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Effect of initiation feature on microstructure-scale fatigue crack propagation in Al–Zn–Mg–Cu

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High fidelity measurements of constituent particle or corrosion topography nucleated fatigue crack growth rates (da/dN) are established for 7075-T651 in humid air. Values of microstructure-scale da/dN are determined by microscopy of programmed load-induced crack surface markers, rather than surface-only measurements. Both pristine and corroded specimen da/dN from various applied stress levels are successfully correlated using continuum-elastic stress intensity (ΔK or ΔK and K_{max}) or dislocation-based (Bilby–Cottrell–Swindon) crack tip opening displacement (cyclic ϕ and ϕ_{max}), with the former accounting for the gradient of elastic stress concentration due to the initiating feature. Values of da/dN vary by an order of magnitude at each fixed driving force due to microstructural influences that result in a locally irregular crack front. Grain-scale models using stress intensity closure or slip-based crystal plasticity do not capture experimental da/dN variability. Due to an inadequate mechanistic basis, mechanics-inspired models of da/dN do not predict multiple growth regimes that are typical of environment enhanced cracking. An elastic ΔK-based description of long crack da/dN data for a given alloy-environment can be transformed to a continuum elastic–plastic ϕ_c basis to provide a mean crack growth rate description. Coupling mean rates with a statistical description of microstructure sensitive variability, and dislocation or crystal plasticity-finite element modeling of component ϕ_c for non-continuum cracking, will enhance prognosis in the MSC regime.

1. Introduction

Next-generation prognosis methods aim to increase the safety and reliability of aircraft structures by coupling real-time state awareness and realistic loading/environmental spectra with environment-sensitive fatigue properties predicted from an understanding of microstructure and damage physics [1,2]. Emphasis has been placed on understanding the mechanical driving forces and material response for fatigue crack formation and growth in high strength aluminum alloys; as governed by distributions of initiation feature, microstructure, fatigue environment, and time dependence [3–8]. Additionally, the deleterious effect of localized corrosion (pitting, exfoliation, intergranular attack, etc.) on fatigue formation and early growth must be integrated into a realistic aircraft prognosis capability [9–17]. Fatigue crack growth rate (da/dN) measurement and modeling focused on elastic stress intensity range ($\Delta K = K_{max} - K_{min}$) similitude is well established in the damage tolerant fracture mechanics framework [7,18–21]. However, the corrosion–fatigue interaction requires consideration of the microstructurally small crack (MSC) problem. While much progress has been reported in MSC propagation measurement and modeling [22–24] challenges remain.

Extensive research has delineated the unique features of MSC behavior, particularly for aluminum alloys, which include: cracking below the apparent long crack threshold level of ΔK, oscillating-increased da/dN with retardation by grain boundaries, high da/dN variability associated with microstructure and measurement inaccuracy, and slip band cracking character dictated by grain-level plasticity [25–28]. Researchers have proposed elastic ΔK [6,7,29–36] and elastic–plastic dislocation based crack tip opening displacement (ϕ) [22,37–49] approaches to describe MSC da/dN. While this phenomenology is extensive, quantitative-predictive models of MSC propagation kinetics must be better developed. Deficiencies are associated with: (a) uncertain crack tip damage mechanisms lacking both direct validation of quantitative feature crystallography and local-failure criteria that capture the interaction between plastic strain accumulation, tensile stress, and environmental factors, (b) the data used to analyze MSC growth are generally obtained via surface measurement techniques that do not describe the two-dimensional character of crack growth with microstructural interactions [29–32,50–53], (c) MSC growth rate models contain one or more adjustable parameters, and (d) MSC

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parameters based on ϕ have not been developed for complex component stress states and geometries. Most models for MSC da/dN are based on a plasticity perspective. However, the data used to assess such models and constants have been obtained for fatigue in moist environments; environmental effects have not been explicitly factored into existing MSC models. Accordingly, validations are suspect pending comparisons with MSC data obtained for high vacuum or inert gas environments.

Given these uncertainties the present work focuses on MSC propagation in Al–Zn–Cu–Mg. Companion studies investigated fatigue crack formation from a single-pit in pre-corroded specimens at various stresses [54], and the role of stressing environment on crack formation and MSC propagation [55]. Coupling quantitative characterization of corrosion damage, associated fatigue crack formation sites, and underlying microstructure with solid mechanics analysis of local driving forces provided an experimental basis to model crack formation [54]. At an engineering level, this work validated the linear-elastic fracture mechanics (LEFM) assumption that formation life (N_i) is nil at moderate-applied stress levels and demonstrated that empirical modeling reasonably predicts crack formation life at lower stresses [54]. A science-based foundation for prognosis of corrosion initiated fatigue requires coupling this work with a similar analysis of microstructure-scale crack progression. Once the governing mechanics are established for humid air crack formation and MSC growth, then the important effects of cyclic loading in realistic operating environments (e.g., salt water spray at lower altitudes and low-temperature high altitude flight) can be quantified and analyzed. The sum of these observations can inform and justify the assumptions inherent in modeling of crack formation and microstructure-scale crack growth (1–1300 μm), thus extending current damage tolerant prognosis techniques into the small crack regime.

The goals of the current work are threefold. First, obtain quantitative (da/dN versus crack length, da/dN variability) and qualitative (crack shape progression, fracture surface morphology) descriptions of MSC growth from realistic crack formation sites and over the 1–1300 μm size scale and at various stress levels. Crack formation sites include: constituent particles, controlled ellipsoidal corrosion pits, and a broadly corroded alloy surface. Second, quantitatively evaluate various mechanical driving force models used to correlate small crack growth from engineering and mechanistic viewpoints. Third, extend the method for predicting corrosion nucleated fatigue crack growth below a typical airframe damage tolerant flaw size (≈1300 μm) [56,57]. The approach utilizes programmed loading to produce crack surface marker-bands at set cycle intervals, allowing comprehensive analysis of crack growth relative to the underlying microstructure.

2. Experimental methods

A 50.8 mm thick, rolled plate of 7075-T651 (Al-5.7 Zn-2.53 Mg-1.66 Cu-0.263 Fe-0.06 Si-0.026 Mn-0.19 Cr-0.02 Ti; wt%) from the DARPA Structural Integrity Prognosis System (SIPS) program heat was investigated [2]. The average grain size ranges were 1–2 mm, 50–74 μm and 8–19 μm, in the longitudinal (L), transverse (T) and short-transverse (S) directions, respectively, with a partially recrystallized microstructure [2]. The monotonic tensile yield strength (σ_{ys}) was 508 MPa (L-oriented), ultimate tensile strength (σ_{UTS}) was 598 MPa, and plane strain fracture toughness was 33 MPa $\sqrt{m}$ (L–T) [58]. Fatigue experiments with pre-corrosion were performed on flat-uniform gauge specimens machined with specimen thickness (parallel to S) centered 7.0 mm from the plate surface and tensile axis parallel to L. Gauge length was 30.5 mm, thickness was 7.6 mm and width was 19.1 mm. Also used were two-holed specimens (machined in the same orientation) with specimen thickness centered 19 mm from the plate surface. The specimens were 47.4 mm wide and 5.7 mm thick, with two 4.8 mm diameter holes aligned perpendicular to the tensile direction, 0.13 mm from either L–S surface [2]. The hole surfaces were electro-polished and not pre-corroded.

Prior to fatigue testing, isolated-controlled pitting and EXCO-solution exposures were used to corrode the L–S uniform-gauge surface at the mid-length position along L. Specimens were masked using solvent resistant polyvinylchloride and rubber electro-plating tape (3M-470) and peelable butyl rubber lacquer (Miccro-super XP 2000). Three equally spaced 370 μm diameter holes were drilled through the electro-plating tape to allow solution contact with the metal surface. The corroding solution consisted of 0.1 M $AlCl_3$ + 0.86 M NaCl to which HCl was added to lower the pH to 2. A potentiostat provided a constant current of 4.5 mA for 15 min to the three holes simultaneously, and equal current (1.5 mA) at each hole was assumed. This protocol was applied at about 23 °C to both L–S surfaces of the fatigue specimen to produce six semi-ellipsoidal pits; each approximately 230 μm deep (T_P in the T direction), 630 μm in surface length (S_P) along S, and 630 μm height (L_P) along L. To mimic pit heights on broadly corroded surfaces, controlled pits with either elongated ($S_P/L_P < 0.8$) or short ($S_P/L_P > 1.2$) L_P-dimensions were formed by varying the hole size in the electroplating tape and adjusting current. The specimen and relevant orientations are shown in Ref. [54]. For select specimens, one L–S surface (no corner exposure) was exposed to EXCO solution for 3 h in accordance with ASTM G-34 [59]. This corrosion protocol was chosen based on prior work showing that EXCO corrosion was repeatable, produced consistent fatigue lives at various severities, and mimicked 7075-T6511 corrosion seen in sea-coast aircraft operating environments [9,10,60,61]. After corrosion, all specimens were rinsed and ultrasonically cleaned for 15 min in methanol, dried with nitrogen, then stored at ambient temperature in a desiccator with anhydrous calcium sulfate. Post-test energy-dispersive X-ray spectroscopy (EDX) was performed on four pits and an EXCO specimen. Only trace amounts of Cl ($K-\alpha$ = 2.622 keV) were present, with the spectra otherwise consistent with the composition of 7075 [55].

Constant amplitude uniaxial fatigue testing was performed on pre-corroded specimens in accordance with ASTM E466 [62]. A micrometer was used to center the specimen in a hydraulically actuated grip equipped with 90° diamond serrated wedges. Bending calibrations, performed in accordance with ASTM E1012, resulted in acceptable measured-maximum bending strains of 2.6% and 3.1% of the total applied strain for the top and bottom grips, respectively [63]. For baseline loading, maximum applied tensile stress (σ_{max}) was either 100, 200 or 300 MPa, stress ratio ($R = \sigma_{min}/\sigma_{max}$ where σ_{min} is minimum-applied tensile stress) was 0.5, and frequency (f) was 30 Hz. Testing was performed at room temperature (23 °C) in water vapor saturated N_2 (RH > 85%) maintained in a plexiglass chamber O-ring sealed to the specimen. Two-holed specimens (tested by Northrop Grumman [2]) were fatigued at 23 °C in laboratory air (RH ≈ 55%) at σ_{max} = 276 MPa, R = 0.5 and f = 5 Hz. The test matrix is shown in Table 1.

A programmed loading sequence created resolvable bands on the fatigue crack surface [9,10,60,61,64–66]. The example shown in Fig. 1 for a corroded specimen is typical of all stress levels investigated. The σ_{max} for marker-band loading equaled the baseline σ_{max}, but with R = 0.1 and f = 10 Hz; such cycles were applied after every 5000, 10,000, or 300,000 baseline (R = 0.5) cycles for σ_{max} = 300, 200 and 100 MPa, respectively. For short and elongated L-dimension pits, marker loads were applied every 20,000 cycles. A full description of the corroded specimen marker sequence is given in Ref. [54]. Marker loading for the uncorroded two-hole specimens consisted of up to 40 repetitions of: 1000 baseline loads (σ_{max} = 276 MPa, R = 0.5) followed by 10 marker cycles at σ_{max} = 276 MPa