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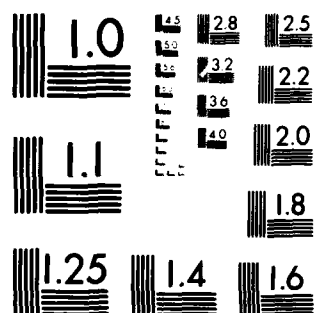
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MATERIALS REPORT 110

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**FATIGUE-CRACK GROWTH IN D6AC STEEL
HEAT-TREATED TO DIFFERENT VALUES
OF FRACTURE TOUGHNESS**

by

N. E. RYAN

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DOCUMENT CONTROL DATA SHEET

Security classification of this page: Unclassified

1. Document Numbers

- (a) AR Number:
AR-001-734
- (b) Document Series and Number:
Materials Report 110
- (c) Report Number:
ARL-Mat-Report-110

2. Security Classification

- (a) Complete document:
Unclassified
- (b) Title in isolation:
Unclassified
- (c) Summary in isolation:
Unclassified

3. Title: FATIGUE-CRACK GROWTH IN D6AC STEEL HEAT-TREATED TO DIFFERENT VALUES OF FRACTURE TOUGHNESS

4. Personal Author(s):

N. E. Ryan

5. Document Date:

April, 1979

6. Type of Report and Period Covered:

7. Corporate Author(s):

8. Reference Numbers

- (a) Task:
Air 72/8
- (b) Sponsoring Agency:

9. Cost Code:

35 1610

10. Imprint:

11. Computer Program(s) (Title(s) and language(s)):

12. Release Limitations (of the document)

Approved for public release

12-0. Overseas:

N.O.	P.R.	I	A	B	C	D	E
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13. Announcement Limitations (of the information on this page):

No Limitation

14. Descriptors:

Fatigue (materials)
Steels
Crack propagation

Fractography
Fractures (materials)
Fracture strength

15. Cosati Codes:

1106
2012

16.

ABSTRACT

Measurements of fatigue-crack growth in D6AC steel, heat-treated to various fracture-toughness levels, show that, under conditions where the growth mechanism produces striation markings and the crack extension per cycle is a linear function of a power of the stress intensity range, the rate of fatigue-crack growth is independent of fracture-toughness. At peak stress intensity values greater than $0.7 K_{IC}$, the rate of growth accelerates as the critical stress intensity is approached and tensile modes of crack extension occur. At peak stress intensity values less than $18 \text{ MPa m}^{1/2}$, fractographic examination indicates a change in mechanism as the crack extension per cycle decreases more rapidly with decreasing peak stress intensity, approaching a threshold value of stress intensity dependent upon heat-treatment.

The implications of these results for the choice of materials for service applications are discussed.

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1. INTRODUCTION

Linear-elastic fracture mechanics, with its concept of a stress-intensity factor $K\alpha\sigma_0a^{1/2}$ (where a = length of a pre-existing crack and σ_0 = the stress applied remote from the crack), provides a basis for the quantitative analysis of various crack-growth and fracture processes. The maximum, or critical, stress intensity factor, K_{1C} of a material relates to the onset of *unstable* crack growth to fracture and thus measures "fracture toughness". Sub-critical values of the stress intensity provide a measure of the mechanical 'driving force' for *stable* (sub-critical) crack-growth processes.

Fatigue cracking is a stable crack-growth process whose rate, according to Paris and Erdogan¹, can be represented by the relationship

$$da/dN = C\Delta K^m$$

where

da/dN is the crack extension per cycle

ΔK = the cyclic amplitude of stress intensity, $K_{\max} - K_{\min}$

C a material constant and m an exponent ≈ 4 .

Experimentally determined values of m commonly lie between 2 and 4 although values up to 8 have often been noted; wide variations in C are also obtained²⁻⁶. Factors influencing the values of C and m have received considerable attention over the past decade in the hope of formulating laws for predicting rates of fatigue-crack growth. The physical processes contributing to fatigue-crack extension are, however, not clearly understood and a number of theoretical expressions for fatigue-crack growth have been derived from bulk physical and mechanical properties, e.g. elastic modulus E , proof or yield stress σ_y , ultimate strength σ_U , fracture toughness K_{1C} and/or fracture strain ϵ_F . Good reviews of the various fatigue crack-growth laws have been given by Pelloux⁷ and Schwalbe⁸. Theoretical studies, in general, have concentrated on producing expressions to fit simple growth-rate relationships of the Paris-Erdogan form. Meanwhile, more careful experimental studies have shown that the relationship between fatigue crack-growth rate and ΔK cannot be attributed to one single mechanism. Departure from a simple power law was noted by Liu⁹ in 1964. Current analyses show that the relation between $\log da/dN$ and $\log \Delta K$ is a sigmoidal curve with three characteristic regions^{10,11,12}:

- (i) a region, corresponding to low growth rates and low ΔK , where a 'fatigue-threshold' cyclic stress intensity factor (K_{th}) is approached below which cracks might be considered as non-propagating,
- (ii) a central region where the linear relationship between $\log da/dN$ and $\log \Delta K$ is maintained, and
- (iii) a region where the rate of growth accelerates as the ΔK or $K_{\max}$ value approaches K_{1C} .

In the work described here, the fatigue-crack growth rates in D6AC steel, heat-treated to produce a variety of mechanical properties have been examined. Particular attention was given to the influence of fracture toughness, since recent studies^{13,14} have shown that the fracture toughness could be varied (by different heat-treatment schedules), from 40 to 105 MPa.m^{1/2} without significantly changing the tensile properties.

2. EXPERIMENTAL

2.1 Steel Composition

D6AC steel manufactured to specification AMS6438 had the following chemical composition.

C	Mn	Si	P	S	Cr	Ni	Mo	V	Fe
0.44	0.75	0.22	0.004	0.002	1.08	0.7	1.0	0.1	rem. wt %

2.2 Test Specimens and Heat Treatment

Standard tensile and ASTM. T_2 , 19 mm thick, compact-tension, fracture-toughness specimens were machined from a single batch of 32 mm thick bar and heat-treated to four different schedules:

- A. Austenitize 930°C
Intermediate 'Ausbay' quench 520°C
Oil Quench
Double temper at 550°C
- B. Austenitize 930°C
Intermediate 'Ausbay' quench 520°C
Salt Quench 210°C
Double temper at 550°C
- C. Austenitize 930°C
Intermediate 'Ausbay' quench 520°C
Air cool
Double temper at 550°C
- D. Austenitize 930°C
Intermediate 'Ausbay' quench 520°C
Oil Quench
Double temper at 290°C.

2.3 Mechanical Properties

The mechanical properties, resulting from the above heat-treatments, are given in Table 1.

TABLE 1

Heat Treatment Schedule	Yield Strength 0.1% Proof MPa	Ultimate Tensile Strength MPa	Reduction in Area %	Elongation %	Fracture Toughness K_{Ic} MPa. $m^{1/2}$
A	1434-1454	1600-1620	47-51	14-17	98-105
B	1440-1460	1600-1630	48-51	14-17	78-86
C	1450-1470	1630-1650	44-48	13-15	42-48
D	1510-1518	1850-1910	32-35	11-12	54-60

Heat-treatments A, B and C gave essentially the same tensile strength properties but different toughness values, while heat-treatment D gave higher tensile properties and a medium fracture toughness level.

2.4 Fatigue Tests

2.4.1 Pre-cracking

Initial pre-cracks, about 3 mm long, were introduced into the compact-tension specimens prior to tests to determine toughness and fatigue crack-growth rates. The final extension of these pre-cracks was made with a cyclic $K_{\max}$ of $\sim 15 \text{ MPa.m}^{1/2}$ applied for 5 to 10×10^4 cycles.

2.4.2 Test Atmosphere

A constant environment was preserved by maintaining a flow of medically dry air ($\sim 0.5 \text{ mg/litre H}_2\text{O}$) through a Perspex box surrounding the test specimens.

2.4.3 Testing Procedures

With the exception of a few tests at amplitudes of stress intensity less than $15 \text{ MPa.m}^{1/2}$, all fatigue tests were carried out at approximately 3 Hz using a variable-speed, eccentric-driven, fatigue machine which applied sine wave tension-tension loading. Tests at values of $K_{\max}$ less than $15 \text{ MPa.m}^{1/2}$ were carried out in an Amsler 'Vibrophore' machine at approximately 100 Hz. The applied cyclic amplitudes of stress intensity, ΔK , were essentially equal to $K_{\max}$ as only a low $K_{\min}$ was used (simply to maintain specimen alignment). Little, if any, mean load effects were introduced as the stress ratio ($R = K_{\min}/K_{\max}$) was always much less than 0.1.

2.4.4 Crack-Growth Measurements

Fatigue-crack extension was monitored during tests by observing the polished specimen surface through a microscope with a $\times 20$ graduated eyepiece. Crack-growth rates determined in this manner were later checked against other values determined fractographically. These quantitative fractographic measurements were facilitated by producing periodic "markers" on the fracture surface by changing the load level. Markers produced by decreasing the load level, by as much as 50 percent, and increasing the cyclic frequency from 3 to 20 Hz for 10,000 cycles, produced distinct bands (up to 0.3 mm wide) on the fracture surface. When the load levels were increased, by more than 20 percent, characteristic 'tide' markings were apparent on the fracture surfaces. Measurements of these progression marking were made using a profile-projecting, tool-makers microscope, or from enlarged photographs of the fracture surfaces.

2.5 Fractography by Electron Microscopy

In addition to the optical fractography, detailed scanning and transmission electron microscopy of the various fracture surfaces was undertaken. Two-stage carbon replication was used for the transmission work. Fracture surfaces corresponding to crack growth at particular stress-intensity levels were selected for examination so as to obtain evidence of changes in the mechanism of fatigue crack growth.

3. EXPERIMENTAL OBSERVATIONS

3.1 Rate of Growth of Fatigue Cracks

Crack-growth rates, determined from at least three specimens from each heat-treatment batch, are shown in Figure 1. The characteristic sigmoidal form of the $\log da/dN$ versus $\log \Delta K$ curve is exhibited for each toughness level or toughness/strength condition.

A notable feature of the curves is that the rate of fatigue-crack growth in the linear region is not significantly influenced by toughness or strength level. The growth rate accelerates as the maximum cyclic stress intensity increases beyond about 0.7 times the respective K_{IC} values; as stress intensity amplitudes decrease below $18 \text{ MPa.m}^{1/2}$, the curves diverge and the proportionality between $\log da/dN$ and $\log \Delta K$ breaks down. At growth rates less than $5 \times 10^{-5} \text{ mm per cycle}$, a threshold stress-intensity amplitude is approached. Fatigue-cycling was not carried out at growth rates of less than $10^{-5} \text{ mm per cycle}$ but extrapolation suggests that the microstructural

changes which result in differences in toughness, and/or strength, also influence the value of the threshold stress intensity.

3.2 Fractographic Examination

Detailed fractographic examinations were made of fracture surfaces of specimens given heat-treatments A (high toughness), C (low toughness but the same strength properties as A) and D (intermediate toughness and higher strength). Specimens given heat-treatment B (i.e. different in toughness but similar in strength to A and C) did not show sufficient differences in fracture topography from those exhibited for heat-treatment A to warrant the same degree of attention.

Inspections of fracture surfaces were made within zones of the following crack growth rates:

- (a) $da/dN < 2 \times 10^{-5}$ mm/cycle; $K_{max}/K_{1C} \sim 0.15$
- (b) da/dN of 7 to 9×10^{-5} mm/cycle; $K_{max}/K_{1C} \sim 0.25$
- (c) da/dN of 2 to 3×10^{-4} mm/cycle; $K_{max}/K_{1C} \sim 0.3 - 0.4$
- (d) da/dN of 5 to 8×10^{-4} mm/cycle; $K_{max}/K_{1C} \sim 0.5 - 0.6$
- (e) da/dN of 2 to 4×10^{-3} mm/cycle; $K_{max}/K_{1C} \sim 0.7 - 0.8$
- (f) Region of unstable over-load fracture; $K_{max} > K_{1C}$

Typical scanning and transmission electron micrographs of the fracture surfaces at the above rates of crack propagation are shown with the relevant $da/dN/\Delta K$ curve for heat-treatments A, C and D in Figures 2, 3 and 4, respectively.

3.2.1 High-Toughness Steel (Heat Treatment A)

In zone (a) of the high-toughness steel ($K_{1C} \sim 100 \text{ MPa.m}^{1/2}$), where a threshold ΔK is approached, a highly reflective fracture surface was obtained. Scanning microscopy, Figure 2, revealed shallow 'feathery' features similar in appearance to the underlying martensitic structure of the steel. Transmission electron microscopy failed to provide any evidence of fatigue 'striation' markings. The apparent relationship between the fracture surface topography and the underlying microstructure could not be substantiated even with replicas taken from lightly etched fractures.

In zone (b), at the commencement of the linear region of the $\log da/dN - \log \Delta K$ curve, the feathery appearance persisted but the surface was more rumpled with patches of shallow secondary cracking clustered in rows across the direction of crack growth (Fig. 2).

In zone (c), in the centre of the linear region, a 'lumpy' fracture surface was obtained and further indications of parallel rows of shallow secondary cracks, transverse to the growth direction, were noted. Transmission electron microscopy of replicas from this zone revealed clear evidence of striation markings whose spacing approximated to the growth rate of $2-3 \times 10^{-4}$ mm per cycle.

In zone (d), the 'lumpy' fracture surface persisted and the secondary cracks became more widely spaced. Replica studies again revealed the presence of the characteristic fatigue striations. In D6AC martensitic steel, the fatigue striations are discontinuously distributed and are not as clearly defined as those in aluminium alloys.

For zone (e), where the growth rate accelerates as the cyclic stress-intensity amplitude approaches K_{1C} , the fracture mode changes, giving rise to the appearance of ductile 'dimples', indicative of void formation and coalescence. These 'dimples' were especially obvious in the transmission electron microscopy of carbon replicas of the fracture surface, Figure 2. Locally, these features were similar to the markings in the unstable, or overload, zone (f) of the fracture surface.

3.2.2 Low-Toughness Steel (Heat Treatment C)

In zone (a), the fracture surface of the low-toughness material ($K_{1C} = 46 \text{ MPa.m}^{1/2}$), was indistinguishable from that of the high-toughness material for the same zone, (see Fig. 3). In

the linear region, where da/dN ranged from 1 to 2×10^{-4} mm per cycle, striations were prevalent. Above $K_{max} = 30 \text{ MPa}\cdot\text{m}^{1/2}$, i.e. for $K_{max}/K_{IC} > 0.66$, the striations were supplanted by regions having the appearance of cleavage fracture. The relative proportion of this 'cleavage-like' fracture increased as K_{IC} was approached. The overload region exhibited a totally 'cleavage-like' fracture surface, characteristic of this low-toughness D6AC steel¹³.

3.2.3 High Strength, Medium Toughness Steel (Heat Treatment D)

Examination of the fracture surfaces of high-strength, medium-toughness (UTS = 1900 MPa, $K_{IC} = 56 \text{ MPa}\cdot\text{m}^{1/2}$) steel again revealed a highly reflective fracture surface in zone (a), i.e. for slow crack-propagation (see Fig. 4). Features on the fracture surface, both in scanning electron micrographs and at magnifications up to $\times 2500$ in carbon replicas, appeared to be associated with the 'feathery' martensitic structure. However, examination of replicas from etched fracture surfaces at high magnification did not show this relationship.

In the linear zone (c), the 'striation' mechanism predominated and as the K_{max}/K_{IC} ratio exceeded 0.7, evidence of ductile dimpling or void formation increased, interspersed with patches of striation markings. The fracture surface in the overload region consisted entirely of ductile dimples varying from 0.2 to $5.0 \mu\text{m}$ in diameter.

4. DISCUSSION

The rates of fatigue-crack propagation in D6AC steel, heat-treated to different strength and toughness levels, are almost identical for amplitudes of stress intensity between 18 MPa^{1/2} and 70 to 80 percent of the plane-strain fracture toughness. Within this range, the rate of fatigue-crack growth per cycle, da/dN , is proportional to $\Delta K^{2.3 \text{ to } 2.5}$. This observation agrees with similar work reported by Feddersen *et al.*¹⁵ for D6AC steel with $\sigma_y = 1455$ to 1500 MPa, UTS = 1600–1650 MPa, and K_{IC} varying between 55 and 105 MPa^{1/2}. They stated that the "toughness level, K_{IC} , is not a variable in fatigue-crack propagation provided K_{max} is substantially (as much as 20 ksi.in^{1/2} [22 MPa^{1/2}]) below K_{IC} for the material." These results, therefore, do not support theories^{16,17} that the rate of fatigue-crack growth is inversely proportional to fracture toughness, K_{IC} .

The increase in yield strength from 1440 to 1510 MPa (4.7 percent) between heat-treatment A and heat-treatment D material is, in the present work, not sufficiently large to enable any conclusions to be drawn regarding the influence of yield strength upon rates of fatigue-crack growth in martensitic steels. However, more comprehensive studies by Richards and Lindley¹², Hahn *et al.*¹⁸, Clark and Bates¹⁹ and Barsom *et al.*²⁰ all indicate that substantial changes in yield strength are not reflected by corresponding variations in rates of fatigue cracking.

Analysis of the fracture surfaces reveals that the crack-growth mechanism which produces striation markings persists over the region where $da/dN \propto \Delta K^m$ and growth is virtually independent of K_{IC} . Moreover, this mechanism is apparently insensitive to the microstructural changes accompanying variations in K_{IC} .¹³ Below about 18 MPa^{1/2}, a fatigue threshold is approached and a change in the appearance of the fracture surface is seen. Crack growth in this region has been described by Cooke *et al.*²¹ as structure-sensitive. However, no direct relationship between the surface topography and the underlying microstructure could be found in the present study.

Above 0.7 K_{IC} , tensile fracture modes (void formation and coalescence or micro-cleavage) are associated with accelerated rates of fatigue-crack growth. Micro-void coalescence was observed in the steel of higher strength and medium toughness (1860–1910 MPa, 58 MPa^{1/2}) and in the steel of high-toughness (100 MPa^{1/2}), 1600–1650 MPa strength level. Facets indicative of micro-cleavage were found in the steel of low-toughness (46 MPa^{1/2}), 1600–1650 MPa strength level. Observations of this kind have also been reported by Donahue *et al.*¹¹ and Richards and Lindley¹².

When the volume of material ahead of a crack is exposed to cyclic-strain 'conditioning' (which for martensitic steels leads to cyclic 'softening' and local flow stresses very much less than the tensile yield values²²), it is clearly inappropriate to apply monotonic tensile properties to theoretical predictions for fatigue-crack growth. Furthermore, the expression devised by Donahue *et al.*¹¹, viz $da/dN = 4A/\pi\sigma_y E [K^2 - K_{th}]$, is applicable, as stated, to a wide range of

materials in which the constant A approximates to $2\sigma_y/E$. The dominant material parameter affecting crack growth is then Young's modulus (E), consistent with the observations reported by Hahn *et al.*¹⁸ and Barsom *et al.*⁶ Thus yield strength, like toughness, bears no clear correlation with fatigue-crack growth, at least not in the central region of the crack growth relationship, i.e. where striation formation and crack extension are not influenced by tensile cracking modes.

The practical significance of the present results and considerations is that, although the resistance to fatigue-crack propagation is unlikely to be improved significantly by optimizing mechanical properties, high toughness materials are to be preferred for service under fatigue conditions. This is not because of any inherent resistance to fatigue-cracking but rather because of the extended range of cyclic stress amplitudes over which da/dN is approximately proportional to ΔK^2 before tensile cracking processes (dependent upon K_{IC}) increase crack growth rates.

5. CONCLUSIONS

1. Fatigue testing, under constant amplitude load cycling, of D6AC steel, having various strength and toughness parameters, yields curves having a three-stage, sigmoidal shape when $\log da/dN$ is plotted against $\log \Delta K$.
2. In the central region of these curves, the rate of fatigue crack-growth da/dN is proportional to $\Delta K^{2.4}$ and independent of fracture toughness, for toughness values ranging from 46 to 100 MPa.m^{1/2} at essentially the same tensile strength.
3. Variations of yield and ultimate strength do not produce any significant change in the rate of fatigue-crack growth in the range where da/dN is proportional to $\Delta K^{2.4}$.
4. In the linear region of the curves, cracking proceeds by a mechanism which gives rise to striation marking upon the fracture surface.
5. At amplitudes above $0.7 K_{IC}$, tensile modes of cracking (void coalescence and micro-cleavage) cause progressively accelerated crack growth as K_{IC} is approached.
6. At amplitudes of stress intensity less than 18 MPa m^{1/2}, threshold values of K are approached. Rates of growth in this region are sensitive to changes in material properties and a different mechanism of growth is indicated by a change in fracture surface topography.

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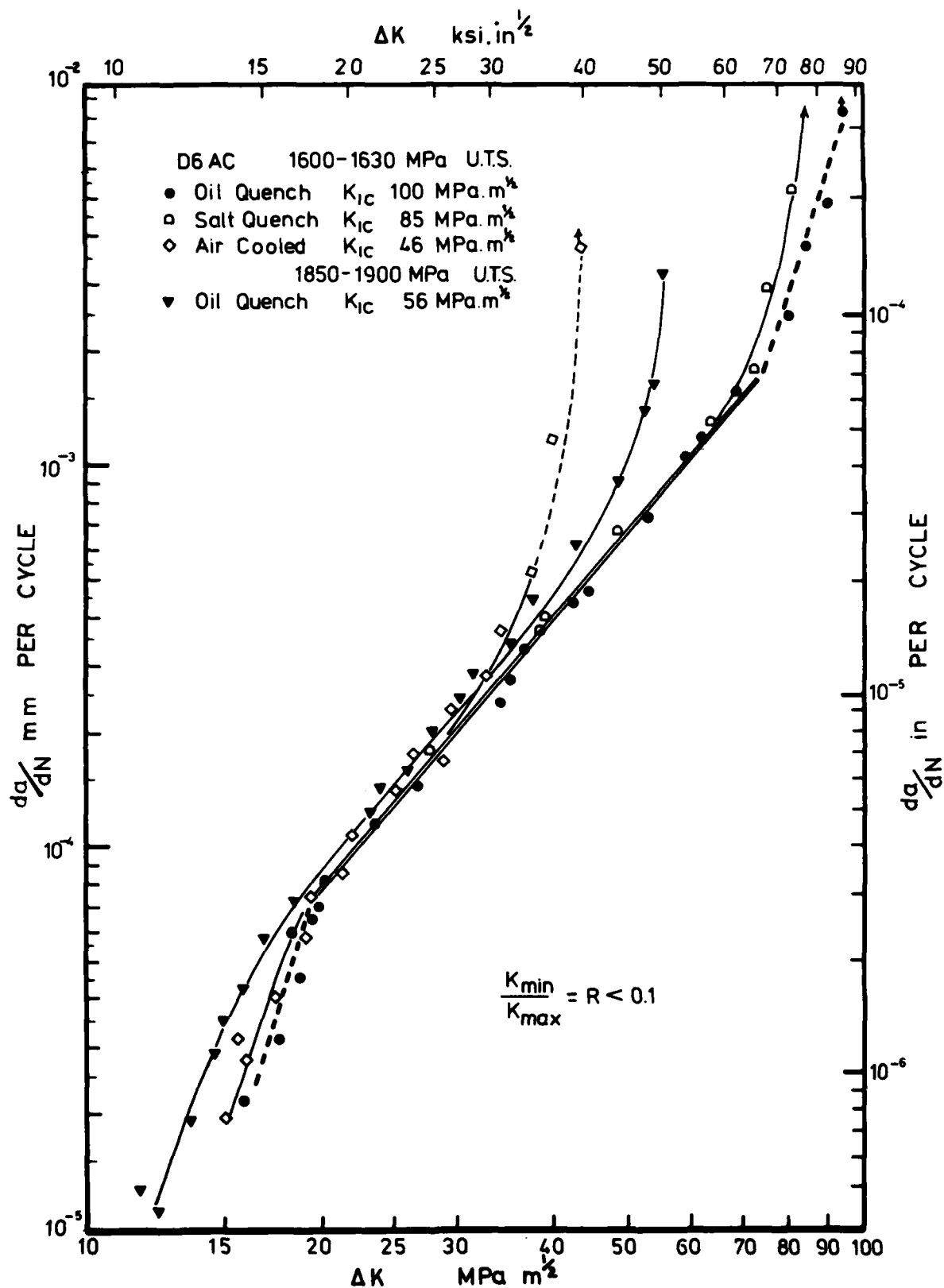


Fig. 1. Rates of fatigue – crack growth, da/dN , in D6AC steel, heat treated to different strength and toughness (K_{1c}) values, as a function of stress – intensity amplitudes, ΔK .

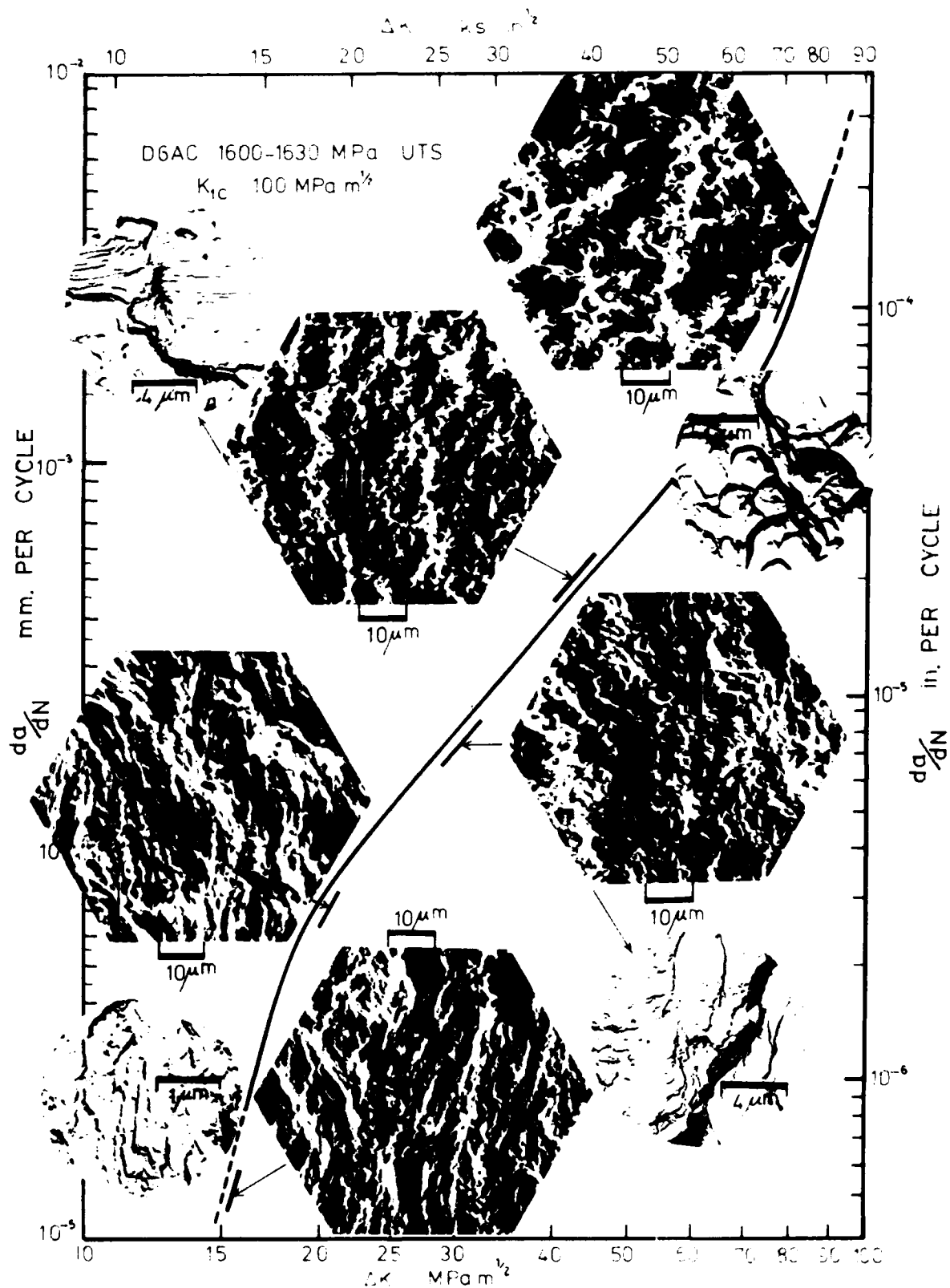


Fig. 2. Fractographic features exhibited on the fatigue-fracture surface of D6AC steel heat treated to U.T.S. = 1600 – 1630 MPa, and K_{1c} = 100 $\text{MPa m}^{1/2}$ (heat treatment A). Hexagonal photographs are SEM fractographs and circular photographs are TEM fractographs using germanium – shadowed carbon replicas.

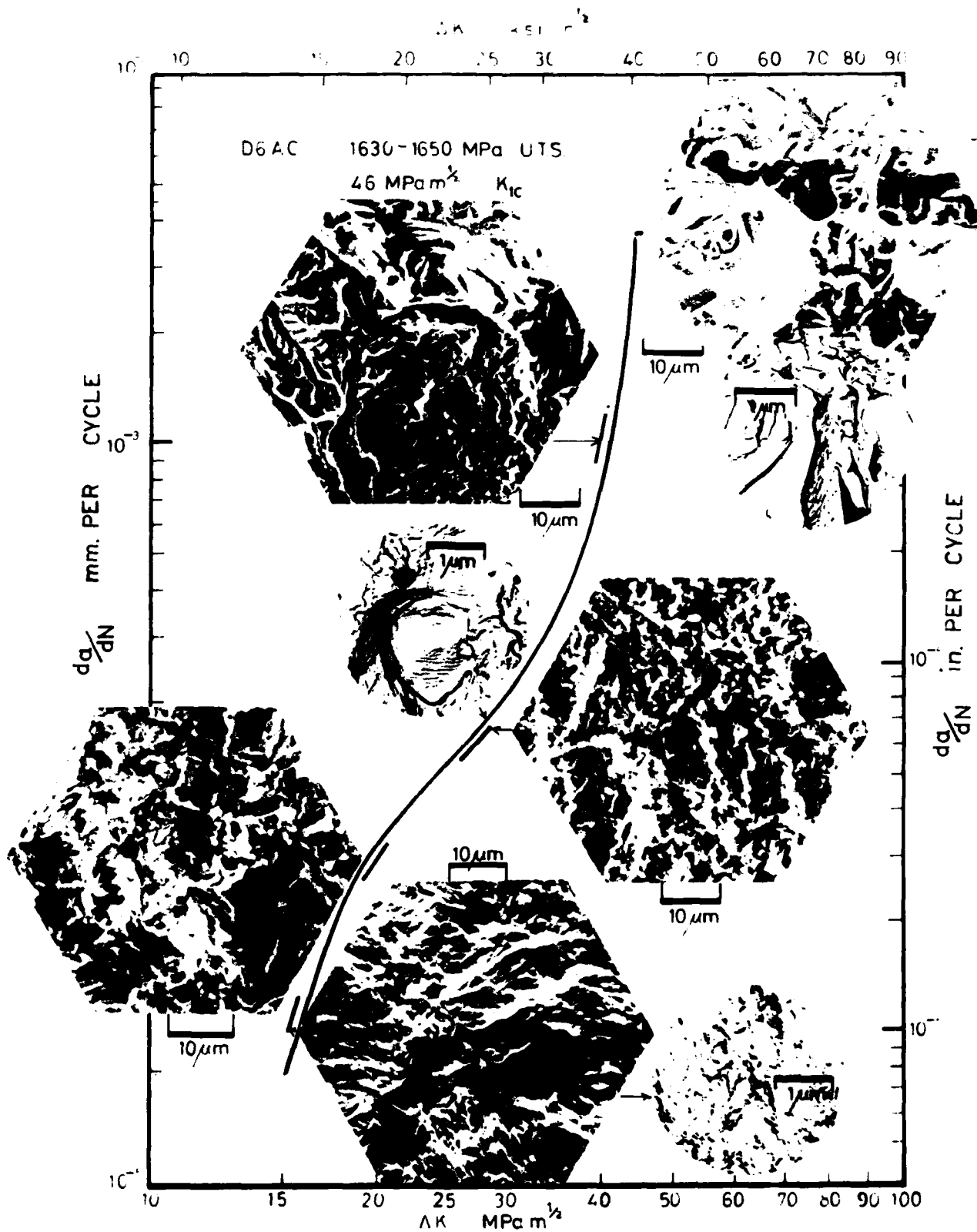


Fig. 3. Fractographic features exhibited on the fatigue-fracture surface of D6AC steel, heat treated to U.T.S. 1630 – 1650 MPa and $K_{1c} = 46 \text{ MPa m}^{1/2}$ (heat treatment C)
 Hexagonal photographs are SEM fractographs and circular photographs are TEM fractographs using germanium – shadowed carbon replicas.

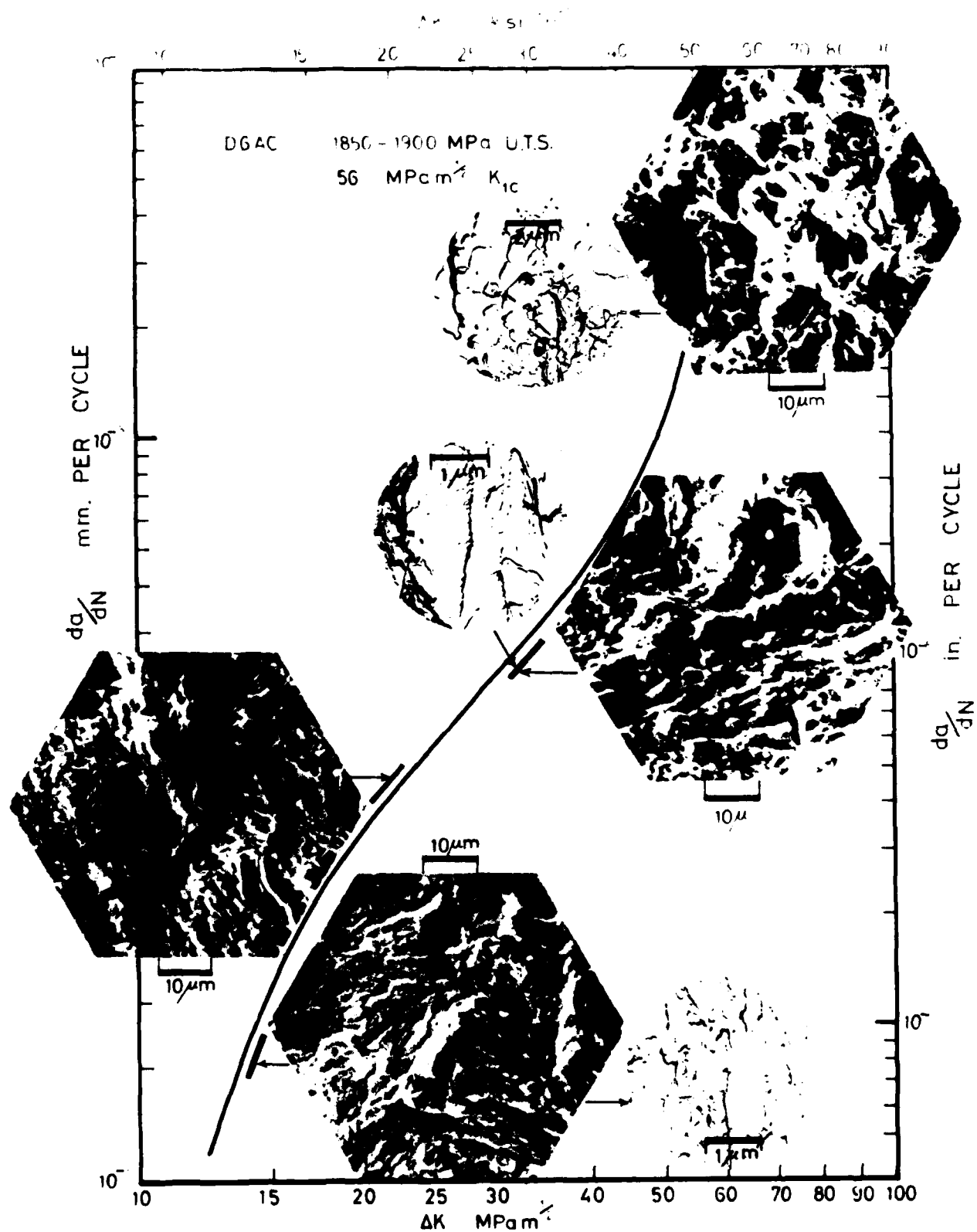


Fig. 4. Fractographic features exhibited on the fatigue-fracture surface of D6AC steel, heat treated to U.T.S. = 1850 - 1900 MPa, K_{1c} = 56 MPa m^{1/2} (heat treatment D).

Hexagonal photographs are SEM fractographs; circular photographs are TEM fractographs using germanium - shadowed carbon replicas.

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