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Effect of corrosion on the fatigue life and fracture mechanisms of 6101 aluminum alloy wires for car manufacturing applications

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ABSTRACT

An innovative solution for the automotive industry is to replace the copper used for wiring harnesses with aluminum alloys, such as the aluminum–magnesium–silicon 6101 alloy. Wiring harnesses are composed of thin strand arms obtained by a wire drawing process. These strands are susceptible to exposure to a corrosive environment and fatigue solicitations simultaneously. The fatigue endurance of this alloy was studied using the stress-life approach for three metallurgical states representative of three cold-drawing steps. Fatigue tests performed in corrosive media tests highlighted a strong decrease of the 6101 alloy lifetime due to fatigue–corrosion interactions and a modification of failure modes.

1. Introduction

Considering the rarefaction of fossil fuels and the need to reduce greenhouse gas emissions, automotive manufacturers are looking for innovative solutions to reduce both the cost and weight of wiring harnesses. One promising solution is the substitution of copper with aluminum alloys, such as the aluminum–magnesium–silicon 6101 alloy, which is currently used for high-power electrical applications. Wiring harnesses require thin strand arms, which are obtained by wire-drawing. This metalworking process consists of a reduction in the rod or wire diameter by passing it through a series of drawing dies interspersed, if necessary, with aging heat treatments to release the stresses induced by the process. The different microstructures obtained after each step of the manufacturing process strongly depend on the thermo-mechanical process characteristics, which also affect the mechanical properties and the corrosion behavior of the material. The strands are susceptible to exposure to several corrosive environments, e.g., de-icing road salt, coolant, and windshield washer fluids, and fatigue solicitations due to motor and engine vibrations. These combined damages can lead to a premature degradation of wiring harnesses. Many studies have focused on the role of pre-corrosion damage on the residual fatigue life and the effect of localized corrosion defects on fatigue crack initiation in aluminum alloys. Most such studies have focused on the 2xxx and 7xxx series and fatigue crack initiation from pitting corrosion defects [1–11]. These alloys have shown a large reduction in fatigue life due to corrosion pits, providing potential stress concentrator sites at which fatigue cracks can initiate. The pit size (average depth and width) seems to be an essential parameter; the fatigue life is successfully predicted by 2D AFGROW (Fracture Mechanics and Fatigue Crack Growth Analysis software tool) calculations based on initial crack sizes similar to those of pits [4]. Dolley et al. have shown, for AA 2024-T3, that the fatigue lives were reduced by more than one order of magnitude after 384 h of pre-corrosion in a 0.5 M NaCl solution compared to those of uncorroded specimens. The reduction of fatigue life was related to the time of exposure to the corrosive environment and pit size [11]. Burns et al. have shown that the fatigue life of a 7075-T6511 alloy was reduced substantially by EXCO pre-corrosion, but this was nearly independent of exposure time after initial-sharp degradation, scaling with the evolution of pit-cluster size and initial stress intensity range [12]. Liao et al. have studied the remaining fatigue life of 7075-T6511 aircraft wing skins containing natural exfoliation corrosion. It was found that the relative humidity did not have a significant effect on the fatigue life of naturally exfoliated specimens. Exfoliation above a "critical" level (undetermined in the paper of Liao et al. due to the limited number of tests) could significantly decrease the fatigue life of the naturally exfoliated specimens. When the exfoliation damage is not very severe, the discontinuities present in the fillet of the samples, where the stress concentration Kt was equal to 1.23, could override the exfoliation damage and become the primary crack origin. Above the "critical" level, the cracks primarily nucleated from corrosion pits and intergranular corrosion defects [13]. Pauze has studied fatigue

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crack initiation on localized corrosion defects for a 2024-T351 aluminum alloy, i.e., intergranular corrosion defects and pits. Both defects acted as preferential initiation sites for transgranular fatigue cracks [14].

Finally, the fatigue life duration for pre-corroded samples is essentially related to the fatigue crack initiation step on preexisting localized corrosion defect. A possible effect of H-uptake during pre-corrosion treatment on crack propagation rates is possible but it is probably negligible compared to the strong decrease of the duration of the crack initiation step.

However, for the tests performed directly in the corrosive media, the same conclusion cannot be advanced seeing that corrosion and fatigue mechanisms could interact as well during crack initiation as during crack propagation.

Most of literature results about fatigue–corrosion interactions have shown that the reduction in the fatigue life in NaCl solutions could be attributed to premature crack initiation from surface corrosion defects [15–17].

The characteristics of these localized corrosion defects such as pit geometry [18,19], micro-topographic strain concentration and local stress distribution around pit [20] are essential in determining the pit-to-crack transition such as fatigue life tests performed on pre-corroded samples. Mutombo and du Toit have characterized the fatigue–corrosion endurance behavior of a welded 6061-T651 aluminum alloy in a 3.5% (by weight) NaCl solution at various applied stress amplitudes [21]. The reduction in the fatigue life in the corrosive solution compared to the fatigue life in air was also most strongly related to the presence of pits, which nucleated on second-phase particles or precipitates and acted as preferential fatigue crack initiation sites. Pauze [14] has shown for a 2024 aluminum alloy that fatigue–corrosion interactions only acted on the crack propagation regime. The crack initiation occurred on corrosion defects which were strictly the same as those obtained during continuous immersion in corrosive media without applied stress. The effect of fatigue corrosion synergy only occurred during the crack propagation with the competition between the growth rate of the transgranular fatigue–corrosion crack and the growth rate of the intergranular stress corrosion crack. The propagation rate of the intergranular corrosion defect whose kinetic is relatively important during the first hours is quickly followed by a slow transgranular fatigue–corrosion propagation. The effect of the corrosive media is relatively low (factor 2–5) during this second step and occurs for low ΔK values. Finally transgranular fatigue–corrosion propagation is comparable to the propagation obtained in air [14].

In the present work, the reference metal was a 9.5 mm wire rod of AA 6101 (T4 metallurgical state); this rod was cold-drawn to manufacture the strand arms. The steps of the cold-drawing process used in the industry are summarized in Fig. 1. An intermediate aging heat treatment (185 °C–10 h) following the cold-drawing from a 9.5 mm diameter wire rod to a 1.34 mm diameter strand was applied. Finally, a final cold-drawing step provides a thin strand with a 0.51 mm final diameter (T9 metallurgical state). The influence of a pre-corrosion step on the fatigue and fatigue–corrosion behaviors of both T4 (the initial material) and T9 (the metallurgical state used) samples were studied. In parallel, some T4 samples were aged at 185 °C for 10 h in the laboratory to specifically study the influence of the intermediate heat treatment (T4 + 185 °C–10 h) on the material properties.

In a previous work [22], the corrosion behavior of the aluminum–magnesium–silicon 6101 alloy has been studied in chloride solution in terms of the different metallurgical states resulting from the wire drawing process (T4, T4 + 185 °C–10 h, and T8 metallurgical states; T8 corresponds to T4 + cold drawing until a 1.34 mm-diameter rod was obtained, followed by the 185 °C–10 h aging treatment; T9 corresponds to the same procedure as T8 except with an additional cold drawing step, which does not modify the corrosion morphology observed compared to the T8 state). Regardless of its metallurgical state, AA 6101 was found to be susceptible to localized corrosion. Scanning electron microscopy (SEM) observations of the electrode surfaces obtained after potentiokinetic polarization tests have shown that all metallurgical states were susceptible to two types of localized corrosion morphologies, i.e., pitting corrosion (Fig. 2(a) and (d)) and matrix dissolution around particles (Fig. 2(b)). Many thin corrosion filaments were observed around and between pit cavities according to literature results (Fig. 2(a) and (d)) [23]. These observations are consistent with the literature results. The main intermetallics present in Al–Mg–Si–Fe alloys are Fe-rich particles such as Al_3Fe, α-Al_8Fe_2Si, β-Al_5FeSi and π-$Al_8FeMgSi_6$ [24–26] and Mg_2Si precipitates [27].

Pits in Al–Mg–Si alloys often initiate on intermetallic particles. Blanc et al. [28] have demonstrated using interferometry and surface observations that Mg_2Si particles suffer preferential Mg dissolution just after a few seconds of immersion in aggressive media, in agreement with other authors [29,30]. Preferential initiation for deep corrosion was then observed at Mg_2Si sites [28]. Concerning Al–Fe–Si precipitates, they are known to be cathodic respective to the aluminum matrix [26,31]. During immersion in an aggressive media, the galvanic cells created between these intermetallics and the matrix initiate the dissolution of the surrounding matrix [32].

For the T4 + 185 °C–10 h metallurgical state only, observations of the electrode surfaces revealed a third form of localized corrosion, i.e., intergranular corrosion (Fig. 2(c)). As a reminder, for T4 + 185 °C–10 h, numerous precipitates arrange into a continuous film, with needle (average dimension = 35 nm × 140 nm) or globular shapes observed at the grain boundaries, unlike in other metallurgical states, which could explain the susceptibility to intergranular corrosion of the T4 + 185 °C–10 h samples. Therefore,

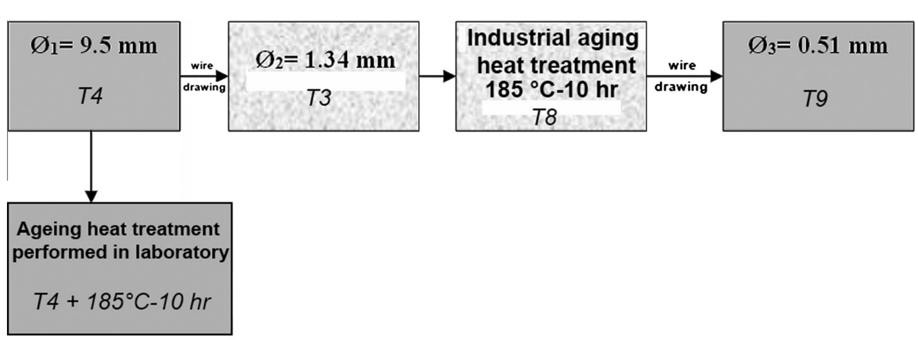

Fig. 1. Schematic presentation of the wire-drawing process used to manufacture the strand arms and resulting metallurgical states.