INDENTATION CRACKING OF BRITTLE COATINGS

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ABSTRACT

Cracking due to indentation of a thin, brittle coating on an elastic–plastic substrate is analysed. Solutions for indentation of elastic–plastic materials are coupled to solutions for channel cracking of thin coatings. As a result a relationship is obtained between spacing and length of radial cracks in the coating around the indentation.

KEYWORDS

Indentation, coated materials, residual stresses, coating strength.

INTRODUCTION

For years, hardness measurements based on various indentation methods have been used for extracting material information for metals. As it has become common to improve the hardness, the corrosion resistance *etc.* of metals by depositing thin coatings on the surface of the metal, it is of interest to explore to what extend the well–known and convenient indentation tests can be used for extracting information on such systems.

The elastic properties of coatings and layered materials, possibly with delaminations, can be extracted from the unloading process following indentation by load measurement and depth–sensing (*e.g.* Doerner and Nix, 1986 and Gao *et al.*, 1992).

The interface fracture resistance in a system where a diamond coating was bonded to a titanium substrate was measured in Drory and Hutchinson (1993, 1996). Upon indentation in such systems it was observed that failure took place by interface decohesion. As the interface fracture resistance increases compared to the fracture resistance of the coating, cracking of the coating will be observed rather than interface decohesion.

The present paper concerns radial cracking of a coating at indentation diameters which are large compared to the coating thickness. A specific example considered is a hardened steel substrate with a TiAlN coating of thickness 2.3 μm . This is an example of a system where the

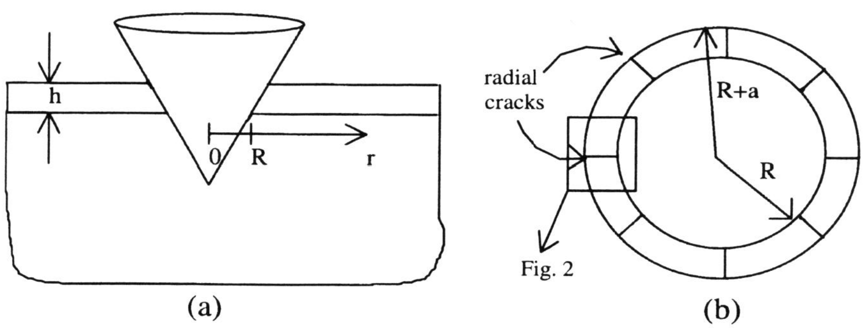

Fig. 1. Geometry of system analysed; axisymmetric indentation of a ductile substrate with a hard, brittle coating, (a) side view, (b) top view.

bonding is so good that interface decohesion does not take place to a significant extend, but, instead, a number of radial cracks through the coating spread out from the indenter. In Fig. 1 the geometry is sketched; R is the indentation radius, h is the thickness of the coating, a is the length of the radial cracks and N is the number of radial cracks so that $L = 2\pi R/N$ is the crack spacing at the edge of the indent. Fig. 2 shows a SEM of a detail of the crack pattern in the indented steel/TiAlN system as indicated in Fig. 1.

Fig. 2. SEM microscopy of a detail of the crack pattern in a TiAlN coating on a steel substrate as indicated in Fig. 1.

SUBSTRATE DEFORMATION

It is assumed that the indenter and the state of deformation in the substrate is axisymmetric throughout the process of indentation with the geometry of the indenter taken to be spherical (Brinnel indentation) or conical (Rockwell C indentation). Non–symmetric states of deformation and stress in a debonded diamond coating initially bonded to a titanium substrate were observed in Drory and Hutchinson (1993, 1996). The non–symmetry was caused by compressive stresses in the debonded coating resulting in coating buckling. This type of instability was analysed in Jensen and Thouless (1995) and Thouless *et al.* (1994).

The effect of the coating on the substrate deformation is neglected which requires that the indentation depth is considerably larger than the thickness of the coating. Non–linear behaviour of the substrate is thus allowed for.

The stress component in the coating which is of interest for analysing radial cracks is the circumferential stress, $\sigma_{\theta\theta}(r)$. This stress component can be evaluated as the sum of the stresses caused by the surface displacements of the substrate as a result of the indentation, and the equi–biaxial residual stresses, σ_0 . By the assumption $h << R$,

$$\sigma_{\theta\theta}(r) = \sigma_0 + \frac{E}{1 - \nu^2}\left[\frac{u(r)}{r} + \nu\frac{du(r)}{dr}\right] \tag{1}$$

where $u(r)$ denotes the radial displacement of the substrate at the interface between the substrate and the coating which is a function of the distance, r , from the line of axisymmetry. The function $u(r)$ is evaluated numerically in Hill *et al.* (1989) in the case of Brinell indentation and in Bower *et al.* (1992) for various indenter geometries. In (1), E and ν are the Young's modulus and Poisson's ratio of the coating, respectively, taken here to be identical to the values for the substrate, and with $\nu = 0.3$ in the following.

A fit of the radial surface displacement of the substrate was provided in Drory and Hutchinson (1996) on the form

$$\ln(u/R) = b_0 + b_1(r/R) + b_2(r/R)^2 + b_3(r/R)^3 \tag{2}$$

with the coefficients $b_0,..b_3$ tabulated for various values of yield stress, σ_y , and hardening index, n , of the substrate in a Ramberg–Osgood true stress–logarithmic strain curve in uniaxial tension $(\tilde{\sigma}, \tilde{\epsilon})$

$$\tilde{\epsilon} = \frac{\tilde{\sigma}}{E} + \frac{3}{7}\frac{\sigma_y}{E}\left(\frac{\tilde{\sigma}}{\sigma_y}\right)^n \tag{3}$$

The approximation (2) was believed to be most accurate for $r > 2R$ in Drory and Hutchinson (1996), but is used here also for $r < 2R$ as a closed form expression for $u(r)$ is convenient.

The numerical results of Hill *et al.* (1989) for the surface displacement in case of a spherical indenter were fitted by the expression

$$\frac{u}{d} = \frac{b_0 + b_1(r/R) + b_2(r/R)^2}{b_3 + b_4(r/R) + b_5(r/R)^2} \tag{4}$$

for $r < 2R$ with $(b_0, ..b_5)$ given by $(-.702, 2.212, -.776, -12.4, 16, -8)$ for a uniaxial behaviour of the substrate of the type $\bar{\varepsilon} \propto \bar{\sigma}^n$ with $n = 4$. Here, d is the indentation depth.

CHANNEL CRACKING OF BRITTLE COATINGS

A brittle coating under a homogeneous tensile stress state may fail by the spreading of channel cracks (Hutchinson and Suo, 1992). This is a failure mode where parallel cracks through the coating with a spacing, L, propagates in a steady state manner. If $G(L, h)$ denotes the energy release rate of a such crack, the energy release rate required for steady state propagation, G_{ss}, is given by the smallest of the values (Thouless, 1990; Delannay and Warren, 1991 and Hutchinson and Suo, 1992)

$$G_{ss}h = \int_0^h G(L, \bar{h})d\bar{h} \tag{5a}$$

$$G_{ss}h = 2\int_0^h G(L, \bar{h})d\bar{h} - \int_0^h G(2L, \bar{h})d\bar{h} \tag{5b}$$

corresponding to whether all channel cracks extend together (5a), or whether new cracks nucleate between already propagated cracks (5b). The function $G(L, h)$ was given in Benthem and Koiter (1972) for identical elastic properties of the coating and the substrate (see also Murakami, 1986). In Beuth (1992) the effect of elastic mismatch between the coating and the substrate on the limit values $G(\infty, h)$ was studied, and in Hu and Evans (1989) and Beuth and Klingbeil (1996) the effect of plastic yielding in the substrate on $G(\infty, h)$ was studied. These limit values correspond to the energy release rate of an isolated crack in a bimaterial system of infinite extend.

In Fig. 3 the variation of normalised crack density is shown as a function of the largest principal tensile stress, σ, in the coating which will be perpendicular to the crack faces. The figure is drawn for identical elastic properties of the substrate and the coating, and small scale yielding of the substrate is assumed. The fracture criterion

$$G_{ss} = \Gamma_f \tag{6}$$

is imposed where Γ_f should be regarded as the toughness of the coating. Within a precision of 0.5%, the curve in Fig. 3 can be fitted by the expression

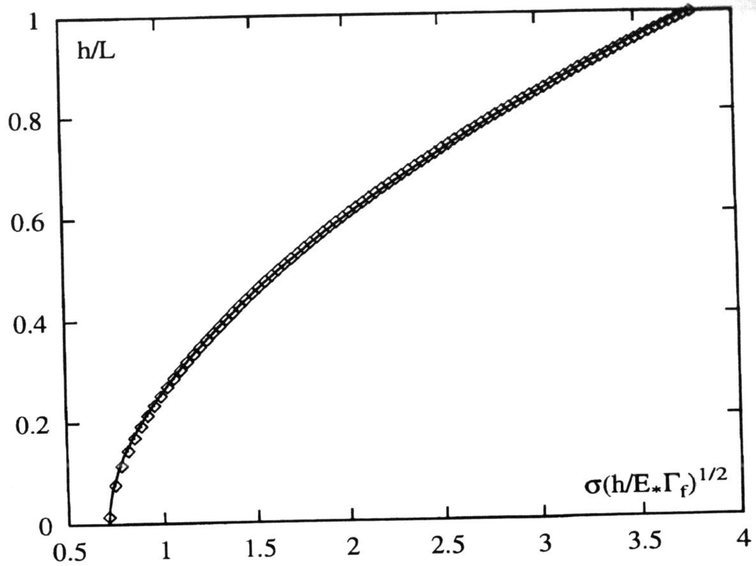

Fig. 3. Crack density *vs.* principal stress, σ. The solid line denotes results based on a numerical evaluation of (5b) while the points denote the fit given by (7). Here, $E_* = E/(1-\nu^2)$.

$$\frac{h}{L} = .52\left[\sigma\sqrt{\frac{h(1-\nu^2)}{E\Gamma_f}} - .71\right]^{.58} \tag{7}$$

which is included in the figure for comparison. This closed form expression will be useful in the following. By Fig. 3, channel cracks are arrested when the stress satisfies the threshold condition

$$f\left[\sigma\sqrt{\frac{h(1-\nu^2)}{E\Gamma_f}}\right] \equiv \frac{h}{L} = 0 \tag{8}$$

This condition will be used to identify the circumferential stress in the coating at the radius where the radial cracks arrest.

INDENTATION CRACKING

The results of the previous section are now applied to a situation where a non–homogeneous stress state exists in the coating, and where the channel cracks do not spread out parallel but rather radial from the indenter. This approximation is accurate if the indentation radius, r, is large compared to the crack length, a, and the crack spacing, L.

The stress, σ, in the previous section is thus identified as the circumferential stress in the coating due to indentation, $\sigma_{\theta\theta}(r)$. By writing the residual stress as

$$\sigma_0 = c\frac{E}{1-\nu^2} \tag{9}$$

the condition for channel crack arresting (8) can be written as

$$\left(c + \frac{u(r)}{r} + \nu\frac{du}{dr}\right)_{r=R+a} = 0.71\left(\frac{Eh}{(1-\nu^2)\Gamma_f}\right)^{-1/2} \tag{10}$$

with $u(r)$ given by (2) or (4). Now, for a given value of c, (10) is used to provide a relationship between channel crack length, $a = r - R$, and the non–dimensional quantity for the coating

$$\sqrt{\frac{Eh}{(1-\nu^2)\Gamma_f}} \tag{11}$$

An example of the relationship is shown in Fig. 4. Having established this relationship, the crack density can be evaluated by (7)

$$\frac{h}{L} = .52\left\{\sqrt{\frac{Eh}{(1-\nu^2)\Gamma_f}}\left[c + \frac{u}{r} + \nu\frac{du}{dr}\right]_{r=R} - .71\right\}^{.58} \tag{12}$$

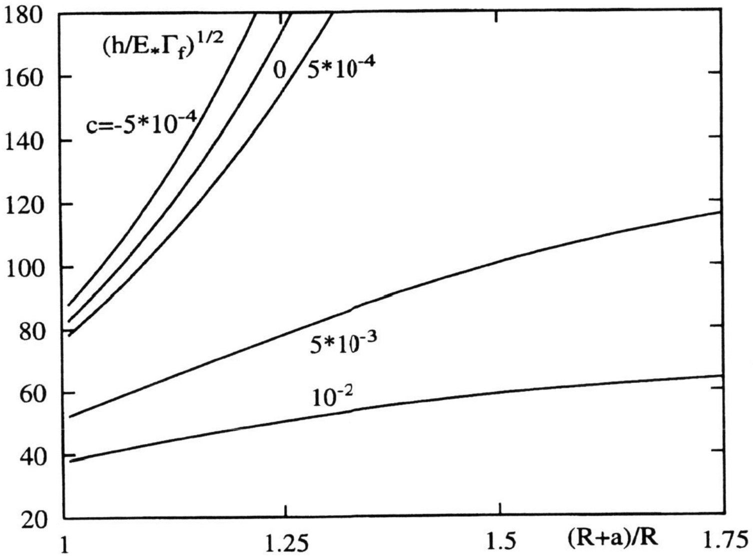

Fig. 4. Variation of normalized crack length with non–dimensional parameter for coating. Here, $E_* = E/(1 - \nu^2)$. Rockwell C indentation of substrate with $\sigma_y/E = .005$ and $n = 100$.

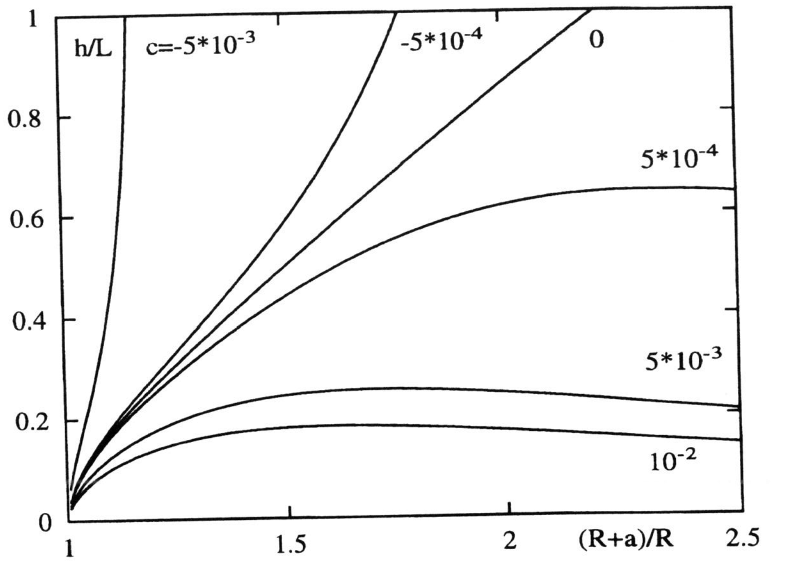

Fig. 5. Crack density *vs.* normalized crack length. Rockwell C indentation of substrate with $\sigma_y/E = .005$ and $n = 100$.

thus providing a relation between crack length and crack density with the residual stress as the only parameter for a given material pair. In Fig. 5, the variation of crack density, h/L, with normalized crack length, $(R + a)/R$, is shown.

DISCUSSION AND CONCLUSION

By Figs. 4 and 5 it is seen that relatively small residual stresses, say $|c| < 10^{-4}$, has insignificant influence on the crack length–density relation and the variation of the parameter (11) with crack length.

For compressive residual stress levels, channel cracks cannot propagate beyond a critical crack length, which can be identified from (10) as the value of $r = a + R$ for which $u/r + \nu du/dr = -c$ so that the parameter (11) becomes unbounded. This critical crack length will be zero when the residual compressive stress exceeds a certain level so that radial cracks at indentation are suppressed completely. In Fig. 5 this is seen by the curves for $c < 0$ having a vertical asymptote.

For tensile residual stresses in the coating, these tensile stresses may themselves cause channel cracking if they exceed the threshold condition (8). In Fig. 5 this would seen by the curves for $c > 0$ levelling off to a horizontal asymptote as r tends to infinity.

Plastic deformation of the substrate with a large plastic zone compared to the coating thickness will affect the results for the energy release rate, G , leading – through Eqs. (5) – to the energy release rate required for steady state propagation of channel cracks. Elastic mismatch between coating and substrate will also have an effect on G_{ss} . The effects can be estimated qualitatively by comparing the limit values $G(\infty, h)$ in Benthem and Koiter (1972) to the results of Beuth (1992), Hu and Evans (1989) and Beuth and Klingbeil (1995). Plastic deformation of the substrate has the effect of increasing G_{ss} so that the curves in Figs. 4 and 5 would be shifted towards lower values for the toughness of the coating.

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