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On the quench sensitivity of 7010 aluminum alloy forgings in the overaged condition

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ABSTRACT

The quench sensitivity of an overaged 7010 alloy forging was characterized by tensile and Vickers hardness tests, as well as scanning electron microscopy. Longitudinal tensile specimens, excised from a rectilinear open die forging were cooled from the solution treatment temperature following thirty-two different cooling paths including interrupted and delayed quenches. SEM analysis of the microstructure showed that quench precipitates were (i) Al_2CuMg (S) which nucleated heterogeneously on grain boundaries and (ii) $Mg(Zn,Cu,Al)_2$ (η) on grain boundaries, dispersoid bands, subgrain boundaries as well as in the aluminum matrix. The quench sensitivity of the alloy's yield strength and Vickers hardness was modeled simultaneously by quadruple-C curves, using an improved methodology for Quench Factor Analysis. The four C-curves used in the model represented loss of solute by (i) precipitation of S on grain boundaries, and precipitation of η (ii) on grain boundaries and dispersoid bands, (iii) on subgrain boundaries and (iv) in the matrix. The model yielded coefficient of determination (R^2) values of 0.967 and 0.974 for yield strength and Vickers hardness, respectively. The model and the implications of the results are discussed in this paper.

1. Introduction

Wrought Al–Zn–Mg–Cu alloys are used extensively in aerospace applications due to their high strength-to-density ratio. Heat treatment of these alloys involves a solution treatment, subsequent quenching, and finally artificial aging that may involve several stages depending on the desired temper. Quenching the parts at the highest rate possible, as in cold water, retains more solute and vacancies in solution, which increases the yield strength attainable after aging. However, high cooling rates associated with water quenching result in the generation of inhomogeneous thermal stresses leading to distortion and residual stresses. Conversely, low cooling rates that provide reduced levels of thermal stress issues produce non-strengthening quench precipitates, such as η ($Mg(Zn,Cu,Al)_2$), S (Al_2CuMg) and T ($Al_2Mg_3Zn_3$) in the Al–Zn–Mg–Cu system, which not only reduce the strength attainable after aging but also lead to stress corrosion cracking [1]. Therefore, process engineers strive to design cooling processes with an optimum balance between strengthening potential and residual stress. To accomplish this task, the effect of different cooling paths on the strength as well as the microstructure needs to be characterized and modeled.

Aluminum alloy 7010 was developed for applications requiring high strength, high fracture toughness, exfoliation resistance and stress corrosion cracking resistance in thick sections [2]. 7010 was developed as an alternative to 7050, which has a higher Cu content (2.0–2.6 wt%) than 7010 (1.5–2.0 wt%). The quench sensitivity of 7050 has received considerable attention since its inception in 1971 [1]. However, the quench sensitivity of 7010 was given less attention. In the studies involving 7010, the focus was either on the characterization of the type of quench precipitates or modeling the quench sensitivity by the Quench Factor Analysis introduced by Evancho and Staley [1]. To the best of the authors' knowledge, there has not been any research that combines characterization of quench precipitates and simultaneously provides a model to predict property loss due to the quench precipitates. This study is intended to fill this gap in the literature.

2. Experimental details

A rectilinear open die forging of 7010 alloy was manufactured by HDA Forgings Ltd. (now Mettis Aerospace, Ltd.), Redditch, UK, on a 20 MN draw down hydraulic press. The forging temperature was in the range of 390–400 °C. This forging was similar to a

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production item that receives extensive machining and ultimately forms part of the wing spar assembly in the Airbus A330/A340. The rectilinear forging had dimensions of 3045 mm (L, longitudinal) × 158 mm (LT, long transverse) × 125 mm (ST, short transverse). The chemical composition of the forging is given in Table 1. Tensile specimens with 6 mm diameter and 30 mm gage length were excised from the forging. The long axis of the specimens corresponded to the L direction of the forging. Specimens were solution treated in an air-recirculating furnace at 475 °C for 50 min. Solution treatment was followed by 32 different quench paths, both interrupted and delay quenched [3] to obtain a wide interval of yield strength and hardness values. Quenches were interrupted at seven temperatures for various durations by inserting specimens into a salt bath filled with a mixture of KNO_3 and $NaNO_2$. The hold temperatures and times are given in Table 2. In delayed quench experiments, the specimens were initially cooled in still air until the target temperatures (400, 350, 300, 250, 200 °C) were reached, and subsequently quenched in cold water. For each interrupted and delayed quench path, two tensile specimens were used along with the dummy specimen into which a type-K thermocouple was placed in the geometrical center, allowing the collection of the time–temperature data. In addition, two specimens were quenched directly in cold water from the solution treatment temperature. The specimens were then naturally aged at room temperature for 5 days. Subsequently they were artificially aged at two stages, 120 °C for 10 h followed by 173 °C for 8 h, to attain the overaged condition.

Tensile and Vickers hardness tests were conducted on each specimen. A Zwick tensile tester was used at an engineering strain rate of 0.001/s and 0.2% offset yield strength (σ_Y) values were recorded. A total of sixty-five tensile tests were conducted. Three Vickers hardness tests (H_v) at 20 kg load were conducted for each specimen and their average was taken as the representative hardness value.

Certain interrupted quench specimens were investigated metallographically to identify the type of quench precipitates at various hold temperatures. The samples were polished by the standard metallographic polishing methods. Subsequently, the samples were prepared for electron microscopy on a Vibromet II vibratory polisher, using a 0.2 μm colloidal silica suspension. A JEOL 700F scanning electron microscope (SEM) equipped with a field emission gun (FEG) at Colorado School of Mines was used in this study. Additional analysis by Energy Dispersive Spectroscopy (EDS) was performed on a Phillips XL30 field emission SEM equipped with EDAX at the University of Florida Major Analytical Instrumentation Center.

Table 1
Chemical composition (in wt%) of 7010 used in this study.

Si	Fe	Cu	Mg	Zn	Zr	Al
0.03	0.06	1.69	2.44	6.26	0.14	Balance

Table 2
Temperatures and hold times for interrupted quench specimens.

T (°C)	Hold time (s)
425	333, 1000, 5000
375	33, 100, 333, 1000
325	33, 100, 333, 1000
275	100, 333, 1000
260	10, 100, 333, 1000
240	10, 100, 333, 1000
225	333, 1000, 5000

3. Results and discussion

3.1. Characterization of quench precipitates

The microstructure of the specimens is presented in Fig. 1, which shows the presence of Fe-bearing coarse phases (constituents) on grain boundaries. However, the number density of these coarse phases is small, due to the low Fe content (0.06 wt%) of the alloy. Robson [4], in his delayed quench experiments on 7050 plate with 0.15 wt% Fe, found that globular S phase nucleated on coarse phases only five degrees below the S solvus temperature. Because the number density of the constituent particles is low, the number of nucleation sites on the constituents can be expected to be low as well. Consequently, loss of solute by precipitation on coarse phases and its effect on strength or hardness were not characterized in this study.

Quench precipitates in a specimen held at 425 °C for 1000 s are shown in Fig. 2. Note that there are two types of precipitates, both nucleated on grain boundaries: coarse globular precipitates as well as much finer precipitates. Energy dispersive spectroscopy analysis indicated that the coarse globular precipitates did not contain Zn while the finer precipitates contained both Zn and Cu. Therefore, coarse globular particles were interpreted as the S phase and the finer precipitates as the η phase. This is in contrast with the results of Archambault and Godard [5] who investigated the kinetics of η precipitation 7010 100 mm thick plate at various temperatures ranging between 420 and 225 °C. Archambault and Godard did not find any S phase and stated that all quench precipitates were the η phase. However, Godard et al. [6], in a later study, investigated the effect of holding time at six different temperatures ranging from 400 to 150 °C and characterized different quench precipitates in the 7010 plate. Godard et al. did find S phase especially at higher temperatures. It should be noted that the globular precipitates in 7010 have a similar size and morphology to the S phase precipitates found on grain boundaries and constituents by Robson [4] in 7050. The solvus temperatures for S and η phases for the chemical composition of the alloy in Table 1 were calculated by using the ThermoCalc® software to be 463 and 440 °C, respectively. Hence a hold of 1000 s at an undercooling of 38 °C was

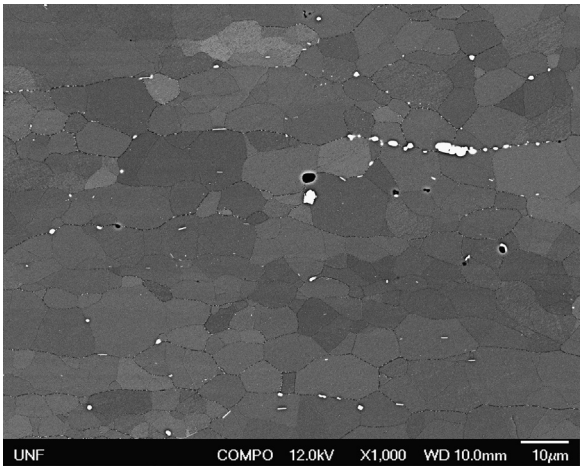

Fig. 1. Backscattered electron micrograph depicting the typical microstructure of the 7010 specimens at low magnification, demonstrating grain size and coarse constituents common to the samples in this study. Long axis of the photograph corresponds to the longitudinal axis of the tensile specimens and the original forging.