ROZPRAWY INŻYNIERSKIE · ENGINEERING TRANSACTIONS · 36, 4, 707—737, 1988
Polska Akademia Nauk · Instytut Podstawowych Problemów Techniki

DAMAGE MECHANICS EXPERIMENTAL BACKGROUND

M. BASISTA (WARSZAWA)

The present paper is a synthetic review of available experimental results concerning damage nucleation and growth in various materials at diverse loading conditions. An emphasis is placed on the physical complexity of the damage processes. To this end, an influence of the state of stress, type of applied load, material mesostructure and temperature are studied in detail. Characteristic features attributed to the damage process, such as damage-induced material inelasticity and anisotropy are stressed. The second part of the review will be devoted to selected damage models published so far.

1. Introduction

Real engineering materials contain numerous mesostructural flaws such as microcracks, voids, inclusions, pores, second-phase particles and other stress inhomogeneities even in the virgin, unstressed state. During the process of straining new microdefects appear while the already existing ones grow and link up. Nucleation and growth of these microdefects affect the mechanical properties of the material in the macroscale and reduce considerably its ultimate strength. Their coalescence in advanced stage of the deformation process leads to the nucleation of a macrocrack which brings the loaded element to final failure.

In the literature, the term "damage" or "damage process" has been attached to the phenomena of nucleation, growth and initial coalescence of a multitude of microdefects in the material structure. However, one should distinguish between damage and fracture. The first notion is intentionally referred to the weakening of material bearing capacity that manifests itself, among others, in the degradation of the elastic moduli and reduction of the material strength, whereas fracture generally means material disability to carry the load. In this sense, damage denotes a preceding phase of the fracture process. It is also a common case that damage and fracture appear simultaneously in the same material as a damaged zone often occurs in front of the macrocrack tip and it aids in the macrocrack propagation.

Obviously, these two phenomena differ in scale — damage develops at the mesoscale of the material structure while fracture is identified with a macroscopic crack running through the body. Following Krajcinovic [1], let us at this point introduce a simplified classification of the volume scales with corresponding typical defects at each level of the analysis, Tabl. 1. A material defect may be regarded as a microdefect in the damage mechanics sense if it has the same size as some mesostructural feature of the material, e.g. the grain size. Otherwise, it belongs to the fracture mechanics methods of analysis. In the case of plain concrete, for example, the microcracks are smaller than the characteristic radius of a mesoscale representative element being of the order of 10 mm [2]. In the majority of brittle rocks, the initial microcrack length does not exceed 1 mm [3, 4], while at the onset of structural failure it usually falls within the range of 3—7 mm [5]. For the hot-pressed ceramics like Si_3N_4, the mean grain size is of the order of 1 μm, so is the scale of microcracks that grow at the grain boundaries [6].

Fracture mechanics deals with one well-developed macrocrack or with regular arrays of such macrocracks assumed to be embedded into homogeneous continuum. Thus it is unable to account for the growth of the multitude of randomly distributed microdefects for which the surrounding material is strongly inhomogeneous due to the energy barriers produced, for example, by the grain boundaries.

Table 1. Volume scales with correseponding, characteritic material defects (from [1]).

Scale	Material	Defects
micro	atoms, molecule chains	vacancies, dislocations
meso	ensemble of grains	microcracks, voids
macro	specimen, structure	macrocracks, shear bands

Much of the inelastic behaviour of solids, especially brittle ones, is attributed to the accumulation of internal damage in their structure. Therefore is seemed quite appealing to try to employ the theory of plasticity, which is already fairly well developed, as a framework for analytical modelling of the damage process. However, as pointed out by Krajcinovic [1], such an approach fails for two reasons at least. Firstly, damage is generally connected with the loss of interatomic bonds, whereas plastic flow is related to the slip of matter through the crystalline lattice with no significant change in the number of bonds. Secondly, these two dissipative processes differ in that damage induces a different slope of the unloading path in the stress-strain curve as compared with the initial segment of that curve in loading, Fig. 1, [7, 8].

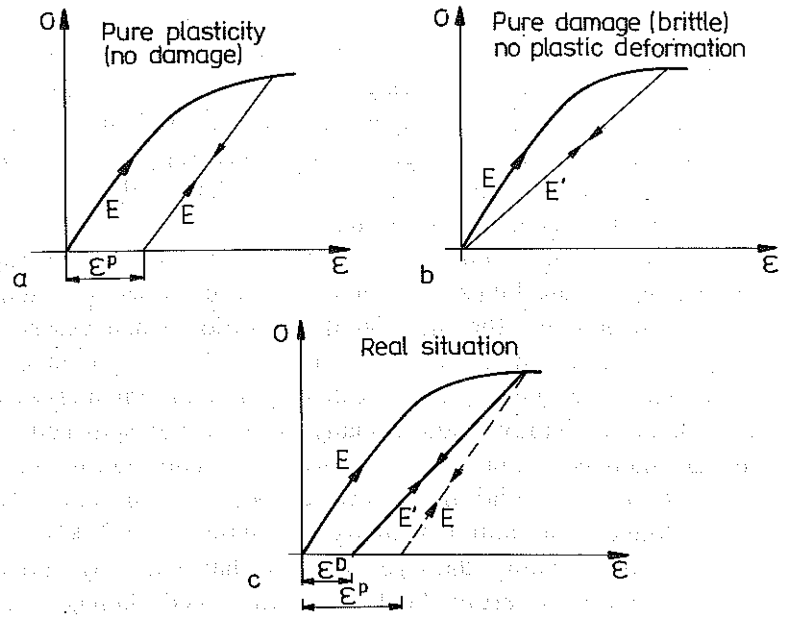


FIG. 1. Damage versus plasticity (figure based on the papers by DRAGON and MRÓZ [9], HULT [8]).

It may then be concluded that neither the fracture mechanics nor the theory of plasticity provide a suitable basis to investigate the damage process. Consequently, an urgent need has arisen for systematic, separate studies of the problem in hand, both on the theoretical and experimental plane.

Damage is a complex phenomenon strongly dependent on the material, type of applied loading, state of stress and temperature. From the purely geometrical standpoint, the class of microdefects in question can be divided into two groups:

microcracks: roughly planar in shape, growing mostly on the grain boundaries, typical of brittle solids (rocks, concrete, ceramics, metals at certain temperature levels);

voids, being of volumetric nature, growing within the grains or on grain boundaries; observed mainly in ductile materials (majority of metals and alloys).

As far as different types of damage are concerned, one can distinguish brittle (or elastic-brittle) damage, ductile (or elastic-plastic) damage, creep damage, fatigue damage, damage of initially anisotropic materials like composites, and spall damage which is due to impulsive loads of short duration. Each type will be analyzed in detail later on.

Except for very ductile metals that manifest a dilute and random concentration of almost spheroidal voids, the nucleation and evolution of microdefects in most of the engineering materials exhibit pronounced anisotropy that can in many cases be related to the prinicpal directions of the stress- or strain tensor. This damage-induced anisotropy is an important feature of the whole phenomenon and cannot be disregarded if theoretical modeling of material behavior is to be realistic.

The origins of damage mechanics go back to the pioneering paper by Kachanov [10] in the late fifties, who first introduced a separate mathematical quantity to account for the material deterioration when describing the tertiary creep of metals under uniaxial tension. It was a macroscopic scalar variable that represented microcrack density in a specimen cross-section. That early work of Kachanov inspired many authors and spawned a lot of diverse damage models within the framework of continuum mechanics, so that a new branch of solid mechanics known as Continuum Damage Mechanics was found Continuum Damage Mechanics (CDM)(1) envelops a wide gamut of the existing damage theories that start by introducing into constitutive equations certain field variables, called damage variables, that reflect the current state of internal material deterioration in an average sense. If a model aims at describing damage as a process, the defined damage variable requires formulation of an evolution law (damage law) that would govern the growth of microdefects. Besides numerous variations of the original Kachanov's concept postulating the damage measure in the form of a scalar parameter, various vectorial and tensorial models were proposed, motivated by the fact that damage process reveals directional character that can hardly be described using solely a scalar variable.

The present paper, being an extended and updated version of the author's earlier publication [13], is intended to be a comprehensive state-of-the-art work summarizing the most remarkable results achieved in the field of Continuum Damage Mechanics. Its main objective is to provide a synthetic yet detailed review both of the experimental observations and the existing theoretical models of damage processes in various materials at diverse loading circumstances. Although most of the paper will be devoted to damage of brittle and ductile materials under quasi-static loadings, some attention will also be paid to damage in creep conditions, fatigue damage, spall damage, and damage of composites. However, it is by no means claimed that this report comprises all the research effort done in the field of Continuum Damage Mechanics. Inevitably, the selection of papers and the objectivity of the review are, to some extent, biased by the author's own opinions upon the problem considered.

---

(1) Janson and Hult [11] were probably the first who used the term CDM, as remarked by Krajcinovic [12].

2. Experimental observations

2.1. *Damage in brittle solids*

Damage in brittle materials such as rocks, ceramics and concretes subjected to quasi-static loadings takes usually the form of sharp, flat microcracks that grow either in the cleavage mode or result from grain boundary sliding. Microcracks show in this case a clearly defined orientation, thus damaged brittle material must be considered as markedly anisotropic.

It is commonly reported in the literature [6, 14, 15, 16] that in the uniaxial tensile tests microcracks develop mainly in the planes that are

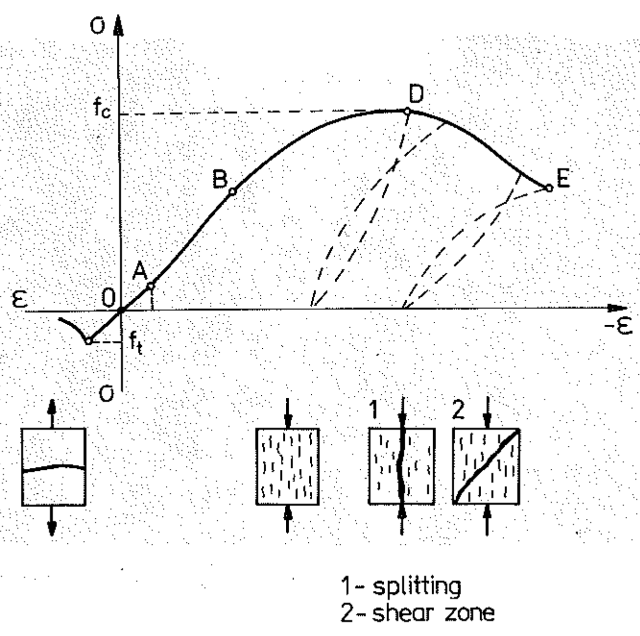


Fig. 2. Damage-fracture process of rock-like materials under uniaxial compression (after Dragon and Mróz [9], Murakami [16]).

nearly perpendicular to the load axis. Eventually, they coalesce into one continuous macrocrack that traverses across the entire cross-section of the specimen [17, 18].

An apparently different mechanism of damage is oserved in brittle solids subjected to uniaxial compression. At low stresses, a typical stress-strain curve for rock-like materials, Fig. 2, is nonlinear and concave upward (OA) due to the closure of pre-existing microcracks and pores [4]. After that, almost linearly elastic behaviour of the rock is found (AB), with no significant damage activity within the material. In the regions (BD) of stable inelastic deformation, intensive microcracking is reported [4] predominantly in the planes that are oriented parallel to the direction of compressive stress [5, 19]. Beyond the instability point D, the microcracks begin to interact and coalesce. This process accelerates and a dominant fault is formed which means the onset of final fracture of the specimen. Two modes of gross fracture are observed in the analyzed case: splitting [20] and localized shear zone [14, 15, 21]. It is perhaps worthy to comment upon that any CDM based theory for the behaviour of rocks under uniaxial compression should not be extended over the post-critical branch (DE) of the $\sigma-\varepsilon$ curve shown in Fig. 2 since the assumptions of the CDM cease to be valid in this region as the macroscopic, distinct crack has already arisen.

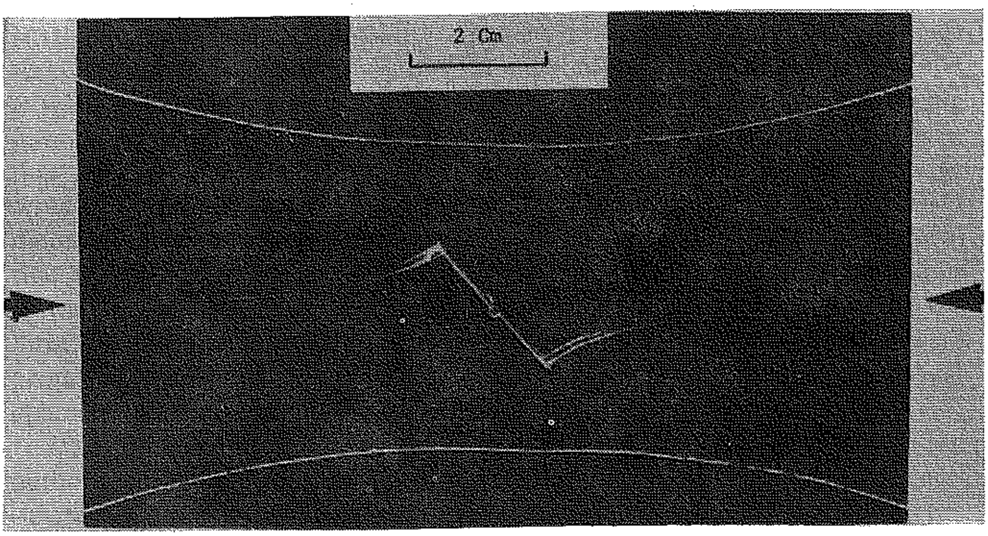


FIG. 3. Kinked microcrack in resin CR 39 subjected to uniaxial compression (from [22]).

A qualitative evidence for the above-stated orientation of the microcrack pattern in brittle solids under uniaxial compression may be found in the papers by HORII and NEMAT-NASSER [22, 23](²) who examined the damage growth by simulating the pre-existing flaws in the model plates made of amorphous brittle materials such as resin CR 39 and glass. Their experiments confirmed that kinked microcracks that nucleate at the tips of pre-existing flaws become parallel to the direction of maximum far-field compression after some curving at an early stage of the loading process, Fig. 3. Additionally, an influence of a lateral confinement that accompanied the dominant compression was also reported. The lateral tension, when imposed

(²) See also the earlier papers by BRACE and BOMBOLAKIS [25] and HOEK [26].

upon the specimen, made the microcrack growth unstable after a certain crack length had been reached. Conversely, when the lateral compression existed, the crack growth was stable and stopped after some finite length, unless the maximum principal compression increased. It was also found that the larger cracks moved first as compared to the smaller ones.

Final overall splitting due to coalescence of cracks was observed.

The well-known effect that in concretes, rocks and ceramics subjected to uniaxial compression the microcracks are aligned parallel to the axis of loading has encountered substantial difficulties as regards its physical explanation. It seemed as if the microcracks could open against compressive stresses (for details consult [19]). In order to overcome these difficulties, some authors claim that it is always a tensile deviatoric stress that causes the nucleation and growth of microcracks [3, 27, 28]. One can easily check that experimentally observed facts of the microcrack orientation both in compression and tension are then preserved. In our opinion, however, this is but a formal assumption which does not help very much in the elucidation of the problem. It can readily be clarified if one recognizes that the materials in question are in fact inhomogeneous. Material inhomogeneity induces local tensile stresses needed to initiate the microcrack propagation under the external compression.

In polycrystalline rocks fluctuating tensile stresses arise from material property mismatches between grains or from contact between the grains with irregular boundaries [5]. In the case of concrete, local tensile stresses

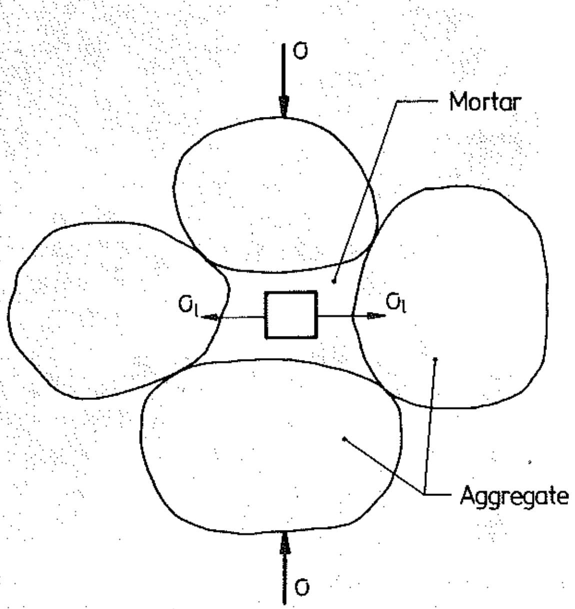


FIG. 4. Local tensile stresses in concrete due to squeezing of aggregates (from [19]).

may be generated at the tips of the interface microcracks due to sliding of the aggregates along the microcrack faces [24]. Splitting stresses may also result from squeezing out of the aggregates as shown in Fig. 4. In ceramics, the grain boundary sliding can produce high tensile stresses when blocked at triple points and cavities are thereby created, Fig. 5

The local tensile stresses responsible for the microcracking in concrete, rocks and ceramics may significantly differ from the externally applied stresses due to the inhomogeneity of these materials. This non-uniform redistribution of the external stresses within the material should be taken into account when modeling damage in brittle solids.

Precise observations indicate that the microcrack growth is governed not only by the state of stress but also by the distribution of planes of

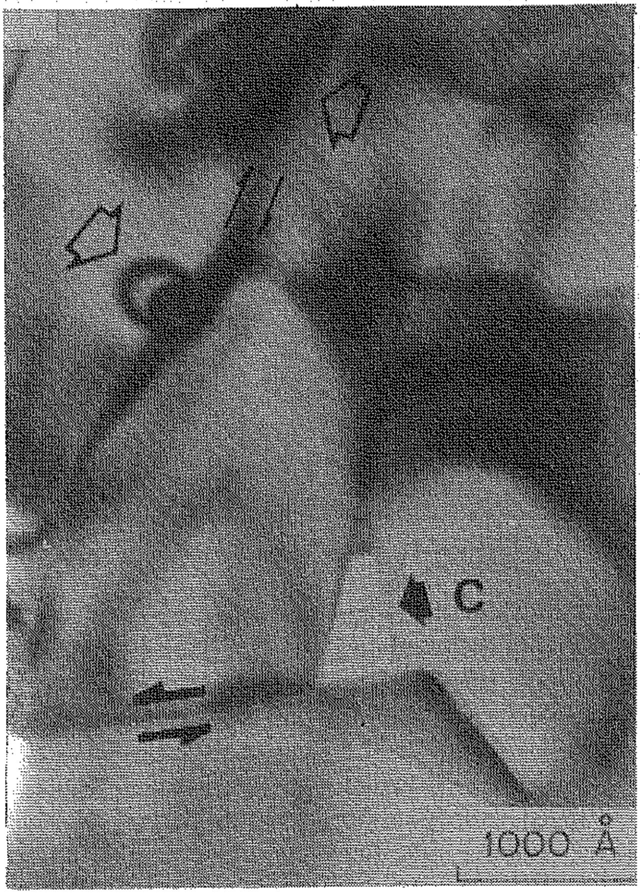


FIG. 5. Cavitation in ceramic HP-Si_3N_4 (from [6]); arrow C indicates cavity nucleation at triple point; double arrows denote grain boundary sliding; empty arrows show stress concentrations resulting from grain boundary sliding.

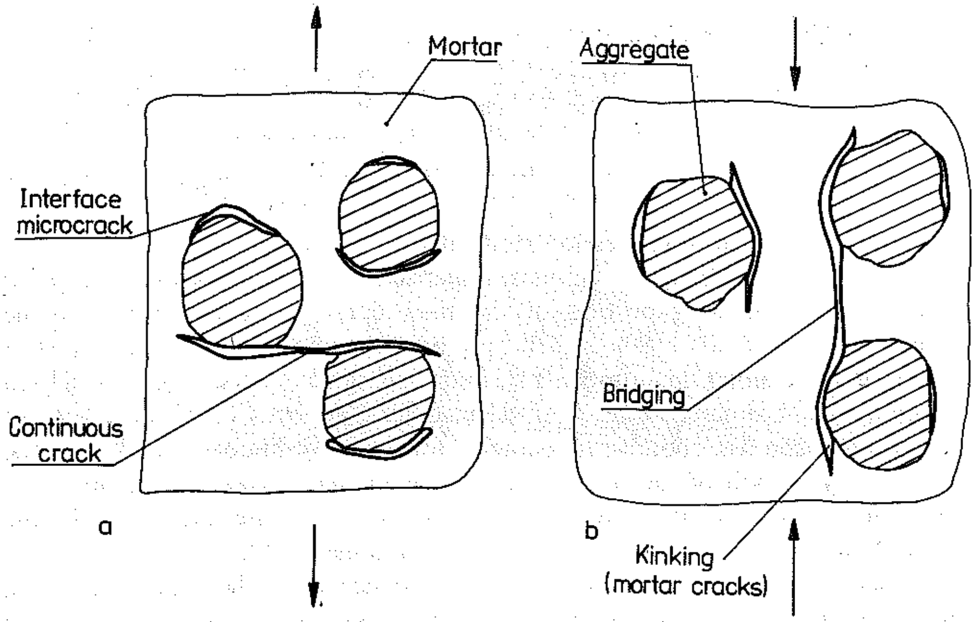


FIG. 6. Damage mechanism in concrete: a) uniaxial tension, b) uniaxial compression.

lower toughness in the material mesostructure. Moreover, the energy barriers produced by the grain boundaries or aggregates can arrest the propagating microcrack. Let us illustrate it on the example of plain concrete. It is known that initial microdefects already exist within concrete prior to the first application of loading [29]. Microscopic inspection of a non-loaded specimen reveals that pre-load microcracks appear most frequently at the aggregate-matrix interfaces which are the weakest links of the concrete mesostructure, Fig. 6a. The occurrence of the initial microcracks results from the mix water migration (bleeding) during the forming process, as well as from volume changes during hardening (shrinkage) [30]. Apply now an external tensile loading. Since the critical stress intensity factor K_{I}^{c} takes the lowest value for the aggregate-matrix interface, Tabl. 2, the damage commences in this region mainly at those aggregate facets that are oriented perpendicularly to the load axis. The destabilized microcrack grows along the aggregate facet until it comes to the facet edge where it is forced to stop by a superior toughness of the mortar. As the external tension increases, more and more microcracks become active and a deviation from the elastic bridge between the bond microcracks forming continuous crack patterns. the interfacial microcracks branch into mortar, coalesce with adjacent microcracks and form a continuous transverse macrocrack. The macrocrack grows further in an unstable manner up to the final separation of the specimen.

Table 2. Critical values of K_{I} factors for plain concrete (from [2]).

	Aggregate (K_{IC}^{A})	Cement matrix (K_{IC}^{M})	Aggregate-matrix interface (K_{IC}^{IF})
$K_{IC}/K_{IC}^{\mathrm{Matrix}}$	1	1	0.5

In the uniaxial compression test, up to some 30% of the ultimate compressive strength f_c, no significant increase in microcracking is reported and the corresponding portion of the $\sigma-\varepsilon$ curve is nearly linear. In the stress interval of 30%—80% f_c, the pre-existing interface microcracks start to propagate along favourably oriented aggregate facets, Fig. 6b. Upon reaching approximately 80% f_c, mortar cracks appear. The mortar cracks bridge between the bond microcracks forming continuous crack patters. A runaway growth of the macrocracks through the mortar is next observed. If such a crack is not arrested by a coarse aggregate on its path, the overall fracture (in most cases-splitting) is imminent [1].

There is still a lack of experimental evidence from the damage-oriented tests for brittle materials in the multiaxial stress states. Instead, a common assumption is made, similarly as in the uniaxial cases, that microcracking be always driven by the tensile stresses [19, 5, 6]. Microcracks are thus expected to grow in those planes that are aligned nearly perpendicular to the directions of principal tensile stresses. Obviously, such an assumption leads to the orthotropic symmetry of the microcrack distribution which is not necessarily true. Therefore, any damage model of that type is a hypothesis-based theory rather than a rational image of the real brittle material behaviour.

2.2. *Damage in metals*

Mechanisms of damage in metals and alloys subjected to quasi-static loadings are of a more complicated nature since they essentially depend on the temperature. Below some 0.4 T_M, where T_M denotes the melting temperature, polycrystalline metals may degrade either by cleavage or in a ductile way. At higher temperatures creep damage is the principal mechanism of the internal structure degradation [31, 32].

Let us focus on the damage mechanisms for metals within the range of test temperatures lower than 0.4 T_M. There exist two competing types of these mechanisms: trans- or intergranular cleavage and ductile damage. A so-called cleavage-fibrous fracture transition temperature is the main factor that separates them. Transgranular cleavage is brought about by the microcracks that can arise due to one of the following micromechanisms: intersections of the slip bands with grain boundaries, intersections of slip bands

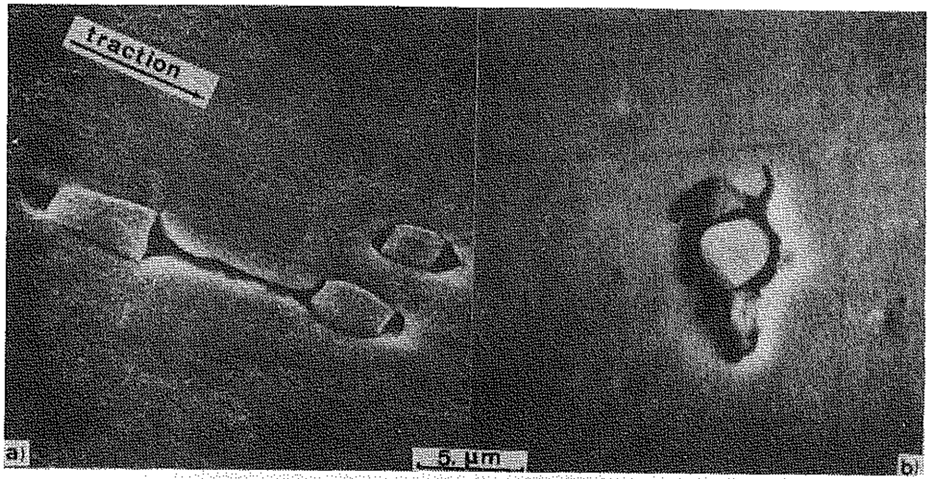


FIG. 7. Void formation by particle — matrix decohesion in copper; a) uniaxial tension, b) biaxial tension; (from [38]).

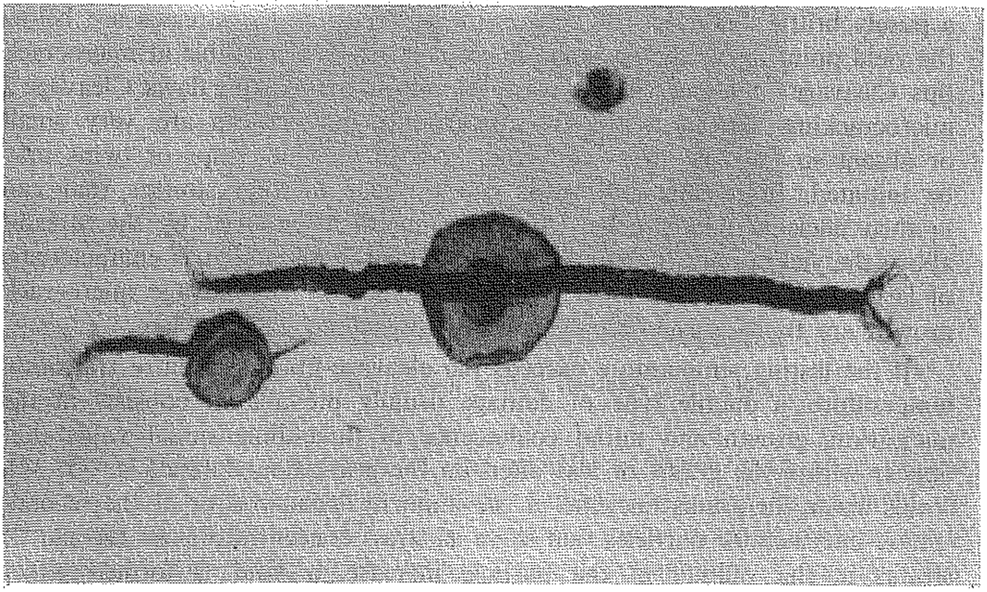


FIG. 8. Cracking of second-phase particle in high-strength steel (from [44]).

themselves, and twin intersections. Exact descriptions of these mechanisms are beyond the scope of this paper and may be found elsewhere [33]. Brittle separation along grain boundaries (intergranular cleavage) is caused by the cracking of brittle particles localized at the grain boundaries, due to the dislocation pile-up [15, 34].

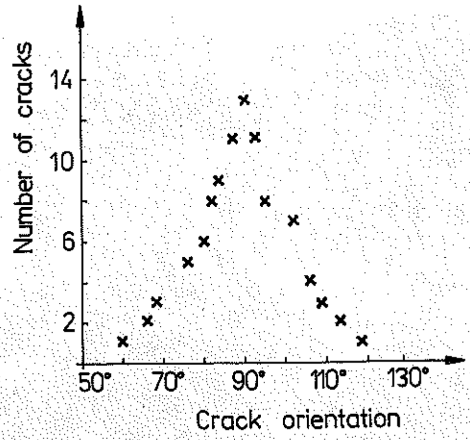


FIG. 9. Frequency of crack orientation relative to tensile axis (from [45]).

ASHBY [31] distinguishes "cleavage 1", "cleavage 2", "cleavage 3" within the grains as well as "brittle intergranular damage". Cleavage 1 means a purely brittle damage with no general plasticity as the stresses are below the yield point. Nevertheless some microplasticity is admissible at microcrack tips. Cleavage 1 microcracks always develop from pre-existing flaws. Cleavage 2 microcracking appears when the pre-existing microcracks are very small or absent so that the stress can reach a level at which slips or twins may be initiated. Both slips and twins can produce internal stresses that can in turn nucleate microcracks. Cleavage 3 mode is induced by the stresses above the yield point. It requires higher temperatures and it is characteristic of large plastic strains present. Preferred planes of the cleavage microcrack growth are those normal to the tensile stress direction.

If the testing temperature exceeds the cleavage-fibrous fracture transition temperature, a ductile type of damage usually prevails. It consists in the nucleation, growth and initial coalescence of three-dimensional voids that appear within the grains or less often on the grain boundaries [34, 35, 37]. Voids inside the grains nucleate at the impurity inclusions or at the second--phase intermetallic particles that serve as stress concentrators. As the plastic strains in the vicinity of the particle increase, a void is formed either by decohesion along the matrix-particle interface owing to the fact that a hard particle cannot deform as easily as the matrix, Fig. 7a, b, [38, 39, 40, 41, 42, 43] or by cracking of the particle itself, Fig. 8, [37, 44]. Void nucleation due to particle microcracking was studied experimentally by GURLAND [45, 46]. He observed that in a spheroidized steel carbide particle micro-cracking preferentially occurs at right angles relative to the tensile stress axis, Fig. 9, whereas in uniaxial compression these microcracks tend to lie parallel to the load direction.

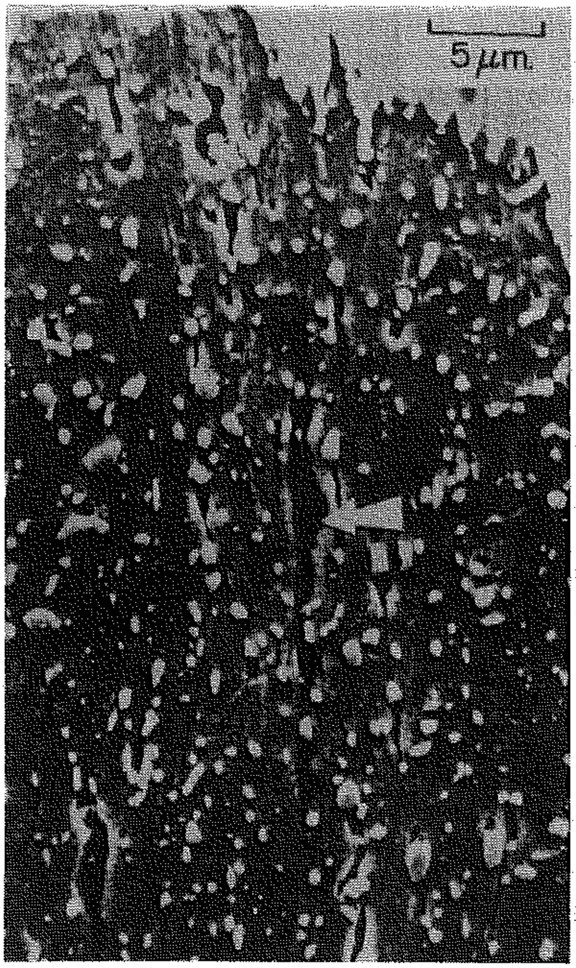


FIG. 10. Void growth in steel specimen (from [36]).

If the fracture toughness of the matrix is lower than that of the inclusion or inclusion-matrix interface, damage may also be initiated by fracturing of the matrix [47]. Once nucleated under uniaxial tension, voids elongate by slip mostly in the tensile stress direction, Figs. 7 and 10. They enlarge with an increasing stress until some critical plastic strain is attained at which the adjacent voids link up by a transverse local necking. This leads to the macroscopic fracture path localized along the plane normal to the direction of tension [36, 31]. If the temperature is raised, a grain boundary sliding occurs and the voids nucleate mainly on the grain boundaries. A mechanism of void formation in this case consists in blocking of the grain boundary sliding at the triple points, ledges or tougher particles that gives rise to the concentration of stresses and, subsequently, to the nucleation of voids. It is worth mentioning that the grain boundary voids nucleate mainly on those boundaries that are parallel to the direction of maximum principal tensile stress [48].

2.3. *Creep damage*

In the past few years, the problem of material damage in creep conditions has intensively been investigated both experimentally and theoretically, e.g. [49, 50, 51, 52, 53, 48, 54, 55, 56, 57].

It has been experimentally confirmed that in polycrystalline metals subjected to creep régimes, damage initiation and accumulation concentrate on the grain boundaries aligned perpendicularly to the maximum tensile stress axis [54, 58, 59, 60]. Two different types of creep-induced microdefects were observed: volumetric cavities, almost spherical in shape, which are often called "*r*-type voids", Fig. 11a, and sharp, wedge-shaped microcracks known in the literature as "*w*-type microcracks", Fig. 11c. Some authors distiguish also a third group of microdefects called "angular voids", Fig. 11b, but these appear very rarely in comparison with the *r*-type voids or *w*-type microcracks. Following a comment by Murakami [15], the appearance of either type of these defects depends on the material, state of stress and temperature. Sometimes, the *r*-woids and *w*-microcracks nucleate within the same material but then *w*-microcracks occur more easily at lower temperatures and higher stress levels. A prerequisite for the initiation of both the *r*-voids and *w*-microcracks is the grain boundary sliding [15, 55]. The *r*-voids arise as a consequence of the stress concentrations induced by the disturbances in the grain boundary sliding process at certain microstructural irregularities along grain interfaces. Evolution of the *r*-voids advances according to two distinct mechanisms, namely that of diffusion and condensation of vacancies along the grain and that of strain-controlled growth; for further information consult Evans [32]. A potential source for the *w*-type microcracks nucleation are the triple junctions. They are able to block the grain boundary sliding thus to cause local stress concentrations that lead to the nucleation of microcracks [15, 16, 32, 44].

As regards creep damage in polycrystalline ceramics, three stages are again recognized, that is: cavity initiation, cavity growth to form microcracks and coalescence of microcracks into macrocrack. Observations on the sintered and hot-pressed ceramics assembled in a review paper by Tsai and Raj [6] suggest that the cavities nucleate at the triple junction pockets due to large tensile stresses brought about by the grain boundary sliding, Fig. 5. Glassy regions are often present at the triple junctions and at grain interfaces owing to the way the ceramics are processed. Therefore, the cavities that nucleated

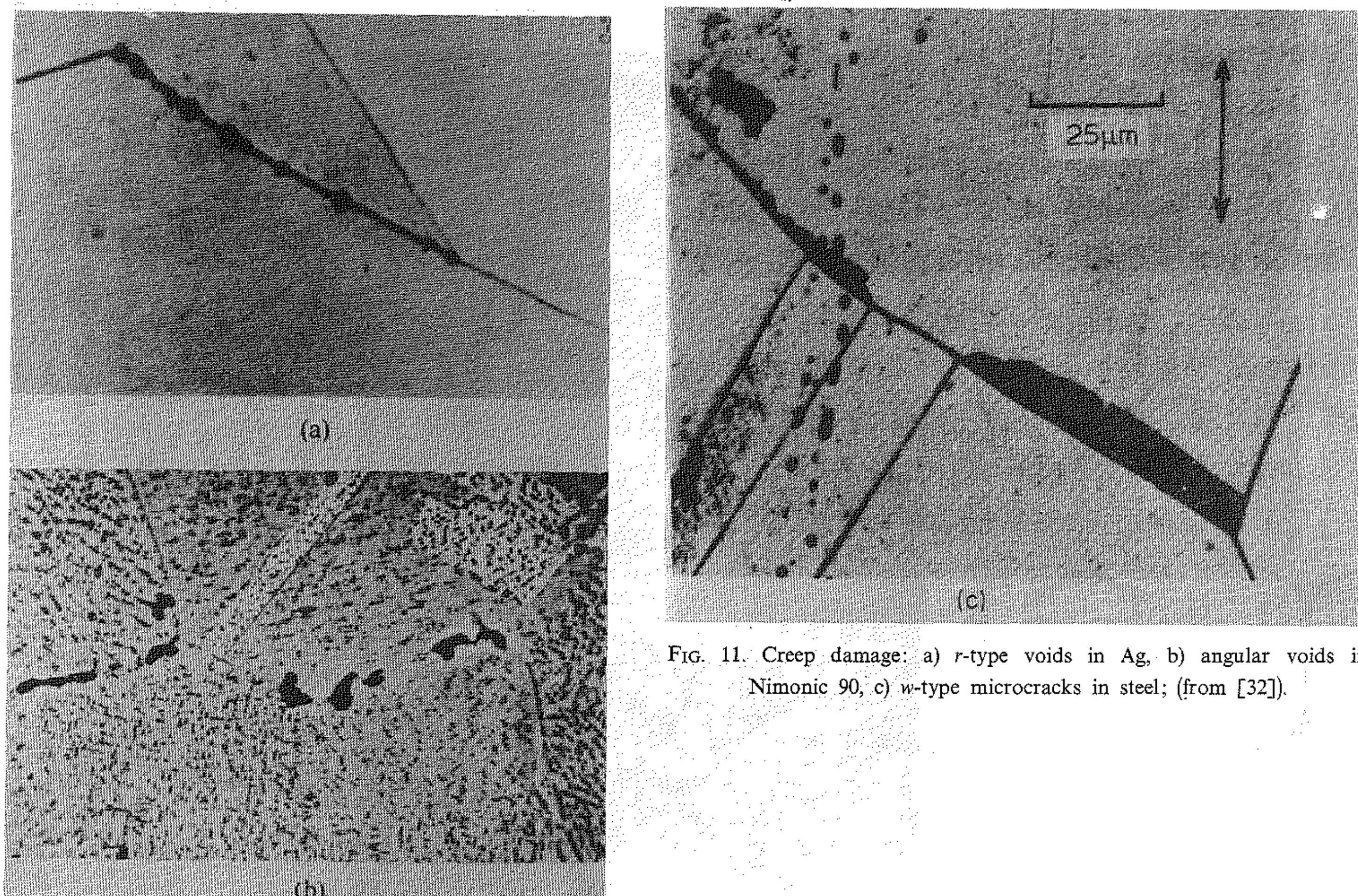


FIG. 11. Creep damage: a) r-type voids in Ag, b) angular voids in Nimonic 90, c) w-type microcracks in steel; (from [32]).

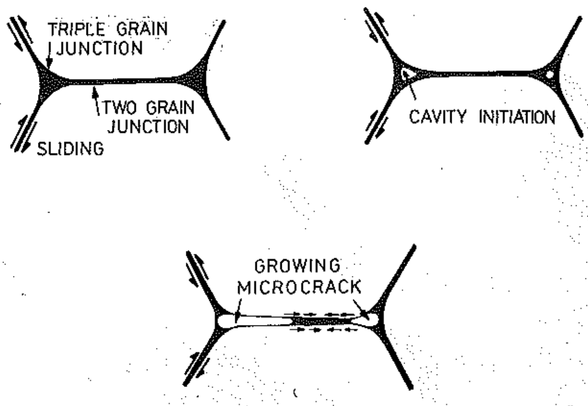


FIG. 12. Mechanism of microcrack initiation in ceramics (from [6]).

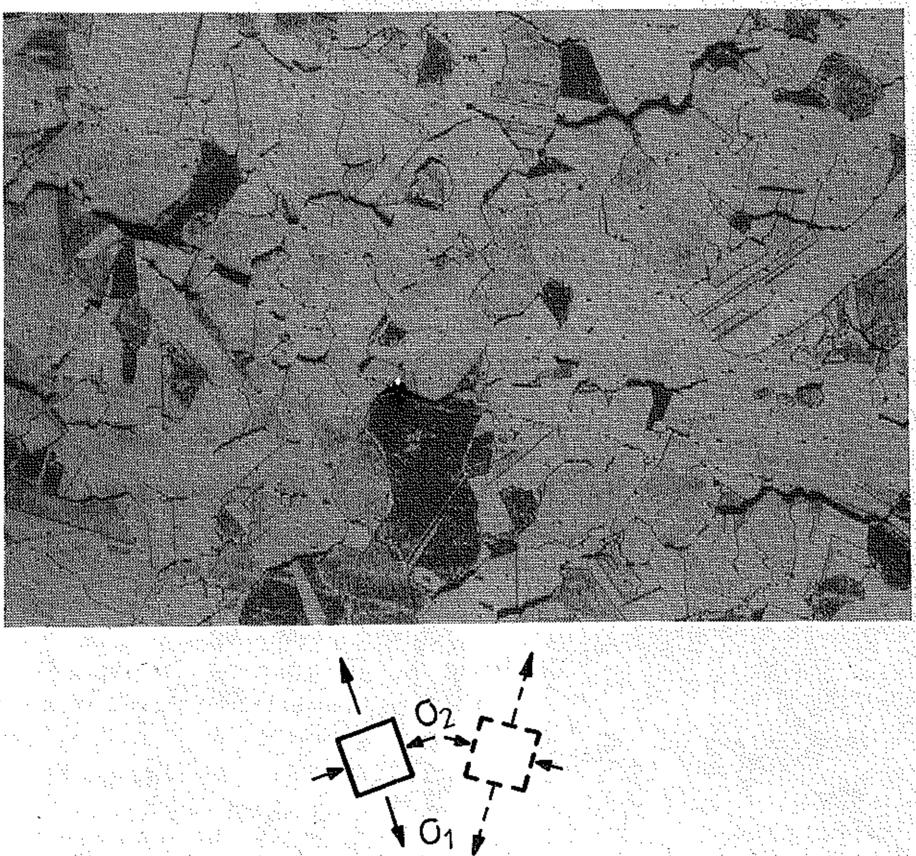


FIG. 13. Anisotropy of creep damage in copper (from [57]).

at triple junctions grow within the thin glass film between two grains and a microcrack is formed next as sketched in Fig. 12. After some growth along the two grain junction, the microcrack blunts, apparently when faced with an energy barrier produced by the adjacent grain, while new microcracks appear continuously on the other grain boundaries. Final fracture is caused by the linkage of blunted microcracks into a macrocrack [16].

It has already been stressed in this section that available experimental data reveal an anisotropic character of the creep damage process both in metals and ceramics. This feature of the creep damage may even better be seen from the creep tests carried out under non-proportional loading as noted by Murakami [15]. Dyson et al. [48] examined the grain boundary cavitation in Nimonic 80A under tension — reversed torsion. They found that solely those cavities which were located on the grain facets perpendicular to the tensile stress axis experienced a substantial growth in either state of stress. Similar experiments were performed by Trąmpczyński *et al.* [56, 57] on copper tubes acted upon by a constant tension and cyclic torsion. Independent developments of two families of microcracks on the planes corresponding to the maximum principal tensions were observed, Fig. 13. The oriented nature of creep damage has thereby been confirmed.

2.4. *Fatigue damage*

Under the action of cyclic loadings upon metals, microcracks can be initiated at free surfaces as a consequence of local plastic deformations. During the fatigue process, plastic extrusions and intrusions appear and the latter often transform into microcracks, Fig. 14. Experiments show that fatigue damage evolves in two essentially distinct stages. In stage I, the microcracks follow the slip bands, thus they tend to lie in the direction of maximum shear, i.e. at approximately 45 degrees to the stress axis. As the microcrack approaches a certain characteristic length a_T, a transition from stage I to stage II is observed. It has been noticed that a_T is structure dependent and is of the order of $10 \div 15$ grain diameters [62]. Stage II is identified with the propagation of the conventional fatigue macrocrack which grows at the right angle to the applied tensile stress direction, Fig. 15 [63, 64]. In the low-cycle fatigue, i.e. when the number of cycles to failure N_F does not exceed 10^4 [65], the stage II cracking plays a dominant role in the fatigue damage-fracture process. In the high-cycle fatigue ($N_F > 10^5$), a major part of the fatigue life is spent on the formation of intrusions and initiation of the stage I microcracks [16].

A typical feature observed in the stage II are the striations at the crack tip left by subsequent plastic deformations, Fig. 16. A mechanism governing the stage II crack growth is called "plastic blunting process", and

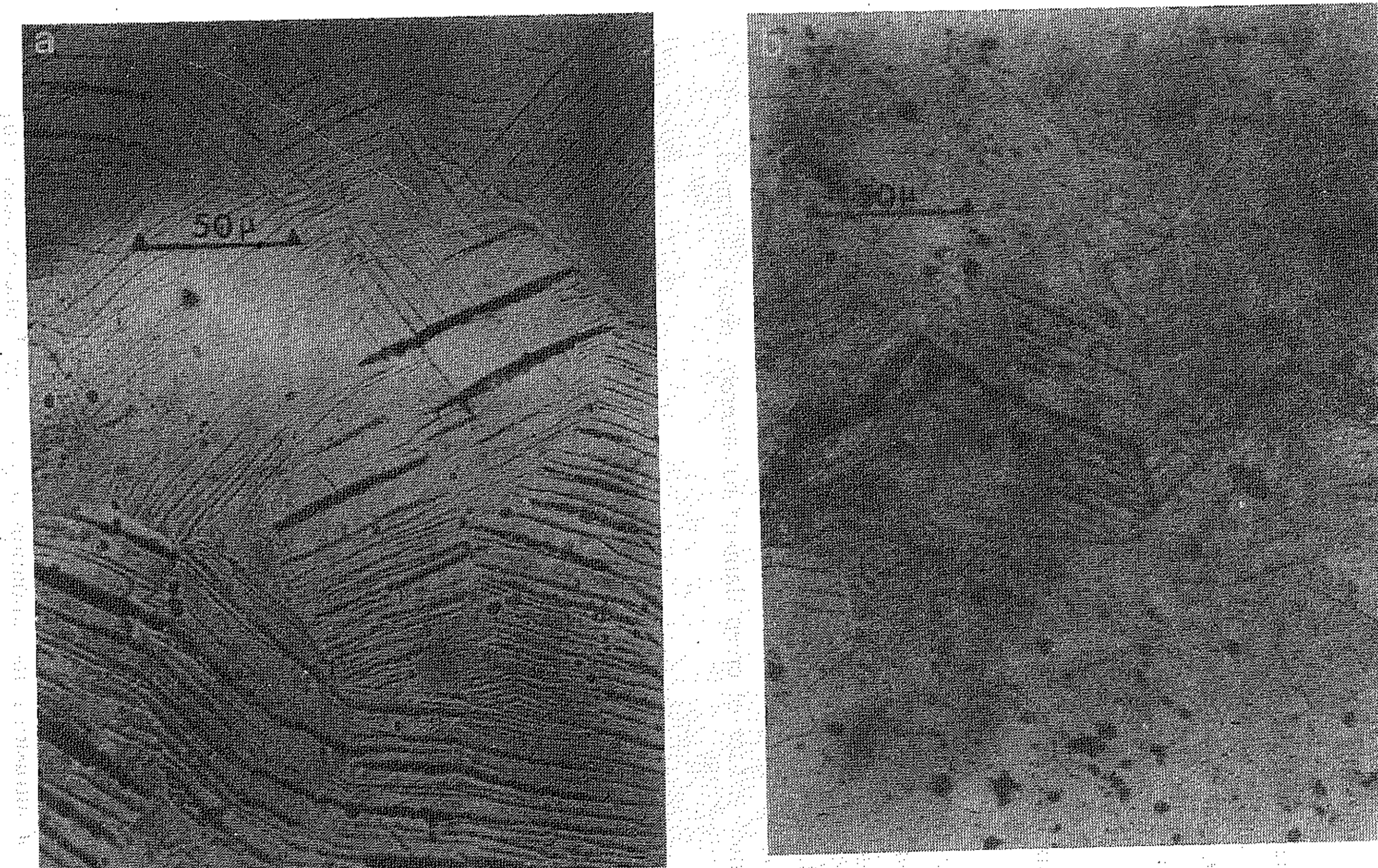


FIG. 14. Fatigue microcracks in Al-alloy; a) intrusions and extrusions, b) slip band microcrack-stage I; (from [61]).

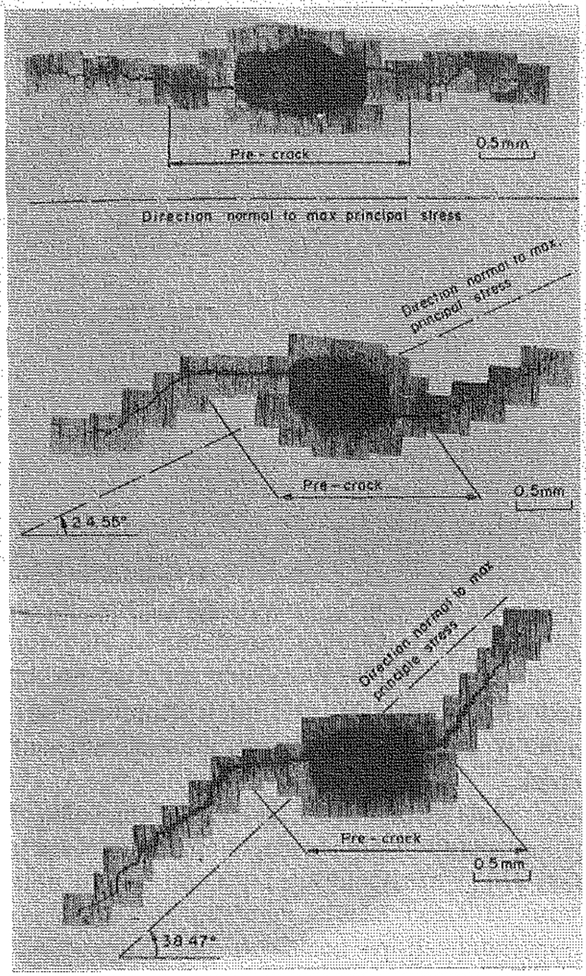


FIG. 15. Fatigue crack growth in stage II (from [65]).

FIG. 16. Fatigue strations in Al-Cu-Mg alloy (from [61]).

has been explained elsewhere [64]. Overall fracture in the fatigue process takes form of single or double shear planes after a sufficient growth of the stage II tensile crack.

Notches, inclusions and second-phase particles can also be the sources of fatigue damage, especially in high-strength materials since they are typical stress concentrators [61].

2.5. Dynamic damage

We shall now consider the problem of damage produced by the impulsive loads of high intensity and short duration, such as projectile impact, air-shock loading, explosion and alike. Dynamic loading of this kind induces a so-called spall damage which has some characteristic features as compared with damages caused by the quasi-static loads [66]. Spall damage is a process of gradual internal degradation of a material implied by the rarefaction waves that follow the compressive waves resulting from the impact. In general, it consists of the same basic phases as static damage, i.e. nucleation, growth and coalescence of microcracks or voids, while overall failure is realized by a complete separation of a target into disjoint elements. Ductile spall damage that takes form of roughly spherical voids, has been observed in copper, soft aluminium and tantalum [66]. Brittle spallation which is

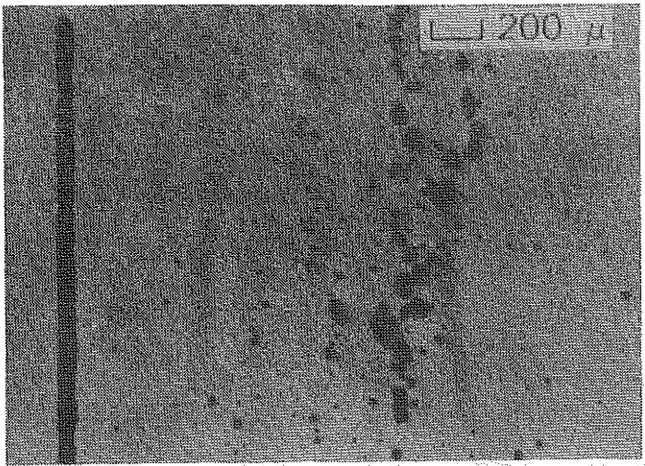


FIG. 17. Ductile spall damage in aluminium target (from [67]).

associated with the development of planar microcracks occurred in less ductile materials, such as Armco-iron, beryllium, polycarbonates [67, 68]. In one-dimensional wave propagation problems, the spall damage usually occupies a major part of the target volume. However, the concentration of both voids and mirocracks reaches a maximum value in narrow zones perpendicular to the impact direction and localized near the center of the target, due to intersection of several rarefaction tensile waves, Figs. 17 and 18, [67, 69].

In the case of ductile spalling, the voids enlarge, coalesce and form a macrocrack that runs thorugh the heavily damaged material leading to the full separation of the impacted specimen. In brittle materials, a process of fragmentation occurs as the microcracks begin to interact and link up. Fragments of various sizes are produced by the intersections of mirocracks having different lengths. Since the fragmentation necessarily involves some voided spaces among the separated parts, the stress in the effective cross--section must increase in order to sustain the external load. This in turn leads to the nucleation of more and more microcracks that form next new fragments, and the whole process reaches a self-accelerating cataclysmic stadium which ends by the complete disintegration of the material [67].

2.6. *Damage in composites*

Damage mechanisms in initially anisotropic materials such as composites are significantly different from those found in isotropic and homogeneous solids that were reviewed in the preceding sections. Faced with the overwhelming complexity of the damage phenomena in composites, we shall in this

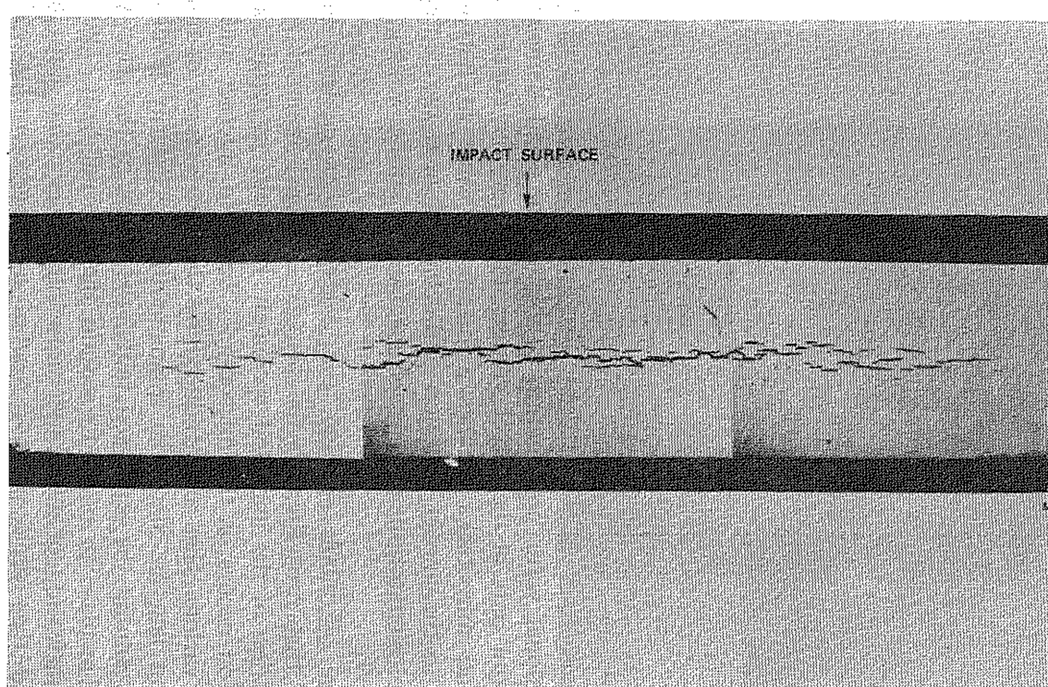


FIG. 18. Brittle spall damage in polycarbonate (from [68]).

section only touch the problem in question, addressing the prospective reader to the vast body of available literature in this field, e.g. [70, 71, 72, 73].

Composites may be assembled in various groups, depending on the classification criterion assumed. We may, for example, conceive: polymer-, metal- and ceramic-matrix composites. Also, articulate versus fibrous composites are distinguished. Another classification divides them into laminate and fiber composites. As an example of particular materials, the following can be mentioned: carbon-fiber reinforced epoxy resin, glass-fiber mat laminates, etc.

Due to the very nature of composites being a combination of two diverse constituents, new types of internal degradation appear in this class of materials. In composite laminates subjected to static tension, damage can develop according to three principal mechanisms: debonding along the fiber-matrix interface, local fracturing of the fibers resulting from the fiber microflaws growth, microcracking within the matrix. The first signs of damage usually take the form of separations between the filaments and the matrix in regions where the filaments are perpendicular to the load axis, Fig. 19 [74]. This behaviour is called debonding. Once it has taken place, the debonding damage is intensified by affecting other fibers inclined at smaller angles until, at some load, microcracking in the matrix appears. It was observed that

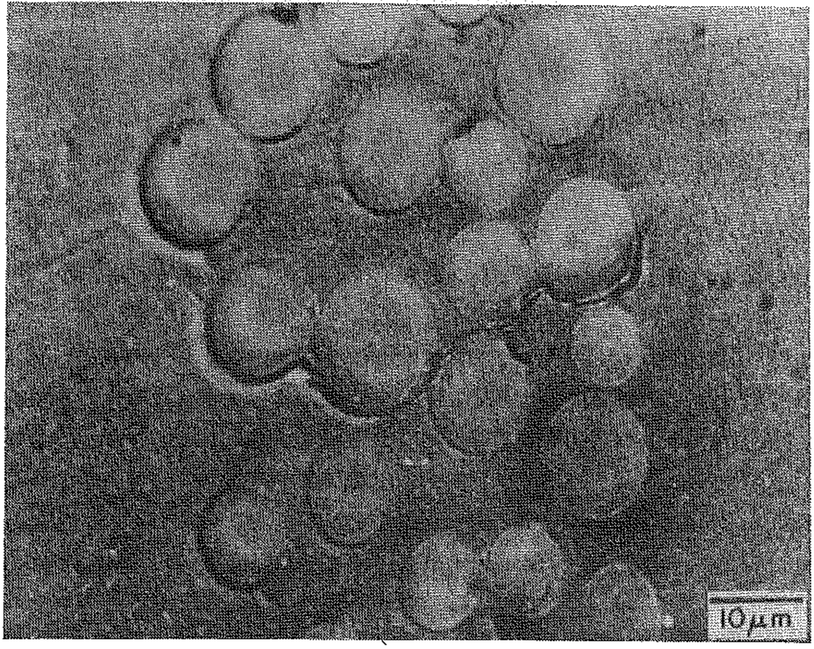


FIG. 19. Fibre debonding in mat-polyester resin laminate; tension axis-vertical (from [74]).

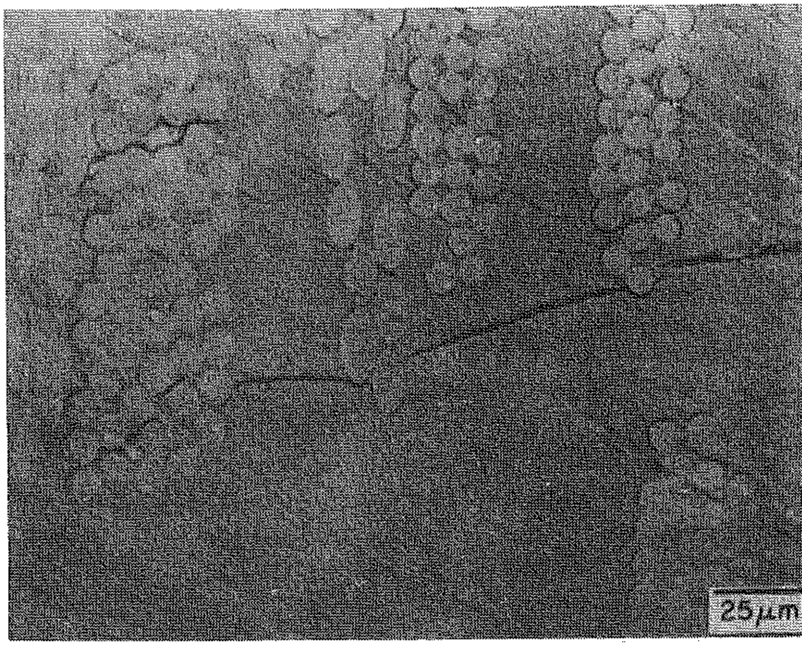


FIG. 20. Matrix microcracking in mat-polyester resin laminate; tensile load axis-vertical (from [74]).

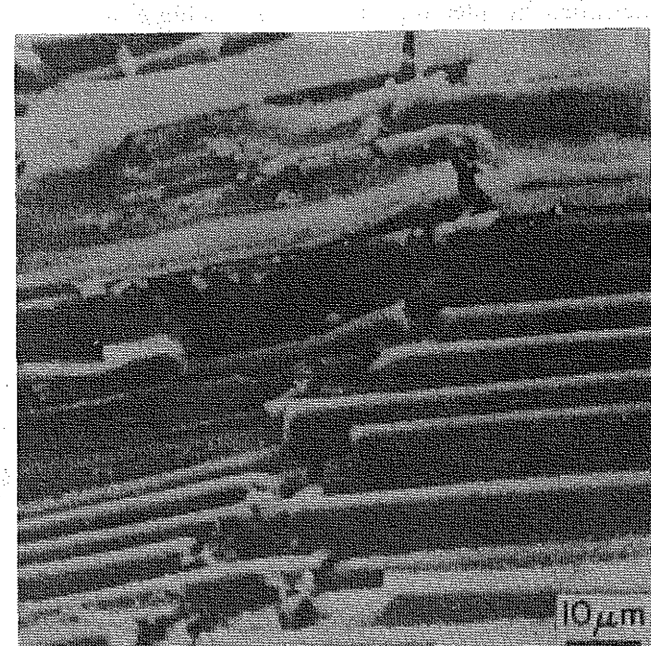
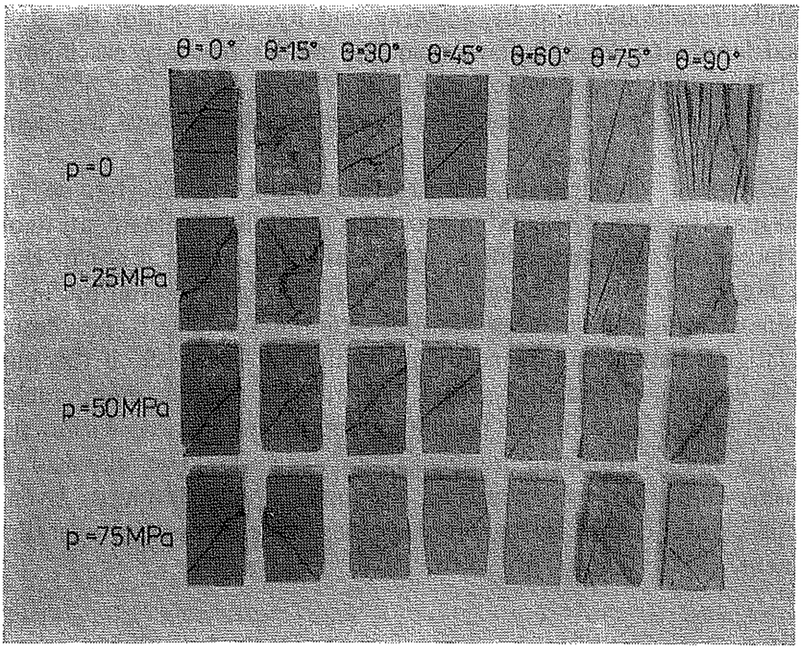


FIG. 21. Compression-induced damage in composite due to fiber buckling; load axis horizontal (from [74]).



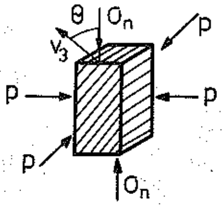


FIG. 22. Failure modes of glass-fiber/epoxy resin in compression test with confining pressure (from [75]).

in the resin-matrix laminates the matrix microcracking originated from the existing sites of debonding [74]. These microcracks were also predominantly perpendicular to the tensile load direction, Fig. 20. Debonding commenced at some 30% of the ultimate tensile strength while matrix microcracking occurred at about 70% of this strength. As the tensile load increased, the nucleated microcracks grew and formed a macrocrack that brought the specimen into final separation.

Under quasi-static compression, composites such as carbon-fiber reinforced plastics suffer a local buckling of fibers. It primarily occurs on the specimen surface and develops later on within the layers affecting more and more individual fibres. Fiber buckling produces a macroscopic crack that propagates normal to the compressive load axis, Fig. 21.

Just to get an idea of other possible mechanisms of the final fracture of composites under confined and unconfined compression, we recall the experimental results obtained by BOEHLER and RACLIN [75] for the glass-fiber reinforced laminates, Fig. 22. Two principal modes of the gross failure were observed: "parallel mode" which can readily by seen for $\theta = 45°$, $60°$, $75°$, and "across mode", generally noted for the remaining values of the inclination angle θ.

Consider now damage in composites induced by the fatigue process. In metal-matrix composites, fatigue cracks generally nucleate at two sites: free surfaces and filament-matrix interfaces [76]. It is known that free surfaces are the common sources of the fatigue crack initiation in metals. In certain composites, such as aluminium reinforced with beryllium, this type of fatigue damage commencement dominates. However, if the filaments are brittle enough, the second source prevails, i.e. fatigue microcracks may be initiated at the sites of the filament fracture and evolve within a metal thus ductile matrix to form a macrocrack, Fig. 23. The classical fatigue mechanisms is further recognized as the macrocrack develops. Eventually, composites exhibit dominant transverse planes of weakness and overall failure is imminent.

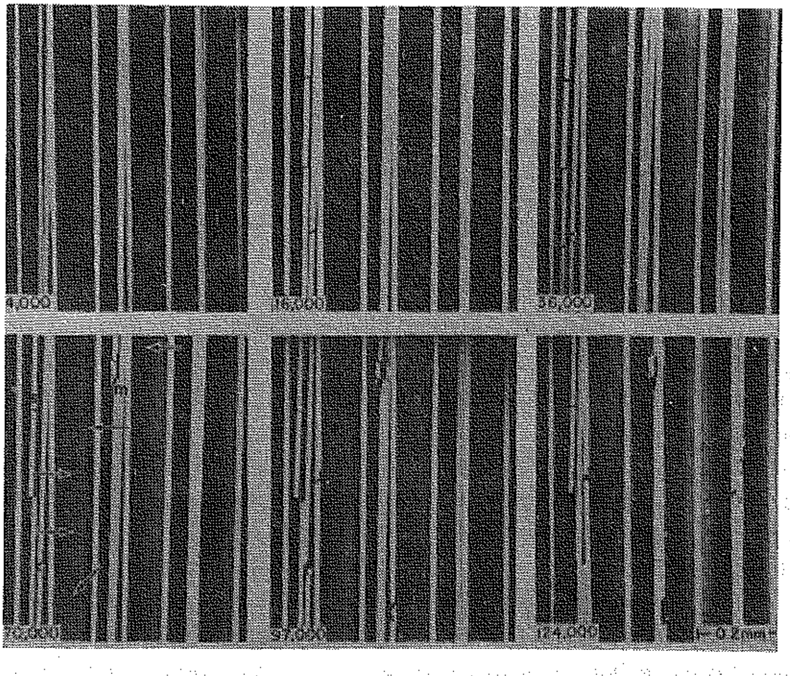


FIG. 23. Development of fatigue damage in aluminium reinforced with boron; inserted: number of cycles, single (s) and multiple (m) fractures of filament; arrows indicate fatigue macrocrack within matrix (from [77]).

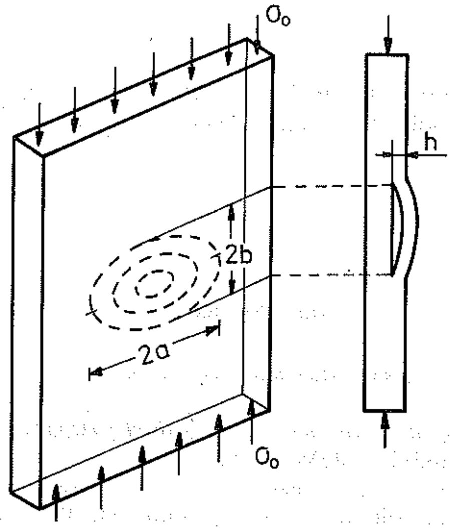


Fig. 24.

Another damage mechanism can be observed in composites exposed to dynamic loadings. For instance, a graphite-fiber reinforced epoxy laminate when subjected to a low-speed impact revealed damage mechanism that consisted in delamination with simultaneous local buckling of individual layers as sketched in Fig. 24 [77].

3. Remarks

Even a superficial review of the existing experimental data concerning the damage process in engineering materials reveals its complexity and diverse nature. One general conclusion comes out quite clearly: damage develops in all materials and all loading régimes. A strong dependence on the material mesostructure, type of applied load, state of stress, temperature and environmental conditions is observed. In the majority of materials, especially, those exhibiting some brittleness, a pronounced anisotropy of the microcrack nucleation and evolution is found.

The second part of this report will be devoted to the selected damage theories published so far. An appraisal will be attempted of the capability and limitations of these theories to record properly the mechanical behaviour of damaged solids.

Acknowledgement

This work was carried out during a short stay of the author at Institut de Mécanique de Grenoble. The hospitality and advice of Prof. J. P. Boehler are appreciated.

References

1. D. Krajčinovic and D. Sumarac, *Micromechanics of the damage processes*, CISM Lecture Notes, Udine 1986.
2. Yu. V. Zaitsev, *Modeling of deformation and strength of concrete using fracture mechanics methods* [in Russian], Stroiizdat, Moscow 1982.
3. P. Tapponier and W. F. Brace, *Development of stress-induced microcracks in Westerly granite*, Int. J. Rock Mech. Min. Sci. Geomech. Abstr., **13**, 103—112, 1976.
4. C. H. Scholz, *Microfracturing and the inelastic deformation of rock in compression*, J. Geophys. Res., **73**, 1417—1432, 1968.
5. L. S. Costin, *A microcrack model for the deformation and failure of brittle rock*, J. Geophys. Res., **88**, 9485—9492, 1983.
6. R. L. Tsai and R. Raj, *Crêep fracture in ceramics containing small amounts of a liquid phase*, Overview 18, Acta Metall., **30**, 1043—1058, 1982.
7. J. Lemaitre and J. L. Chaboche, *Aspect phenomenologique de la rupture par endommagement*, J. Méc. Appl., **2**, 317—365, 1978.
8. J. Hult, *Continuum damage mechanics: theory and applications*. Introduction and general overview, CISM Lecture Notes, Udine 1986.
9. A. Dragon and Z. Mróz, *A continuum model for plastic-brittle behaviour of rock and concrete*, Int. J. Engng. Sci., **17**, 121—137, 1979.
10. M. L. Kachanov, *On time to rupture in creep conditions* [in Russian], Izv. Ak. Nauk SSR, Otd. Tekhn. Nauk., No. 8, 26—31, 1958.
11. J. Janson and J. Hult, *Fracture mechanics and damage mechanics — a combined approach*, J. Méc. Appl., **1**, 69—84, 1977.
12. D. Krajcinovic, *Continuum damage mechanics*, Appl. Mech. Rev., **37**, 1—6, 1984.
13. M. Basista, *On continuum models of damage* [in Polish], IFTR Reports, 40, 1984.
14. J. C. Jaeger and N. G. W. Cook, *Fundamentals of rock mechanics*, Chapman and Hall, London 1979.
15. S. Murakami, *Anisotropic damage in metals*, in: Failure Criteria of Structured Media, J. P. Boehler [ed.], Balkema Publ., 1988.
16. S. Murakami, *Anisotropic aspects of material damage and applications of continuum damage mechanics*, CISM Lecture Notes, Udine 1986.
17. R. H. Evans and M. S. Marathe, *Microcracking and stress-strain curves for concrete in tension*, Matériaux et Constructions, **1**, 61—64, 1968.
18. J. G. Rots, P. Nauta, G. M. A. Kusters and J. Blaauwendraad, *Smeared crack approach and fracture localization in concrete*, Heron, **30**, No. 1, 1985.
19. M. Ortiz, *A constitutive theory for the inelastic behavior of concrete*, Mechanics of Materials, **4**, 67—93, 1985.
20. C. Fairhurst and N. G. W. Cook, *The phenomenon of rock splitting parallel to the direction of maximum compression in the neighbourhood of a surface*, Proc. 1st Congr. Int. Soc. Rock Mech., Lisbon, **1**, 687—692, 1966.

21. J. LEMAITRE and J. L. CHABOCHE, *Mécanique des materiaux solides*, Dunod, Paris 1985.
22. S. NEMAT-NASSER and H. HORII, *Compression-induced nonplanar crack extensions with application to splitting, exfoliation, and rockburst*, J. Geophys. Res., **87**, B8, 6805—6821, 1982.
23. H. HORII and S. NEMAT-NASSER, *Compression-induced microcrack growth in brittle solids: axial splitting and shear failure*, J. Geophys. Res., **90**, B4, 3105—3125, 1985.
24. H. HORII and S. NEMAT-NASSER, *Brittle failure in compression: splitting, faulting and brittle-ductile transition*, Phil. Trans. R. Soc. London, A **319**, 337—374, 1986.
25. W. F. BRACE and E. G. BOMBOLAKIS, *A note on brittle crack growth in compression*, J. Geophys. Res., **68**, 3709—3713, 1963.
26. E. HOEK, *Rock fracture under static stress conditions*, Council for Scientific and Industrial Research Report MEG 383, National Mechanical Engineering Research Institute, Pretoria, South Africa, 1965.
27. A. DRAGON, *On phenomenological description of rock-like materials with account of kinetics of brittle fracture*, Arch. Mech., **28**, 13—30, 1976.
28. J. RACLIN, *Sur l'orientation des fissures fragiles dans un milieu anisotrope*, C. R. Acad. Sc., Paris, t. 285, Série B, 345—348, 1977.
29. T. T. C. HSU, F. O. SLATE, G. M. STURMAN and G. WINTER, *Microcracking of plain concrete and the shape of the stress-strain curve*, J. Amer. Concr. Inst., **60**, 209—224, 1963.
30. YU. V. ZAITSEV, *Inelastic properties of solids with random cracks*, in: Z. P. BAZANT [ed.], William Prager Symposium on Mechanics of Geomaterials: Rocks, Concretes, Soils, preprints, Northwestern University, Evanston 1983.
31. M. F. ASHBY, *Micromechanisms of fracture in static and cyclic failure*, in: Fracture Mechanics, R. A. SMITH [ed.], Pergamon Press, Oxford, 1—27, 1979.
32. H. E. EVANS, *Mechanisms of creep fracture*, Elsevier 1984.
33. N. J. PETCH, *Metallographic aspects of fracture*, in: Fracture an Advanced Treatise, Vol. 1, H, LIEBOWITZ [ed.], Academic Press, New York 1968.
34. J. F. KNOTT, *Fundamentals of fracture mechanics*, Butterworth, London 1973.
35. F. A. MCLINTOCK, *Ductility*, p. 255, A.S.M., 1968.
36. G. LE ROY, J. D. EMBURY, G. EDWARD and M. F. ASHBY, *A model of ductile fracture based on the nucleation and growth of voids*, Acta Metall., **29**, 1509—1522, 1981.
37. S. H. GOODS and L. M. BROWN, *The nucleation of cavities by plastic deformation*, Overview 1, Acta Metall., **27**, 1—15, 1979.
38. J. M. JALINIER and J. H. SCHMITT, *Damage in sheet metal forming*, Part I, II, Acta Met., **30**, 1789—1809, 1982.
39. J. H. SCHMITT, J. M. JALINIER and B. BAUDELET, *Analysis of damage and its influence on the plastic properties of copper*, J. Mater. Sci., **16**, 95—101, 1981.
40. J. H. SCHMITT, R. ARGEMI, J. M. JALINIER and B. BAUDELET, *On the existence of initial damage in sheet metal*, J. Mater. Sci., **16**, 2004—2008, 1981.
41. J. DUFAILLY, J. LEMAITRE, J. M. JALINIER, J. H. SCHMITT and B. BAUDELET, *Determination of the relative density changes in the presence of high strain gradient*, J. Mater. Sci. Letters, **15**, 3162—3165, 1980.
42. K. E. PUTTICK, *Ductile fracture in metals*, Philosophical Magazine, **4**, 964—969, 1959.
43. H. C. ROGERS, *The tensile fracture of ductile metals*, AIME Trans., 218, 498—506, 1960.
44. V. F. ZACKAY, W. W. GERBERICH and E. R. PARKER, *Structural models of fracture*, in: Fracture — An Advanced Treatise, Vol. 1, H. LIEBOWITZ [ed.], Academic Press, New York 1968.
45. J. GURLAND, *Observations on the fracture of cementite particles in a spheroidized* 1.05% C *steel deformed at room temperature*, Acta Metall., **20**, 735—741, 1972.
46. J. GURLAND, *Fracture of metal — matrix particulate composites*, in: Composite Materials, Vol. 5: Fracture and Fatigue, L. J. BROUTMAN ed., Academic Press, New York 1974.

47. P. PERZYNA, *Internal state variable description of dynamic fracture of ductile solids*, Int. J. Solids and Structures, **22**, 797—818, 1986.
48. B. F. DYSON, M. S. LOVEDAY and M. J. RODGERS, *Grain boundary cavitation under various states of applied stress*, Proc. Roy. Soc. London, A **349**, 245—259, 1976.
49. B. F. DYSON and F. A. MCLEAN, *Creep of Nimonic* 80A *in torsion and tension*, Metal. Sci., **11**, 37—45, 1977.
50. D. R. HAYHURST and F. A. LECKIE, *The effect of creep constitutive and damage relationships upon the rupture time of solid circular torsion bar*, J. Mech. Phys. Solids, **21**, 431—446, 1973.
51. F. A. LECKIE and D. R. HAYHURST, *Constitutive equations for creep rupture*, Acta Metall., **25**, 1059—1070, 1977.
52. S. MURAKAMI and N. OHNO, *Creep damage analysis in thin-walled tubes*, in: Inelastic Behaviour of Pressure Vessels and Piping Components, T. Y. CHANG and E. KREMPL [eds.], PVP-PB-028, ASME, 55—69, New York 1978.
53. S. MURAKAMI and N. OHNO, *A continuum theory of creep and creep damage*, Proc. IUTAM Symp. Creep in Structures, Leicester U. K., 422—444, Springer 1981.
54. I. W. CHEN and A. S. ARGON, *Creep cavitation in* 304 *stainless steel*, Acta Metall., **29**, 1321—1333, 1981.
55. R. LAGNEBORG, *Creep: mechanisms and theories*, in: Creep and Fatigue in High Temperature Alloys, J. BRESSERS [ed.], 41—71, Elsevier Applied Science, London 1981.
56. W. A. TRĄMPCZYŃSKI and D. R. HAYHURST, *Creep deformation and rupture under non-proportional loading*, in: Creep in Structures, 388—404, Springer 1981.
57. W. A. TRĄMPCZYŃSKI, D. R. HAYHURST and F. A. LECKIE, *Creep rupture of copper and aluminium under non-proportional loading*, J. Mech. Phys. Solids, **29**, 353—374, 1981.
58. B. F. DYSON, *A unifying view on the kinetics of creep cavity growth*, in: Creep and Fracture of Engineering Materials and Structures, B. WILSHIRE and D. R. J. OWEN [eds.], Pineridge Press 1981.
59. F. GAROFALO, *Fundamentals of creep and creep-rupture in metals*, Macmillan, New York 1965.
60. S. MURAKAMI and T. IMAIZUMI, *Mechanical description of creep damage state and its experimental verification*, J. Méc. Thér. Appl., **1**, 743—761, 1982.
61. D. BROEK, *Elementary engineering fracture mechanics*, Noordhoff, Leyden 1974.
62. V. M. RADHAKRISHNAN and Y. MUTOH, *On fatigue crack growth in stage I*, in: The Behaviour of Short Fatigue Cracks, K. J. MILLER and E. R. DE LOSS RIOS [eds.], Mech. Eng. Publ., London 1986.
63. A. F. BLOM, A. HEDLUND, W. ZHAO, A. FATHULLA, B. WEISS and R. STICKLER, *Short fatigue crack growth behaviour in* Al 2024 *and* Al 7475, ibid.
64. M. RONAY, *Fatigue of high-strength materials*, in: Fracture — An Advanced Treatise, Vol. 3, H. LIEBOWITZ [ed.], Academic Press, New York 1971.
65. A. T. YOKOBORI jr., T. YOKOBORI, K. SATO and K. SYOJI, *Fatigue crack growth under mixed modes I and II*, Fatigue Fracture Engng. Mater. Struct., **8**, 315—325, 1985.
66. L. DAVISON and A. L. STEVENS, *Thermomechanical constitution of spalling elastic bodies*, J. Appl. Phys., **44**, 668—674, 1973.
67. L. SEAMAN, D. R. CURRAN and D. A. SHOCKEY, *Computational models for ductile and brittle fracture*, J. Appl. Phys. **47**, 4814—4826, 1976.
68. L. SEAMAN, D. R. CURRAN and D. A. SHOCKEY, *Development of a microfracture model for high rate tensile damage*, in: Creep and Fracture of Engineering Materials and Structures, B. WILSHIRE and D. R. J. OWEN [eds.], 345—364, Pineridge Press, Swansea 1981.
69. D. A. SHOCKEY, D. R. CURRAN, L. SEAMAN, J. T. ROSENBERG and C. F. PETERSEN, *Fracture of rock under dynamic loads*, Int. J. Rock Mech. Sci. Geomech. Abstr., **11**, p. 303, 1974.

70. L. J. Broutman [ed.], *Composite materials, Vol. 5. Fracture and fatigue*, Academic Press, New York 1974.
71. G. K. Sih and V. P. Tamuz [eds.], *Strength and fracture of composite materials* [in Russian], Zinatne, Riga 1983.
72. K. L. Reifsnider [ed.], *Damage in composite materials*, ASTM STP 775, New York 1982.
73. Z. Hashin and C. T. Herakovich [eds.], *Mechanics of composite materials*, Pergamon, 1983.
74. M. J. Owen, *Fatigue damage in glass-fiber reinforced plastics*, in: Fracture and Fatigue, L. J. Broutman [ed.], 314—339, Academic Press, New York 1974.
75. J. P. Boehler and J. Raclin, *Failure criteria for glass-fiber reinforced composites under confining pressure*, J. Struct. Mech., **13**, 371—393, 1985.
76. J. R. Hancock, *Fatigue of metal-matrix composites*, in: Fracture and Fatigue, L. J. Broutman [ed.], 371—414, Academic Press, New York 1974.
77. H. Chai, K. D. Babcock and W. G. Knauss, *Modeling of growth of delamination defect in composite plate subjected to impact load* [in Russian], in: Strength and Fracture of Composite Materials, 45—47, Zinatne, Riga 1983.

Streszczenie

MECHANIKA USZKODZEŃ: PODSTAWY EKSPERYMENTALNE

Praca jest syntetycznym przeglądem wyników doświadczeń nt. powstawania i rozwoju uszkodzeń w materiałach kruchych i ciągliwych przy różnych typach zewnętrznego obciążenia. Wykazano, że uszkodzenie jest zjawiskiem złożonym, bezpośrednio zależnym od stanu naprężenia, rodzaju przyłożonego obciążenia, budowy materiału, temperatury. Podkreślono, że uszkodzenie wywołuje anizotropię oraz niesprężyste zachowanie się materiałów, zwłaszcza kruchych. Przegląd wybranych teorii uszkodzenia będzie przedmiotem drugiej części pracy.

Резюме

МЕХАНИКА ПОВРЕЖДЕНИЙ: ЭКСПЕРИМЕНТАЛЬНЫЕ ОСНОВЫ

Работа является синтетическим обозрением результатов экспериментов, касающихся возникновения и развития повреждений в хрупких и тягучих материалах, при разных типах внешнего нагружения. Показано, что повреждение является сложным явлением, непосредственно зависящим от напряженного состояния, рода приложенного нагружения, строения материала, температуры. Подчеркивается, что повреждение вызывает анизотропию и неупругое поведение материалов, особенно хрупких. Обозрение избранных теорий повреждения будет предметом второй части работы.

POLISH ACADEMY OF SCIENCES
INSTITUTE OF FUNDAMENTAL TECHNOLOGICAL RESEARCH.

Received November 18, 1987.