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## Effect of varying conditions of exposure to an aggressive medium on the corrosion behavior of the 2050 Al–Cu–Li alloy

Mathilde Guérin<sup>a,b</sup>, Eric Andrieu<sup>a</sup>, Grégory Odemer<sup>a</sup>, Joël Alexis<sup>b</sup>, Christine Blanc<sup>a,\*</sup>

<sup>a</sup> Université de Toulouse, Institut CARNOT CIRIMAT, UPS/CNRS/INPT, ENSIACET, 4 allée Emile Monso, 31030 Toulouse Cedex 04, France

<sup>b</sup> Université de Toulouse, LGP, ENIT/INPT, 47 Avenue d'Azereix, 65016 Tarbes, France

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### ABSTRACT

The corrosion behavior of aluminum alloy 2050 was studied as a function of varying conditions of exposure to an aqueous NaCl solution. Two metallurgical states were considered, i.e., an as-received alloy (NHT) and aged samples (HT). After continuous immersion tests, the NHT samples were susceptible to intergranular corrosion, whereas intragranular corrosion was observed for the HT samples. For the NHT samples, cyclic corrosion tests, with alternate immersion – emersion periods, induced subgrain boundary dissolution, and for the HT samples, grain and/or subgrain boundary sensibilization was induced. Hydrogen content measurements suggested a hydrogen enrichment of the alloy under cyclic corrosion tests.

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### 1. Introduction

Corrosion is considered a major threat to the structural integrity of aircraft structures and may significantly reduce the lifetime of such structures. To improve the corrosion resistance of aircraft metallic structures and to reduce their weight, new aluminum alloys, called 3rd generation aluminum–copper–lithium alloys, were developed. The 2050 alloy is one of the most promising solutions because this material combines low density, good mechanical properties and good corrosion resistance. Indeed, the addition of Li results in the reduction of the weight of the alloy because the density is reduced by 3% for each 1 wt% Li added to Al, whereas the elastic modulus is increased by almost 6% [1–3]. However, the presence of copper-rich second phases can produce micro-galvanic coupling, which leads to a decrease of the corrosion resistance of the alloy.

Previous studies focused on the corrosion behavior of Al–Cu–Li alloys demonstrated that these alloys were susceptible to different types of corrosion according to the thermomechanical treatments applied [4–8]. Thus, depending on their thermomechanical history, the Al–Cu–Li alloys may be susceptible to pitting and to intragranular or intergranular corrosion in relation to the distribution and to the type of intermetallic precipitates or dispersoids controlling the

local electrochemical potentials. Galvanic coupling, which occurs between second phases, usually rich in copper, and the matrix is a common mechanism responsible for pitting [9–12]. As an example, Al–Cu–Fe–Mn intermetallic particles, which are encountered in Al–Cu–Li alloys, cause the dissolution of the surrounding matrix because of their nobler electrochemical potential. The effect of the precipitation of hardening phases in Al–Cu–Li alloys on their corrosion behavior has also been thoroughly studied [4,9,13–16]. The results of these studies indicated that the electrochemical potential of  $T_1$ -Al<sub>2</sub>CuLi precipitates, which are the most common hardening precipitates encountered in these alloys, differed from that of the matrix, leading to a galvanic coupling phenomenon. For a low immersion time in chloride media, the corrosion potential of  $T_1$  precipitates was found to be more cathodic than that of the matrix, whereas for a longer immersion time, the corrosion potential of  $T_1$  precipitates evolves, as well as that of the matrix, leading to an inversion in polarity [14]. The coupling between  $T_1$  phases and the matrix was then primarily considered responsible for the corrosion susceptibility of Al–Cu–Li alloys, with intergranular and/or intragranular corrosion, depending on the thermomechanical treatment applied to the material [7]. Furthermore, Smith and Scully proposed that hydrogen could be trapped at the incoherent interfaces of  $T_1$  particles [17,18]; these authors considered that  $T_1$  particles are likely to be one of the high-energy trapping states in Al–Li–Cu–Zr alloys and pointed out the error in assuming that hydrogen cannot play a major role in cracking of Al-based alloys due to the low lattice solubility. Therefore, it was of interest to consider the hydrogen –  $T_1$  particles interactions to explain the

\* Corresponding author. Address: Université de Toulouse, CIRIMAT, UPS/CNRS/INPT, 4 allée Emile Monso, BP 44362, 31030 Toulouse Cedex 4, France. Tel.: +33 (0)5 34 32 34 07; fax: +33 (0)5 34 32 34 98.

E-mail address: [christine.blanc@ensiacet.fr](mailto:christine.blanc@ensiacet.fr) (C. Blanc).

properties of corroded samples when hydrogen was produced during the corrosion processes. Of course, this was dependent on the quantity of hydrogen produced and therefore on the exposure conditions.

Indeed, the conditions of exposure to an aggressive medium are another major factor affecting the corrosion susceptibility of metallic alloys [19–24]; cyclic corrosion tests with wet and dry phases modified the corrosion susceptibility of aluminum alloys compared with continuous immersion tests [24]. Larignon et al. showed that, after cyclic corrosion tests, a high hydrogen amount was detected inside a corroded 2024 aluminum alloy [24]. The influence of the exposure conditions to the aggressive medium is particularly true for aircraft structures because they are cyclically exposed to corrosive environments throughout their life. Indeed, when the plane is on the tarmac, the plane is exposed to a corrosive environment and sometimes to high temperatures; however, during a flight, the exposure duration to the corrosive environment decreases. Additionally, for some parts of the structure, the temperature falls to roughly  $-50\text{ }^{\circ}\text{C}$ , whereas other parts of the structure are maintained at room temperature.

The aim of this work is to study the effect of varying conditions of exposure to an aggressive medium on the corrosion susceptibility of the 2050 alloy. To apply testing conditions similar to real exposure conditions, corrosion tests with alternate immersion and emersion phases were performed in the laboratory. Because the corrosion susceptibility of the alloy strongly depends on its precipitation state, the 2050 alloy was studied in both its as-received state, which is associated with a low volume fraction of precipitates, and in an aged state, which is characterized by a large precipitation of  $T_1$  phases. Comparison of the results obtained for the two metallurgical states will help in understanding the influence of  $T_1$  precipitates on the corrosion morphology depending on the exposure conditions to the aggressive medium and will be helpful to obtain data about hydrogen trapping at  $T_1$  precipitates. Therefore, these two metallurgical states were exposed to continuous immersion and to cyclic corrosion tests to study the effect of environmental exposure conditions on their corrosion behavior.

## 2. Experimental

### 2.1. Material

The material for this study was an aluminum–copper–lithium AW (Airware) 2050 alloy (Al base, 3.86% Cu, 0.86% Li; wt%) provided by Constellium (France). The material was received as a 50 mm thick plate obtained by hot rolling, followed by solutioning, water quenching, stretching and natural ageing at room temperature, leading to the T34 metallurgical state. In this metallurgical state, according to the literature [25], hardening  $T_1$  precipitates were not present in the grains. Therefore, to promote the formation of hardening  $T_1$  precipitates in the grains, a piece of the plate received a heat treatment corresponding to an artificial ageing treatment at  $155\text{ }^{\circ}\text{C}$  for 30 h. In the following text, samples in the T34 metallurgical state and aged samples are called NHT (non-heat treated) and HT (heat-treated), respectively. The two metallurgical states are of interest from an industrial point of view. Friction stir welding (FSW) joints of Al–Cu–Li alloys constitute promising solutions to replace the riveted structures of aluminum alloys. The welded joints could be obtained by welding two plates in the T34 metallurgical states; after the welding process, the joints are submitted to an artificial ageing, without solutioning before the ageing, to increase their mechanical properties [7]. Therefore, the HT alloy corresponds to the metallurgical state of the base metal after the post-welding heat-treatment. Due to the rolling process, the microstructure of the material presented grains

strongly elongated in the rolling (longitudinal) direction. The average grain sizes in the longitudinal (L), transversal (LT) and short transversal (ST) directions were 500, 350 and  $60\text{ }\mu\text{m}$ , respectively. Electron backscattered diffraction (EBSD) observations (Fig. 1a; see the experimental section below for EBSD explanations) showed the strong morphological anisotropy of the material. Scanning Electron Microscope (SEM), using a JEOL 7000F field emission gun SEM, allowed coarse Al–Cu–Mn–Fe particles to be observed both for NHT and HT samples. Their density and distribution (both in the grains and at the grain boundaries) were similar for the two metallurgical states because the artificial ageing at  $155\text{ }^{\circ}\text{C}$  could not change the distribution of coarse particles. These particles were not studied with more details because they mainly influence the susceptibility to pitting corrosion of the alloy. The finer precipitation, i.e. the distribution of  $T_1$  precipitates, was described in the experimental results and discussion part.

### 2.2. Corrosion tests

All corrosion tests were performed in a 0.7 M NaCl solution maintained at a temperature of  $25\text{ }^{\circ}\text{C}$  using a Julabo refrigerated circulator in a laboratory room at a controlled temperature ( $25\text{ }^{\circ}\text{C} \pm 1\text{ }^{\circ}\text{C}$ ). This chloride concentration is representative of the chloride content for the solutions used during Mastmaasis Wet Bottom test, an industrial corrosion test used by Constellium. For the most part, corrosion tests corresponded to four exposure conditions with two cyclic corrosion tests and two continuous immersion tests (Table 1). For all the tests, during the immersion periods, the samples were hanged in the corrosion cell by using a thin nylon yarn. Cyclic corrosion tests (CR 72 h and CF 72 h) consisted of three 24 h cycles; each cycle was composed of a 3 h immersion in the electrolyte, which was followed by a 5 h dry period in air, then a 2 h immersion period and, finally, a 14 h dry period in air. During the air exposure periods (emersion phase), the samples were hanged by using the nylon yarn in a sample glass store placed in the laboratory room ( $25\text{ }^{\circ}\text{C} \pm 1\text{ }^{\circ}\text{C}$ ) for the CR 72 h tests and, for the CF 72 h tests, the samples were hanged in a container maintained at a negative temperature equal to  $-25\text{ }^{\circ}\text{C} \pm 1\text{ }^{\circ}\text{C}$ . For comparison, two continuous immersion tests (CI 15 h and CI 72 h) were also performed with durations of 15 h and 72 h. The pH was monitored during the continuous immersion tests. It was equal to 5.7 at the beginning of the test and became equal to 7.65 after 72 h for both NHT and HT samples. The duration of the first continuous immersion test represented the cumulated immersion time for cyclic corrosion tests, and the duration of the second one represented the total duration for the cyclic tests. For a few tests, to amplify the corrosion damage, cyclic tests consisted of seven 24 h cycles; these tests were called CR 168 h and CF 168 h (Table 1). For these tests, a 168 h continuous immersion test (CI 168 h) was considered as a reference. All corrosion tests were performed on cubic samples (10 mm edge) extracted from the core of the plate. The plane exposed to the aggressive medium was the LT–ST plane to observe the propagation of the corrosion defects in the rolling direction, which is the most critical direction for corrosion propagation because of the morphological texture of the material. Before corrosion tests, the samples were mechanically abraded up to 4000 grade, then polished with  $3\text{ }\mu\text{m}$  and  $1\text{ }\mu\text{m}$  diamond paste, rinsed in distilled water and air-dried. After corrosion tests, the samples were cut along the L–ST plane to obtain cross-sections for the analysis of the propagation of the corrosion in the bulk of the samples (Fig. 1b). Cross-sections were polished and then observed using optical (OM) and scanning electron microscopes (SEM; see experimental part below). To quantify the corrosion damage, a statistical analysis was performed from many polished cross-sections. Considering the cumulative length of the corroded surface observed in the ST direction, which was estimated at

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