Experimental investigation of fatigue crack propagation in 2xxx aluminum alloy with local yield strength gradient at the crack path

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ABSTRACT*. The effect of controlled overaging on fatigue crack propagation (Stage II) of clad aluminum alloy 2024-T3 was studied experimentally. Two types of heat treatment were utilized resulting a) in local uniform reduction of yield strength and b) in local yield strength gradient at the crack path. Precracked specimens were subjected to cyclic loading to assess the influence of the induced yield strength profile at the crack path due to overaging on fatigue crack propagation. The crack propagation behavior was compared with crack growth data resulting from reference T3 material. The experimental results have shown that uniform strength reduction enhances the fatigue crack propagation at intermediate ΔK values compared to the reference material. On the contrary, in material with yield strength profile in the form of gradient, crack growth rates were slightly higher with respect to the reference material behavior. The experimental analysis performed presents useful experimental data of material damage tolerance behavior under cyclic loading at structural regions with variations of yield strength.*

1. INTRODUCTION

In ductile metals crack propagation rates in the Paris region are inherently dependent on the amount of plasticity and strain hardening effects at the crack tip [1-3]. In structural areas inhomogeneous material (e.g. welds regions), strength variation in the form of gradients may exist at local scale. In such areas the material potential to strain harden or produce plastic deformation may be influenced due to variation of strength. Hence, crack growth characteristics may diverge from the original behavior. In work [4] the reduction of yield strength caused by overaging in 2024 aluminum alloy was found to influence crack growth characteristics of the material. Reifsnider et al. in [5] found that crack propagation rates of high strength aluminum alloys of 6xxx and 7xxx series were influenced by the slope of the yield strength gradient ahead of the crack tip. Despite the above evidence, a systematic investigation on the effect of yield strength gradient on fatigue crack propagation including also alloys from the 2xxx series is missing.

In the present work, the effect of yield strength profile on 2024 material i) in the form of uniform reduction with regard to reference material and ii) as local yield strength gradient on fatigue crack growth (FCG) was studied experimentally. For the investigation a controlled heat treatment process has been used to achieve appropriate overaging conditions to locally modify the material's yield strength. Specifically two types of heat treatment were selected, (HT1) which results in local uniform reduction of yield strength and (HT2) which results in local yield strength variation at the crack path.

2. MATERIAL

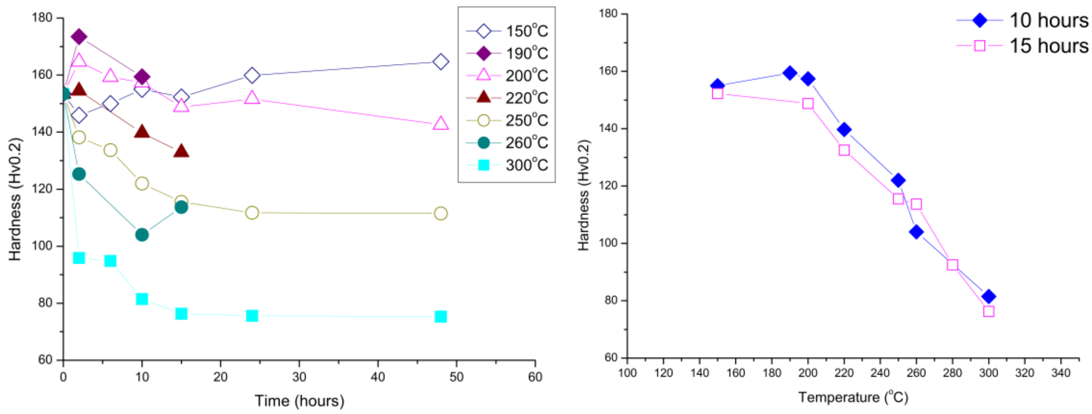
In the experimental investigation clad Aluminum alloy 2024 has been used in T3 condition, which includes heat treatment, control stretching and natural aging. In Table 1 the chemical composition of the alloy is given. It was received in sheet form of 3.2mm thickness with a clad surface thickness of 0.125mm.

Table 1. Chemical composition (wt.%) of Aluminum alloy 2024-T3

Al	Cu	Mg	Mn	Si	Fe	Cr	Zn	Ti	Other Each	Other Total
90.7 - 94.7	3.80 - 4.90	1.20 - 1.80	0.30 - 0.90	max. 0.50	max. 0.50	max. 0.10	max. 0.25	max. 0.15	max. 0.050	max. 0.15

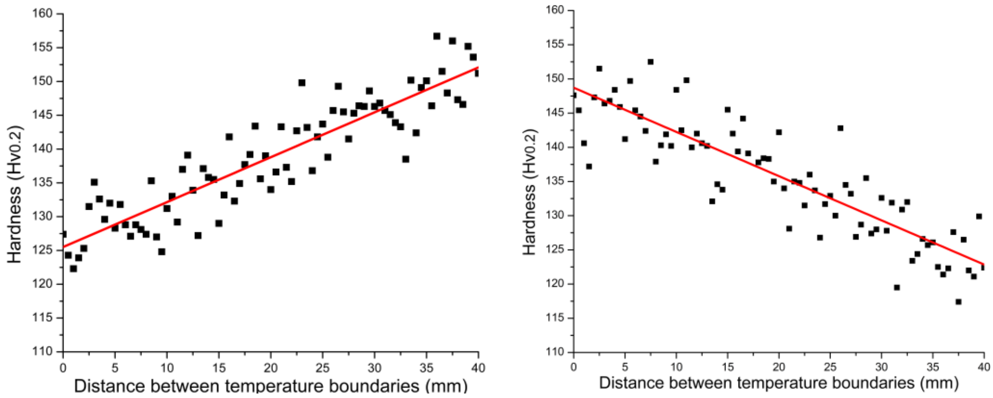
3. HEAT TREATMENT

Appropriate heat treatment processing has been used in order to achieve the overaging conditions HT1 and HT2 in the material. In order to select the parameters (temperature and time) for HT1 and HT2 treatments small material samples were artificially aged at different temperatures. The determined overaging curves were used to obtain the microhardness variation with regard to aging parameters. From the obtained overaging curves (see Fig. 1) the HT1 and HT2 conditions were selected. HT1 aging was used in two different temperatures. The temperatures were 250^{o}C and 300^{o}C for 15 hours. HT2 treatment included exposition of the samples to a temperature gradient between two temperature boundaries from 300 to 200^{o}C to achieve a gradual decrease of micro-hardness between the boundaries. The results of the hardness profile in the material after HT2 are shown in Fig. 2.



(a) (b)

Figure 1. Overaging curves of clad aluminum alloy 2024-T3 showing (a) variation of hardness with heat treatment time (b) variation of hardness with temperature

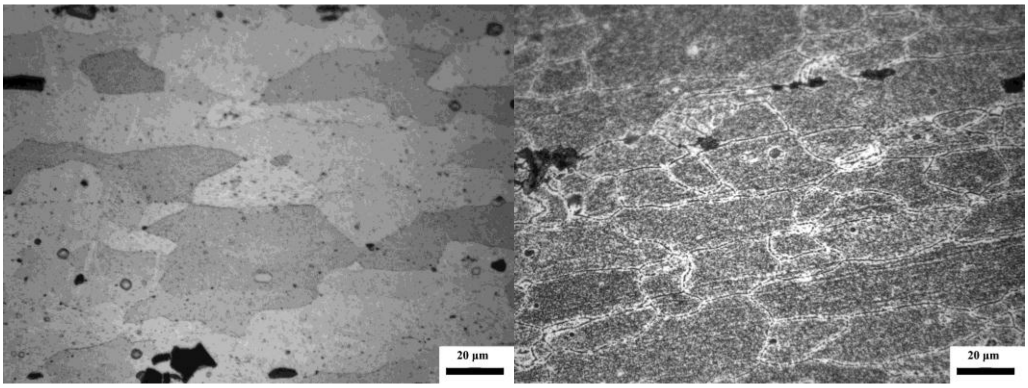


(a) (b)

Figure 2. Hardness profiles of specimens after HT2 (a) specimens with linear increase of hardness with regard to the crack tip position (b) specimens with linear decrease of hardness with regard to the crack tip position

The microstructure of the material in the as received condition as well as after HT1 (300^oC for 15 hours) is shown in Fig. 3. The T3 microstructure consists of elongated grains, parallel to the rolling direction. The dark particles are Al_7Cu2Fe and Al_2Cu phases as obtained from the performed SEM/EDS analysis. The results are in agreement with the literature [6,7].

In Fig.3(b) the microstructure of the alloy after HT1 is presented. Metastable phase-S΄ (Al_2CuMg) has been dissolved from the grain boundaries, and has been precipitated within the grains. Precipitate-free zones (PFZs) are also evident at the grain boundaries. Inside the grains, coarsening of the metastable phase-S΄, can also be observed. The Al_7Cu2Fe and Al_2Cu phases seem to be unaffected.



(a) (b)

Figure 3. Microstructure of Aluminum 2024 (a) T3 condition (b) after HT1 at 300 ^{o}C for 15 hours

3. TESTING

3.1 Tensile tests

Tensile tests were carried out (ASTM E 8M-01) on subsize specimens at T3 condition and specimens with HT1 treatment at 250 and 300^{o}C. In Table 2 the mechanical properties of the material in T3 and HT1 condition are given. Also, in Fig. 4 the tensile stress-strain curves are shown. HT1 treatments at 250^{o}C and 300^{o}C lead to a significant decrease of the yield strength that reaches 40% of the initial value. Simultaneously, a decrease of tensile strength is observed. The elongation at fracture is reduced to 60% of the initial values.

Table 2. Tensile test results (average values)

	Yield Strength $R_{0.2}$ (MPa)	**Elongation at fracture A_{25} (%)**	**Tensile Strength R_m (MPa)**
T3 condition	*370*	*19*	*465*
Specimens after HT1 at 250 ^{o}C	*260*	*10*	*355*
Specimens after HT1 at 300 ^{o}C	*150*	*11*	*280*

The hardness measurements presented in Fig.1 and yield strength data from Table 2 were used to derive the empirical correlation between and yield strength and hardness values. The empirical relation was used in the form:

$$S_y = 3*Hv - 90 \qquad (1)$$

The above relation was used to evaluate the yield strength profile values at the crack path from microhardness measurements in the FCG specimens.

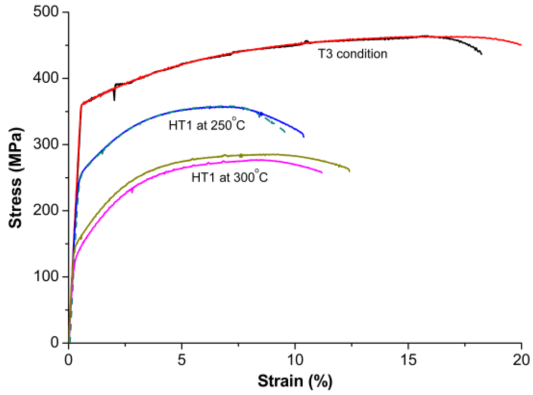
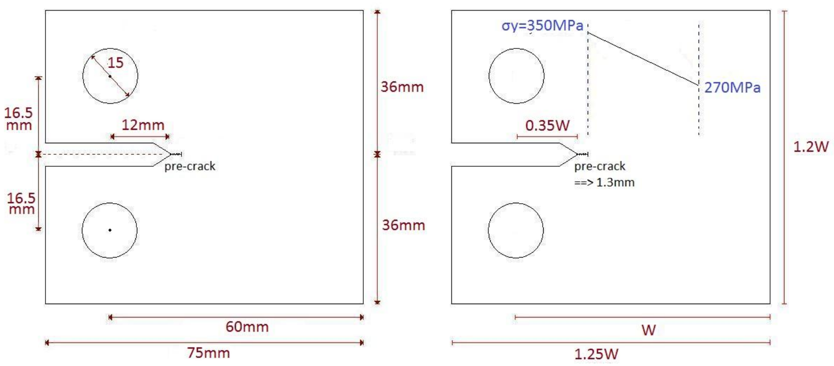


Figure 4. Stress-Strain curves of clad 2024 material in T3 and HT1 (250°C and 300°C) conditions

3.2 Fatigue crack growth tests

Fatigue crack propagation tests were conducted using compact tension C(T) specimens in accordance with ASTM E647-00. In Fig. 5(a) the dimensions of the C(T) specimen used in the fatigue tests is displayed. The tests were carried out on a 100KN servo-hydraulic fatigue machine at room temperature with a constant stress ratio of R=0.1. The maximum stress was σ_{max}= 10MPa and the frequency 5Hz. Crack length measurements were made using a crack opening displacement (COD) gauge and subsequent data evaluation by implementation of the compliance method.



(a) (b)

Figure 5. (a) C(T) specimen configuration according to ASTM E 647-00 (b) strength gradient boundaries in front of notch tip resulting from HT2

The crack growth characteristics were examined in T3, HT1 and HT2 conditions. The C(T) specimens with HT2 treatment were appropriately machined so that the axis at the notch tip perpendicular to the notch plane coincides with a boundary of strength gradient (Fig. 5b). Two types of specimens corresponding to HT2 condition were tested. Specifically, specimens with increasing and decreasing yield strength gradient with respect to the tip notch ranging from 155 to 120 Hv (see Fig. 2b).

In Fig. 6 the fatigue crack growth results of 2024 material in T3 state as well as HT1 heat treatment are compared. Crack growth rates were measured at an intermediate ΔK region ranging from 11 to 25 $MPa\sqrt{m}$.

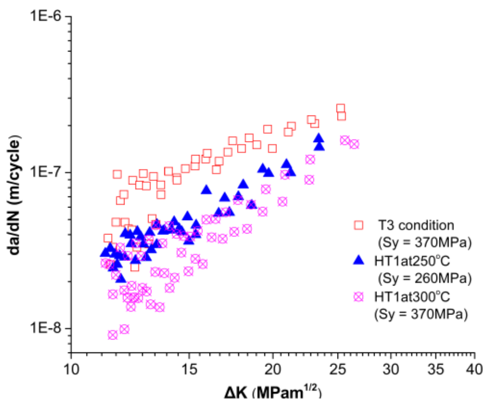


Figure 6. Fatigue crack growth rates (da/dN-ΔK) of 2024 material in T3 and HT1 conditions

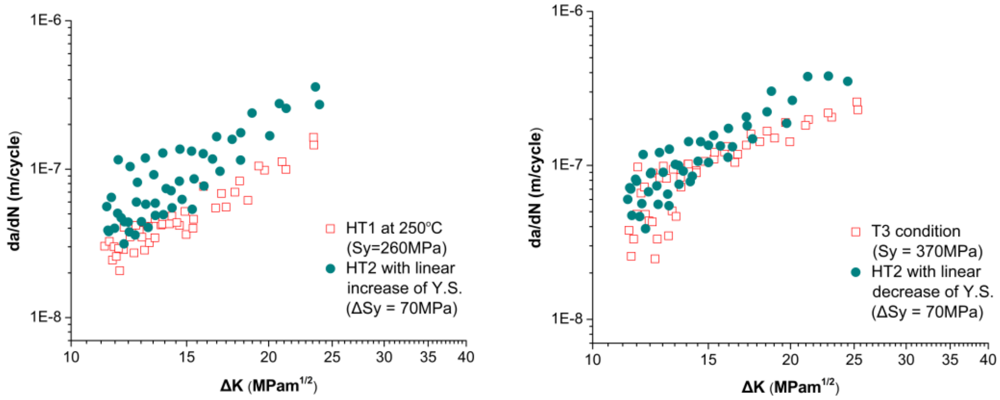
The results indicate a pronounced effect of HT1 treatment on FCG rates. Crack growth resistance is enhanced in HT1 specimens compared to T3 and increases further with the magnitude of temperature of HT1.

In the case of HT2 treatment an opposite effect is observed. The reference material behavior, in this case is the state of the material prior to entering the strength profile. Hence, HT2 treatment with increasing yield strength gradient (Fig. 2a) is compared to HT1 as reference (same yield strength at notch tip) and HT2 with decreasing yield strength gradient (Figs 2b or 5b) is compared to T3 condition as reference (same yield strength at notch tip of 350MPa).

In Figs 7(a) and 7(b) the da/dN vs ΔK crack growth data of HT2 specimens with increasing and decreasing yield strength gradient from the notch tip are compared to the reference behavior as described previously. In Fig. 7(a) the FCG results of 2024 material with HT2 treatment with linear increase of the local yield strength are compared to HT1 (250^{o}C).

Crack growth rates of HT2 specimens were higher with regard to the reference behaviour in the whole ΔK range examined.

In Fig. 7(b) the FCG results of HT2 specimens with linear decrease of yield strength are compared to the reference behaviour (T3 state). Again a degradation of crack growth resistance in HT2 specimens is observed which increases with increasing ΔK values.



(a) (b)

Figure 7. Fatigue crack growth rates (da/dN-ΔK) of 2024-T3 material (a) in HT1 at 250^{o}C and HT2 conditions with increasing yield strength with regard to the crack tip (b) in T3 and HT2 condition with linear decrease of the local yield strength with regard to the crack tip position conditions

4. DISCUSSION

The FCG behavior observed in HT1 specimens is divergent from the finding in the works [8,9] where intermediate ΔK crack growth rates in 2xxx aluminum alloys have been found to be unaffected by artificial aging process. However, in these works artificial aging was limited to peak aging conditions whereas in the present work overaging seems to trigger significant strain hardening due to continuous formation of dislocation loops and significant yield strength reduction. Coarsening of precipitates due to HT1 and HT2 treatment within the grains has been observed but their possible influence on stage II growth is not obvious and is further under investigation. Stage II crack growth according to established theories is dependent on closure effects (plasticity or surface roughness), residual stresses and strain hardening.

To assign a specific mechanism responsible for the crack resistance increase in HT1 specimens further investigation that takes into account the following parameters:

HT1 material at 300^{o}C has lower yield strength than HT1 at 250^{o}C, which has lower strength than material in T3 state. The lower yield strength may contribute to higher small scale plasticity and possible closure of crack surfaces as the crack propagates, but this effect requires further investigation with validation obtained by ΔK_{eff} experiments.

Of significant interest is the fact that tensile tests reveal that HT1 in 250^{o}C and 300^{o}C demonstrate an increasing strain hardening rate behavior with increasing overaging. This mechanism is caused by the formation of dislocation loops around strengthening particles and is assisted with strength decrease due to overaging. Material strain hardening rate at the crack tip may therefore be more significant in HT1 in 300^{o}C, less in 250^{o}C and even less in T3 state (see Fig. 4). This in combination with lower yield strength can be an explanation for reduced crack growth rates in HT1 materials which suffer more severe cyclic hardening at the crack tip with regard to T3 material.

In the case of yield strength gradient (HT2 state) the justification for increased crack growth rates is difficult due to the complexity induced by the strength variation at the crack path. It requires a better understanding of the underlying mechanisms that influence crack growth.

5. CONCLUSIONS

The effect of overaging conditions leading i) to uniform reduction and ii) to gradient of yield strength on fatigue crack growth rates of 2024 T3 aluminum alloy was investigated experimentally. The experimental findings showed that:

- Local uniform yield strength reduction at the material crack path (HT1) enhances the fatigue crack propagation behavior compared to the reference material.
- In material with local yield strength gradient at the crack path (HT2), crack growth rates were dependent on the slope of gradient at the crack tip. The crack growth rates were higher with respect to reference material in the case of increasing gradient. In the case of decreasing gradient the crack growth rates were not influenced at lower ΔK values but diverged to higher values than the reference material at higher ΔK values.

Acknowledgments

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