QUANTITATIVE FRACTOGRAPHY OF FATIGUE-TESTED INCONEL IN 713 LC AND NIMONIC 86

U.W. Hildebrandt*

On cast high temperature alloy Inconel IN 713 LC and forged Nimonic 86 mixed fatigue tests (LCF and HCF) were carried out. The influence of HCF/LCF predamage on the fatigue fracture behaviour, particulary the distance between fatigue striations has been determined. There are correlations between the location of crack origins and fatigue parameters for cast material. Predamage of forged material Nimonic 86 increase the width of striations in a characteristic way. The degree of predamage can be calculated by observation and measurement of the distance between striations.

INTRODUCTION

In failure analysis there are qualitative informations concerning crack initiation and propagation. The SEM investigation of the fracture surface enables to determine distinct crack initiation, different stages of fatigue, striations and residual fracture. But often it is useful to know also quantitative properties, i.e. defined areas of fracture, position of crack origin relative to the surface of a component or specimen, width of striations and others. With this knowledge an estimation of stresses, temperatures crack propagation velocities is possible and a more detailed failure mode can be derived (1-3). In this work the influence of LCF respectively HCF-predamage of HCF (LCF) microfracture quantities was studied (for two different materials, the cast alloy Inconel 713 LC and a forged material Nimonic 86).

* Brown, Boveri & Cie AG, Mannheim
Central Materials Laboratory

EXPERIMENTAL

Table 1 shows the chemical analysis of Inconel 713 LC and Nimonic 86 samples. These specimens were LCF (HCF) predamaged and fractured by HCF (LCF). Tables 2 and 3 show the test parameters for the two materials. The encircled values represent the degree of predamage for the applied load. There are three kinds of specimens for each alloy: the LCF-predamaged (33-75% N/N_j) HCF-samples, the HCF predamaged LCF samples (26-75% N/N_f) and HCF-specimens tested at different mean stresses.

RESULTS

HCF-fracture, cast alloy Inconel 713 LC.

The microfractographical characteristics of HCF-specimens without any predamage are given schematically in figure 1. After the crack has initiated it propagates concentric till the surface is reached. This primary circle has transcrystalline characteristics, striations are not observed.

When the crack reaches the surface of the specimen, propagation will continue in a multiaxial mode with fracture paths till residual fracture occurs.

The radius of the primary circle, i.e. the distance r (crack origin - sample surface) depends on given stresses σ_a. The higher the stress the smaller the circle. Figure 2 shows the fracture surface for three different stresses.

In all cases cracks are initiated on a pore or pore agglomerations as indicated in figure 3. The above correlation is only valid, if pores are statistically scattered throughout the specimen. This is common to all cast alloys.

For a defined LCF predamage the fracture behaviour changes. Depending of the degree of predamage (same HCF-stress) the origin of the crack is shifted towards the specimen surface. In the case of only LCF-load the crack always starts at the specimen surface. Figure 4 shows 8 samples where this effect is true. For HCF only the crack is initiated near the specimen center, when the stress is lower than $\sigma_{0,2}/5$ ($\sigma_{0,2}$= 570 MPa). For $\sigma_{0,2}/3$ the crack origin is just below the sample surface (= 0,5 mm). A 33% LCF-predamage specimen (table 2) shows an increase in shift of the crack origin. Even for relatively small σ_a = 120 MPa this crack origin is located just below the surface (figure 5). But by applying stresses of the order of the yield stress (i.E. LCF-damage) the crack initiates on the surface. This is in agreement with other observations for LCF-fracture behaviour.

The same is true for integrated fatigue fracture areas F_F (specimen cross section - residual fracture region), see figure 6. For HCF the relative fatigue area F_F/F decreases from 50% at 120 MPa to 20% at 180 MPa. The LCF-predamage decreases the area considerably. At 120 MPa the area of predamage is reduced from 50% (HCF) to 45% (33% LCF-predamage) and to 30% (75% predamage).

LCF-fracture, cast alloy Inconel 713 LC.

Several samples were HCF-predamaged and LCF-tested to fracture. Depending on the degree of predamage the crack origins are influenced in a typical way. For a low degree of predamage (≈ 20%) there are no HCF incipient cracks. Cracking starts from the edge of the fracture surface at several points. This is to be expected (see figure 7). For a higher degree of HCF-predamage cracking starts from HCF incipient cracks which are localized at pores (figure 8). Additionally a tendency for splitting and an increase in width of striations with increasing predamage has been observed (figure 9). Because of an extensive degree of surface oxidation no further quantitative measurements were possible.

HCF-fracture, forged alloy Nimonic 86.

Several samples (table 3) were HCF-predamaged and LCF-tested respectively LCF-predamaged and HCF-tested. HCF-tests were performed in push-pull mode (R = -1).

For HCF-testing without predamage the specimens exhibit a crack origin at the edge of the fracture surface. This can be expected because there are no pores in the forged material. It was not possible to observe an influence of stress on the position of the crack.

Three specimens were predamaged by LCF with different strain amplitudes. In all cases the cracks initiate at the sample surface exhibiting radiating fracture paths. A characteristic striation structure was observed and will be discussed later. A typical micrograph is shown in figure 10.

LCF-fracture, forged alloy Nimonic 86.

Figure 11 shows a typical view of the fracture region. The crack is initiated at the sample surface. The width of striations increases with increase in distance from the crack origin.

DISCUSSION

For all Nimonic 86 specimens the width of striations Δa, as a function of their distance from crack origin and degree of predamage were determined.

Apparently high HCF-stresses, high strain amplitudes and a high degree of relative exhaustion are effective in the same direction: i.e. cause an increase in Δa. This is shown in figure 12.

The parameter δ is defined as:

$$\delta = (\text{strain amplitude } \varepsilon_a) \times (\text{relative cycle number } N/N_F) \times (\text{HCF mean stress } \sigma_a)$$

The dimension of δ is the same as for a specific energy which is equivalent to the damage energy brought into the specimen volume. With this parameter all the measured values of Δa fall on straight lines shown in figure 12.

Depending on the relative distance from crack origin d/D the striation width increases with increasing damage parameter δ. This is valid for HCF-fracture as well as for LCF-fracture. In the latter case Δa is about a factor 2 higher than for HCF. Further on it can be seen, that only based on measurement of Δa it cannot be decided whether there is HCF or LCF-fracture without precise determination of d/D and degree of predamage.

APPROACH TO FAILURE ANALYSIS

Based on figure 12 it is possible to estimate the degree of predamage by measuring the width of striations as a function of d/D. In general reasonable assumptions concerning the service parameters can be made. Together with the morphology of the fracture surface it is possible to decide whether there is HCF- or LCF-fracture. This shows that a set of parameterisized curves can be determined by measuring Δa. By measuring or by reasonable estimation of two of the factors of δ, i.e. σ_a and ε_a, the degree of HCF-predamage can be calculated. With a better knowledge about the failure mode it is possible to improve and simplify failure analysis.

SYMBOLS USED

$\sigma_{0,2}$ = yield stress (MPa)
F_F = fatigue fracture area (mm^2)
F = specimen cross section = 78,54 mm^2
Δa = striation width (μm)
d = distance between position of crack origin and striation
D = specimen diameter = 10 mm
δ = damage parameter (MPa)
ε_a = strain amplitude (%)
N = cycle number
N_i = cycles to first crack
N_f = cycles to fracture
$\sigma_a{}^*$ = ½ ($\sigma_+ + \sigma_-$) = stress amplitude (MPa)
σ_m = mean stress (MPa)
r = radius of primary circle

REFERENCES

(1) Dammer, A., Steffens, H.D. and Wielage, B., Proc. of 11th session "AK Rastermikroskopie" DVM, 7th, 8th June, pp. 251.

(2) Kolednik, O. and Stüwe, H.P., Z. Metallkunde Bd 73 ,1982, H. 4, pp. 219.

(3) El-Sondani, S.M., Metallography 7, 1974, pp. 271-311.

Table 1 - Chemical analysis of investigated materials.

Element	IN 713 LC	Nim 86
C	0.07	0.03
Si	< 0.10	0.40
Mn	< 0.10	0.06
Cr	11.7	24.7
Ni	bal	bal
Fe	0.02	1.49
Ti	0.74	0.02
Al	5.9	< 0.05
Mo	4.4	10.75
Nb + Ta	2.1	-
Zr	0.09	0.03
B	0.009	n.d.

table 2: test parameters for the investigated materials
Inconel 713 LC

fracture	spec. No.	LCF ($T = 850^\circ C$, $\dot{\varepsilon} = 6$ % min^{-1})					HCF ($T = 850^\circ C$, 50 Hz, $R = 0$)			
		$2\varepsilon_a$ (%)	N	N/N_i (%)	σ_a^* (MPa)	N_i	N	N_f	N/N_f %	$\sigma_a = \sigma_m$ (MPa)
HCF	628							$2x10^6$	100	135
	627							$1x10^7$	100	135
	689							$2,47x10^7$	100	121
	636							$2,0\ x10^5$	100	185
HCF	686	0,31	7500	(75)	201			$3,32x10^6$	17	121
	681	0,31	3300	(33)	196			$3,01x10^7$	140	121
	692	0,56	333	(33)	284			$8,75x10^6$	26	121
	693	0,56	333	(33)	287			$3,4\ x10^4$	5	152
LCF	632	0,35		24		2345			(59)	121
	633	0,35		33		3346			(26)	121
	665	0,82		105		342			(75)	152
	697	0,81		70		219			(33)	152

table 3: test parameters for the investigated materials
Nimonic 86

fracture	spec. No.	LCF ($T = 850^{\circ}C$, $\dot{\varepsilon} = 6\ \%\ min^{-1}$)					HCF ($T = 850^{\circ}C$, 50 Hz, $R = -1$)			
		$2\varepsilon_a$ (%)	N	N/N_i (%)	σ_a^* (MPa)	N_i	N	N_f	N/N_f %	$\sigma_a = \sigma_m$ (MPa)
HCF	598								100	140
	591								100	200
HCF	1002	0,29	3300	33	194			$3,2 \times 10^5$	6	175
	1005	0,40	1600	50	204			$4,68 \times 10^5$	8	175
	1006	0,40	1600	50	210			$2,65 \times 10^7$	100	140
LCF	996	0,40		78	236	2506	$2,9 \times 10^5$		50	175
	997	0,29		109	210	10920	$1,91 \times 10^5$		33	175
	1000	0,40		64	215	2048	$9,0 \times 10^6$		50	140
	596	0,29		75	181	7474	$5,94 \times 10^6$		33	140

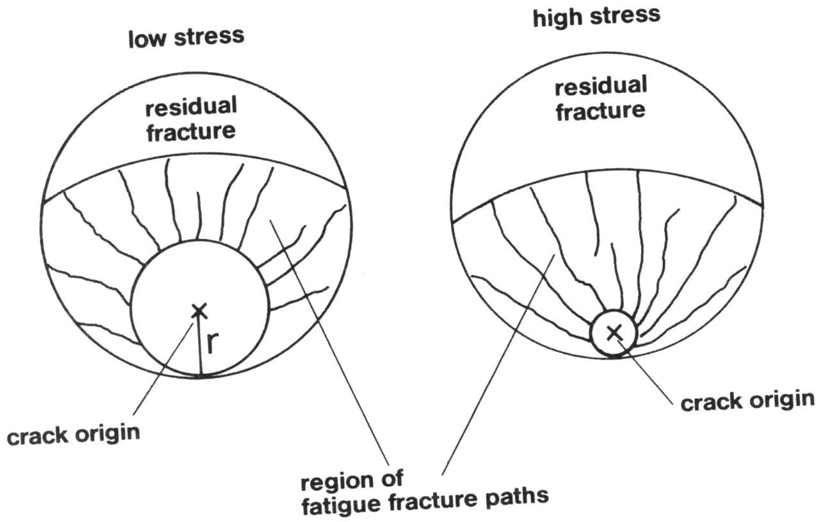

Figure 1 Macroscopic regions of HCF fracture surface of IN 713 LC
low stress high stress

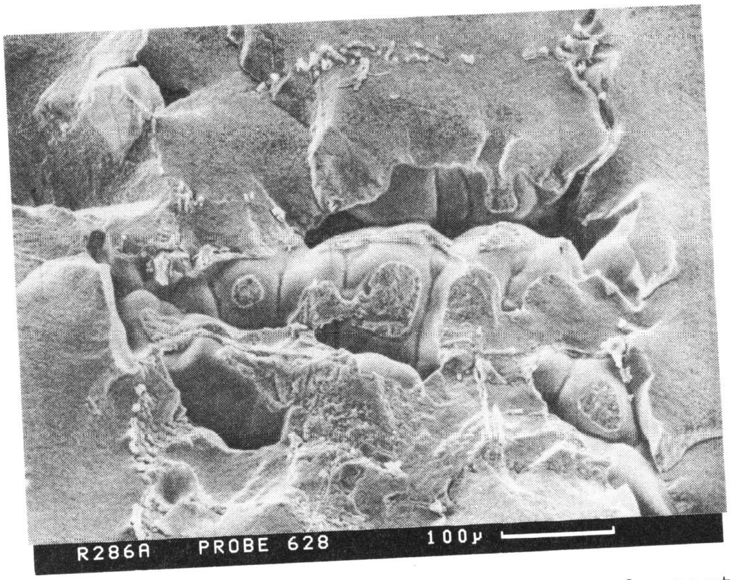

Figure 3 pore as crack origin in the circle center (figure 1) IN 713 LC

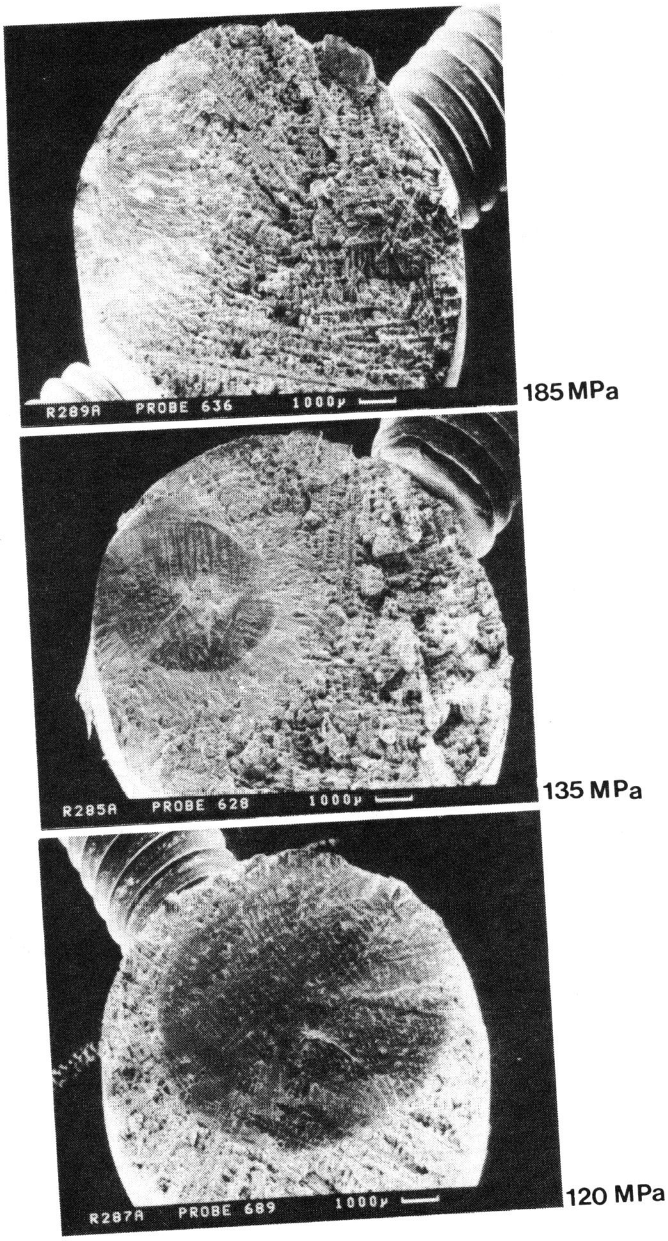

Figure 2 HCF-fracture surface with different radii of primary circle

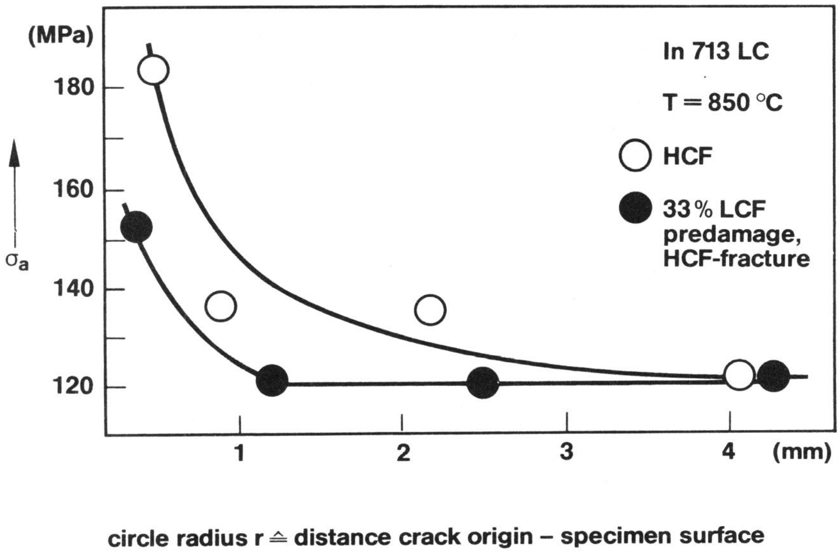

Figure 4 Influence of LCF-predamage and stress amplitude for HCF-load on position of crack origin. Sample diameter: 10 mm

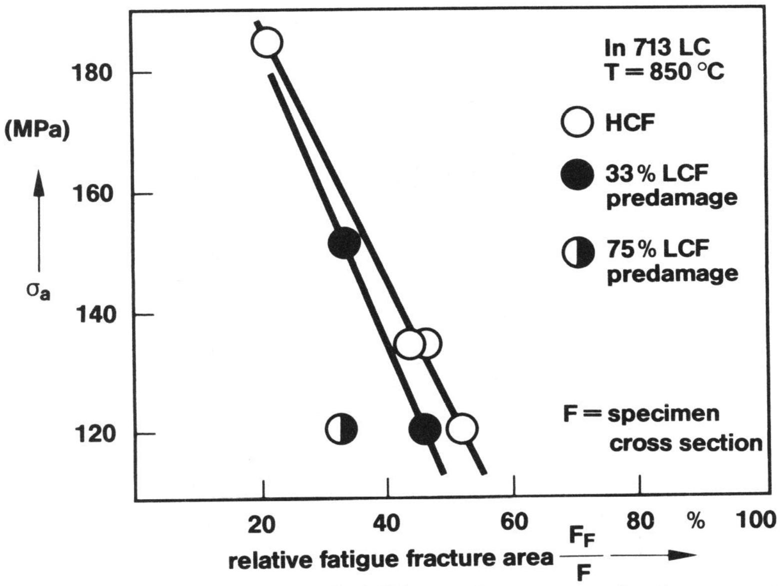

Figure 6 Influence of LCF-predamage and stress amplitude on the relative HCF-fracture area of IN 713 LC

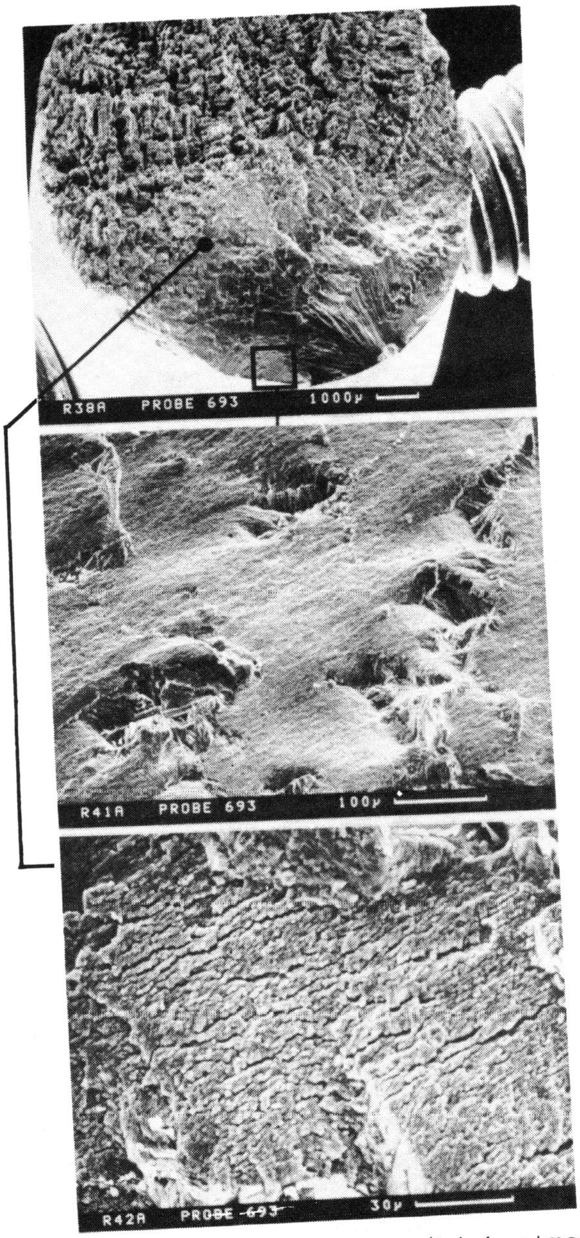

Figure 5 IN 713 LC, HCF fracture (high stress)
LCF-predamage 33 %, striations

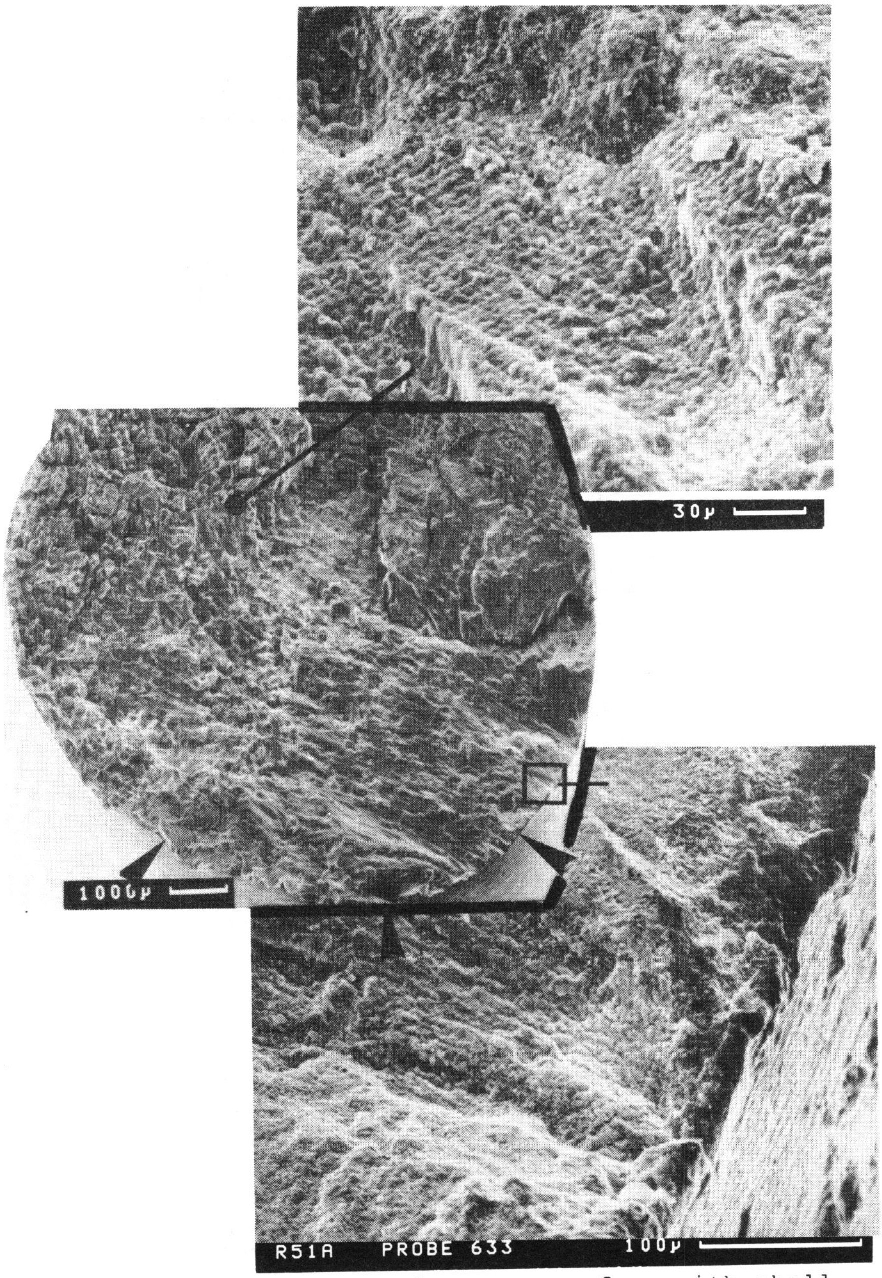

Figure 7 IN 713 LC, LCF fracture surface with shell shaped initial cracks, 20 % HCF-predamage, several crack origins

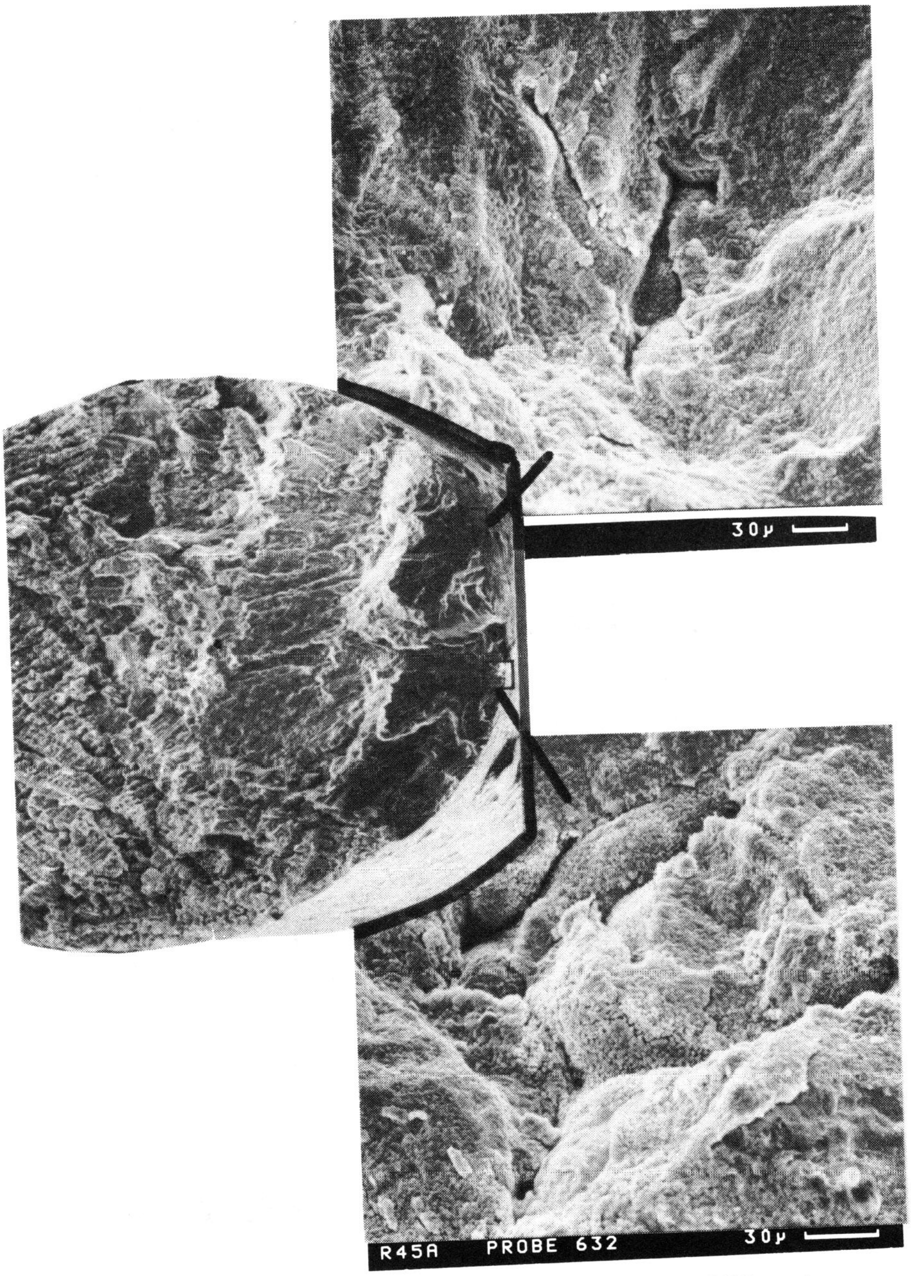

Figure 8 IN 713 LC LCF-fracture surface, HCF pre-damage 59 %,HCF incipient cracks on pores

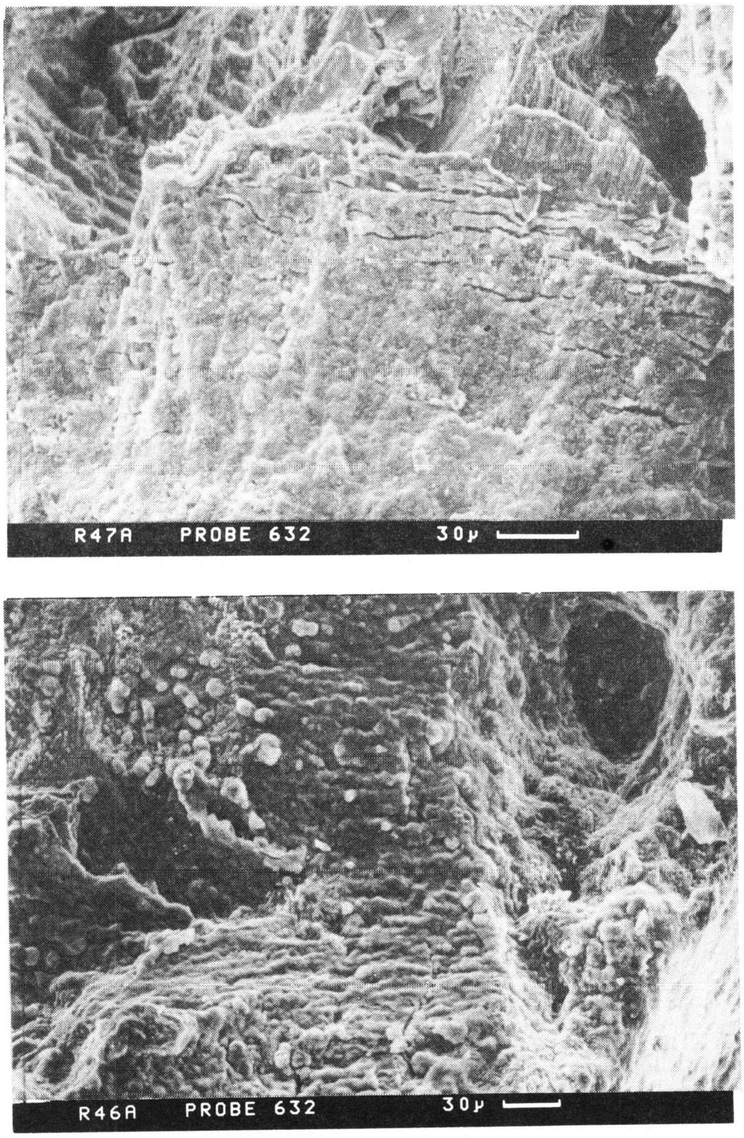

Figure 9 IN 713 LC LCF-fracture with 59 % HCF pre-damage LCF-striations

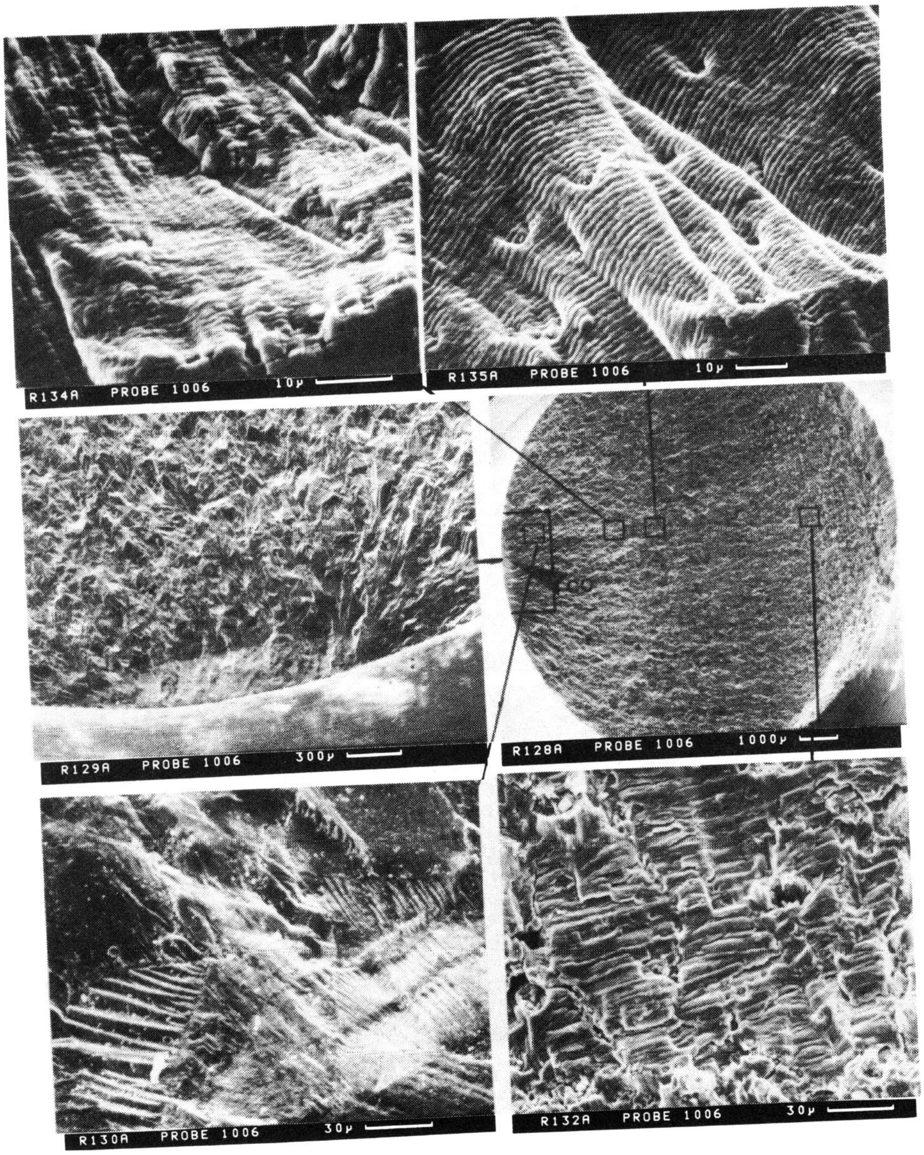

Figure 10 Nim 86, HCF-fracture, crack origin (CO) on the edge, LCF-predamage: 50 %, striations

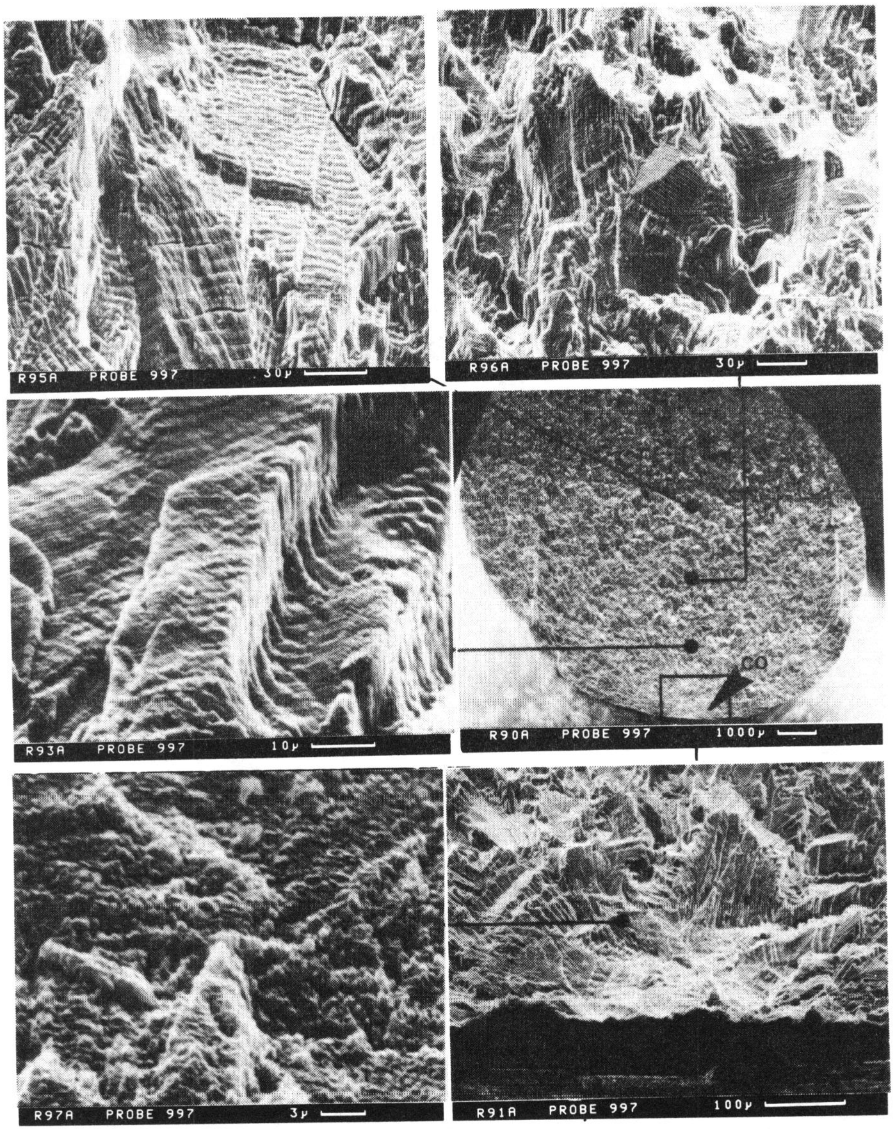

Figure 11 Nim 86, HCF-fracture, crack origin on the edge
HCF-predamage 33%, splitted striations, subsidiary cracks

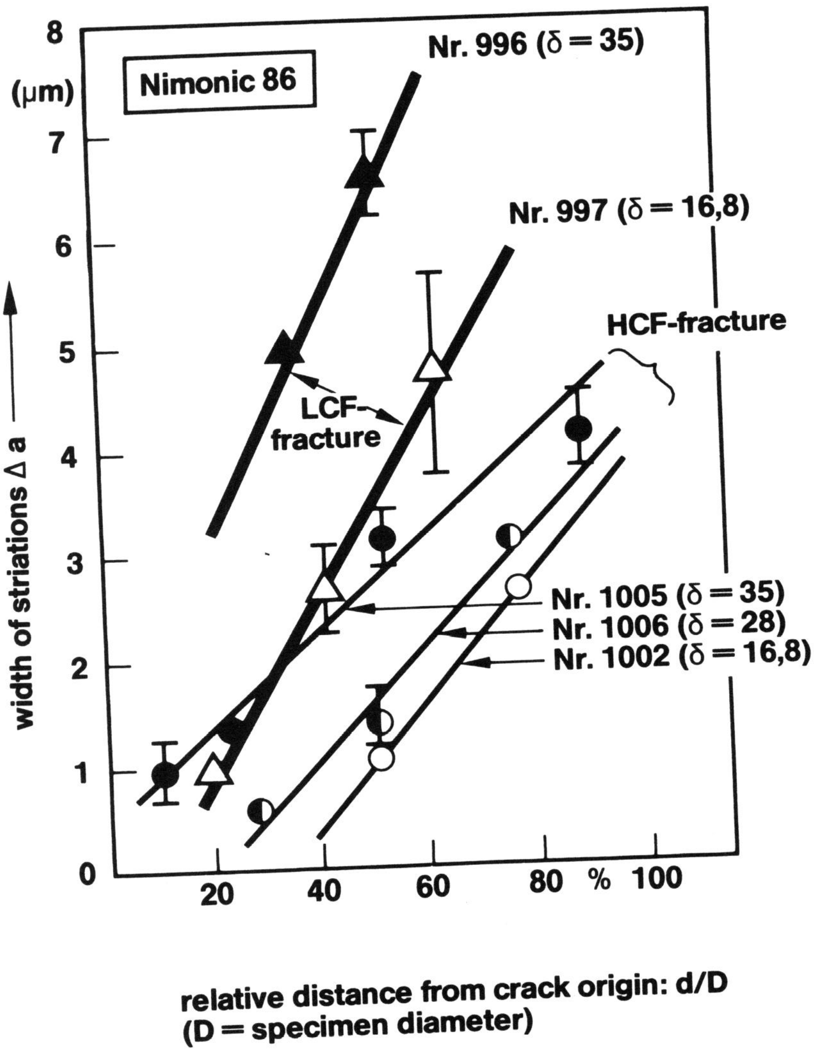

Figure 12 Width of striations as a function of d/D for different loads (δ = predamage)