THE EFFECT OF OVERLOADING ON FATIGUE CRACK PROPAGATION IN TWO ALUMINIUM ALLOYS AND AN AUSTENITIC STAINLESS STEEL.

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ABSTRACT

The effect of single tensile overloads on fatigue crack propagation was investigated for three materials: 7178-T6 aluminium alloy, a high purity Al-Zn-Mg alloy and an austenitic stainless steel. A 'delayed' retardation of growth rate occurred in all three materials. The overload crack-tip plastic zones were mapped by microhardness testing and the effects of net section yielding and post-overload stress relieving indicated that retardation was caused primarily by compressive residual stresses acting upon the overload yield zone. It is proposed that a temporary increase in crack front length can be an additional cause for retardation. The application of a crack-front length correction factor to the theoretical ΔK was sufficient to account for retardation occurring beyond the boundary of the overload yield zone in the Al-Zn-Mg alloy.

KEYWORDS

Fatigue crack propagation; overloads; aluminium alloys; stainless steel.

INTRODUCTION

The retardation in fatigue crack growth rate da/dN following overloading has been widely established. Models for retardation (Matsuoka, Tanaka and Kawahara, 1976; Wheeler, 1972; Willenborg, Engle and Wood, 1971) have considered that the crack length Δa^* affected by an overload can be related to the sizes of the overload and steady-state plastic zones at the crack tip and that retardation is caused by compressive residual stresses and/or crack closure; other mechanisms have been proposed including crack tip blunting (Bathias and Vancon, 1978) and incompatibility of crack front orientation (Schijve, 1974). Experimental ivestigations have shown that Δa^* can in fact be equal to (Von Euw, Hertzberg and Roberts, 1972) smaller than (Rice and Stephens, 1973) or larger than (Inkle, Birkbeck and Waldron, 1976; Schijve, 1976) the estimated monotonic yield zone caused by the overload. Most of the investigations considered only a single material. The aim of the present investigation was therefore to study the retardation effect in a range of materials and to consider the retarded crack length in terms of the overload yield zone. The relevance of crack growth mode and crack-front length was also examined using optical and scanning electron microscopy.

EXPERIMENTAL PROCEDURES

The materials were 7178-T6 aluminium alloy extruded section, high purity Al-Zn-Mg alloy (6.60 wt. - %Zn, 2.70 wt. - %Mg) sheet and annealed En 58B austenitic stainless steel. The Al-Zn-Mg alloy had a grain diameter of 150μm and was tested in both the solution treated condition (45 min at 460°C and air cool to room temperature) and in the fully-aged condition (solution treated, water quenched and 16 hours at 135°C); air cooling was adopted for the unaged material to avoid the formation of quench bands found in a parallel study (Powell, 1979). The stainless steel was tested both with an 'as received' grain size of 18μm and a larger grain size of 100μm; grain growth was achieved by a 3% nominal tensile strain followed by 7 hr at 1080°C.

Fatigue tests were carried out on centre crack coupons having the following dimensions, 7178-T6: 120mm long x 32mm wide x 2.5mm thick, Al-Zn-Mg: 140mm x 45mm x 2.4mm, En 58B: 140mm x 25mm x 3.23mm. All tests were performed under zero-to-tension axial loading (R = 0) at 43 Hz in a laboratory air environment and crack length was measured by a travelling microscope to within 0.05mm. Single tensile overloads were applied to cracked testpieces at a crosshead speed of 2mm/min, which was the rate used to measure the tensile properties in Table 1; the overload frequency was approximately 0.01 Hz.

TABLE 1 Tensile Data

Material	0.2% Proof (MPa)	U.T.S. (MPa)	% Elong.
7178-T6	612	660	10
Al-Zn-Mg Aged	461	486	13
Al-Zn-Mg Unaged	164	315	33
En 58B 18μm Grain	307	663	66
En 58B 100 μm Grain	197	589	60

RESULTS

A summary of the overload test results is given in Tables 2 and 3 for the aluminium alloys and the stainless steel respectively. In all tests, minimum growth rate was not attained until the crack had propagated a distance Δa' from the overload crack length a_o; this is the phenomenon of 'delayed' retardation (Von Euw, Hertzberg and Roberts, 1972). After reaching a minimum, the growth rate then accelerated until the growth rate curve became parallel to the constant amplitude data and the retarded crack increment Δa* was taken to be the distance from a_o to this point. Various levels of overload were employed and the overload ratio p in Tables 2 and 3 is the ratio of the peak overload gross stress to the maximum stress of the baseline cycle; the latter was 75 MPa for the Al alloys and 125 MPa and 90 MPa for the small grained and large grained steel respectively. The estimated width w of the plastic zone was that proposed by Irwin (1960) for plane stress:

$$w = 2r_y = \frac{1}{\pi}\left[\frac{K_{max}}{\sigma_y}\right]^2 \qquad (1)$$

where σ_y is the 0.2% proof stress.

TABLE 2 Summary of Overload Results for Al Alloys

Material	p	a_o (mm)	$w = 2r_y$ (mm)	Δa* / w	Δa' / Δa*
7178-T6	1.50 2.00 2.50	5.4 3.55,5.05 4.1,4.5,5.7 6.15	0.18 0.21,0.30 0.39,0.42,0.54 0.58	3.9 3.3,4.2 7.5,5.9,6.5 1.8	0.07 0.04,0.10 0.02,0.03,0.07 0.01
Al-Zn-Mg Unaged	1.485 1.50 1.65 1.75 2.00 2.03	8.1 5.85 4.85 5.0,5.8 5.85,6.0 4.3	3.77 2.78 2.78 3.30,3.76 4.92,5.06 3.76	1.87 1.67 1.13 1.03,1.33 1.05,0.99 1.01	0.11 0.14 0.11 0.15 0.07,0.08 0.03
Al-Zn-Mg Aged	2.00 2.25	7.5,7.7 5.55,6.15	0.79,0.81 0.74,0.80	3.7,2.8 2.0,3.2	0.09,0.05 0.07,0.01

TABLE 3 Summary of Overload Results for Steel

Material	p	a_o (mm)	$w = 2r_y$ (mm)	Δa* / w	Δa' / Δa*
En 58B 18μm Grain	1.40 1.50 1.625 1.75 1.875 2.00 2.40	4.45 4.6 3.55,4.5 3.5,4.4 4.5 3.55,4.45 3.55	1.45 1.72 1.53,(1.97) 1.78,(2.23) (2.62) (2.35),(2.95) (3.39)	1.03 1.05 0.95,(0.86) 1.12,(0.83) (0.69) (1.19),(0.71) (0.74)	0.25 0.08 0.31,0.29 0.17,0.19 0.42 0.14,0.26 0.10
En 58B 100μm Grain	1.50 1.625 1.75 2.00	4.6 4.0 4.0,4.5 3.5,4.1	(2.15) (2.20) (2.55),(2.87) (2.90),(3.42)	(0.51) (1.00) (1.41),(1.05) (1.03),(0.99)	0.59 0.32 0.21,0.20 0.15,0.21

Figures in parentheses indicate net section yield during overload.

It is seen that there was reasonably good agreement between Δa* and w for the stainless steel (Table 3) but that in the aluminium alloys, Δa* could be up to seven times larger than the estimated plastic zone. Microhardness measurements were made ahead of the crack tip to estimate the extent of the plastic zone. In all the materials the boundary of the strain hardened region was close to the estimated boundary of the plastic zone; the extensive strain hardening capacity of the stainless steel made it particularly suited to this technique and Fig. 1 is a typical plot of microhardness. It was noted that the maximum hardness at tne crack tip (432 HV) was similar to that measured in the necked zone of a fractured tensile testpiece (439 HV). In the stainless steel, the net section stress

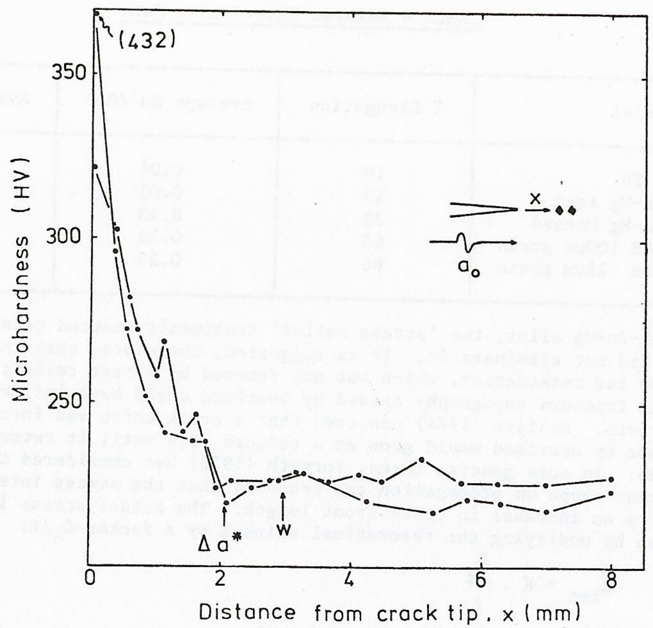

Fig.1 **A microhardness plot at the crack tip in an En 58B testpiece, p=2.0 at a_o=4.45mm.**

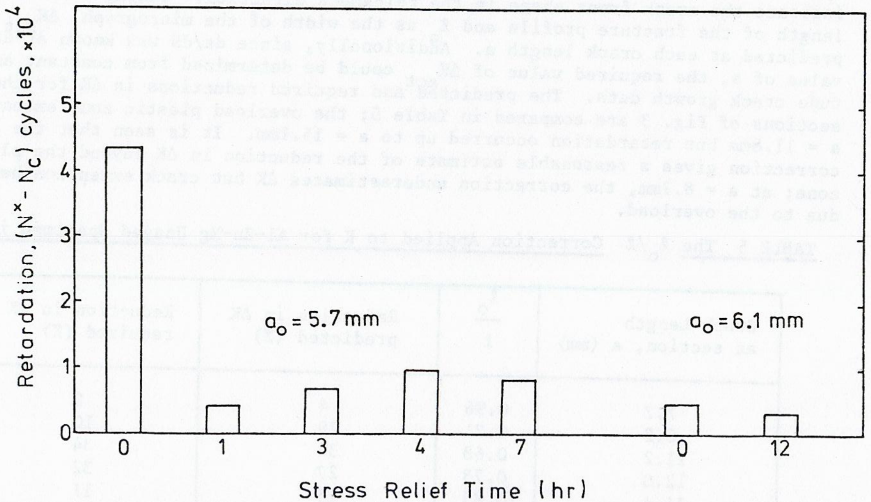

Fig.2 **The effect of stress relief on retardation in the aged Al-Zn-Mg alloy.**

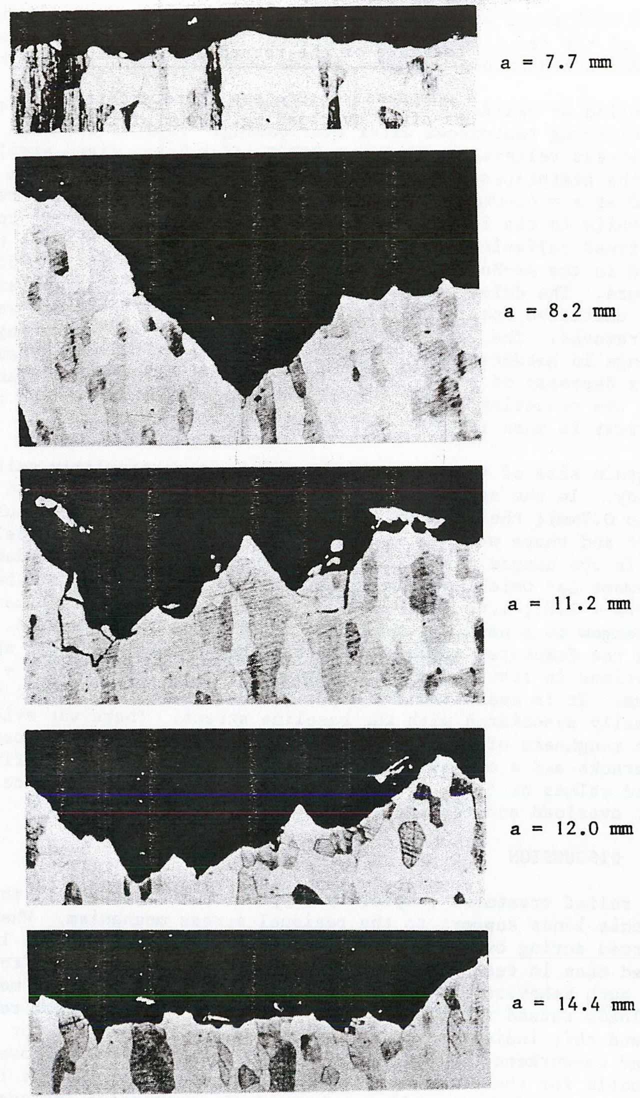

Fig.3 **Sections on the fracture of an unaged Al-Zn-Mg testpiece, normal to the direction of crack growth. (overload p = 1.485 at a_o = 8.1 mm)**

exceeded the 0.2% proof stress in all of the tests on the large grained material and in five of the tests on the small grained material. Nevertheless, equation 1 still provided a reasonable estimate of the retarded crack increment.

The contribution of residual compressive stresses to retardation was investigated by stress relieving testpieces after overloading. Four of the 18μm grain size steel were stress relieved for 1 hr at 900°C after being given similar overloads to four of the testpieces in Table 3, namely p = 1.75 at a = 4.55mm and 3.50mm and p = 2.00 at a = 4.40mm and 3.55mm. In the first three tests, retardation was eliminated while in the final test, it was reduced from 3.3×10^5 to 2.8×10^4 cycles by stress relieving. The effect of a post overload thermal treatment was also studied in the Al-Zn-Mg alloy; testpieces were held at 135°C for periods of up to 12 hours. The delay period was shorter in every stress relieved specimen than in the untreated specimens (Fig. 2) but in none of the tests was the delay completely removed. The 'stress relief' was performed at the ageing temperature but the change in hardness of the testpieces was small, ranging from an increase of 5 HV to a decrease of 4 HV. It is unlikely that these small changes could account for the reduction in delay and an explanation based on the relief of residual stress is more likely.

The large grain size of the Al-Zn-Mg alloy made it particularly suited to fractographic study. In the aged material, overloading frequently caused crack extension (0.1 to 0.7mm); the fracture mode was predominantly intergranular, as in the tensile test and there were often secondary intergranular cracks below the main fracture. In the unaged material, three tests demonstrated retardation over a crack increment Δa* which was larger than the estimated width of the plastic zone w (viz. Δa*/w = 1.87, 1.67 and 1.33). Examination of these specimens showed that they had changed to a partly slant fracture (45° mode) at overload. Sections were taken along the fractures normal to the fracture plane and Fig. 3 shows a sequence of microsections in front of and following an overload of ratio p = 1.485 applied at a = 8.1mm. It is seen that the crack gradually returned to the 90° mode of growth normally associated with the baseline stress. There was evidence of an increase in roughness of the fracture after overload and the microsections show secondary cracks and a marked 'hill and valley' profile. In the five tests, which demonstrated values of Δa*/w closer to unity, there was no evidence of slant fracture at overload and the profiles remained reasonably flat.

DISCUSSION

The stress relief treatment completely suppressed retardation in the stainless steel and this lends support to the residual stress mechanism. When net section yield occurred during overload, a smaller increase in propagation life ($N^* - N_c$) was obtained than in tests where yield was confined to a plastic zone at the crack tip; such behaviour is consistent with the residual stress mechanism. The large overloads caused visible crack tip blunting, which was not removed by stress relieving and this indicates that blunting was not responsible for retardation (Bernard and co-workers, 1977). Matsuoka and Tanaka (1978) proposed that blunting was responsible for the phenomenon of 'delayed retardation' since it would offset crack closure and reduce the effect of residual compressive stresses during the early stages of retardation. The trend in the present work is consistent with their theory; if tensile elongation is taken as an indication of the ability of each material to undergo crack blunting, then increased elongation did correlate with a greater delay before reaching minimum growth rate (i.e. increased Δa'/Δa*) (Table 4). There did not appear to be a correlation between Δa' and the reversed plastic zone w/4 as proposed by Von Euw and co-workers (1972) for 2024-T3 Al alloy.

TABLE 4 Delayed Retardation Data

Material	% Elongation	Average Δa'/Δa*	Average Δa'/w
7178 T6	10	0.04	0.62
Al-Zn-Mg Aged	13	0.05	0.12
Al-Zn-Mg Unaged	33	0.13	0.18
En 58B 100μm grain	60	0.18	0.21
En 58B 18μm grain	66	0.23	0.22

In the Al-Zn-Mg alloy, the 'stress relief' treatments reduced retardation by up to 90% but did not eliminate it. It is suggested, therefore, that there was secondary cause for retardation, which was not removed by stress relief; the marked change in fracture topography caused by overload could have influenced subsequent crack growth. Schijve (1974) proposed that a crack which was forced into the 45°, slant mode by overload would grow at a reduced rate until it returned to the 90°, flat mode. In more general terms, Forsyth (1978) has considered the effect of crack front shape on propagation and proposed that the stress intensity factor was reduced by an increase in crack front length. The actual stress intensity K_{act} was given by modifying the theoretical value K by a factor ℓ_o/ℓ:

$$K_{act} = K \cdot \frac{\ell_o}{\ell} \qquad (2)$$

where ℓ = actual crack front length
ℓ_o = minimum possible crack front length ≡ specimen thickness

The present fractures did not unfortunately reveal the shape of the crack front at specific crack lengths after an overload but microsections such as Fig. 3 did indicate the crack front shape in the thickness direction. Taking ℓ to be the length of the fracture profile and ℓ_o as the width of the micrograph, ΔK_{act} can be predicted at each crack length a. Additionally, since da/dN was known at each value of a, the required value of ΔK_{act} could be determined from constant amplitude crack growth data. The predicted and required reductions in ΔK for the micro sections of Fig. 3 are compared in Table 5; the overload plastic zone extended to a = 11.8mm but retardation occurred up to a = 15.1mm. It is seen that the ℓ_o/ℓ correction gives a reasonable estimate of the reduction in ΔK beyond the plastic zone; at a = 8.2mm, the correction underestimates ΔK but crack extension here was due to the overload.

TABLE 5 The ℓ_o/ℓ Correction Applied to K for Al-Zn-Mg Unaged Specimen in Fig.3

Crack Length at section, a (mm)	$\frac{\ell_o}{\ell}$	Reduction in ΔK predicted (%)	Reduction in ΔK required (%)
7.7	0.96	4	0
8.2	0.71	29	10
11.2	0.68	32	34
12.0	0.73	27	32
14.4	0.85	15	11

Similar results for an aged Al-Zn-Mg testpiece given an overload of p = 2.0 at a_o = 7.5mm are shown in Table 6; the estimated plastic zone extended to a = 8.3mm, while retardation occurred up to a = 10.4mm. The ℓ_o/ℓ correction again accounted for retardation operating beyond the plastic zone boundary.

TABLE 6 The ℓ_o/ℓ Correction Applied to K for an Al-Zn-Mg Aged Testpiece

Crack Length at section, a (mm)	$\frac{\ell_o}{\ell}$	Reduction in ΔK predicted (%)	Reduction in ΔK required (%)
7.1	0.90	10	0
7.6	0.74	26	52
8.9	0.83	17	25
9.5	0.87	13	15

The stainless steel did not demonstrate a marked change in fracture topography after overload and the good agreement between Δa^* and w indicates that crack front length did not play an important role in the retardation process. In 7178-T6, however, Δa^* exceeded w and we might consider the relevance of crack front length to this alloy. Overloads occasionally caused tensile 'pop-in' fracture which would have presented a rough, bowed crack front during subsequent growth. When 'pop-in' did not occur a stretch zone formed, which was of the order of 10μm wide and inclined to the main fracture plane. It is unlikely that the width of this zone was constant across the specimen thickness and consequently, the crack front would have been disjointed. A reduction in ΔK would have operated until the crack front regained its uniformity and such a process could have required a crack growth increment in excess of the plastic zone width.

CONCLUSIONS

A study of the effects of overloading on fatigue crack growth in two aluminium alloys and an austenitic stainless steel indicated that retardation was primarily due to residual compressive stresses acting upon the overload plastic zone. Overloads can also cause a temporary increase in crack front length, which can provide a secondary cause for retardation and should be considered in fatigue crack growth analyses for variable amplitude loading.

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