JOURNAL OF MATERIALS SCIENCE 22 (1987) 2397–2406

Fast fracture of rubber-toughened thermoplastics used for the shells of motorcycle helmets

A. GILCHRIST, N. J. MILLS
Department of Metallurgy and Materials, University of Birmingham, Birmingham B15 2TT, UK

The effects of outdoor ageing and of accelerated UV exposure on the grade of acrylonitrile butadiene styrene copolymer (ABS) used by UK manufacturers of motorcycle helmets were compared. Although the surface layer becomes embrittled the helmets still survive an impact test after conditioning at −20° C because the cracks are arrested before they penetrate the full thickness of the shell. A 4 mm thick rubber-toughened polycarbonate, used in helmets from continental suppliers, shows a transition from "tough" to "brittle" crack propagation behaviour at −15° C. For 4 mm thick ABS this transition had not occurred at the lowest test temperature of −20° C, but if 6 mm sheet is tested the transition is at −15° C. Temperature increases in the 40 sec delay between conditioning a helmet at −20° C and impact testing are shown to be significant.

1. Introduction

One of the authors (N.J.M.) has been involved for a number of years with investigating the failures of motorcycle helmet shells, either for the Accident Research Unit in the University of Birmingham, or for Coroners' Courts, or for the British Standards Institution Committee PSM10 that is responsible for the relevant safety standards. As a result of this experience there was found to be a lack of design information on the fracture mechanics of the thermoplastics used in helmet shells. This is not to say that there had not been many academic studies of the fracture of the materials involved, rather that these data did not exist for the conditions (rates, temperatures, thicknesses) applying to motorcycle accidents.

There are two main thermoplastics being used in helmets on sale in the UK market today. To understand why they are used a brief review of the history of the national standards must be made. From about the time of the introduction of BS 5361:1976 "Protective helmets for vehicle users" and BS 2495:1977 "Protective helmets for vehicle users (high protection)" the main thermoplastic used for the outer shell was unmodified polycarbonate. However, these shells were found to fracture in a brittle way in a number of circumstances such as

(i) there being a brittle layer on the outside that would initiate a crack, namely a paint film or a brittle adhesive such as cyanoacrylate used to attach rubber trim;

(ii) environmental stress cracking affecting drilled holes in the shell, namely petrol etc. passing up a webbing chin-strap to the point where it is riveted to the shell;

(iii) stress concentrations being present, such as a sharp step on the inner surface due to wear in the sliding part of an injection-moulding tool; and

(iv) moulding weaknesses such as a weak weld-plane in the centre of a chin-bar where two opposing streams of melt meet.

As a consequence in particular of (i) and (ii) motorcyclists were warned not to paint helmets, and in February 1981 Amendment 5 to both standards was published. This called for a solvent conditioning, by wiping with a 50% toluene–50% isoctane mix prior to impact testing. This mixture is a powerful stress-cracking agent for polycarbonate [1] initiating cracks at positions, such as drilled holes, where there are residual tensile stresses.

From the middle of 1980 UK manufacturers changed to a particular grade of acrylonitrile butadiene styrene copolymer (ABS) developed by DSM (Geleen, Netherlands). This grade, Ronfalin MST 42, by virtue of its molecular weight and the amount and size distribution of the rubber spherical inclusions in it, has a particularly high toughness in Charpy impact tests. However, continental suppliers of helmets to the UK market changed to a rubber-modified form of polycarbonate (PC). This was a grade ML 3300 supplied by the General Electric Co. of USA. The rubber used has not been disclosed, but there are a limited number of rubbers that are melt-stable at the processing temperature (~280° C) of PC. Thus the rubber phase is believed to be similar in chemical nature to that in ABS.

For fracture mechanics to be used there must be a pre-existing crack in the product. We must therefore enquire how a crack can arise in an ABS or a rubber-toughened PC shell. ABS is well known to be subject to photo-oxidation as a result of a combination of ultraviolet wavelengths in the solar spectrum, and oxygen (or ozone) in the air [2]. Mainly because

0022–2461/87 $03.00 + .12

the UV light only penetrates a short distance in a pigmented plastic, this process only affects an outer layer of thickness 0.1 to 0.2 mm [3]. Oxidation increases the glass transition temperature of the rubber, to a level where it behaves like a glass in an impact. Consequently if the shell is bent so that the outside is in tension, cracks form in the outer layer. Depending on the nature of the ABS used, these cracks may or may not propagate rapidly and cause the shell to break up in the accident. Shell fractures have been observed in ABS grades other than MST 42 in the UK and in a French survey [4]. It was felt that the effects of outdoor exposure, and artificially accelerated UV exposure, on the MST 42 grade should be investigated. Currently the British Standards Institution is starting one-year outdoor exposure tests on helmet shells, as a part of the new BS 6658 : 1985 "Protective helmets for vehicle users". The helmets will be impact-tested at -20° C after exposure, but the results will be for information purposes only.

2. Crack Initiation in aged ABS

2.1. The effects of natural and accelerated ageing

2.1.1. Testing

Ageing experiments were carried out on the MST 42 grade of ABS. There were three methods of ageing.

2.1.1.1. Outdoor exposure in the UK. Blue and black pigmented helmets were faced south with their vertical axis tilted at 45° C towards the equator, and exposed for several years. These experiments were carried out by a UK manufacturer.

2.1.1.2. Artificial exposure to UVB radiation (280 to 330 nm). Most artificial weathering equipment only allows flat-plate specimens to be tested. Therefore an exposure chamber had to be constructed to contain helmets. UVB-313 fluorescent tubes supplied by the Q-Panel Co. (Salford, UK) use phosphors to provide nearly all of their output in the UVB region. The maximum power intensity is at 313 nm. Six 40 W tubes were horizontally 70 mm apart, 50 mm above a row of the helmets, in a closed box. The box was lined with aluminium foil in an attempt to even out the radiation intensity, and the temperature of the top the helmets was thermostatically controlled at 45° C. The rate of oxidation is known to increase rapidly as the test temperature rises, and many commercial chambers operate at 50 to 70° C. The estimated UV intensity is similar to that in commercial Q-U-V equipment, namely 0.8 $\mathrm{W\,m^{-2}\,nm^{-1}}$ at 310 nm wavelength. Both ABS and polycarbonate helmets were exposed for 500 h.

2.1.1.3. Artificial exposure to Xenon lamp radiation and "rain". Hereaus GmbH (Hanau, West Germany) kindly exposed some flat plaques of MST 42 in their Xenotest 1200 machine for periods up to 1100 h. This machine contains a Xenon lamp, which is filtered to produce a spectrum akin to that of daylight. Thus although the intensity at 310 nm, of 0.9 $\mathrm{W\,m^{-2}\,nm^{-1}}$, is similar to that from the Q-U-V tubes, the intensity increases sharply with frequency until 460 nm (3 $\mathrm{W\,m^{-2}\,nm^{-1}}$). The machine was run on a 3 min rain, 17 min dry cycle, so that there was humidity cycling of the ABS. The black panel temperature in the test chamber varies from 30° C (rain) to 70° C (dry), so there is also thermal cycling of the ABS.

2.1.2. Surface appearance after exposure

All of the test methods caused colour changes (yellowing) of the ABS, and some loss of gloss. The exposure to UVB radiation alone did not cause any cracking, whereas the other tests produced a series of parallel cracks along the flow direction followed by the melt as it filled the mould (Fig. 1). The outdoor exposures were for 3 to 5 years. The estimated equivalent of 500 h of UV exposure in an accelerated test is 1 year outdoor exposure. Comment can be made on three aspects of the results.

2.1.2.1. The occurrence of cracks. UVB exposure produced no cracks, although it did degrade the surface layer of the ABS. The reason for the crack formation in the other experiments was probably the transient thermal stresses when "rain" starts. When the surface is cooler than the interior there are biaxial tensile stresses in the surface layer, and these will cause cracking in the weakest direction in the surface layer. There are relatively more strong covalent bonds in the direction of flow, so cracks form parallel to the flow direction. It is unlikely that the cycling in moisture content will cause cracking; ABS only absorbs 0.3% water at 23° C after 24 h.

2.1.2.2. The direction of the cracks. There is a contrast with weathered polyethylene construction workers' helmets, where cracks form perpendicular to the melt flow direction [5]. The difference must relate to the microstructure. In a semicrystalline polymer injection moulding there is likely to be a skin of highly oriented crystals; the weak directions may then be parallel to the surfaces of the lamellar crystals (perpendicular to the flow direction).

2.1.2.3. The spacing and depth of the cracks. The cracks in the outdoor-exposed helmet shells are closer together (30 μm) than those in the injection-moulded plaques exposed to a Xenon lamp and rain (100 μm, Fig. 1). The differences could be as much due to moulding conditions as to the exposure conditions. It is noticeable that the cracks in the outdoor-exposed ABS are wedged open by oxidation debris.

2.2. Crack formation as a result of impact tests

Impact tests were carried out on exposed helmets with a 50 mm radius hemispherical striker, after the helmets had been conditioned at -20° C. There is a delay of about 15 sec between removing the helmet from the freezer and the impact occurring. BS 6658 requires the first impact after 40 ± 5 sec; the older BS 2495 requires it within 3 min. A detailed discussion of temperature control occurs in Section 4.

At one stage it was thought that a direct blow

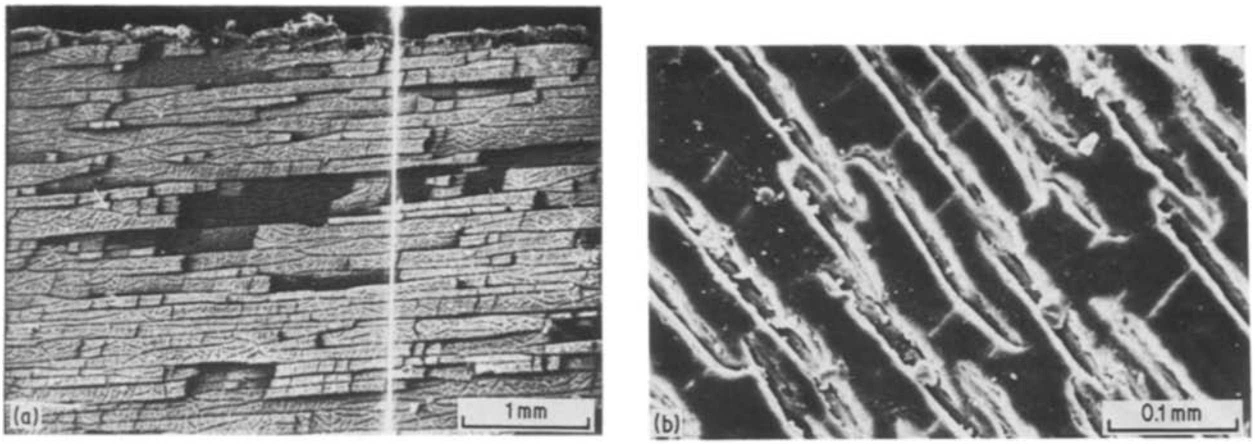

Figure 1 SEM of the surface cracking of Ronfalin MST 42 ABS (a) exposed for 1100 h in a Xenotest with a rain cycle (unpigmented), (b) exposed out of doors for 3 years (black pigment).

on a helmet shell would put the outer surface in compression, so a brittle outer surface layer would not affect the test. However it has been shown, by the use of brittle thermoset layers on the outside, that a hemispherical striker causes the shell to buckle inwards [6] and that there is a radial tensile stress on the outer surface, that has its maximum value at 40 mm from the centre of the contact point. Placing a strain gauge at this point led to a measurement of 6% for the radial tensile strain. In a 125 J impact by a hemispherical striker the polystyrene foam liner is almost completely crushed below the impact point. Typically it deflects by 20 mm [7], and at such a high deflection the shell must buckle. It is arguable that an impact with a flat surface would not produce buckling of the shell, and hence the outer surface would not go into tension near the impact site. However a significant proportion of impacts are with rounded or sharp objects [8], and even an impact with a flat surface will deform the whole shell into an oval shape, with tensile stresses occurring at positions $\sim$90° from the impact site.

The −20° C impact tests produced three types of crack patterns with the ABS helmets:

(a) With Ronfalin MST 42 helmets after 500 h UVB exposure a complete circular crack of diameter 70 mm formed on the outside surrounding the impact site (see Fig. 3a). However, it only penetrated 2.4 mm through the 4 mm thick shell (Fig. 2). SEM of the fracture surface showed that the first 1.6 mm of crack was of the rapid fracture ($> 100\,\mathrm{m\,sec^{-1}}$) type, and the next 0.8 mm was of a slow fracture ($\sim 3\,\mathrm{m\,sec^{-1}}$) type. Correlation of fracture surface appearance with crack velocity is discussed in Section 3.

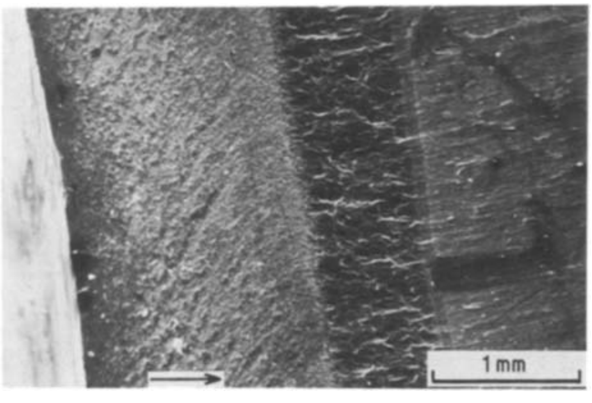

Figure 2 SEM of the fracture surface of a blue MST 42 helmet exposed to UVB for 500 h, then impacted at −20° C. The ring crack changes its surface appearance as it decelerates.

(b) With $\sim$ 10-year-old BS 2001 ABS helmets there is a partial ring crack AA′ as in (a), there are also two radial cracks in the melt flow direction (from the rear to the front of the helmet). Fig. 3b shows a typical crack pattern. The cracks have penetrated the full shell thickness with a high-speed fracture surface appearance. Once they have formed the loading changes to one closer to that of double torsion, so that each of the cracks grows away from the contact point C.

(c) With the outdoor-exposed blue Ronfalin MST 42 helmets there are many small circumferential cracks. These are at radial distances of 20 to 40 mm, with an average radial spacing of 0.9 mm. They are typically 0.3 mm deep, and 5 to 10 mm long on the surface. With the black helmets, however, no cracks were found to form as a result of impact.

The difference between cases (a) and (b) appears therefore to be the way in which an initially rapidly

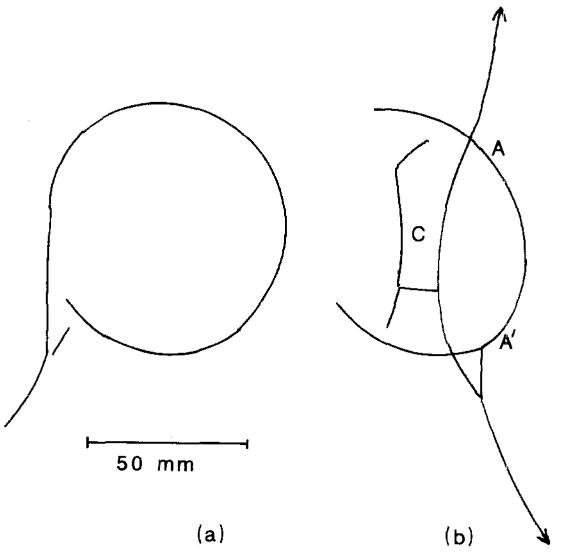

Figure 3 Sketches of cracks on (a) 500 h UVB aged MST 42 helmet, (b) 10-year-old BS 2001 ABS helmet, after −20° C impacts with a hemispherical striker.

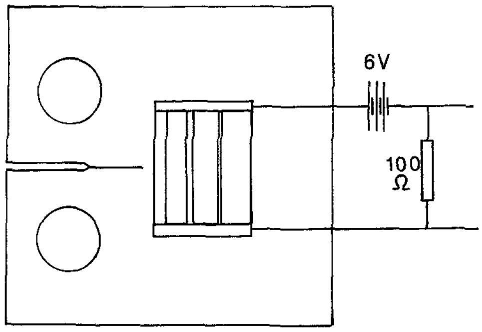

Figure 4 Compact tension test specimen, showing gold-palladium resistance grid for crack velocity measurements.

moving crack decelerates as it penetrates further from the surface and meets un-aged material. Further discussion is left until after Section 3. The peak acceleration levels measured for the MST 42 helmets ranged from 160 to 200 g where 1 g = 9.81 m sec^{-2}. Consequently the helmets would still meet the relevent British Standard, which requires that the acceleration is less than 400 g.

Four polycarbonate ML 3300 helmets were tested at −20° C after 500 h of UVB exposure. There was no sign of any cracks forming as a result of the impacts.

3. High speed, low temperature fracture toughness tests

3.1. Experimental design

The fracture toughness test pieces are the standard "compact tension" design, as for example specified in BS 5447:1977 "Plane strain fracture toughness of metallic materials". The "width" parameter W = 40 mm, but the specimen thickness is that of the helmet shell rather than being a particular fraction of the width. The samples were cut from helmet shells so that the crack direction (Fig. 4) was the melt flow direction in the moulding process. ABS samples cut from aged helmets were flattened by heating to 120° C and applying a small load; however as well as flattening the sample this allows any residual stresses and moulded-in molecular orientation to relax, consequently later samples were tested without flattening. A crack was grown from the sharp saw-cut by fatigue loading at 150 to 200 Hz with a fluctuating load of about ±5 kg and a minimum load just above zero. Fatigue crack growth of about 3 mm was produced in the order of 10 min, the load being reduced to avoid a large yielded zone forming at the crack tip. A set of Au–Pd lines 5 mm apart were then sputter-coated on to one surface, using an aluminium sheet mask (Fig. 4). The line width decreases in the sequence 2, 1, 0.5, 0.25 mm with the intention that there will be roughly equal voltage steps as the growing crack cuts the lines in turn.

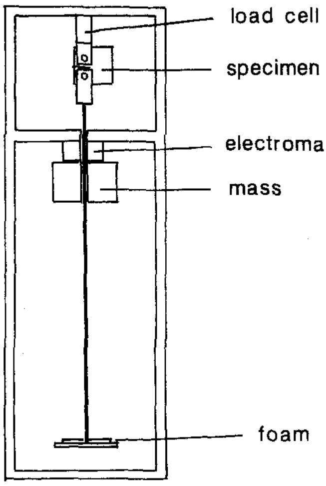

Figure 5 Impact loading frame.

Fig. 5 shows the impact loading frame, which fits inside a temperature-controlled freezer. A 3.27 kg mass is held by an electromagnet until the test starts. It then falls 0.34 m, acquiring 11 J of kinetic energy, before striking a 6 mm thick high-density polyethylene foam layer on top of an aluminium baseplate. The resulting impulse is transmitted through the sample to a quartz piezoelectric load cell. This Model 2103-100 cell (Endevco, California, USA) has a calculated resonant frequency of 9 kHz with the steel upper grip of mass 96 g attached to it. The load-cell output is then amplified with a Kistler charge amplifier that is effectively unfiltered as it has a 180 kHz cutoff filter. The signal then passes to a Datalab DL 902 (Data Laboratories Ltd., Mitcham, UK) twin-channel 8-bit transient recorder, with a sampling frequency of up to 1 MHz.

One problem with such an experiment is that the steel masses can sometimes vibrate on the quartz crystal "spring" to produce spurious "ringing" on top of the force–time trace (Fig. 6). The only reason why such ringing was not a serious problem here is the relative compliance of the polyethylene foam layer. The loading sawtooth (Fig. 6a) takes 800 μsec, an order of magnitude more time than the time for one oscillation (72 μsec). The initial contact oscillations are damped and disappear before the crack growth events. The high-speed crack growth (Fig. 6a) induces ringing of the load cell and top grip *after* the specimen has broken. On the other hand for a "tough" specimen with low-speed growth the events are so slow that no ringing occurs (Fig. 6b).

3.2. Analysis of the results

The data from the transient recorder is transferred to a BBC Model B microcomputer, and stored on disc. The voltage trace from the resistance grid is displayed on the screen in an expanded form, and the positions half-way up each voltage step are read off using a cursor. These data then give the average velocity between the four lines, namely three velocities. The force F at the instant that each line is broken is used to calculate the stress intensity factor K, using

$$K = \frac{FY}{Bw^{1/2}} \tag{1}$$

where w is the width, B the thickness and Y is a function of the crack length to width ratio $a/w \equiv x$:

$$Y = x^{1/2}(29.6 - 185.5x + 655.7x^2 - 1017x^3 + 638.9x^4)$$

This analysis assumes that any yielding at the crack tip remains "small-scale", i.e. much smaller than the

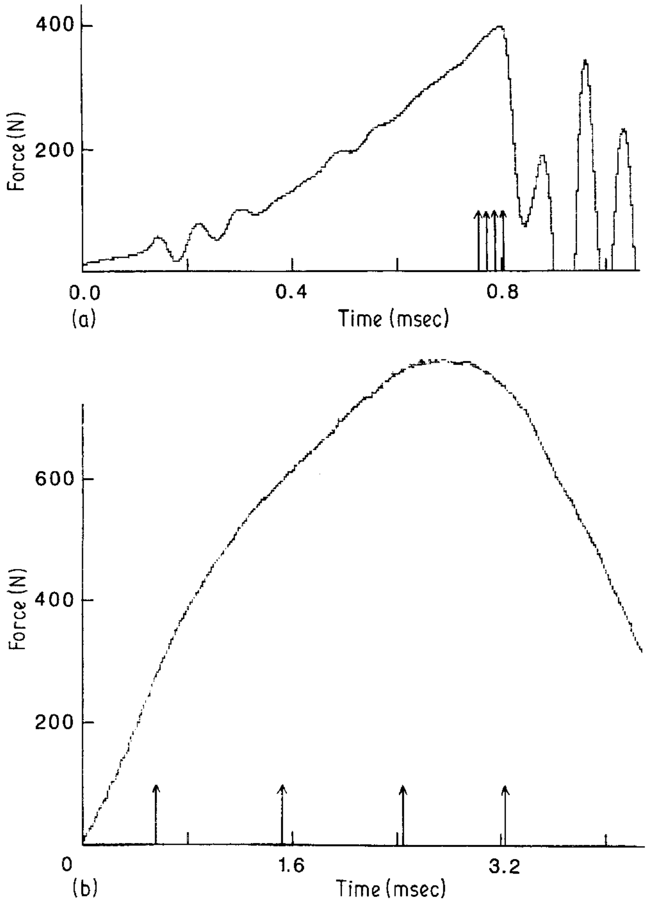

Figure 6 Typical force against time traces for (a) brittle, (b) tough behaviour. The vertical bars indicate when the resistance bands fractured.

crack length. This may not always be the case with a material like ABS.

3.3. Results

The test results can be categorized into two ways, which for want of better names we will call "tough" and "brittle":

(a) "tough" results are typified by a K value which after 10 mm of crack growth exceeds $10\,\text{MN}\,\text{m}^{-1.5}$, and the crack velocity is less than $10\,\text{m}\,\text{sec}^{-1}$, typically $5\,\text{m}\,\text{sec}^{-1}$. As well as the fracture surface appearing whitened there is clear evidence on the sides of the sample of a whitened layer about 1 mm thick on either side of the fracture surface.

(b) For "brittle" results the K value after 10 mm of crack growth is less than $8\,\text{MN}\,\text{m}^{-1.5}$ (PC) or $5\,\text{MN}\,\text{m}^{-1.5}$ (ABS) and the crack velocity exceeds $200\,\text{m}\,\text{sec}^{-1}$. The fracture surface appears the natural colour of the polymer and there is no whitened layer visible on the sides of the specimen.

SEM examination of the fracture surfaces shows detailed differences in appearance (Fig. 7). In the "tough" case the surface appearance is much rougher and there is no sign of the 1 μm rubber spheres. In the "brittle" case the rubber spheres can be seen, or the holes from which they have been torn.

Although further measurements are necessary to fully establish the temperature range over which the tough to brittle transition occurs, there is sufficient evidence for

(a) 6 mm thick MST 42 Ronfalin plaques, and

(b) 4 mm thick ML 3300 polycarbonate helmets

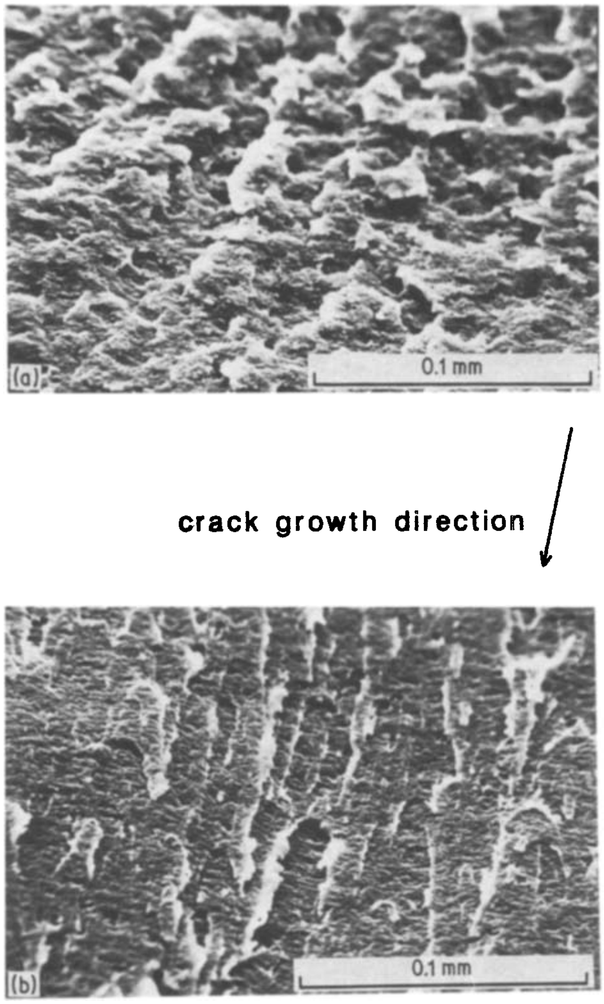

Figure 7 SEM photographs of (a) brittle, (b) tough fracture surfaces of MST 42 compast tension specimens.

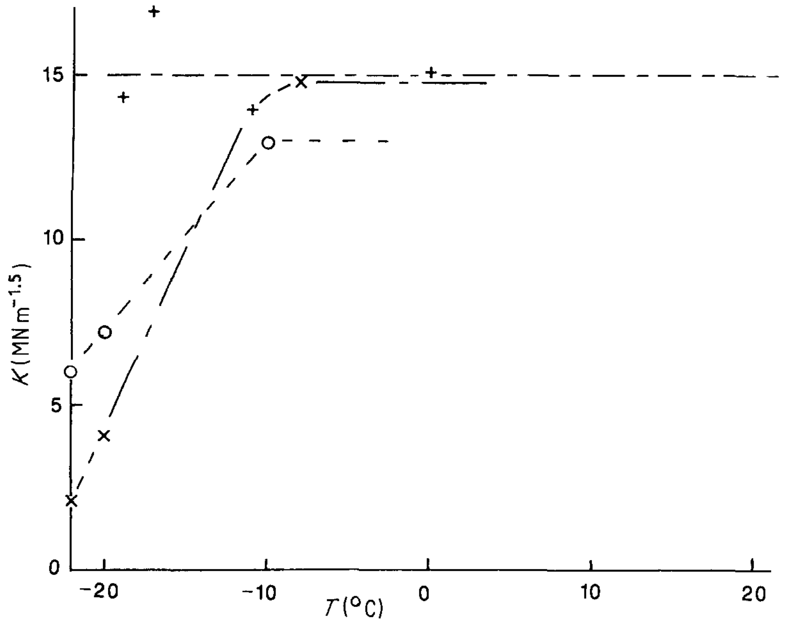

Figure 8 Variation of the fracture toughness after 10 mm of crack growth with temperature for (○) toughened polycarbonate, (4 mm ML 3300), and (+) 4 mm and (x) 6 mm thick MST 42 ABS.

of a transition in the region of − 15° C (Fig. 8). The K values used are those at the third resistance line. On the other hand Ronfalin MST 42 cut from helmets, which is 3.9 ± 0.4 mm thick, does not show "brittle" behaviour at the lowest test temperature of − 20° C. This implies that the transition temperature depends on the specimen thickness. It is the transition to "brittle" behaviour with a K value below 5 MN m$^{-1.5}$ which is the significant feature, rather than the absolute K value. It means that in a typical helmet impact lasting perhaps 5 m sec a crack can run 200 m sec^{-1} × 0.005 sec = 1 m, and therefore the helmet can easily crack into several pieces. The energy absorbed in such a fracture can be calculated using the relationship

$$K^2 \simeq EG_c$$

where E is the Young's modulus of about 3 GN m^{-2} for such plastics at − 20° C under impact conditions. Consequently the critical strain energy requirement for new crack surface $G_c \simeq 8300\,\mathrm{J\,m^{-2}}$ if $K = 5\,\mathrm{MN\,m^{-1.5}}$. For a 200 mm crack in a shell of thickness 4 mm only 7 J of energy is absorbed. This is insignificant compared either with British Standard test energies of 125 to 150 J, or with accident energy inputs. On the other hand a "tough" crack propagation at 5 m sec^{-1} is travelling too slowly to allow the helmet shell to break up in the time scale of the impact.

Studies of fast crack growth in polycarbonate [9] have shown that K rises as the crack velocity approaches a significant fraction of the speed of sound in the plastic. The results calculated here suggest that for "tough" behaviour K rises to the order of 15 MN m$^{-1/5}$ without any significant increase in crack velocity over the initial 5 m sec^{-1}. This is because the thickness of the whitened layer increases as the crack gets longer (Fig. 9). The variability of the initial K values for "tough" ABS in Table I is largely because the first resistance line is a variable distance from the tip of the fatigue crack.

Table I shows that there are no cases where the crack velocity changes from < 10 m sec^{-1} to > 200 m sec^{-1} as the crack grows, or vice versa. The crack sometimes slows down as it approaches the free surface of the compact tension specimen, i.e. as $a \to w$.

There is no evidence that the aged MST 42 material has any lower toughness than the new material. This is because the crack is travelling parallel to the brittle surface, and the fraction of the fracture surface that is brittle in appearance is small. Typically the outer 0.2 mm of the surface appears to be brittle, and this is an insignificant proportion (5%) of the whole.

There is a suggestion that the earlier grade of ABS used in the old BS 2001 helmet has a lower K value and higher crack velocity than the MST 42 grade. However, further study is necessary to confirm this.

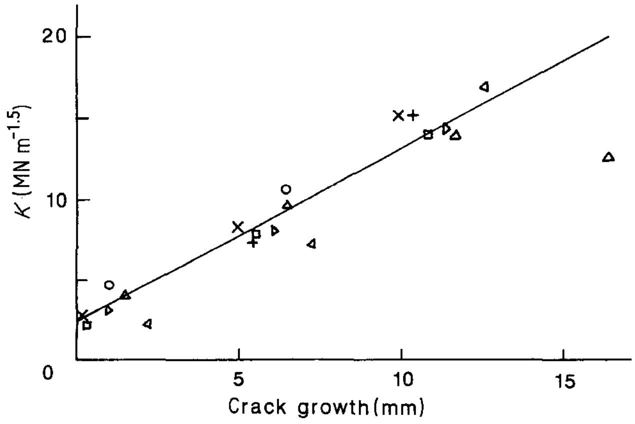

Figure 9 Variation of the K_c value with the distance the crack has grown in the impact test, for "tough" 4 mm thick MST 42 (different symbols apply to different test temperatures). □ ▷, − 18° C; △, − 17° C; ○△, − 10° C, +, 0° C; ×, 23° C.

TABLE I Stress intensity values for helmet shell materials

Material	Colour	Thickness (mm)	Temperature (°C)	Stress intensity, K (MN m$^{-1.5}$)*	Velocity, V (m sec^{-1})*
New materials					
ABS MST 42	Natural	5.9	−20	1.5 (4.1)	305 (271)
		5.9	−22	1.5 (2.1)	416 (384)
		5.9	−8	4.4 (14.8)	2.2 (3.6)
	White	3.6	−19	3.1 (14.3)	6.8 (4.2)
	Blue	3.5	−17	2.2 (16.9)	7.8 (5.3)
	White	4.1	−10	4.7 –	4.1 –
	Blue	3.8	−11	4.0 (13.9)	4.4 (5.9)
	White	4.0	0	2.5 (15.1)	5.4 (6.3)
	Blue	4.0	23	2.5 (15.1)	4.0 (4.0)
Unmodified PC	Natural	5.9	−20	1.9 (4.1)	178 (189)
		6.0	−20	2.4 (3.9)	192 (197)
		6.0	+20	2.0 (2.9)	304 (191)
PC ML 3300	White	3.6	−22	2.9 (6.1)	357 (400)
		4.2	−20	3.6 (7.2)	281 (327)
		3.6	−10	6.1 –	12.3 –
		3.6	−10	4.1 (12.9)	1.8 (2.5)
Outdoor-aged materials (pressed flat)					
ABS MST 42	Black	4.3	−15†	5.0 (13.8)	8.9 (5.8)
	Blue	4.0	−15†	5.8 –	7.8 –
BS 2001 grade	White	3.7	−15†	4.2 (6.9)	42 (11)
		3.7	−15†	3.9 (15.8)	16 (83)

*Initial values; values in brackets are after 10 mm of crack growth.
†Approximate temperature.

4. Temperature changes in helmets after low-temperature conditioning

Experiments were carried out using a 4.2 mm thick Ronfalin MST 42 helmet shell. A fine thermocouple (0.14 mm wire) was attached either to the outer or inner surface of the shell, by sticking a 0.14 mm thick layer of PVC tape over it, then replacing the liner. The helmet was removed from a freezer at −19° C to a laboratory at 22° C. A thin layer of rime was noticed to form on the shell.

In order to avoid the necessity of separate experimental determinations of temperatures for all shell thicknesses, shell materials and laboratory temperatures, an analysis of the one-dimensional heat transfer problem was made. The shell thickness is split into layers in a finite difference model. Each layer has a discrete temperature T_i; usually at least 15 layers are used in the calculation. The one-dimensional heat diffusion equation

$$\alpha \frac{\partial^2 T}{\partial x^2} = \frac{\partial T}{\partial t} \qquad (2)$$

where α is the thermal diffusivity and t time, is solved in finite difference form by the explicit relationship

$$T_i^* = T_i + D(T_{i+1} + T_{i-1} - 2T_i) \qquad (3)$$

where T^* is the "new" temperature after a time interval Δt, and D is given by

$$D = \frac{\alpha \Delta t}{\Delta x^2} \qquad (4)$$

where Δx is the layer thickness. D must be kept below 0.5 for the computations to be stable, so the time interval must be suitably small.

At the outer shell surface there is convective heat transfer to air, with a heat transfer coefficient h determining the heat flow Q:

$$Q = hA(T_s - T_A) \qquad (5)$$

where A is the area of shell, and T_s, T_A the shell surface and air temperatures, respectively. The energy balance of the surface layer of thickness $\Delta x/2$ in finite difference form is

$$T_1^* = T_1(1 - 2D - 2H) + 2HT_A + 2DT_2$$

where

$$H = \frac{h\Delta t}{\varrho C_p \Delta x} \qquad (6)$$

and ϱ is the density, C_p the specific heat of the plastic. At the inner shell surface there is contact with 25 mm of polystyrene foam liner; it can be assumed that this surface is thermally insulated. Consequently the slope of the temperature profile is zero at the inner surface. This is incorporated in finite difference form by putting

$$T_{n+1} = T_n \qquad (7)$$

where a ficticious layer ($n + 1$) lies beyond the inner surface.

The thermal diffusivity of most thermoplastics lies in the range 0.07 to 0.13 mm^2 sec^{-1}, the product of density and specific heat is about 1.2×10^6 J m^{-3} (° C)$^{-1}$ for most polymers. Heat transfer coefficients are less well established; a value of 10 W m^{-2} (° C)$^{-1}$ is appropriate for plastics cooling down from ~ 100° C in still air. If the initial temperature is −20° C then the formation of frost on the surface in moist air may increase the heat transfer coefficient.

Fig. 10 shows the experimental variation with time of the inner and outer surface temperature of the ABS helmet. The best-fit theoretical curves are also shown.

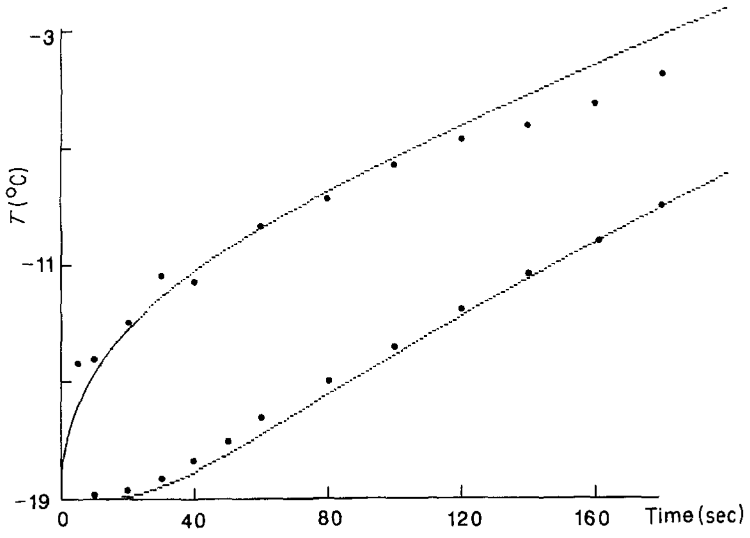

Figure 10 Heating of a 4.2 mm ABS shell outer surface (upper curve) and inner surface (lower curve) with time after removal from $-19°$C freezer to a 22°C laboratory. Points are experimental, curves are calculated best fit for $\alpha = 0.08\,\text{mm}^2\,\text{sec}^{-1}$, $h = 11\,\text{W}\,\text{m}^{-2}\,(°\text{C}^{-1})$, $\varrho C_p = 1.2\,\text{MJ}\,\text{m}^{-3}\,(°\text{C}^{-1})$.

In the theory there are two disposable parameters. The first is the ratio of heat transfer coefficient to specific heat in Equation 6; this determines the slope of the heating curves once a near "steady state" is reached after about 40 sec. If $\varrho C_p = 1.2 \times 10^6\,\text{J}\,\text{m}^{-3}\,(°\text{C})^{-1}$ then a heat transfer coefficient $h = 11\,\text{W}\,\text{m}^{-2}\,(°\text{C})^{-1}$ fits the data for the first 120 sec, but the value of $10\,\text{W}\,\text{m}^{-2}\,(°\text{C})^{-1}$ is more appropriate later. The second parameter is the thermal diffusivity, which determines the temperature differential across the shell. A value of $0.08\,\text{mm}^2\,\text{sec}^{-1}$ seems best. Fig. 11 shows the temperature profile across the shell calculated for these values. Tests carried out in a 20°C laboratory, 40 sec after conditioning, will test a shell with an outer surface at $-12°$C and a 7°C temperature differential across the wall.

If the helmet shell material being tested has a fracture toughness transition in the region of $-15°$C, and if there are cracks in the shell (or if cracks form by the fracture of a brittle outer layer), then the result of a helmet impact test may well differ if the test is carried out in an ambient temperature laboratory *after* conditioning, or in an environment chamber at $-20°$C.

5. Discussion

The events which lead up to the potentially disastrous brittle fracture of a helmet shell will be discussed in order in which they occur. Firstly there is the formation of crack(s) in an aged surface layer. Then there is the propagation of that crack into the unaged substrate. Finally there is the high-speed growth of the crack(s) around the shell.

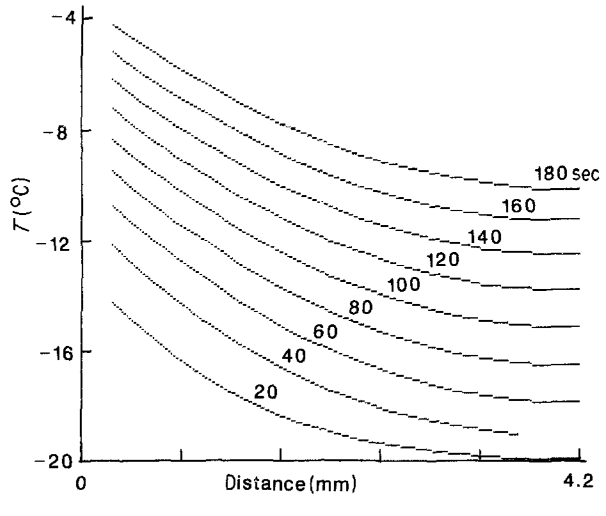

Figure 11 Calculated temperature profiles through the ABS shell for the same parameters as Fig. 10, at 20 sec intervals.

The vast majority of thermoplastic helmets incorporate a rubber phase in the thermoplastic, and this rubber phase has been shown to photo-oxidize under the influence of the UVB part of the sun's spectrum. The greater resistance of black-pigmented ABS to degradation, noted here, has been observed before [3]. User's helmets will be subjected to variable exposure to sunlight, but the upper limit of exposure would appear to be equivalent to one year's continuous exposure. If the helmet is then involved in an accident it is likely that part of the outer surface will be subjected to tensile strains. When a helmet hits a 50 mm radius hemisphere the surface strain is about 6%. The condition for crack formation is therefore that the embrittled outer layer fails at a tensile strain less than 6%. This condition is met for ABS at $-10°$C, and perhaps is met at higher temperatures. However, the nature of the cracks depends on the surface condition prior to impact. For outdoor-weathered ABS with an already cracked surface many cracks form on impact, whereas for an intact UVB-exposed surface a single crack forms. An analysis has been made [10] of the energetics of growth for an infinite array of edge cracks. This showed that the strain energy release rate (a parameter proportional to $K^{1/2}$) reaches a constant value when the crack lengths exceed 0.3 times their separation. This contrasts with the case of a single edge crack, when the strain energy release rate is proportional to the crack length. This explains why the cracks have stopped in the surface of the outdoor-aged ABS. There is a low and equal value of the stress intensity factor K at the tip of each crack, and this value is too small for crack growth to continue. On the other hand the single edge crack in the UVB-exposed ABS does not have to be very long for a high K value to occur. Using the relationship

$$K = 1.12\sigma(\pi a)^{1/2} \qquad (8)$$

between the stress intensity factor K, the surface

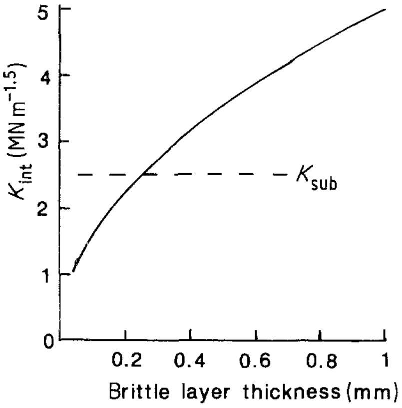

Figure 12 Variation of the stress intensity factor at the brittle layer interface, K_{int}, with the layer thickness, calculated from Equation 8 for a stress of 80 MN m^{-2}. The K level for the initiation of crack growth in MST 42 is also shown.

tensile stress σ and the crack depth a, an estimate of K can be made. For ABS the yield stress at $-10°$ C under impact conditions is of the order of 100 MN m^{-2}. Using a stress of 80 MN m^{-2} in Equation 8 leads to $K = 1.6$ MN m$^{-1.5}$ for $a = 0.1$ mm and $K = 2.3$ MN m$^{-1.5}$ for $a = 0.2$ mm. This shows that, paradoxically, it is better to have a brittle surface layer precracked by thermal cycling [11] than it is to have an intact brittle layer from constant-temperature UVB exposure.

The condition for a crack to propagate from one polymer into another has been analysed by Rolland and Broutman [9], among others. We may take the embrittled surface layer as a separate material, although in reality there is a diffuse rather than a sharp interface between the degraded and the undegraded ABS. Rolland and Broutman proposed that the crack would continue to propagate rapidly if

$$K_{int} > K_{sub} \qquad (9)$$

where K_{int} is the value achieved by the crack as it reaches the interface, and K_{sub} is the critical stress intensity factor for fast (> 200 m sec^{-1}) crack growth in the substrate. Fig. 12 shows how the interface K value calculated from Equation 8 increases with the thickness of the embrittled surface layer. The problem is to evaluate K_{sub}. Rolland and Broutman [9] dealt with relatively thick samples of polymethylmethacrylate and unmodified polycarbonate for which the values of K_{Ic}, the plane strain fracture toughness, are well defined. The problem with ABS is that for the thicknesses used in helmets plane strain fracture is rarely found.

It is necessary to consider the validity of using a critical stress intensity factor as a criterion for whether a crack will grow rapidly in an ABS helmet. The evolution of fracture mechanics concepts has been discussed at length [12], so we can debate which version of fracture mechanics applies to ABS. In the original linear elastic fracture mechanics the criterion

$$K \geqslant K_c \qquad (10)$$

where K is the stress intensity factor of a loaded crack and K_c the fracture toughness of the material, is proposed. If the criterion is met the crack grows at an undefined speed. There have been three major modifications of this criterion:

(i) K_c is a function of the crack velocity V. The justification is that the crack is growing into a small yielded zone, and that the stresses in this zone depend on the time taken for the crack to traverse that zone. K_c in Equation 10 is replaced by a function $K_c(V)$. If the crack growth mechanism changes in type at different regimes of crack speed, then $K_c(V)$ can be a discontinuous function. This certainly seems to be the case for ABS where the fracture surface changes markedly between low- and high-speed fracture (Figs 2 and 7).

(ii) The thickness of the test specimen affects the K_c value. This has been rationalized by saying that K_c decreases from a limit K_{c1}, at low thicknesses to a limit K_{c2} at high thicknesses, the values being the plane stress and plane strain fracture toughnesses, respectively. ABS differs from many plastics in that there is no noticeable thickness reduction near the fracture surface in plane-stress fracture. This is because the multiple crazing in the whitened layer relieves the through-thickness stresses. Here we have been testing ABS of roughly 4 mm thickness because this thickness is used in helmets. The caution therefore is that the K_c values may drop if thicker ABS is tested, and this was certainly found to be the case at $-20°$ C.

(iii) The length of the crack affects the K_c value. This is the most serious criticism, since if K_c values are to be used for design purposes they should apply regardless of the crack length. Some workers have moved on to using measures such as the J integral [13] which can cope with certain cases of large-scale yielding. However, from the practical viewpoint a helmet shell will only fracture into pieces if cracks grow more than 100 mm. It does not matter if the initial crack growth is from a brittle outer surface towards an inner surface 4 mm away, because the crack velocity vector must then turn and run parallel to the shell surface. Examination of ABS shells that have broken in accidents showed that the fracture surface appearance became constant after a distance of about 10 mm. Therefore it is realistic to measure a K_{Ic} value in the laboratory for a similar distance of crack growth.

Finally we come to the likelihood of a crack travelling at speeds exceeding 200 m sec^{-1} in an ABS helmet. The temperature of the shell is of critical importance, because we have seen that in the rubber-toughened polycarbonate ML 3300 there is a transition to brittle behaviour at about $-15°$ C. Previous research on the fracture of ABS has suggested widely differing transition temperatures depending on the nature of the test (and probably on the ABS tested). Newman and Williams [14] performed slow tests on a Borg Warner ABS, in which the ends of a sharply edge-notched specimen, 5 or 13 mm thick, were extended at 2 mm min^{-1}, they found a transition in the K_c value for crack initiation at between -100 and $-80°$ C. In contrast, when the same authors [15] performed Charpy impact tests with a "dead-sharp notch" on extruded ABS that was 4.5, 6.3 or 16 mm thick they found brittle behaviour at −60° C and below, an intermediate behaviour at −40 to 0° C, and ductile behaviour above 0° C. In Williams' book [16] the transition region for Charpy impacts is shown as between −20 and 0° C. The different features of our experiments are that (i) we tested on ABS that is probably tougher than any tested by Williams, (ii) our ABS was thinner at 4 mm, (iii) a fatigue crack is sharper than one cut by a steel tool, (iv) the loading time is considerably shorter in an impact on a compact tension specimen (0.8 m sec) than it is for a Charpy test (~ 10 m sec). Given all these differences the measured transition temperature for 6 mm thick Ronfalin MST 42 of −15° C is compatible with Charpy test results in the literature.

The experiments and analysis of the way in which a helmet warms up after conditioning at −20° C lead us to treat with caution the interpretation of "−20° C" tests. As such temperatures are rarely met outdoors in the UK then the fact that the shell outer surface is at −12° C when impacted is probably irrelevant. However, if tests are intended to simulate performance in colder countries, the conditioning temperature must be lower or the interval between removal from the freezer and impacting must be reduced.

6. Conclusions

1. Outdoor weathering of Ronfalin MST 42 helmets for 3 to 5 years produces surface microcracking. Subsequent impact at −20° C produces in blue-pigmented helmets numerous small surface cracks that do not propagate, and the helmets pass the impact test.

2. Artificial ageing with UVB radiation requires a humidity or temperature cycle if surface microcracking is to be produced. Otherwise the impact outer layer will produce deeper cracks in an impact.

3. Fast crack growth at speeds exceeding 200 m sec^{-1} can be induced in 4 mm thick toughened polycarbonate or 6 mm thick ABS at −20° C, using small compact tension specimens. The test method is suitable for screening plastics for their propensity for high-speed fracture.

4. As the fracture behaviour of certain plastics changes from brittle to tough at about −15° C, delays in impact testing after conditioning at −20° C may give a falsely optimistic impression of the low-temperature impact behaviour.

Acknowledgements

Previous final-year undergraduate projects on aspects of the ageing of ABS, by J. Warlow (1984) and S. C. Aldridge (1985), were of considerable assistance in planning this research.

References

1. M. G. WYZGOSKI and C. H. M. JACQUES, *Polym. Eng. Sci.* **17** (1977) 854.
2. G. SCOTT and M. TAHAN, *Eur. Polym. J.* **13** (1977) 981.
3. A. DAVIS and D. SIMS, "Weathering of Polymers" (Applied Science, London, 1983) p. 103.
4. S. WOJCIECKOWSKI, *Cinque Million Consommateurs* (Paris, February 1984) 31.
5. G. NOEL, 'Study of the natural and artificial aging of protective helmet', *Ann. Inst. Tech. Batiments et Trav. Pub.* No. 139 (September 1979) 93.
6. A. KING and N. J. MILLS, in Proceedings of Plastics and Rubber Institute Conference on Impact Testing, Paper 9 (PRI, Guildford, September 1985).
7. A. GALE and N. J. MILLS, *Plast. Rubber. Proc. Applic.* **5** (1985) 101.
8. H. VALLEÉ *et al.* Proceedings of IRCOBI Conference, Salon de Provence, 1981, (IRCOBI, 69500 Bron, France) p. 176.
9. L. ROLLAND and L. J. BROUTMAN, *Polym. Eng. Sci.* **25** (1985) 207.
10. N. J. MILLS, *J. Mater. Sci.* **16** (1981) 1317.
11. A. BLAGA and R. S. YAMASAKI, *ibid.* **11** (1976) 1513.
12. J. G. WILLIAMS, "Fracture Mechanics of Polymers (Ellis Horwood, Chichester, 1984).
13. N. S. SRIDHARAN and L. J. BROUTMAN, *Polym. Eng. Sci.* **22** (1982) 760.
14. L. V. NEUMAN and J. G. WILLIAMS, *J. Mater. Sci.* **15** (1980) 773.
15. *Idem, Polym. Eng. Sci.* **18** (1978) 893.
16. J. G. WILLIAMS, "Fracture Mechanics of Polymers" (Ellis Horwood, Chichester, 1984) p. 268.

Received 21 July
and accepted 22 September 1986