higher values of K_I, the craze length s and craze opening at the crack tip $2v = 2v_c$ are nearly constant with $s \simeq 40\,\mu m$ and $2v_c \simeq 3\,\mu m$, respectively.

In summary, the Dugdale model gives a good description of the profile and size of the craze zones at static and moving crack tips in PMMA and gives the appropriate information of the viscoelastic material behaviour.

Acknowledgement

The experimental work on which this paper is based was financially supported by the Deutsche Forschungsgemeinschaft (DFG).

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Received 11 December 1979
and accepted 11 January 1980

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A comment on "On the validity of the Dugdale model for craze zones at crack tips in PMMA"

In a publication which investigated the applicability of the Dugdale [1] model for describing craze profiles in PMMA, we concluded that this model was not fully adequate for this purpose [2]. In contrast to these findings, other investigators report that the plastic zone profile described by this model correlates well within the measured craze profiles [3–10]. One of these groups, Döll, Seidelmann and Könczöl [10] are commenting on our publication. They claim that we have errors in both the technique of determining K_I and in the choice of yield strength (σ_{ys}) and elastic modulus (E) used to evaluate the corresponding Dugdale plastic zone profile.

After carefully reviewing our technique for determining K_I it was found that in the preparation of manuscript for the original publication [2] we made the inexcusable mistake of not converting the polynomial expression for the compliance and its derivative to the appropriate S.I. units. Unfortunately, these expressions misled Döll *et al.* [10] in their attempt to reproduce our calculations for comparing our method of determining K_I with those developed by other investigators for the same sample geometry. The corrected expressions for, respectively, Equations 12 (also as an inset on Fig. 6) and 13 in [1] are in metres per Newton

$$C\left(\frac{a}{W}\right) = 5.71 \times 10^{-3} \left\{-0.0356 + 0.4652\left(\frac{a}{W}\right) - 2.3597\left(\frac{a}{W}\right)^2 + 5.936\left(\frac{a}{W}\right)^3 - 7.196\left(\frac{a}{W}\right)^4 + 3.4769\left(\frac{a}{W}\right)^5\right\} \quad (1)$$

$$\frac{\partial C}{\partial\left(\frac{a}{W}\right)} = 5.71 \times 10^{-3} \left\{0.465 - 4.718\left(\frac{a}{W}\right) + 17.808\left(\frac{a}{W}\right)^2 - 28.784\left(\frac{a}{W}\right)^3 + 17.385\left(\frac{a}{W}\right)^4\right\}, \quad (2)$$

It should be noted that the proper unit conversions had been made for all data presented in the paper and we therefore believe that the reported stress intensities are accurate.

For comparison, the possibility of error in the technique of K_I determination employed by Döll *et al.* [10] must also be investigated. Here, comparatively thin (4 mm) CT specimens were used. Loading such thin specimens without proper guides could result in appreciable mode III, or antiplane, shear loading. Under these conditions the actual stress intensity at the crack tip would be the sum of mode I and mode III components. Consequently, the actual stress intensity could be larger than the mode I component which is measured.

The other major area of disagreement was our choice of modulus (E) and yield strength (σ_{ys}) which were used in the Dugdale model to predict the corresponding plastic zone profile. The values for these parameters, $\sigma_{ys} = 7.23 \times 10^7$ Pa and $E = 3.1 \times 10^9$ Pa, were obtained from Rohm and Haas Plexiglass G product literature [11]. To confirm these values we have conducted tensile tests on dumb-bell shaped samples which had been annealed for various lengths of time at a temperature of 70° C. The results of these tests indicate that after annealing for 144 hours, both σ_{ys} and E approach constant values of 7.3×10^7 Pa and 3.2×10^9 Pa respectively.

Döll and co-workers recognized that these parameters show a time dependence and chose to consider this factor and experimentally measure values of E and σ_{ys} for use in the Dugdale model. Their technique involved "performing interference optical measurements on moving cracks/crazes using a special experimental set-up including simultaneous K_I determination". The Dugdale model was then fitted to these measured craze zones and values of E and σ_{ys} were obtained. It was found that these values were dependent on crack speed. To apply the Dugdale model to the static-crack situation, values of σ_{ys} and E were taken for a crack speed, $\dot{a}$, of $10^{-8}\,\mathrm{mm\,sec^{-1}}$. With this technique, the agreement between the predicted Dugdale plastic zone profile and the craze profile was reported to be quite good.

We strongly disagree with this method of analysis. By using this technique of determining σ_{ys} and E, Döll *et al.* have assumed *a priori* that the Dugdale model is accurate for describing craze profiles in PMMA. By equating the plastic zone length, s, and opening at the plastic-zone-crack interface, $2v$, of the model to the corresponding craze parameters, they have essentially fitted the curve of the plastic zone profile to that of the measured craze profile.

Furthermore, when choosing these parameters to apply to a static-craze situation, values were taken for a crack speed, $\dot{a}$, of $10^{-8}\,\mathrm{mm\,sec^{-1}}$, these conditions were reported to be equivalent to a static crack. If this assumption were valid, both σ_{ys} and E should approach some constant value as the crack speed decreases to $10^{-8}\,\mathrm{mm\,sec^{-1}}$. In Figs. 3a and b of the letter by Döll *et al.* [10], the reported experimental results do not indicate this type of asymptotic behaviour. Rather, they indicate that these values decrease continually.

In our original paper we have discussed the importance of annealing the specimen to remove the craze formed by the razor blade pre-crack [2]. If the specimens are not annealed, reloading simply reopens a pre-existing craze. This is unlike the process described by the Dugdale model where a plastic zone initiates and develops with increasing K_I.

Döll and co-workers performed a set of experiments aimed at observing the effects of increasing sample stress or stress intensity on the profiles of static crazes. Although they reported annealing their specimens, from the discussion of their procedure it is not clear whether this took place before or after the razor blade pre-crack. Their experimentally measured craze parameters, s and

2v, exhibit a linear dependence on K_I rather than the parabolic behaviour predicted by the Dugdale model (see Figs. 4a and b of [10]). From preliminary results discussed in our original paper, this seems to indicate that the craze formed by the pre-crack was not annealed out prior to loading.

In conclusion, we maintain our original position that the Dugdale model is not fully adequate for describing craze geometries in PMMA.

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Received 30 April
and accepted 12 May 1980

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Thermal stability of amorphous Co–Fe–B, Co–Si–B and Co–Fe–Si–B alloys

Cobalt containing amorphous alloys exhibit extremely soft magnetic properties and at the same time show favourable mechanical properties [1]. Proper selection of the composition affords materials with magnetostriction approaching zero, e.g. alloys $Co_{74}Fe_6B_{20}$, $Co_{75}Si_{15}B_{10}$, $Co_{70}Fe_5Si_{15}B_{10}$ [2–4]. Their magnetic properties suggest the possibility of applying these materials as improved substitutes for permalloys [1, 5]. They are also characterized by high resistance to oxidation, even at high temperatures and effective cooling of liquid metal can be achieved without any protective gas cover [6].

The present study concerned the crystallization behaviour of these 3 alloys plus 2 more of the Co–Fe–B, type with increased and reduced boron (Table I). It enabled us to characterize the effect of the amount and kind of metalloid on the behaviour of amorphous cobalt-based alloys.

Alloys were prepared from elements of the following purity: Co, Si 99.9%, Fe, B 99%. Samples of 5 to 7 g in weight were remelted several times by argon arc melting. Metallographic observations confirmed that the resulting samples exhibited no segregation, so were suitable for production of ribbons. The resulting ribbons were completely amorphous according to X-ray studies; they were 20 to 30 μm in thickness and several metres in length.

Calorimetric studies were performed using a Perkin Elmer DSC–2 calorimeter in which the course of crystallization during continuous heating at a heating rate of 10 to 80 K min^{-1} was determined. The values of activation energy were obtained from Kissinger's method [7]. The structure of samples both in the initial state and after different crystallization stages was determined with a Philips X-ray diffractometer, using Cu$K\alpha$ radiation.

The exothermic effects associated with the

TABLE I Calculated value of the activation energy of crystallization and temperature of peak I (heating rate = 10 K min^{-1}), for investigated alloys

Alloy	ΔE (kJ mol^{-1})	T_x (K)
1. $Co_{76.8}Fe_{6.2}B_{17}$	274 ± 6	718
2. $Co_{74}Fe_6B_{20}$	257 ± 2	720
3. $Co_{71.2}Fe_{5.8}B_{23}$	235 ± 1	747
4. $Co_{70}Fe_5Si_{15}B_{10}$	324 ± 3	786
5. $Co_{75}Si_{15}B_{10}$	328 ± 9	792

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