newman, Jr., P.R. Edwards, Short-crack growth behaviour in various aircraft materials, AGARD Report No. 767, P.R. Edwards and J.C. Newman, Jr., eds., 1990.