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Effect of plate thickness on the environmental fatigue crack growth behavior of AA7085-T7451

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ABSTRACT

Fatigue crack growth rates (L–T) for 102 and 178 mm plates of 7085-T7451 are established at $R = 0.1$ and 0.5 for a wide-range of ΔK and environments. At R of 0.1, the 178 mm plate exhibits slower crack growth rates independent of environment. The crack growth trends correlate with changes in the fracture surface roughness and compliance-based closure indications, suggesting a dominant role of roughness induced closure. At R of 0.5 compliance-based metrics indicate that minimal closure exists, but plate thickness dependent crack growth behavior persists; grain-scale crack path tortuosity and crack branching govern this behavior. Increased crack path roughness, tortuosity, and branching in the 178 mm plate is attributed to strain localization/cracking along precipitate free zones formed in the thicker plate due to a slower quench rate.

1. Introduction

Cost reduction efforts in aerospace design have driven a transition from "built-up" structures (where separate parts are mechanically joined) to integral or monolithic designs; this requires large scale forgings or thick plate materials with high strength, toughness, and fatigue resistance [1–4]. High speed milling procedures make fabrication logistically- and cost-efficient for high strength aluminum alloys; as such, the major impediment to milling integral components is achieving sufficient materials properties in a thick (up to 200 mm) plate [4]. In addition to the inherent decrease in the degree of deformation (during the rolling process from ingot to final thickness) for increasing plate thickness, the thermal conductivity and heat capacity of the material leads to a gradient in quench rate from the surface (high rate) to the center (low rate). A sufficiently slow quench rate at the plate center deleteriously affects the microstructure evolution (and thus affects material properties – strength, toughness, corrosion resistance, elongation [5]), as well as induced residual stresses [6,7]. This quench sensitivity has been extensively studied in high Zn 7xxx-series alloys [5–18]; the reduction in properties is attributed to the precipitation of coarse η ($MgZn_2$) and S-phase (Al_2CuMg) at heterogeneously distributed transgranular bands of dispersoids and/or along the grain boundaries during the quench. This slow quench induced precipitation does not globally influence the size (but has been shown to decrease the volume fraction) of strengthening precipitates (η') during aging [5,11], however precipitate free zones form due to local solute and/or vacancy depletion [19] proximate to grain boundaries (and potentially at the quench-induced η in the grain interior) [11,13,14]. The deleterious effect of a slow quench on the mechanical properties is typically attributed to a combined effect of a weak interface between the coarse incoherent η and the matrix, reduction of the grain boundary properties, and/or the precipitate free zones.

While 7050 provides reasonable thick plate properties [2,20,21], modern alloys have been developed (e.g. 7085 (patented by Alcoa Inc.) [14,17,22–25] and 7040 [11,26]) to further minimize the quench sensitivity and offer superior thick product properties. Quench rate dependence in 7xxx-series alloys is promoted by high chromium content, incoherent dispersoids (particularly E-phase; $Al_{12}CrMg_2$ [8,16]), high angle recrystallized grain boundaries, and by high solute content [5,8,14,27], each of which promote heterogeneous precipitation of η (and potentially S-phase [17]) during slow quenching. High alloying element content is needed to maintain baseline material properties, as such modern alloys lower the quench rate dependency by both using Zr-based rather than Cr-based dispersoids and increasing the Zn:Mg ratio [14,16,17]. Zr-addition promotes the formation of semi-coherent Al_3Zr dispersoids that are less prone to nucleate η in non-recrystallized structures and result in more regularly shaped and thinner

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sub-grain boundaries [5,8,9,13,14,16]. Increasing the Zn:Mg ratio has been shown to increase or maintain the age-hardening capability while decreasing the solvus of the phases that may form heterogeneously on the grain and sub-grain boundaries, thus reducing the quench sensitivity [16,17,26].

The effect of plate thickness and quench sensitivity on strength, hardness, formability, and fracture toughness of modern 7xxx-alloys for thick gauge applications have been quantified and linked to governing microstructure features [10,12,13,17,22,28–30]. Similar in depth work is lacking for fatigue; characterization of plate thickness dependent fatigue behavior has been largely focused on 7050. Specifically, Deschapelles and Rice show that for plate thicknesses increasing from 25 to 150 mm (at T/2), there is a significant decrease in stress-life (SN) fatigue resistance, whereas for plate thickness increasing from 100 to 220 mm (at both T/4 or T/2) there is only a slight decrease [2]. This behavior was attributed to increases in porosity size with increasing thickness (at a constant T/*x* location) leading to more severe crack formation sites. This trend is not globally applicable, rather will be dependent on the size and distribution of pores in the original ingot and the rolling parameters used to achieve the plate thickness. Thin plate four point bending SN-data of 7050-T7451 (25 mm) and 7085-T7651 (38 mm) [31,32] show similar total fatigue life resistance between alloys, consistent with thick plate (102–152 mm) open-hole SN-data that showed slightly improved behavior for 7085-T7651 [22,30]. *In toto*, literature suggests total fatigue life trends vary with plate thickness due to changes in the microstructural features that govern crack formation; this dependence is common among separate 7xxx-series alloys. While literature convincingly correlates fatigue life with the presence of crack formation features, to be rigorous the fatigue crack growth behavior must also be considered.

Air fatigue crack propagation (FCP) rates of 7050-T7451 (L–T; $R = 0.1$, $f = 10$ Hz) were shown to significantly decrease as plate thickness increased from 35 to 160 mm [21,33]; similar behavior is suggested for 7085-T7651 growth rates for 100–152 mm thick plates (L–T, T/4; $f = 25$ Hz, $R = 0.1$) [22]. (Note, 7085-T7651 growth rates (SENT and MT; $R = 0.1$) are also reported in Ref. [34] however the plate thickness is not specified which precludes data comparison). Such behavior could potentially be due to different grain structure morphology for thicker plates. Starke et al. propose that increasing grain dimensions will lead to both intrinsic- (increased slip reversibility) and extrinsic-enhancements (increased crack path tortuosity reducing FCP by a local crack path deflection and enhanced roughness induced closure) [35]. However, this grain size effect was primarily observed for UHV testing and such intrinsic enhancements would only be pertinent in alloys where the grain boundaries are the principal barriers to slip, which may not be the case in overaged 7xxx-series Al alloys. Effects of dispersoid type, degree of recrystallization, precipitate free zones, and other microstructure features have been suggested for 7xxx-alloys but have not been evaluated in the context of thick plate FCP behavior [35–37]. Such effects of microstructure on high strength aluminum FCP behavior have been shown to be strongly dependent on the loading environment [35,38], which is attributed to a dominant interaction of hydrogen with the cyclic crack tip damage evolution [39].

The effect of water vapor pressure (P_{H2O}) on the FCP of aluminum alloys is generally understood via hydrogen environment embrittlement (HEE), with specific models developed relating fatigue crack growth rates (da/dN) at a given ΔK and R to a governing exposure parameter of water vapor pressure over loading frequency (P_{H2O}/f) [39,40]. The environmental contribution to da/dN is governed by the time-cycle dependent concentration of atomic hydrogen (H) into the crack tip fracture process zone (FPZ) [40]. HEE-based models assume that da/dN is limited by the slowest step in this sequence of mass transport in the occluded crack [41–50], Al–H_2O surface reaction(s) to produce a crack tip surface layer of adsorbed-atomic H [41,42,48], and diffusion of damaging H to the FPZ [51–54]. The dependence of da/dN on ΔK and P_{H2O}/f has been quantified for 7xxx-series alloys [35,45–48,50,55–59], generally for low R-ratios (0.05–0.1) where closure is important. Isolated data is available at high R and low ΔK in several purified gas environments [60,61]. A systematic characterization of the plate-thickness dependent FCP behavior of modern thick-plate 7xxx-series alloys is lacking, particularly for high altitude (i.e. low humidity) environments [62,63] pertinent to airframe applications.

This work characterizes the FCP rates of 102 and 178 mm thick plate of 7085-T7451 in gaseous water vapor environments ranging from high humidity (RH > 90%) to ultra-high vacuum (UHV). Quantitative FCP rates generated at different R-ratios (0.1 and 0.5) are coupled with fractography, metallography of the fracture surface cross-sections, and compliance based closure analysis to interpret the effect of thickness dependent microstructure on the environmental fatigue behavior.

2. Experimental methods

Rolled, 102 and 178 mm thick plates of 7085 in the -T7451 condition were investigated; the nominal alloy composition limits are 7.0–8.0 Zn; 1.2–1.8 Mg; 1.3–2.0 Cu; <0.08 Fe; <0.06 Si; 0.08–0.15 Zr; balance Al [25,30]. At the T/4 through thickness locations, the monotonic tensile yield strengths (σ_{ys}) are 493 and 487 MPa (L-oriented), ultimate tensile strengths (σ_{UTS}) are 523 and 517 MPa, percent elongations are 14% and 11%, and L–T plane strain fracture toughnesses are 46 (at T/2) and 38 (at T/4) MPa$\sqrt{m}$ for the 102 and 178 mm plates, respectively. Fatigue crack growth experiments were guided by ASTM E647-13a, where crack length was calculated from crack mouth opening displacement using compliance [64]. Compact tension (CT) and middle-tension (MT) specimens were machined in the L–T orientation (consistent with common industry practice for airframe applications) centered at the T/4 through thickness location. CT specimen width (W) and thickness (B) were 50.8 and 6.4 mm, respectively. The notch depth was 12.7 mm ahead of the load line ($a/W = 0.25$) and notch height was 3 mm. MT specimen width (W) and thickness (B) were 101.6 and 6.4 mm, respectively. Loading frequency was constant at 20 Hz for CT and 25 Hz for MT testing. Environmental FCP was characterized by two distinct loading formats. First, fatigue crack growth rates (da/dN) were measured at constant ΔK (11 and 18–19 MPa$\sqrt{m}$) and constant R (0.1 or 0.5). At each ΔK level, a single CT specimen (L–T orientation) was used to determine da/dN over a broad range of water vapor over frequency values (P_{H2O}/f) varying from high to low exposure and with each exposure level held constant during an interval (~1 mm) of growth. Linear regression of crack length versus cycles over an interval of steady state growth yielded da/dN. Second, da/dN was measured for a wide range of ΔK using constant R (0.1 or 0.5). All MT testing was performed with a ΔK increasing protocol. For CT testing, following an initial ΔK decreasing ($C = -0.08$ mm^{-1}) segment the ΔK was increased slightly (1–3 MPa$\sqrt{m}$ depending on the test conditions) and a ΔK increasing ($C = 0.2$ mm^{-1}) segment was performed to confirm no load history effects. Growth rates were determined at a specific ΔK using a 7-point incremental polynomial curve fit and differentiation [64]. Estimates of the effective stress-intensity factor range (ΔK_{eff}) can be calculated using the compliance based methods outlined in Appendix X2 (termed ASTM 2% offset) and X4 (adjusted compliance ratio; termed ACR) of ASTM E647-13a. The fundamental distinction between the ASTM 2% and ACR closure corrections methods is that the former assumes that the