# Relation Between Striation Spacing and Fatigue Crack Growth Rate in A1-Alloy Sheets

I. NEDBAL, J. SIEGL and J. KUNZ

*Technical University of Prague, Faculty of Nuclear Science and Pysical Engineering, Department of Materials, Křemencova 10, 110 00 Praha 1, Czechoslovakia*

ABSTRACT

The paper points out the key role of the relation between striation spacing (**s**) and macroscopic crack growth rate (**v**) in applications of fractographic reconstitution of fatigue crack history. The real course of the **v/s** ratio was obtained by fractographic analyses of fatigue fractures of AlCu4Mg1 sheets. Experimental results are discussed and an attempt has been made to link micro- and macro- characteristics of fatigue crack growth and relate them to fractographic findings.

KEYWORDS

Quantitative fractography, fatigue striation spacing, fatigue crack propagation, aluminium alloys, aircraft structures.

## INTRODUCTION

The main input for the fractographic reconstitution of fatigue crack kinetics is the dependence of striation spacing **s** on the fatigue crack length **a**, i.e. the SEM data represented in the form of **s = s(a)**. As previously reported by Nedbal et al. (1984), fatigue crack history reconstitution can be based on the formula

$$N_x = \int_{a_i}^{a_x} \frac{da}{D(s).s(a)} + N_i \,, \qquad (1)$$

which enables the transformation of fractographic results to the crack propagation curve **a = a(N)** and/or to the rate relation **v = v(N)**. Eqn. (1) can be modified to suit the available input data - the different ways of it can be used are summarized and the importance of the factor **D** was pointed out by Nedbal et al.

(1984). The value of **D** reflects a result of stochastic interactions of different micromechanisms on the crack front (Nedbal, 1979) : in the sense of a link between micro- and macroscopic features of the fatigue process, **D** can be represented as the ratio **v/s** (where **v** = **da/dN** is the macroscopic propagation rate and **s** is the striation spacing). Numerous fractographic analyses of failured aircraft structures proved the importance of **D** = **v/s** and demonstrated the influence of **D** on the precision of the fractographic reconstitution of the crack propagation curve **a** = **a(N)** and on the calculated number of load cycles necessary for the forming of a detectable crack (Nedbal et al., 1988). Thus, the accumulation of information on the factor **D** is considered to be prospective for the further development of the quantitative fractography of fatigue fractures.

The view that striation spacing **s** is equivalent to the macroscopic rate **v** of the fatigue crack front was supported by a number of authors (e.g., Mills and James, 1980; Hertzberg, 1983; Klingele, 1984). In fact, however, this equivalence can be accepted as a mere tentative presumption only in a very limited interval of crack propagation rates. Generally, the ratio **v/s** (≠ 1) changes with the fatigue crack growth rate (e.g., Bathias and Pelloux, 1973; Koterazawa et al., 1973; Broek, 1974; Yokobori and Sato, 1976; Wareing and Vaughan, 1977). But so far, the information as to this dependence is both insufficient and unsatisfactory for practical purposes of fractographic reconstitution of fatigue crack history.

## EXPERIMENTAL

Fractographic analyses were performed on fatigue fractures of thin-walled bodies of AlCu4Mg1 alloys type 2024. Central notched specimens of the aircraft fuselage sheets (of width W = 400 mm and thickness B = 1.2 mm or 2.0 mm) were made of two alloys differing in purity. Fatigue tests were carried out at ARTI (Aeronautical Research and Testing Institute, Prague) under constant stress amplitude loading. Fatigue crack propagation rate was measured optically on the specimen surfaces. The set of four specimens for fractography was picked out with the aim to verify the influences of sheet thickness **B**, stress ratio **R** and the purity of alloy (see Table 1).

Table 1. Specimens Characteristics

| Spec. No. | Sheet thickness **B** \|mm\| | Stress ratio **R** \|1\| | Material |
|---|---|---|---|
| 1 | 1.2 | 0.030 | I - AlCu4Mg1 normal purity |
| 2 | 1.2 | 0.030 | II- AlCu4Mg1 reduced contents of Fe and Si |
| 3 | 2.0 | 0.018 | |
| 4 | 2.0 | 0.580 | |

Striation spacing **s** was measured on the SEM micrographs. The results in Figs. 1 and 2 are shown as dependence on stress intensity factor range and they clearly indicate both the influence



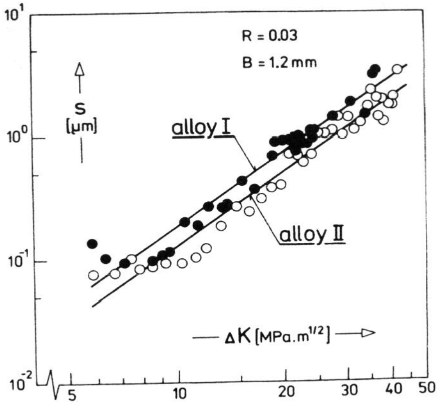



Fig. 1. Influence of alloy purity on s(ΔK).

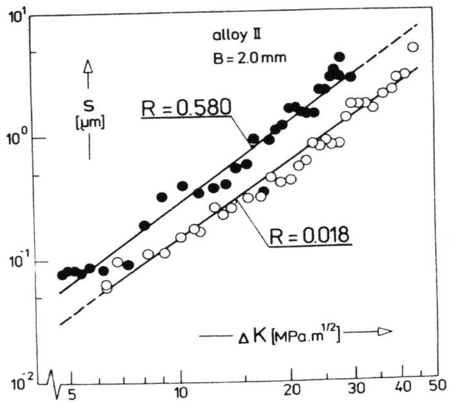



Fig. 2. Influence of stress ratio **R** on s(ΔK).

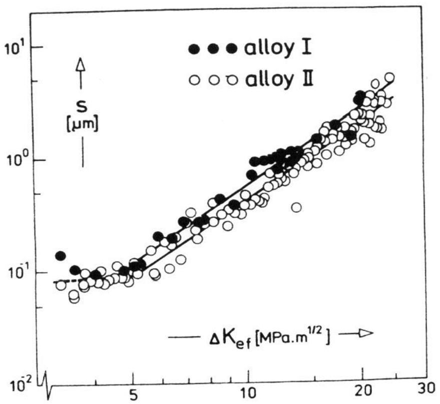
of alloy purity (Fig. 1) and the influence of stress ratio (Fig. 2). If the results of striation spacing measurements are expressed in the form of the Paris - Erdogan law analogy **s** = $C(\Delta K)^m$, exponents for specimens No. 1 to 4 are **m** = 1.987, 1.986, 2.040, and 2.144, respectively, i.e., the slopes of regression lines are practically the same in all the cases studied. Differences in constants **C** (due mainly to differences in **R**) may be eliminated by transforming $\Delta K$ to $\Delta K_{ef}$ (Schijve, 1981) :

$$\Delta K_{ef} = (0.55 + 0.35\,R + 0.1\,R^2).\Delta K \quad (2)$$

Striation spacing **s** vs. $\Delta K_{ef}$ is plotted in Fig. 3-a, where the estimated regression lines for alloys I and II are drawn only for $\Delta K_{ef} \geqslant 5$ MPa.m$^{1/2}$. For $\Delta K_{ef} < 5$ MPa.m$^{1/2}$, the influence of alloy purity is not perceptible and the dependence of striation spacing on $\Delta K_{ef}$ is significantly "weaker" than in the range of higher $\Delta K_{ef}$ (the exponent of the dotted regression line in Fig. 3-a is **m** = 0.18). If we take into account our former results and the high data dispersion, then the configuration of experimental points for the low $\Delta K_{ef}$ might suggest striation spacing independence of the stress intensity factor range. Analogous results for a microalloyed steel were reported by Roven et al. (1987).

The comparison of regression lines from Fig. 3-a with the results of the macroscopic crack rate measurements (carried out at ARTI) is presented in Fig. 3-b. By rearranging the **s** = $s(\Delta K_{ef})$ and **v** = $v(\Delta K_{ef})$ dependences, the plot of **D** = **v/s** vs. $\Delta K_{ef}$ is obtained, as shown in Fig. 4 for three specimens of alloy II and for specimen No. 1 of alloy I. The influence of the different degree of alloy purity is evident : the pronounced increase of **D** in the right branch of the dependence is shifted to a higher $\Delta K_{ef}$ for alloy II with lower contents of Fe and Si.

The shape of **D** = $D(\Delta K_{ef})$ in Fig. 4 can be explained as a simultaneous influence of three basic factors :



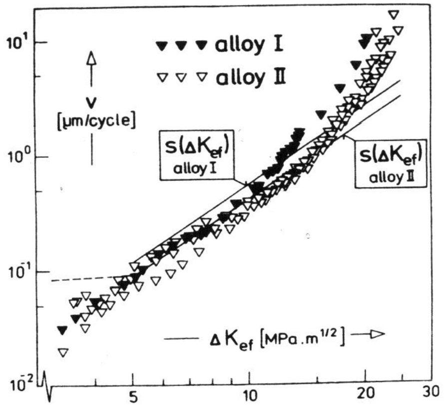




Fig. 3. Experimental data in dependence on $\Delta K_{ef}$ :
a - Microfractographic data from Figs. 1 and 2.
b - Macroscopic data vs. regression lines from Fig. 3-a.

a/ Existence of idle cycles. For one striation to form, it is necessary to apply **n** load cycles, i.e., **(n - 1)** idle cycles have only a latent effect on the crack tip and they do not directly contribute to the crack growth.

b/ Spatial dispersion of local crack growth directions (i.e., deflection of local crack rate vectors from the macroscopically defined direction of fatigue crack propagation). This can be characterized by angles   between the local vectors projection in the picture plane and the direction of macroscopic crack growth (see Fig. 5-a).

c/ Influence of crack growth micromechanisms other than the striating one (e.g., ductile fracture, quasi-cleavage, and/or intermetallic phase decohesion, etc.). The information on the contribution of these "non-striating" mechanisms is recorded in fracture micromorphology (see Fig. 5-b) and it can be quantified by means of the area percentage $p_s$ of striation paths on the fracture surface (Nedbal et al., 1988).

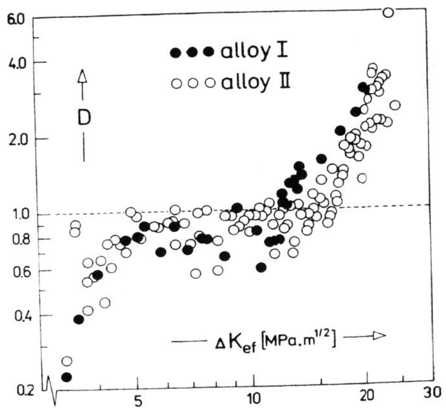



Fig. 4. Influence of alloy purity on $D(\Delta K_{ef})$.

The importance of these three factors in the three branches of $D = D(\Delta K_{ef})$ in Fig. 4 is changing with the value of $\Delta K_{ef}$, i.e., with the fatigue crack growth rate. In the range of low $\Delta K_{ef}$ , where the crack is propagating predominantly by the striating micromechanism, the value of $D = v/s$ is controlled mainly by the occurence of idle cycles. Their influence may not be understood in the macroscopic sense, i.e., as an arrest of the whole crack front in the course of **(n - 1)** cycles. Even if the macroscopic effect of idle cycles can be expressed as

$$da/dN \cong \frac{1}{n} \cdot s \ , \qquad (3)$$

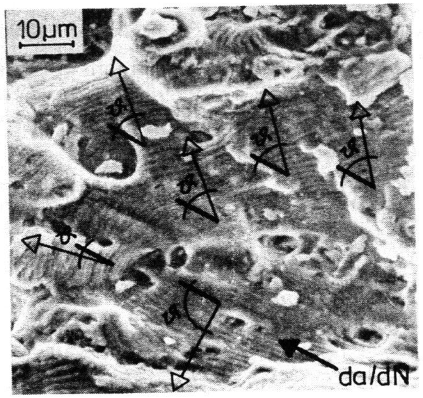
the number **n** has to be considered as a weighted mean value of local values in individual striation paths (see Nedbal et al., 1988). Unfortunately, the direct fractographic determination of **n** is not possible.

The left branch of $D = D(\Delta K_{ef})$ merges into the middle section, where $D$ = const. Here the striation paths are always a dominant feature of the fracture surface and the value $D < 1$ is influenced mainly by the divergence of local growth directions, the influence of idle cycles being scaled down. This can be proved by measuring angles $\vartheta$ : if the average deflection $\bar{\vartheta}$ of the local vectors from the macroscopic growth direction (measured on a chosen crack length **a**) is suitable for

$$da/dN \cong s \cdot \cos\bar{\vartheta} \ , \qquad (4)$$

each load cycle has formed one striation in the observed area of fracture surface (i.e., **n** = 1).

The passage from $D < 1$ to $D > 1$ in the right branch of dependence in Fig. 4 is controlled by an increasing part of crack growth micromechanisms (others than the striating ones) taking a share in the crack front advance. For example, the fractographic analysis of thick-walled body fractures has shown, for alloy type AlCu4Mg1, a drop in the area percentage of striation paths from $p_s$ = 0.9 (for $\Delta K_{ef}$ = 9 MPa.m$^{1/2}$) to $p_s$ = 0.1 (for



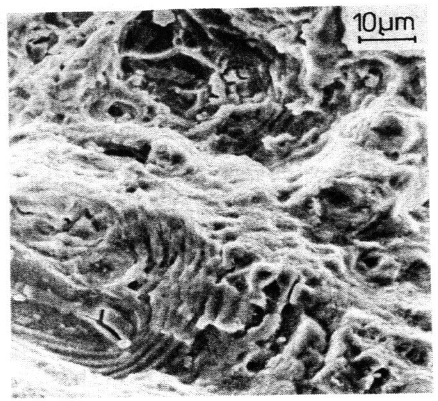




Fig. 5. Fatigue fracture of AlCu4Mg1 alloy :
a - Divergence of local crack rate vectors (for low $\Delta K_{ef}$ range).
b - Coexistence of different micromechanisms (for high $\Delta K_{ef}$ range).

$\Delta K_{ef}$ = 14 MPa.m$^{1/2}$), $p_s$ tending to zero for higher $\Delta K_{ef}$ (Nedbal et al., 1988). In the range of $\Delta K_{ef}$ > 9 MPa.m$^{1/2}$, striation paths are gradually replaced by ductile fracture areas. From the point of view of the local failure processes, these ductile fracture areas correspond to local jumps of small sections of the crack front. Their exact instantaneous rate is not known but it is certain to be higher than the crack growth rate in the adjacent striation paths. These randomly spaced local jumps increase the average macroscopic rate of crack propagation and along with their effect, the influence of the local rate vector divergence in striation paths becomes negligible. Hence, the macroscopic crack growth rate in the ascending right branch in Fig. 4 can be schematically expressed as

$$da/dN \cong s \; . \; f(p_s) \; , \qquad (5)$$

where $f(p_s)$ increases with $p_s$ drop (i.e. with increasing $\Delta K_{ef}$). The values of $p_s$ can be measured on fracture surface as dependence on crack length **a** (or on $\Delta K_{ef}$) but an applicable form of function $f(p_s)$ has to be determined in relation to striation spacing **s** or directly to the factor **D** (an example of dependence $D = D(p_s)$ was presented by Nedbal et al. (1984)).

Consequently, the course of factor **D** may be described on the basis of Eqns. (3) to (5) as

$$D = v/s = \frac{1}{n} \; . \; \cos\bar{\vartheta} \; . \; f(p_s) \; , \qquad (6)$$

where the number of idle cycles **n**, the divergence of local vectors $\bar{\vartheta}$, and the area percentage of striation paths $p_s$ vary with $\Delta K_{ef}$. An analysis of the dependence $D = D(\Delta K)$ or $D = D(\Delta K_{ef})$ can be useful in studying the influence of input factor changes on fatigue crack growth process. Thus, for example, it is possible to estimate that the difference in purity of alloys I and II has exerted no important influence on the number of idle cycles. The principal cause of a gain in specimen life (in the course of macroscopic crack growth stage) is encoded in the shift of the right branch of $D(\Delta K_{ef})$ in Fig. 4 : the higher purity of alloy II has retarded the rise of the proportion of ductile local jumps on the crack front, i.e., the mean macroscopic rate of the crack propagation was lower (see Fig. 3-b).

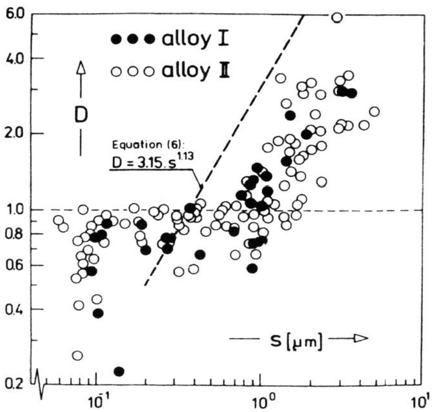



Fig. 6. Factor **D** = **v/s** in dependence on striation spacing **s**.

A direct application of the D-factor in fractographic reconstitution of fatigue crack history prefers dependence $D = D(s)$ (see Eqn. (1)). Thus the data from Fig. 4 were rearranged in Fig. 6 : $D = D(s)$ has qualitatively the same S-shape as



$D = D(\Delta K_{ef})$. Here the influence of alloy purity is not discernible - the input factors affect both variables, i.e., D-factor as well as striation spacing. In addition, the experimental data have a relatively high dispersion corresponding to the stochastic character of interactions of the local processes on the crack front. However, it is again possible to distinguish three branches of $D = D(s)$ : the left one is not too auspicious for fractographic reconstitution of true crack kinetics. Its slope is influenced by an unspecified number of idle cycles and it might be unfavourably affected by the above possibility **s** = const. in the low $\Delta K_{ef}$ range. In the middle section (in the range of **s** = (0.1 to 1.0) um) **D** = 0.8. This **s**-range corresponds to the crack rate range being frequent in failure analyses of airframe structures. The right branch of $D = D(s)$ is very important for the kinetics reconstitution of fast fatigue cracks, and, above all, for cracks in aircraft fuselage skins. For a very rough evaluation of macroscopic crack growth rate **v** in thin-walled bodies ofAlCu4Mg1 alloy, it is possible to use an approximate relation $v \doteq k.s^2$ (where $k \doteq 1$ in the range $v > 1$ μm/cycle).

Experimental data in Fig. 6 can be compared with the result of Foth and Schütz (1984) : based on results of fatigue tests of five different structural alloys (2 types of Al-alloy, Ti-alloy, high-strength steel and Inconel), a universal formula $s = s(v)$ was obtained for the crack growth rata range ($10^{-4} < v < 4.10^{-2}$) mm/cycle :

$$s = 1.5 \; . \; 10^{-2} . \; v^{0.47} \; . \qquad (7\text{-a})$$

Eqn. (7-a) can be transformed as

$$D = v/s = 3.15 \; . \; s^{1.13} \; , \qquad (7\text{-b})$$

where **s** is given in μm and the transformed validity interval is (0.2 < s < 3.3) μm. The line corresponding to Eqn. (7-b) is plotted in Fig. 6 : the relation $s = s(v)$ proposed by Foth and Schütz (1984) does not consider the qualitative changes at the different stages of crack growth. Generally speaking, a search for some universal formula (valid for various materials) can be considered problematic : the analysis of fractographic findings gives support to the strongly held belief in the important role of microstructure, predetermining the character of interacting processes on the crack front and their influence on **v/s** ratio.

The effect of the level of knowledge of the factor **D** is demonstrated by a simple example in Fig. 7. It presents the fractographic input data (e.g., the dependence $s = s(a)$ measured on the fracture surface of the flange-plate failured in the course of a full-scale test of aircraft wing) and the final results, i.e., crack growth curves $a = a(N)$. These curves were calculated on the basis of three different assumptions : 1/ **D** = 1 (i.e., **v** = **s**); 2/ $D = D(s)$ (corresponding to the rough evaluation of an average curve extracted from Fig. 6); 3/ **D**-factor values determined by Eqn. (7) (Foth and Schütz, 1984). Fig. 7 illustrates how these assumptions affect the course of curves $a = a(N)$. Another example including comparison with reference points (**a** ; **N**) was presented by Nedbal et al. (1988). The contribution of more exact information on ratio **v/s** is evident in determining the number of cycles necessary for initiation and growth of a detectable crack, and/or in determining the residual strength of a component at different stages of test or exploitation.

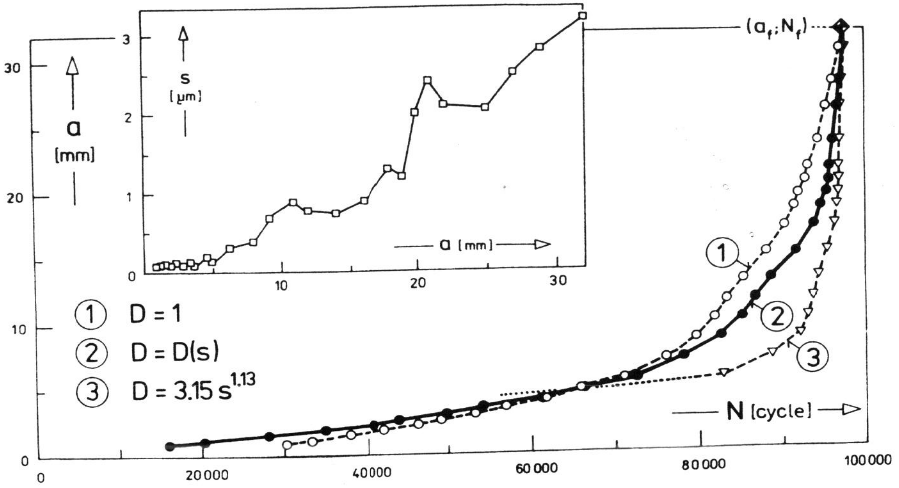



Fig. 7. Reconstitution of crack growth **a(N)** based on microfractographic data **s(a)** and on different assumptions on **D(s)**.

## CONCLUSION

Quantitative fractography of fatigue fractures can lead to a broader and deeper understanding of fatigue crack growth. Fractographic reconstitution of crack front kinetics can provide detailed information - so far unobtainable by any other means - on the course of macroscopic crack growth and its anomalies under heterogeneous conditions of large-scale structures. Unless a marking load spectrum was applied, striation spacing is the principal source of information quantitatively characterizing the failure process. The reliability of data transformation from microvolume to macroscopic scale largely depends upon the level of knowledge available on the relation between macroscopic crack propagation rate **v** and striation spacing **s**. For the reconstitution method summarized above, the link between micro- and macroscopic data is expressed by factor **D** = **v/s**, whose dependences $D(\Delta K_{ef})$ and D(s) were given for thin-walled bodies of AlCu4Mg1 alloy. The relation between **s** and **v** is controlled by quantitative changes of qualitatively differing and simultaneously acting processes on the crack front. Stochastic interactions of local failure processes in the microstructure result in changes in the number of idle cycles, influence the divergence of local crack rate vectors, and change the proportion of different micromechanisms of crack front advance, the effect of these factors varying in dependence on the crack growth stage. Thence, the equivalence **v** = **s** is not a tenable notion - in the case under investigation, the value of ratio **v/s** ranges from ~ 0.2 to ~ 6.0. Further experimental research is thus desirable in order to acquire analogous data even for other materials and for evaluating the contingency influence of body geometry.



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