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Indentation analysis:
applications in the strength and wear of brittle materials

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SUMMARY

Some recent developments in the principles and applications of indentation fracture in brittle materials are surveyed. Attention is focused on 'sharp' indenters, for which precursor 'plasticity' is an essential element of the crack development. A major consequence of this plasticity is a residual contact stress field which exerts a dominant influence on ensuing mechanical behaviour. This influence is discussed in relation to strength and wear properties of brittle ceramics. Emphasis is placed on the advantages of the indentation method as a means of producing controlled cracks for evaluating material fracture parameters and for gaining insight into flaw micromechanics.

1. INTRODUCTION

Solids with large components of covalency and ionicity in their atomic bonding (ceramics, glasses, semiconductors) tend to be highly brittle at room temperature. The key to engineering design with this class of materials is thus the containment of crack growth. Nowhere is this susceptibility to fracture more apparent than in contact phenomena; for example, the inadvertent impingement of a single dust particle on to the pristine surface of an optical fibre can lead to a strength degradation of more than an order of magnitude. It is in this context that the widely expanding discipline of 'indentation fracture mechanics' (Lawn & Wilshaw, 1975) may be seen as establishing a basic scientific framework for analysing a broad range of strength- and wear-related properties.

In this paper we shall outline some of the more recent developments in indentation analysis, with particular emphasis on the microscopical aspects. Our discussion will focus on the damage patterns produced in ideally 'sharp' (i.e. plastic) as distinct from 'blunt' (elastic, e.g. Hertzian) contact. The former, quite apart from their greater topicality, offer certain advantages as a tool for materials testing: the patterns contain information on the modes of deformation, as well as on the fracture; since indentation fracture can be produced on the most pristine of surfaces above some threshold in the loading, the processes of crack nucleation, as well as propagation, can be investigated; the patterns are subject to a high degree of control in their geometry, scale and placement, and may be quantified entirely in terms of characteristic surface dimensions in conjunction with the contact load; the nature of the damage relates more closely to that which pertains in 'real' strength degradation and wear mechanisms under typical component fabrication and service conditions.

In our survey we shall first outline the mechanics of crack evolution during an indentation cycle, and then describe certain applications to practical properties. For the first of these, the

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vital interrelation between deformation and fracture processes will be a central theme. The most important manifestation of this interrelationship is the residual driving force exerted by the deformation zone on the cracks. For the applications, strength properties of pre-indented test pieces are considered, both for the insight they provide into the general micromechanics of flaws and for the element of control that may be exercised in the evaluation of material fracture parameters. A brief description of erosive wear properties is also included.

2. INDENTATION PATTERN IN SHARP CONTACT

2.1. *General features*

The indentation fracture patterns produced by sharp indenters (e.g. Vickers, Knoop) have been well characterized (Lawn & Swain, 1975; Evans & Wilshaw, 1976; Swain & Hagan, 1976; Hagan & Swain, 1978; Puttick, 1978; Arora *et al.*, 1979; Marshall & Lawn, 1979; Lawn *et al.*, 1980a). The general features of this type of pattern are shown in the schematic of Fig. 1. The indenter is loaded on to the surface at normal force P (or, equivalently, at incident kinetic energy U_K in impact loading). Immediately beneath the contact the material deforms irreversibly, giving rise to a 'plastic' enclave within the elastic half-space surround matrix; the scale of this deformation is quantified by the characteristic dimensions a and b of the hardness impression and plastic zone, respectively. Two types of crack, both with essentially penny-like geometry, grow out of the deformation zone: median/radial cracks, on planes defined by contact normal and impression diagonal, characterized by dimension c_r (Lawn *et al.*, 1980a); lateral cracks, closely parallel to the specimen surface and centred on the base of the deformation zone, characterized by dimension c_l (Marshall *et al.*, 1982). The object of the fracture mechanics approach to the contact problem is to determine relations between the quantities defined in Fig. 1 and appropriate material parameters such as toughness, K_c, hardness, H, and Young's modulus, E.

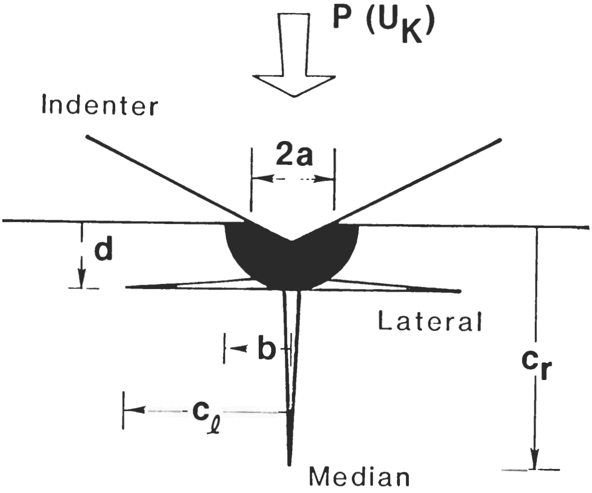

Fig. 1. Schematic of deformation/fracture pattern associated with a sharp indenter.

Important modifications to the axial symmetry of the indentation pattern can ensue if a tangential component is added to the loading, as in sliding or rolling, or oblique impact. Some of these modifications are discussed in other papers in this volume.

2.2. *Deformation zone*

The deformation that occurs in highly brittle materials (especially those with a large component of covalent bonding) is typified by relatively large ratios of hardness (defined here as the mean contact pressure) to modulus, H/E. In silicon, for instance, $H/E \approx 0{\cdot}08$, implying stress levels approaching the theoretical limiting strength of the structure. At these levels the conventional descriptions of slip and yield by classical dislocation glide must be regarded with caution. Hill & Rowcliffe (1974) prefer to consider the deformation process in silicon in terms of 'block' slip, in which shear takes place homogeneously and catastrophically across surfaces of maximum shear stress. A feature of this mode is that it does not depend critically on the existence of crystallographically favourable easy-glide planes. Accordingly, the concept is consistent with the recent identification of shear lines in indented soda-lime glass by Hagan (1980).

An important manifestation of such high-stress modes of deformation is the strong localization about the contact site (i.e. $b \gtrsim a$ in the notation of Fig. 1). Materials in this category tend to show strong depth recovery at the unloaded impression, due to 'elastic springback' (Lawn & Howes, 1982). However, this recovery is never complete, in which case there exists a state of residual stress in the material. The intensity of such residual fields can be high, as is readily apparent from the TEM photograph of an indentation site in silicion (Fig. 2). (These fields are perhaps more commonly observed by virtue of their attendant birefringence patterns in transmission optical microscopy—see Fig. 4 later.) As we shall demonstrate, residual stresses can play a crucial role in the subsequent mechanical response of indented surfaces.

At sufficiently high rates of loading dynamic effects can become important (Evans & Wilshaw, 1977). Indeed, under impact conditions the plastic work rate can be so intense as to cause surface melting, in even the most refractory of ceramics (Lawn *et al.*, 1980b). Figure 3 shows evidence of this in glass and alumina impacted with silicon carbide particles.

Quantitative analysis of the deformation zone is a complex problem in solid mechanics. Elastic/plastic models based on the notion of an expanding internal cavity under pressure

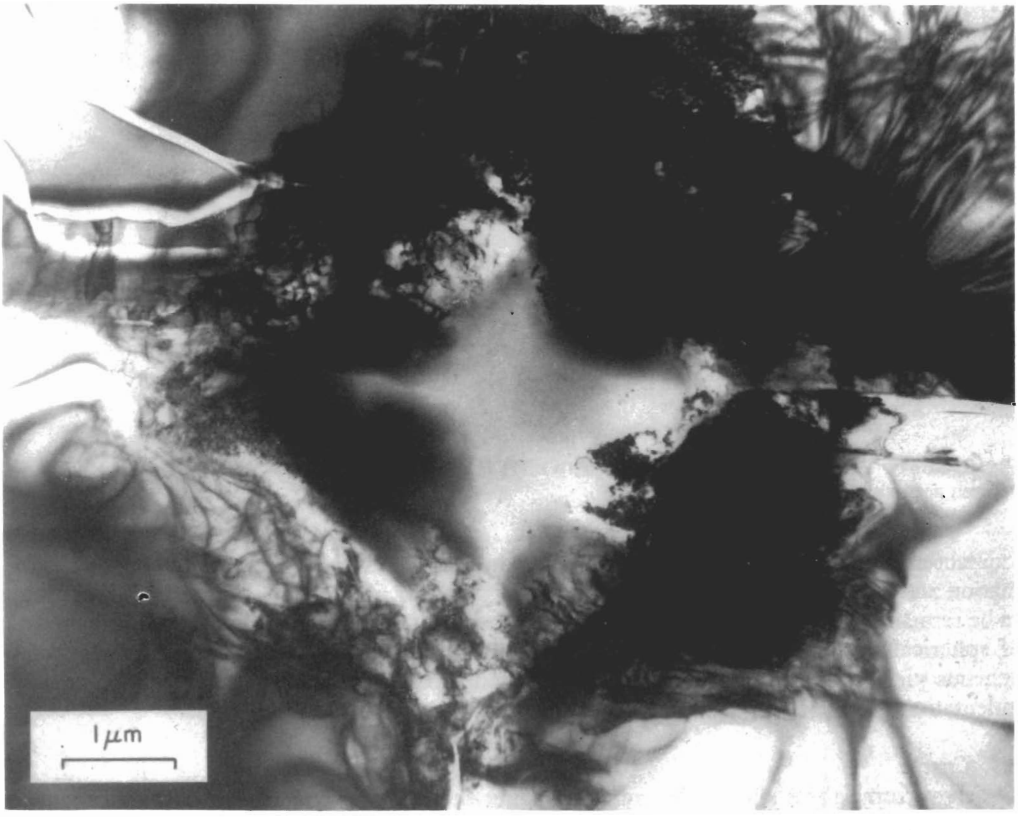

Fig. 2. TEM of foil section in Vickers-indented silicon. Plane of specimen surface and of foil is (110). Note confinement of deformation about contact area.

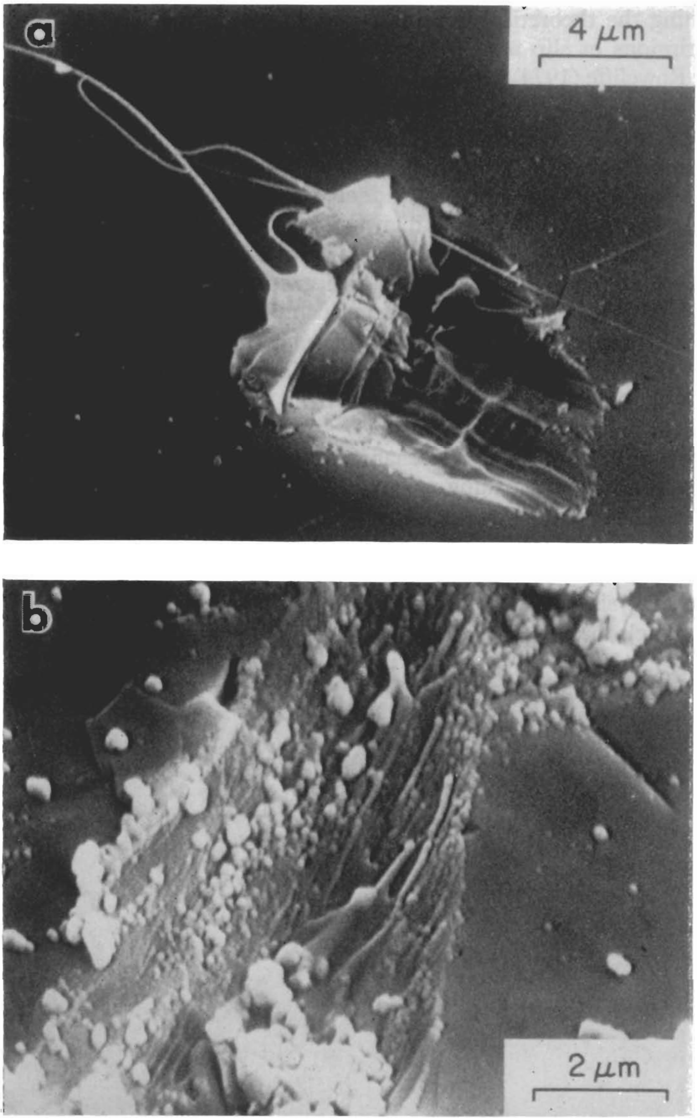

Fig. 3. SEM views of surface of (a) glass, (b) Al_2O_3 (H.P.), impacted with 150 μm SiC particles at 90 m s^{-1}. Molten zones are evident.

(contact area), with immediate plastic and remote elastic surround volumes (embedded deformation zone), provide the most common starting point for obtaining a solution. It needs always to be remembered, however, that such models, based as they are on the underlying assumptions of spherical symmetry in the deformation geometry and a well-defined, continuous, homogeneous yield process, must inevitably be limited in their capacity to represent the inelastic indentation response of brittle solids.

2.3. *Cracks*

Above some threshold in the contact loading the crack system of Fig. 1 initiates spontaneously. Once developed, the radial and lateral cracks become highly stable. The physical processes actually responsible for the initiation remain somewhat obscure, although there is growing evidence that the shear modes referred to in the previous subsection provide the essential crack nuclei (Hagan, 1980; Dabbs & Lawn, 1982).

Although the regular penny-like geometry of the well-developed cracks would seem to suggest a reasonably straightforward growth history through the contact cycle, direct observations reveal that this is far from so. Figure 4, a sequence of subsurface views of Vickers indentation in glass, illustrates the point. It is clear from this sequence that a significant portion of the radial crack evolution occurs during the unloading half-cycle (Marshall & Lawn, 1979). An important clue to such behaviour is the intense residual birefringence which remains about the contact site. It is the irreversible component of the stress field which drives the cracks outward into their ultimate circular symmetry; the reversible component is actually compressive over much of the prospective crack area, and accordingly acts as a constraint to growth while the load is applied (Lawn *et al.*, 1980a). The identification of the residual field as the critical component in the crack driving force is reinforced by the observation that the crack system can continue to expand (in the presence of a moist environment) long after the indentation cycle has been completed.

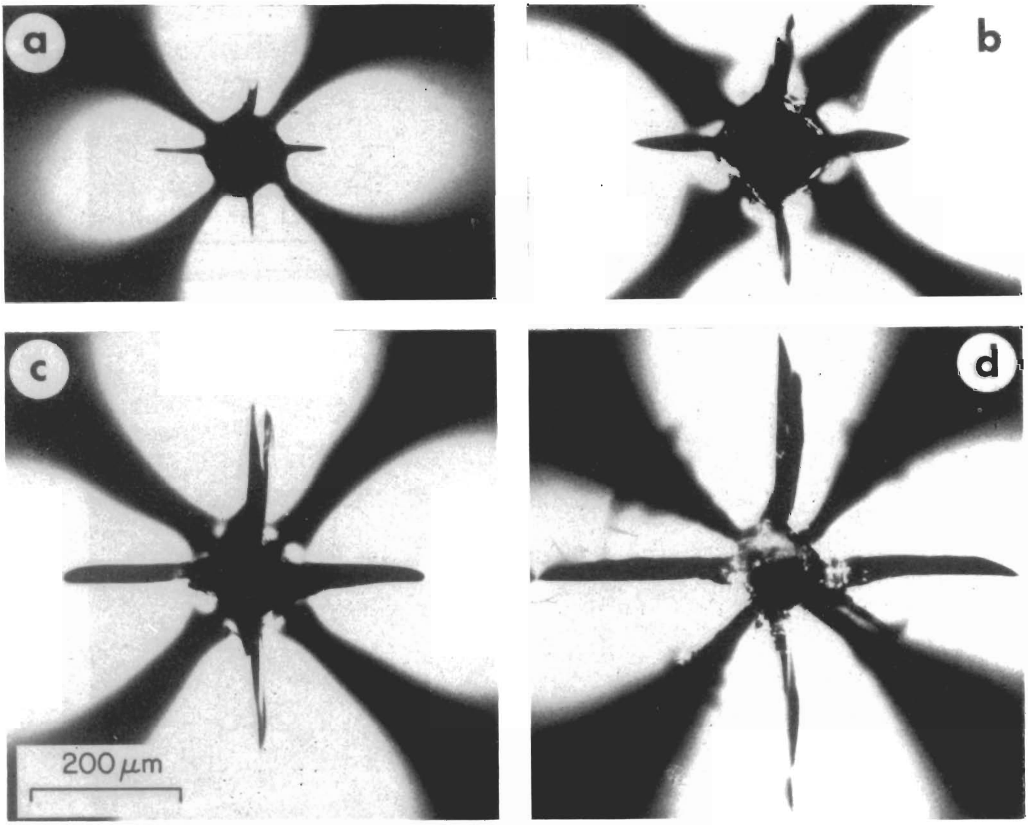

Fig. 4. Subsurface view, in crossed polars, of Vickers indentation in soda-lime glass in inert environment (N_2 gas). Showing full cycle, at loading (a) 47 N and (b) 90 N, and unloading (c) 30 N and (d) 0. Lateral cracks faintly visible in (d).

A closer look at the tip and interface regions of the indentation cracks reveals that the plasticity elements so critical to the initiation processes play absolutely no role in the micromechanisms of propagation; the brittle crack is atomically sharp, and extends in strict accordance with the classical picture of sequential bond rupture (Lawn *et al.*, 1980c; Lawn, 1983).

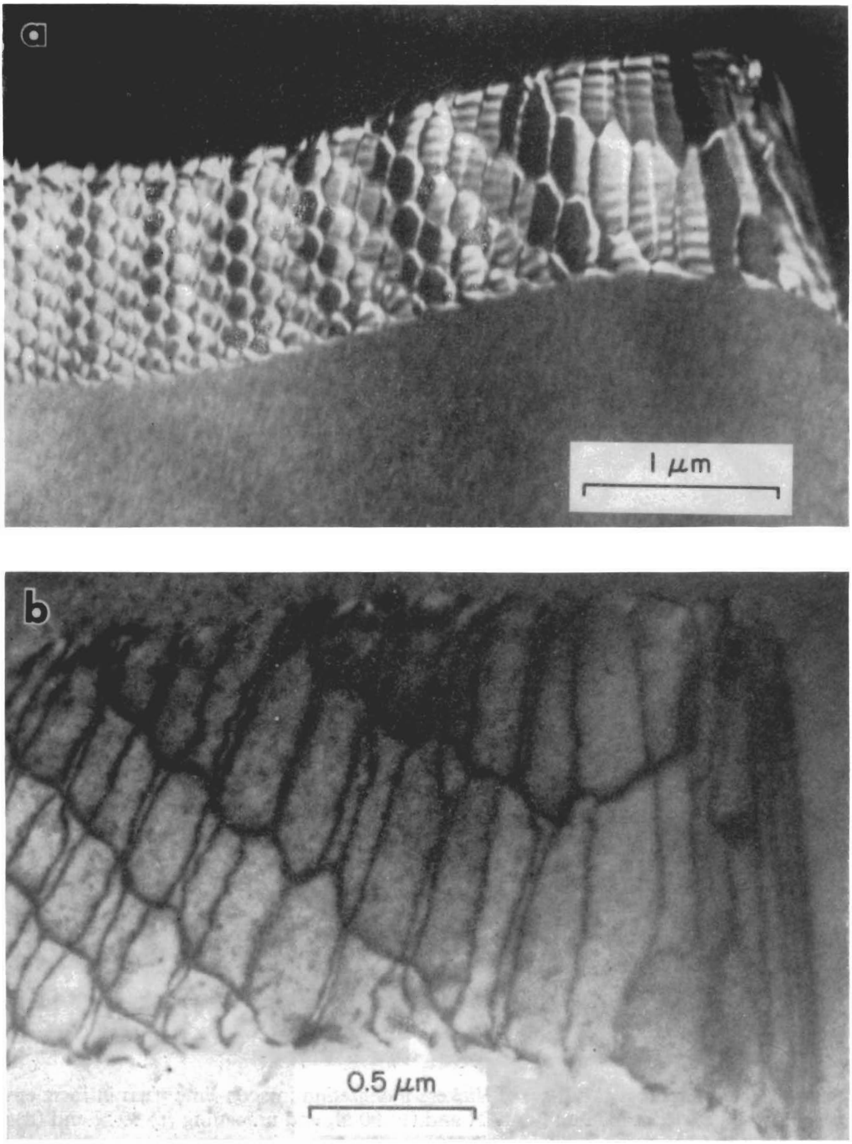

Fig. 5. TEM of ribbon segment of radial crack in sapphire: (a) dark field and (b) bright field conditions. Note misfit dislocation and stacking fault contrast at interface, absence of crack-tip plasticity.

Figure 5, showing TEM micrographs of a radial crack segment in an aluminium oxide foil, is a typical example of the configurations observed. There is clear evidence of network dislocations at the crack, but these are readily demonstrated to be associated with lattice misfit at a healed interface (Hockey, 1982). No slip dislocations are observed to emit from the crack tip. These observations are important because they establish the basis for theoretical descriptions of fundamental crack growth laws (Lawn, 1983).

The symmetry of the fully grown indentation cracks allows for straightforward fracture mechanics analysis. Accordingly, the residual driving force acting on the cracks may be quantified in terms of an appropriate stress intensity factor (Marshall & Lawn, 1979; Fuller *et al.*, 1982);