THIRTEENTH EUROPEAN ROTORCRAFT FORUM

II.4
Paper No. 39

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Cascina Costa, ITALY

September 8-11, 1987

ARLES, FRANCE

ASSOCIATION AERONAUTIQUE ASTRONAUTIQUE DE FRANCE

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ABSTRACT

The fatigue behaviour of composites is not classical and not yet well known; today a lot of works are in progress with the aim of understanding fatigue phenomenon in composite materials. Moreover, it is not easy to determine the behaviour of defects in complex structure with an applied dynamic load. The problem, we are dealing with, is faced in two different ways, but strictly connected each other:

- mechanical modelling, developed with flat specimens with different lay-up sequences
- very accurate non destructive testing methods, in order to have a defective situation as precise as possible of structures under testing.

1. SHORTCOMINGS OF A CLASSICAL APPROACH ON COMPOSITE MATERIALS FATIGUE

In order to apply the "classic" or "metal" approach to characterize the composite fatigue behaviour, the composite materials must be considered as homogeneous, i.e. not formed by different plies. Consequently, with this approach, it is impossible to analyze the different damage mechanisms which appear in multi-layer laminates, occurring at the interfaces between adjacent laminae with differently oriented reinforcements.

Experiments carried out with this classical approach show important differences between these two kinds of materials.
First of all, for composite materials, it is not always possible to define the fatigue stress limit; fig. 1 shows the S-N curve for glass fibre/epoxy prepregs (1). It's evident the absence of an asymptotic behaviour of the curve at high numbers of load cycles: the monotonic decreasing shape of Wohler curve denotes a continuous loss of fatigue strength. This fact is generally true for various glass fibre composite materials (2).

For composite materials it's also impossible to separate experimentally "nucleation range" from "propagation range" of defects as we do for metallic alloys. The diagram of fig. 2 shows the absence of an abrupt transition between a damage initiation phase and a damage propagation phase (2); it is also impossible to define the damage onset.

Consequently, the difference between "safe life" and "fail safe" design criteria is not so clear for composites as for metals.

Another macroscopic aspect that points out the need of a critical examination of the classical approach for composite material is shown in fig. 3; in fact, while for light alloys, at fixed cyclic stress, the range between first damage and failure is of 1 or 2 orders of magnitude in terms of numbers of cycles, for composite materials the range is of 4 or 5 orders of magnitude (3), (4). Furthermore, fig. 3 represents the developement and growing of several modes of damage and not of a single damage as it is for metals.

As a matter of fact, composites are non homogeneus materials because at least three phases can be distinguished in them. fibres, matrix and interface.

The fatigue experiments on composite specimens show the existance of two families of mechanical properties: the "matrix dominated" and the "fibre dominated". The first ones are evident in tension-tension tests on 90° and ±45° specimens and in bending tests on unidirectional specimens; in these cases matrix itself plays a primary role in load carring. The second ones are evident when fibres are the primary load carrying phase (0° specimens).

Now, the polymeric matrices more used in aerospace composites are the epoxy ones (5). Since the epoxy matrix is very brittle, with a low strain at fracture, the reinforcing action of fibers is important; in particular high modulus fibres prevent matrix from high strain. So, under the same stress and with the same epoxy matrix, carbon fibres "protect" matrix better than glass fibers do, so as S glass fibres "protect" it better than E glass fibers do (see fig. 4, (2)).

According to the above statements, multi-layer laminates cannot be treated by a classical approach: in this "frame" delamination, which is one of the most significant damage in these materials, cannot be explained, such as the extension of the typical damage does not play a significant role. The developement of a "laminawise" approach to fatigue is necessary to get over these limits in order to take into account the roles of different damage modes present in composite materials and the influence of the continuous degradation, produced by cyclic load. Moreover, a precise measurement of damage patterns is needed, in order to find the correlations between damage onset and growth and the mechanical properties decay.

2. DAMAGE MECHANICS OF MULTIDIRECTIONAL LAMINATES

The composite material fatigue behaviour is the result of the overlap and of the integration of several modes of damages, which affect the matrix till they transfer to the fibres such a stress as to cause their failure. In order to get more information about damage morphology and residual properties, as already done by other authors (6), (7), (8), multidirectional laminates with different lay-ups ($[\pm45/0/90]_S$, $[+45/-45/0]_S$ and $[0/90]_S$), are examinated. From a macroscopic point of view, these laminates show all the damages observable on real structures, in simple tension-tension fatigue tests: edge delamination, intralaminar matrix cracking and local delamination originated from matrix cracks (9),(10). Cyclic loading causes the onset and the spreading of different damages; the consequent effect on mechanical properties is stiffness decay (fig. 5).

In fatigue testing of these laminates, non destructive techniques for damage monitoring are used, in order to evaluate the mechanical properties decay versus the damage area. The non destructive techniques must be sensitive enough to determine accurately the ammount of each mode of damage, just to evaluate each single contribution to stiffness loss. A significative radiographic view of a specimen under test is shown in fig. 6.

Fatigue testing of multidirectional laminates shows matrix cracks, parallel to the fibres, arising in laminae orientated out of the direction of the load. The effect of these cracks is the decay of elastic properties (E_{22}, G_{12}) of the interested laminae, with the consequent decay of the stiffness of the laminate (11),(12),(13). Both analytical and numerical models are used to explain and predict the decay of the average modulus E_x versus crack density. The agreement between these data and the experimental ones is fairly good (fig. 7).

It may be observed that diffused matrix cracks tend to saturate, to reach a stable density value (13),(14) (fig. 5).

During fatigue testing, local delaminations, originated from matrix cracks (15), spread all over the specimen for all the lay-ups; only in the quasi isotropic specimen ($[\pm45/0/90]_S$ lay-up), we observe also the growth of edge delamination .

By other authors, it has been underlined and proved the necessity of a detailed analysis of the damage resistance properties of laminates with respect to the parameters typical of the fracture. The "strain energy release rate" G, for given laminate configuration and applied strain, is the parameter chosen for these studies (16),(17).

In O'Brien analytical model (18), which deals with stiffness loss and strain energy release rate for free edge delamination, stiffness decreases linearly as delamination size increases:

(1) $$E_x = E_o - \left(E_o - E^*\right)\frac{a}{b}$$

where E_0 is the laminate modulus before delamination, a/b is the ratio of delamination size to specimen width and E^* is the modulus of the laminate completely delaminated.
Using the definition

(2) $$G = -\frac{dU}{dA}$$

where U is the strain energy and A is the delamination area, we obtain (11):

(3) $$G = \frac{\varepsilon^2}{2} t \left(E_0 - E^*\right)$$

where ε is the applied strain and t is the specimen thickness. As this model doesn't consider suitably the interaction of delamination with the preesistent matrix cracks, it doesn't describe exactly the behaviour of glass fibre/epoxy laminates.

Another analytical model is proposed to predict the onset and the growth of matrix cracking: its fundamental assumption is that all the matrix cracks may be treated as a single equivalent flaw of area A. Substituting in (2), we obtain:

(4) $$G = -\frac{t}{w}\frac{\varepsilon^2}{2}\frac{dE}{dD}$$

where w is the length of the matrix crack and D is the crack density. From such results, G can be evaluated and correlated with the accumulation of matrix cracks.

Using the above assumptions, the static tests results on the progressive accumulation of matrix cracking have been analyzed to obtain the critical strain energy release rate corresponding to increasing crack density. The obtained plotting, which can be defined as "crack resistance curve" in accordance to fracture mechanics is shown in fig. 8 (9). The growing G for onset of new cracks is congruent with the observed cracks saturation. The presence of this saturation density value may be so explained: as the dE/dD term (fig. 7) is very close to zero when the crack density is near to the saturation value, the strain energy release rate will not be sufficient to generate any other crack, even with increased stress.

The application of this model to the onset and growth of matrix cracks in glass fibre laminates gives good results, showing a substantial overlap of the experimental points obtained with different lay-up laminates.

3. MATRIX CRACKING IN UNIDIRECTIONAL TENSION BANDS

In helicopter dynamical components, it is not rare the use of unidirectional tension bands to carry relatively high loads. These structures are generally high stiffness/high strength bands, designed to work only in tension, but, due to the variation of fibre tension through the thickness of the structure (strongly dependent on the curvature); significant shear stresses can be predicted in the matrix in the transition zone between two differently curved parts.

Moreover, from the non destructive inspection of tension bands we find that the most characteristic defect is "delamination" (or better, a thin void between two adjacent plies) located near the middle of the laminated ring (fig. 9).

Combining the above exposed situations, it seems that a matrix crack in the transition zone is the most likely mode of failure of such a structure.

To better investigate this problem, an unidirectional band, like it is represented in fig. 10, is analyzed with the finite element model (19). Applying to this example the considerations about composite fatigue behaviour above mentioned (par. 2), it is confirmed the dependence of G on the extension of the damaged area (see results of 19).

Some thick unidirectional specimens, whith and without delaminations (fig. 11), are tested statically by multiple application of a flexural load, inspected by xeroradiography after each load phase, and then loaded up to failure : even if the amount of collected data is not sufficient for a quantitative evaluation of damage growth vs. applied stress, these tests confirm (qualitatively) that: even if the presence of little defects "drives" the failure mode, the static strength is not highly affected by it. I.e., in presence of little defects, even multiple but on different planes, the load threshold for delamination onset is not sensitively reduced but delaminations arise in the planes of defects. On the contrary, the presence of bigger defects, even if in a lower number, reduces the load threshold and this reduction is roughly proportional to defective area.

In this optic, it is clear that it is really important to develop non destructive techniques that can allow not only the detection but also an accurate measurement of defects and damages both on components of a structure and on the assembled structure, as a whole.

4. NON DESTRUCTIVE INSPECTION OF SPECIMENS AND REAL STRUCTURES

The first goal of this job is the development of a powerful source of data in order to have a better understanding of damage growth phenomena during cyclic loading tests performed on real components (or very similar specimens), just like it has been done on flat specimens, and to develop an "in service inspection" technique (obviously with some differences due to the different "boundary" conditions of the "in laboratory" and of the "on field" inspection).

In the following, two cases are presented:

- radiographic inspection of tension bands of main rotor hub of EH 101, using xeroradiography as image collection medium

- US inspection, with electronic treatment of data, for the same tension bands, assembled in the whole rotor hub (the section of the analyzed hub has known standard defects introduced)

Obtained results are shown and commented in detail in pictures from 11 to 17.

5. CONCLUDING REMARKS

The classical interpretation of fatigue does not fit enough the fatigue composite structures behaviour, that looks like a "continuous degradation" driven by a very complex phenomenology strictly connected to damage mechanims and to the different phases present in composite materials.

An approach based on damage mechanics can be helpful for a better understanding of this behaviour, but it needs an accurate mathematical modelling, that considers the different damage modes and material phases, supported by fairly reliable techniques for the damage accumulation monitoring, and this introduces a quite new role of non destructive inspection.

6. ACKNOWLEDGEMENTS

The authors are greatly endebted to the colleagues of the Technology Development Department (of the Design and Development Service at Costruzioni Aeronautiche G. Agusta Gallarate - Italy) and to its chief Carlo Zanotti.

The proficuous help of professor V. Giavotto (Aerospace Engineering Division, Politecnico di Milano) and of Krautkramer firm is also to be acknowledged.

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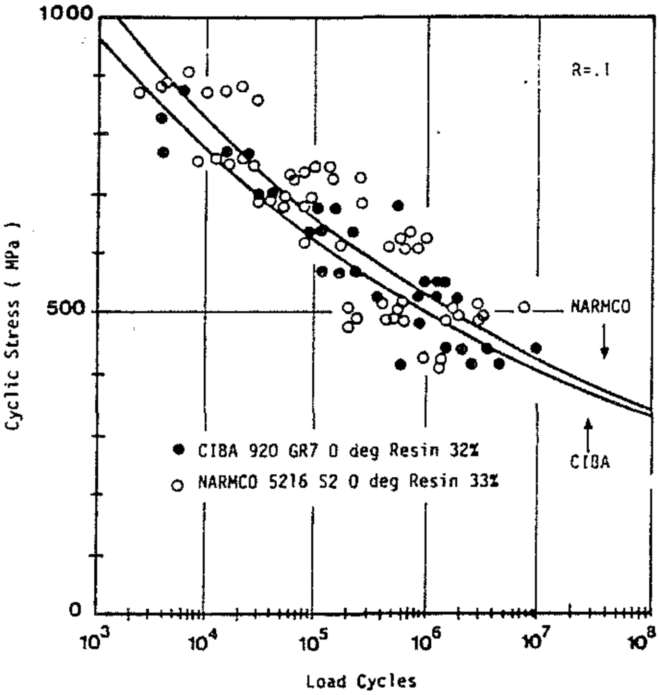

fig. 1: fatigue of unidirectional Glass/Epoxy in tension; failure, rapresented by the points, is defined as the appearance of visible matrix cracking or delamination (1).

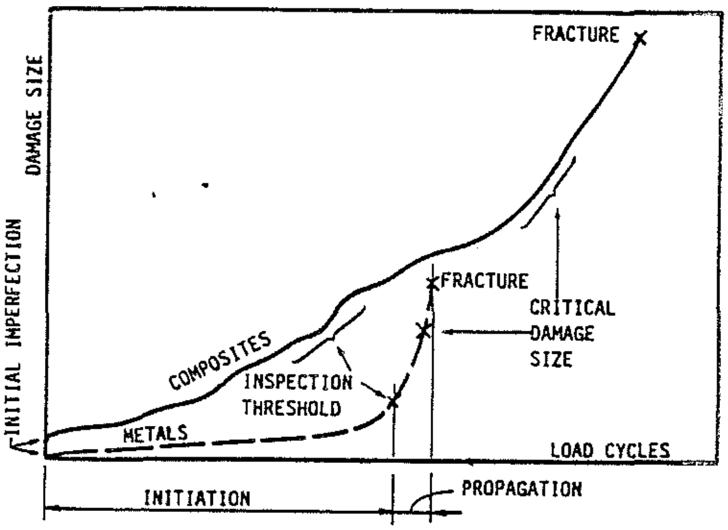

fig. 2: damage developement in metals and composites under cyclic loading (2).

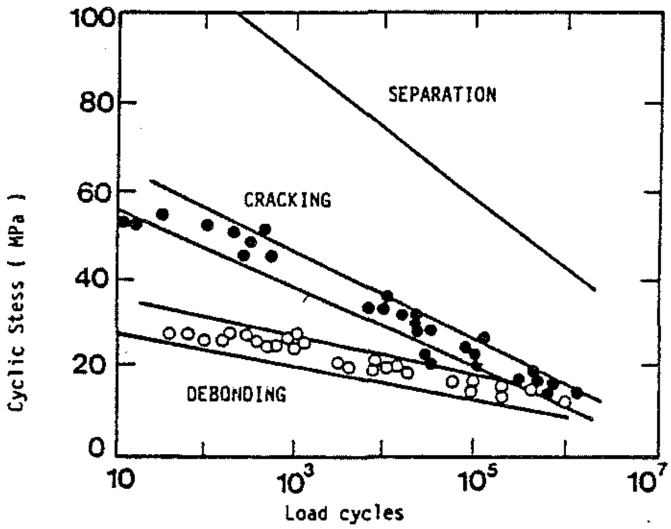

fig. 3: damage extension in short fibre Glass/Polyster (4).

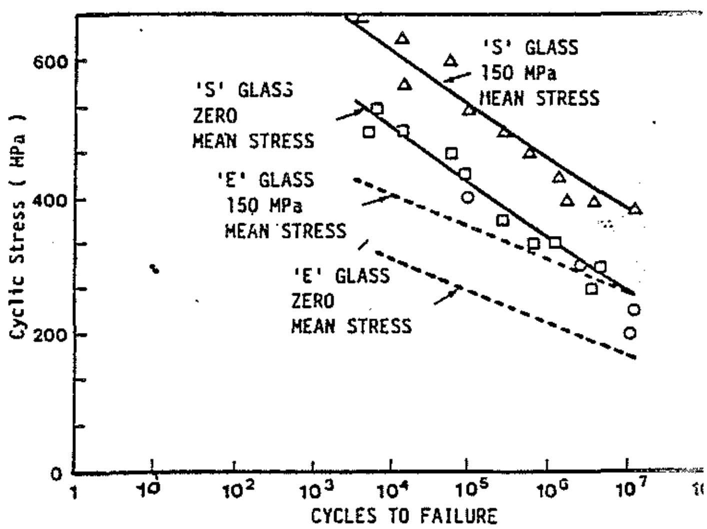

fig. 4: Effect of fibre stiffness on composite fatigue (2).

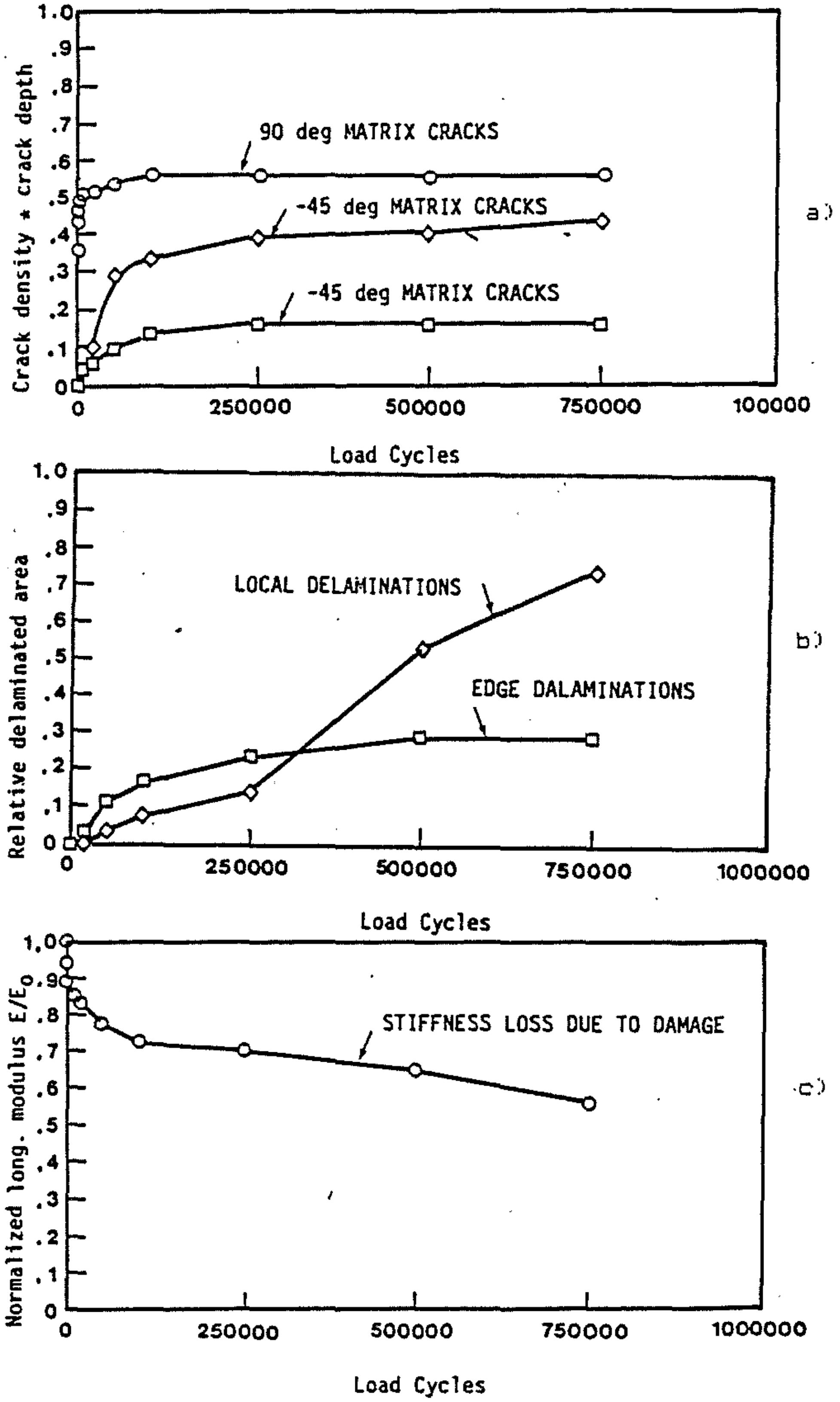

fig. 5: a) damage accumulation in a [+45/-45/0/90]s specimen (11)
b) delamination development in specimen of fig. 5a
c) stiffness loss in specimen od fig. 5a

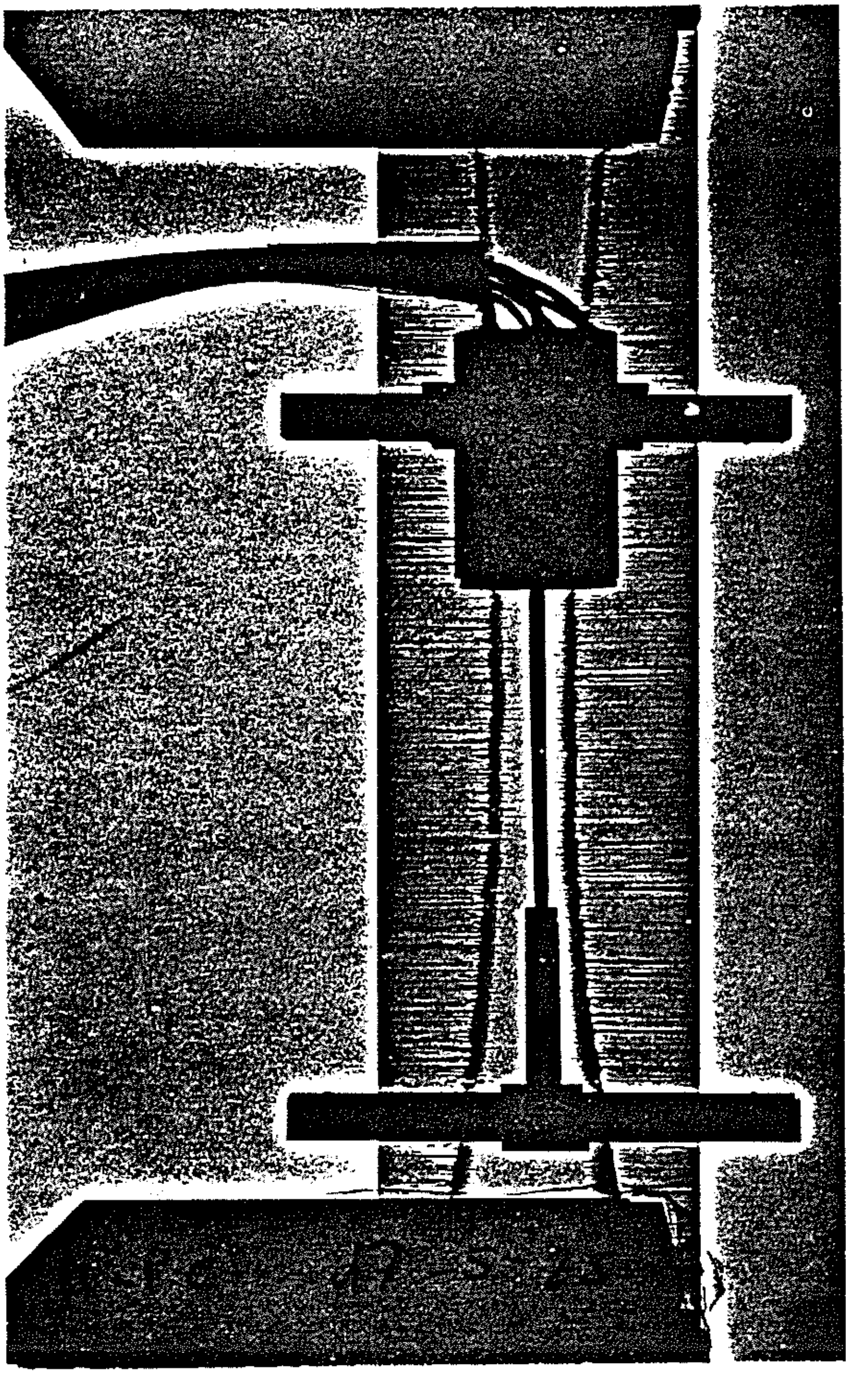

fig. 6: Xeroradiographic image of a graphite multidirectional specimen; the enhancement of damage is obtained by the use of a die penetrant radio-opaque solution.

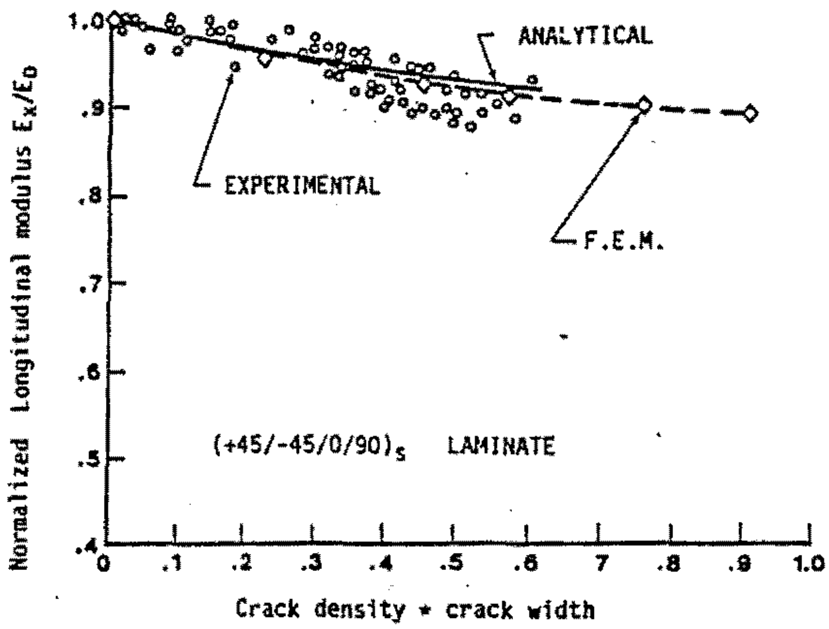

fig. 7: Stiffness loss in a quasi-isotropic laminate: comparison between experiment and analysis (11)

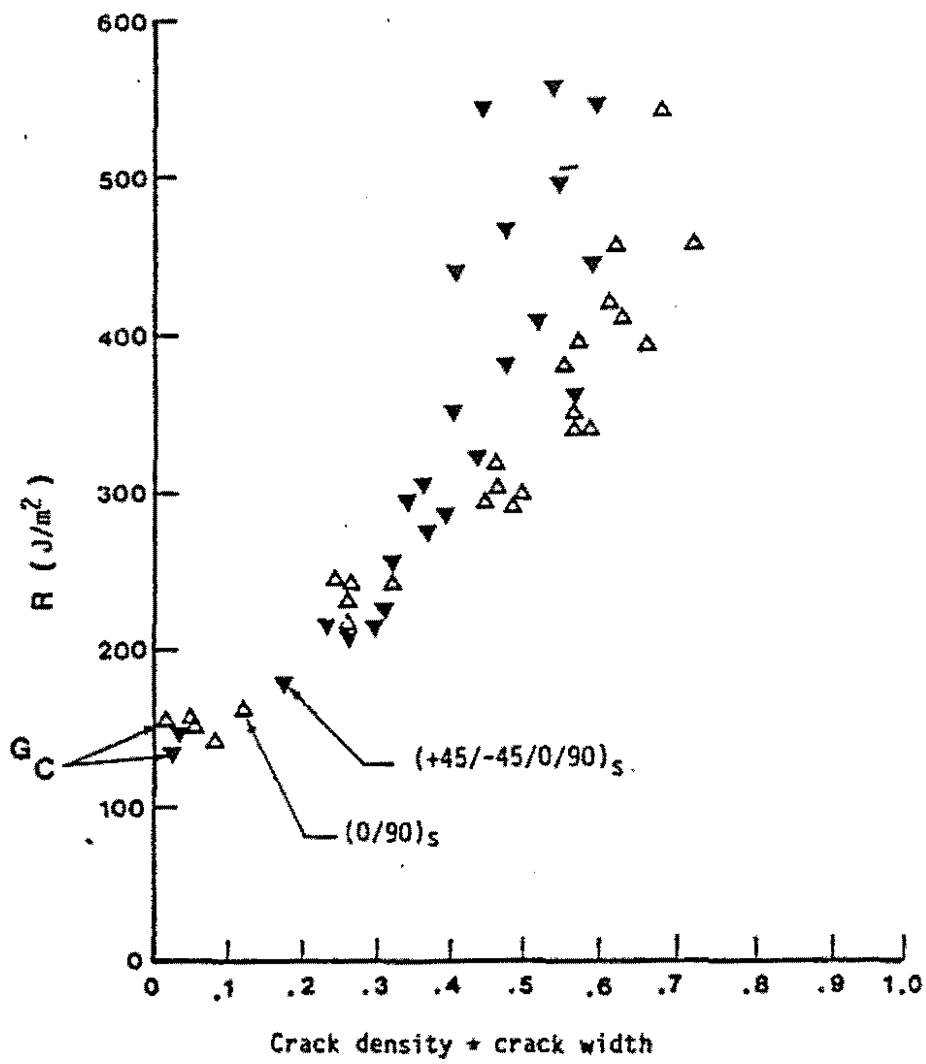

fig. 8: Matrix crack resistance curve: R-curve (11)

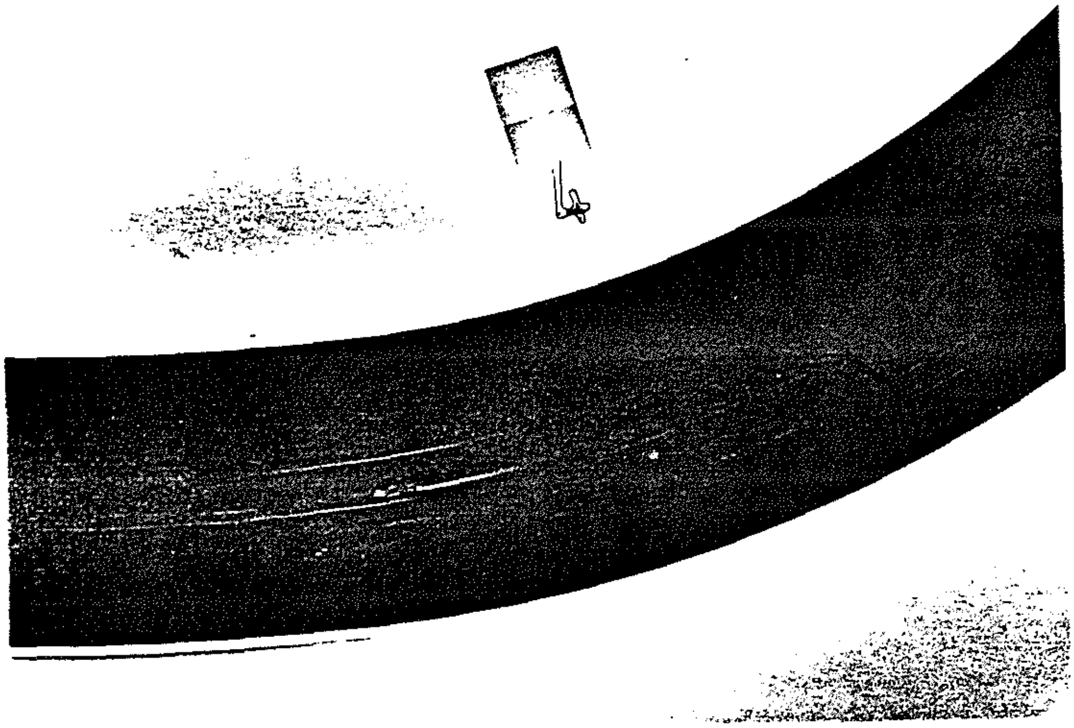

fig. 9: Xeroradiographic image of a tension band with delamination in the middle.

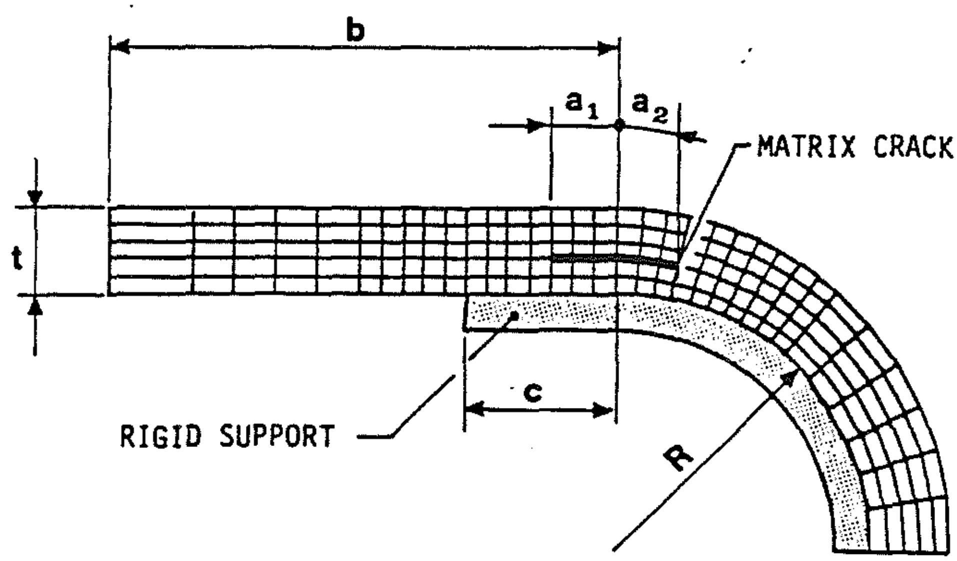

fig. 10: F.E. model of cracked tension band.

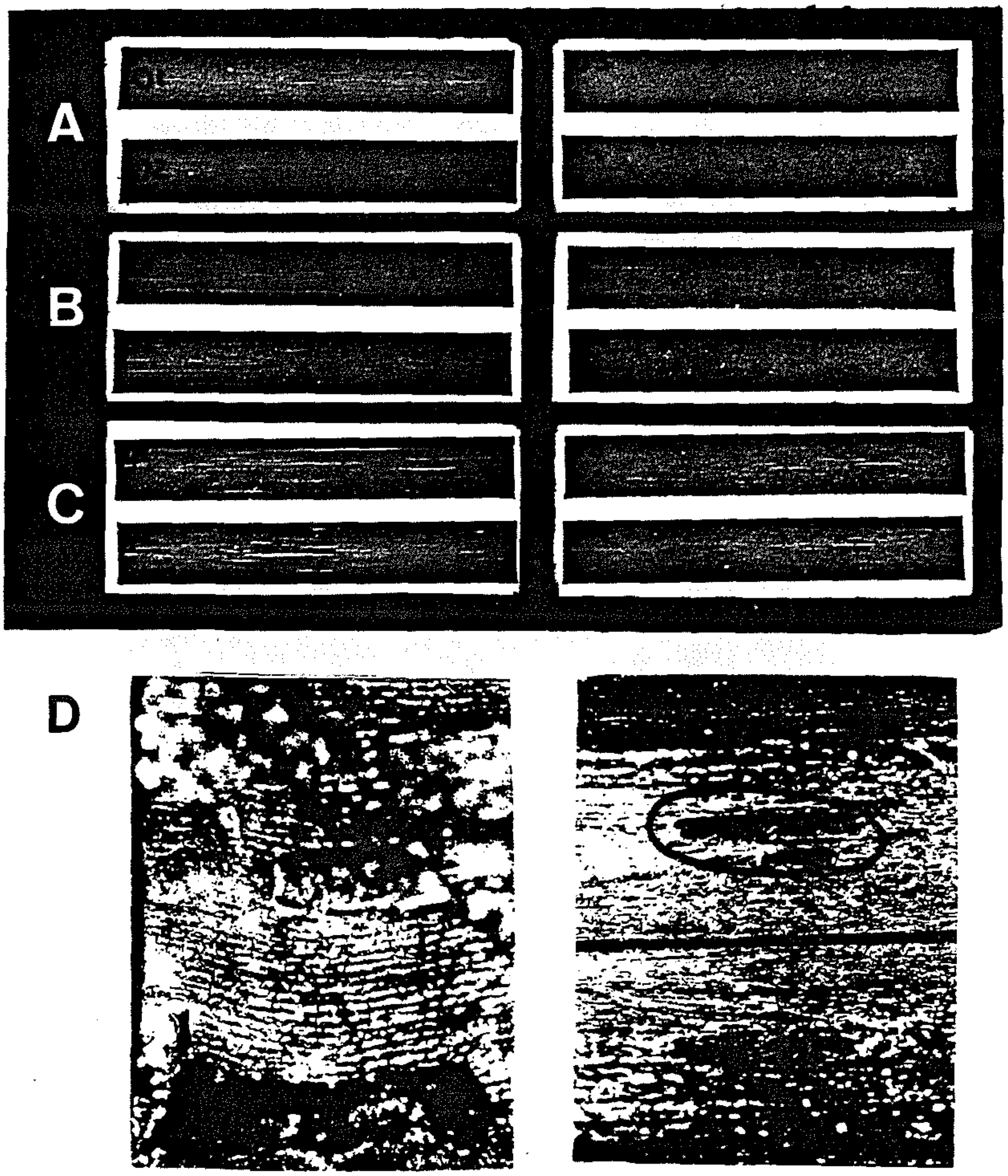

fig. 11: Images of thick unidirectional specimen with (01 & 02) and without (03 & 04) delaminations. Xeroradiographic views show the defective status of specimens before testing (A) and damage patterns after the first bending stress application (B) and after the second one (C).
In D, two interesting aspects of the failure are presented: it is evident (left side) that failure follows "interplies" surfaces , and (right side) that voids presence "drives" the failure.

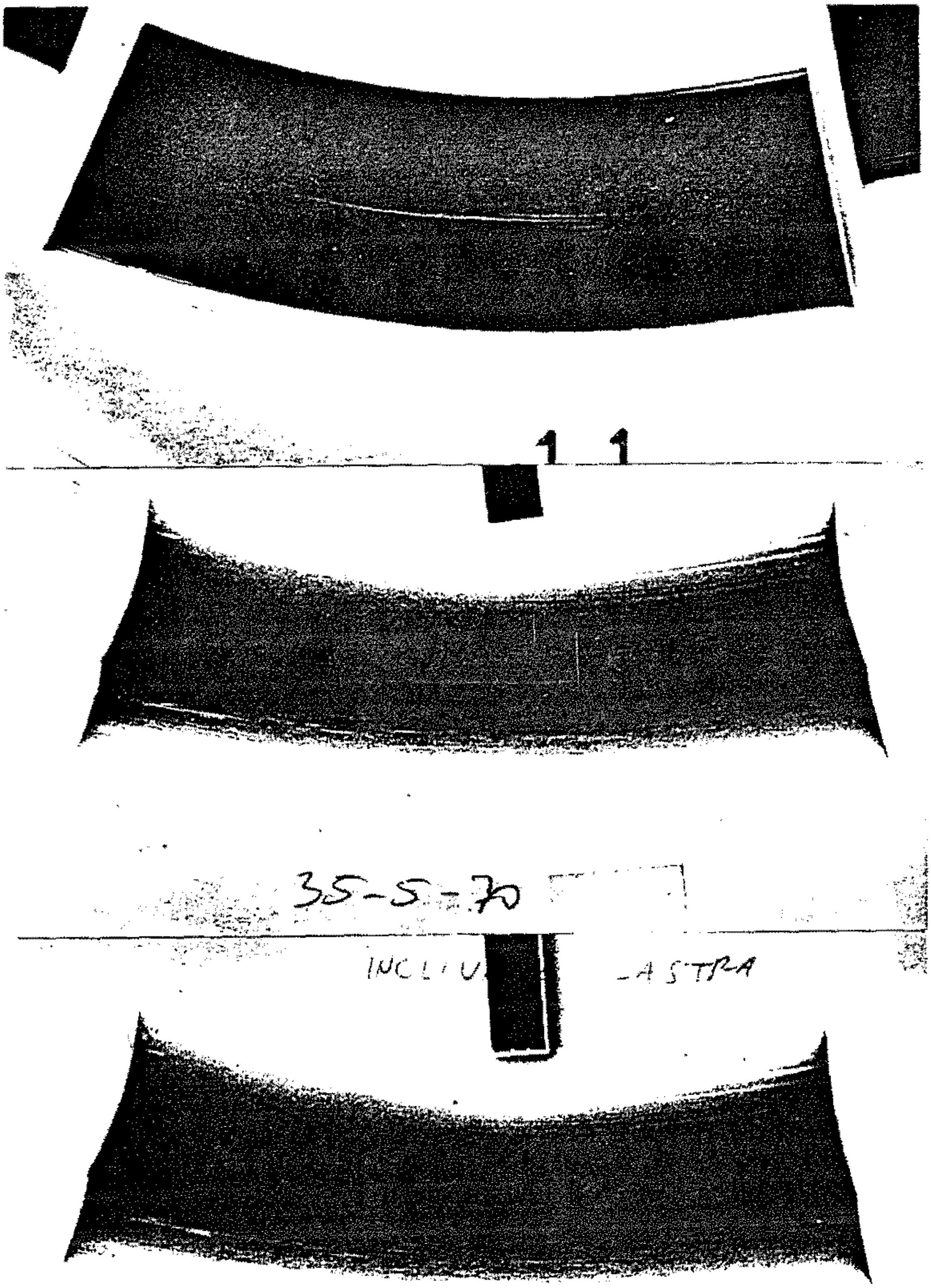

fig. 12: Xeroradiographic images on different projections allow a good understanding of defective patterns: in this situation it is possible to find a good method for defect/damage measurement.

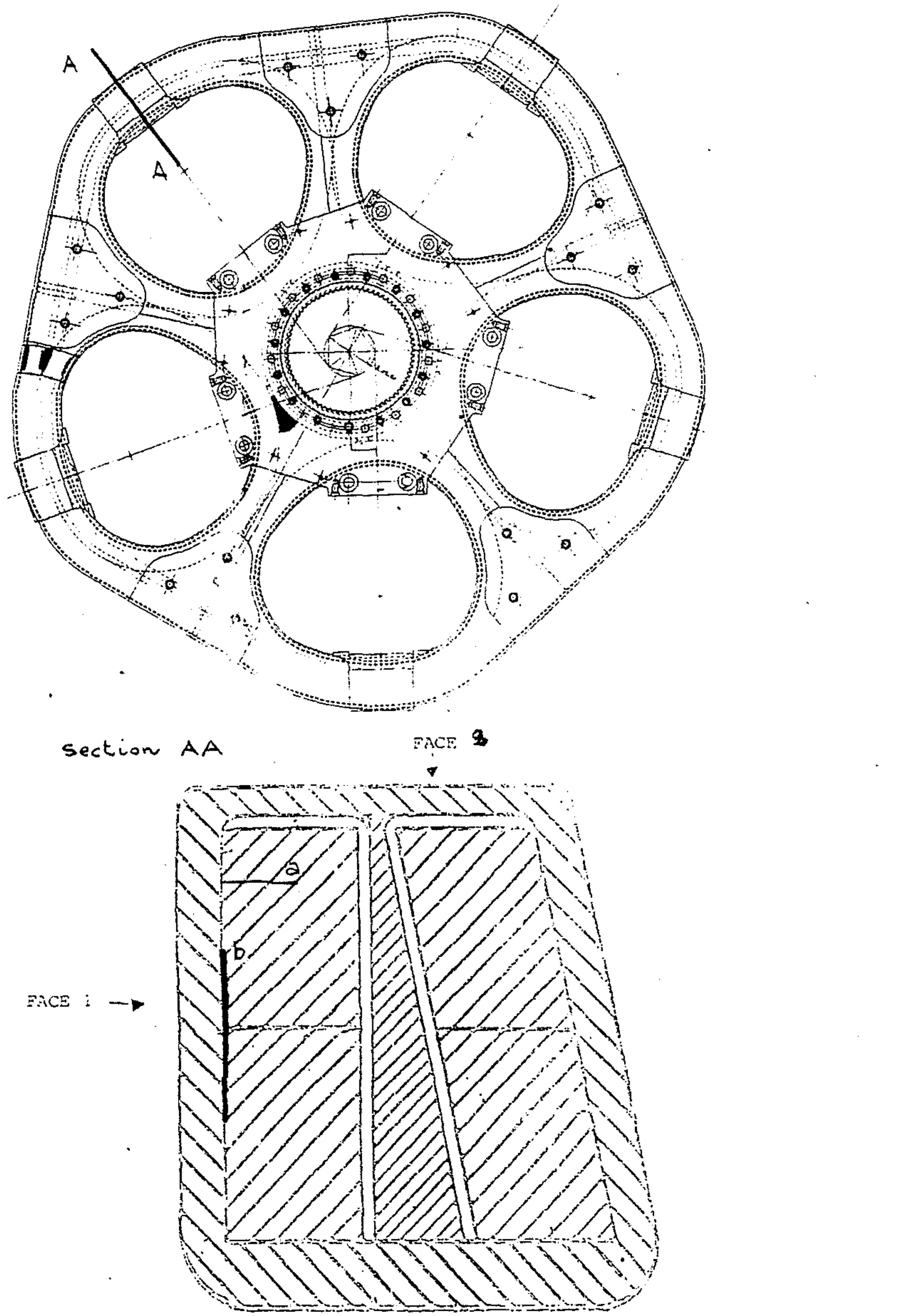

fig. 13: Scheme of the EH 101 main rotor hub section used like sample for U.S. testing with thin cuts reproducing defects:

a) simulate a matrix crack (delamination) in the graphite tension band

b) simulate a debonded area between internal structures and the box

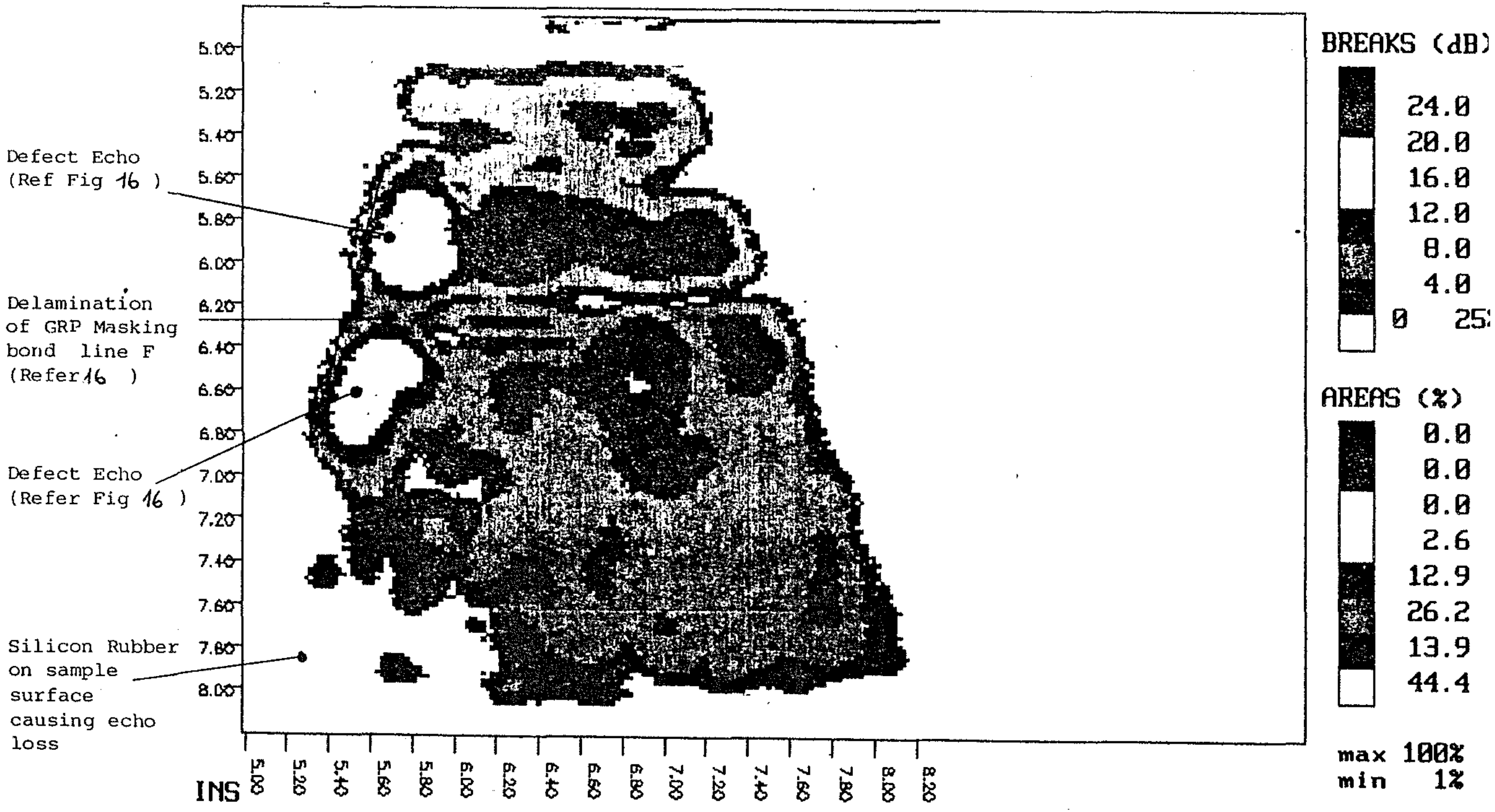

fig. 14: U.S. C-scan (pulse-echo technic) of the sample: plane "view" taken from face 1.

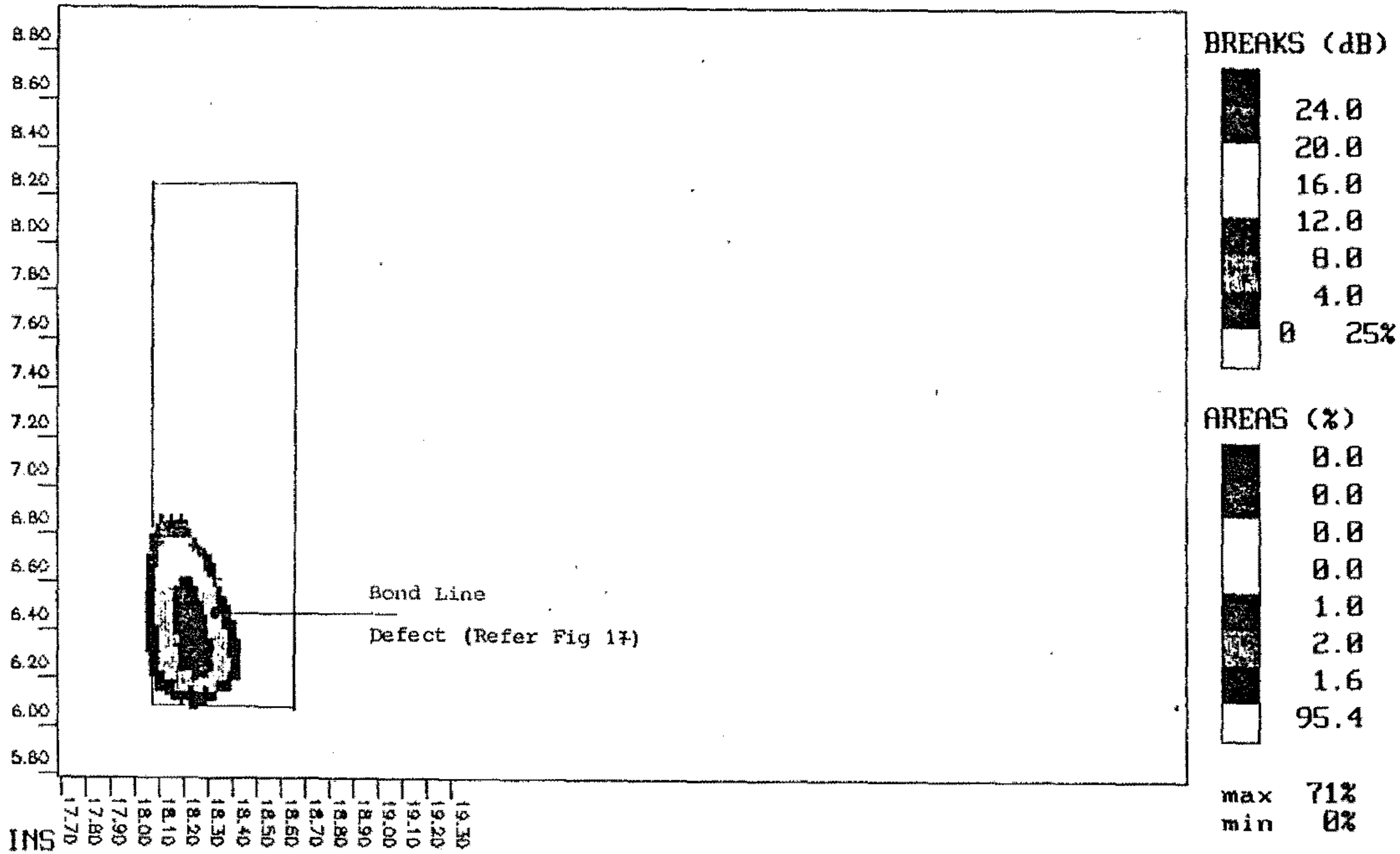

fig. 15: U.S. C-scan (pulse-echo technic) of the sample: side view.

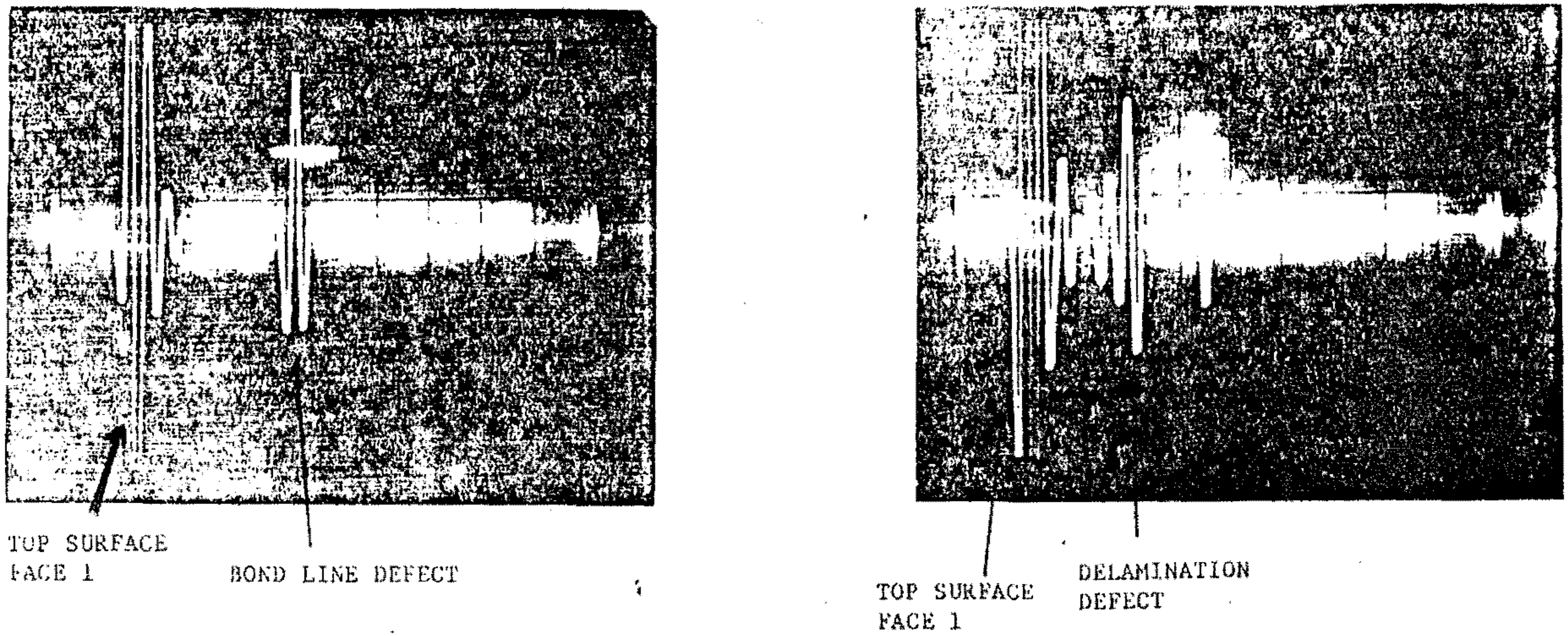

fig. 16: A-scan display of bond line defect and of delamination shown in fig. 14.

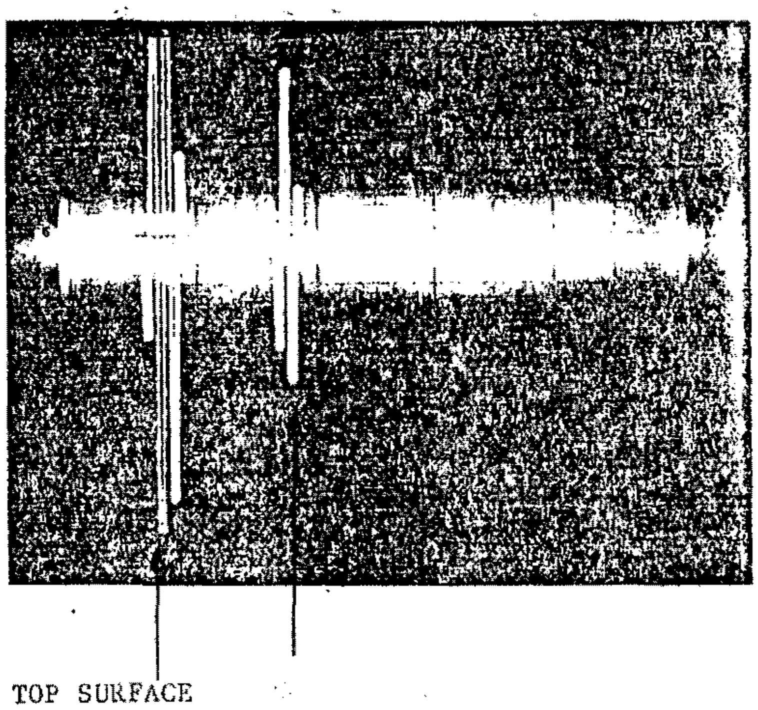

fig. 17: A-scan display of defects shown in fig. 15.