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Heat-Pulse Effects on Glass

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Glass components of technological devices that are subject to heat pulses of short duration may suffer damage. At the same time, pulsed heating of glass is a feature of some new processing techniques. An experimental investigation of some of the relevant effects produced in three common glasses of widely different properties is described. Heat pulses were generated in a gas discharge tube under conditions of controlled ambient temperature in the range 20 to 600° C. The intensity of the pulses could be varied up to a level at which considerable evaporation of the glasses occurred. Cracking, weakening, strengthening, and phase separation effects are described and discussed in relation to the stress conditions, and to physical and chemical changes at the surface. Their significance with regard to practical applications is considered.

1. Introduction

Glass tubes, lenses, and windows are often employed in electrical apparatus, such as gas discharge tubes and lasers, which operate in a pulsed mode, and in which the glass components may be subjected to intense heat pulses causing physical damage. The general effects have been previously described and discussed by the present authors [1, 2].

Pulsed heating of glass and other materials may be used in fabrication or processing techniques which exploit the confinement of transient high temperatures to the surface layers of the material. An example, relevant in the present context, is that of electron beam milling (see, for example, the review by Brown and Nichols [3]), in which material is removed by evaporation under the high heat flux of a focused electron beam, or the similar techniques of laser beam milling [4].

As shown previously [1], a heat pulse will produce temperature and stress effects largely confined to a surface layer having a thickness of $n\sqrt{(Kt)}$, where K is the thermal diffusivity, t is the duration of the pulse, and n is a number in the range 1 to 5. If the heat pulse is sufficiently intense to raise the surface temperature to the point where viscous flow can relieve the transient compressive stress, a tensile stress will appear at the surface during cooling. Under the semi-infinite conditions relevant to the present technique, this stress will reach the magnitude:

$$\sigma = \frac{E\alpha\, \Delta T_\mathrm{V}}{1 - \nu} \qquad (1)$$

where: E is Young's modulus; α is the coefficient of thermal expansion; ν is Poission's ratio; and $\Delta T_\mathrm{V} = T_\mathrm{V} - T_\mathrm{a}$, T_V being the minimum temperature at which complete viscous relief of the compressive stress occurs, and T_a the ambient temperature.

This equation predicts that thermal stresses decrease with ambient temperature. The range of magnitude of the tensile stress for commonly used glasses may be obtained by taking ΔT_V to be of the order of 10^3. Then, for α lying in the range 0.5×10^{-6} to $10^{-5}/°$ C (fused silica to soda glass), the stress σ will lie in the range of about 10^4 to 10^5 lb/in.^{2} (1 lb/in.$^2 = 7 \times 10^{-4}$ kg/mm^2). Since practical glassware has its strength in the lower end of this range, the heat pulse may be expected, in general, to crack the surface of the glass; although at a sufficiently high ambient temperature cracking will cease. It was the aim of the present work to investigate experimentally the influence of the ambient temperature on surface cracking and related effects.

2. Apparatus and Technique

A scale diagram showing the main features of the apparatus is given in fig. 1. A fused-silica discharge tube of 1 cm bore, 30 cm long, passed between two graphite electrodes in a manner intended to prevent material particles from the electrodes entering the tube during a discharge. The air pressure in the tube was maintained at ~0.5 torr, and the 465 μF capacitor, charged to a voltage in the range 1 to 7 kV, discharged spontaneously through the tube, via a resistor of 1.3 Ω, when the switch was closed. The current reached an initial maximum in ~10 μsec, then decayed exponentially with a time-constant of 720 μsec. The voltage across the tube, initially of the order of 100 V, fell off during the discharge to about 12 V, such that the power dissipated in the tube decayed exponentially with a time-constant of 404 $\pm$ 20 μsec for at least three periods. The tube behaved electrically, in the current range 2000 to 5000 A, as a simple resistance of 0.31 Ω. Ambient temperature in the range 20 to 1000° C, generated by the centrally situated furnace, had no influence on these characteristics.

Small glass specimens of dimensions 11 × 5 × 2 mm, intended for subsequent examination, were exposed, as shown in fig. 1, resting on the tube at the middle of the furnace. Specimens for strength measurements consisted of 4.5 mm diameter rods of 10 cm length, supported axially within the discharge tube by fused-silica end-pieces and mounted symmetrically in the region occupied by the furnace. The strength was measured in three-point bend over a central, 3 cm length of the specimen, at a strain rate of 1.5×10^{-4}/sec. Three glasses, commonly used in laboratory work, were studied; they were soda glass, borosilicate glass of the Pyrex type, and fused silica (coefficients of expansion nominally 10^{-5}, 3×10^{-6}, and 5×10^{-7}/° C, respectively).

It was hoped that heat-pulse conditions at the specimen would be similar to those generated in the much shorter, 5 cm tube employed in previous work at an ambient of room temperature [1]. However, the thresholds for cracking on the three glasses indicate that, for a given capacitor voltage, surface peak temperature was only 60 to 70% of that achieved in the short tube, and that maximum intensity dissipated about 15 joule/cm^2 at the specimen. No calibrations, of the type previously made, were carried out, but, from the data obtained and with the assumption that the heat pulse at the specimen had the same waveform as the power dissipated

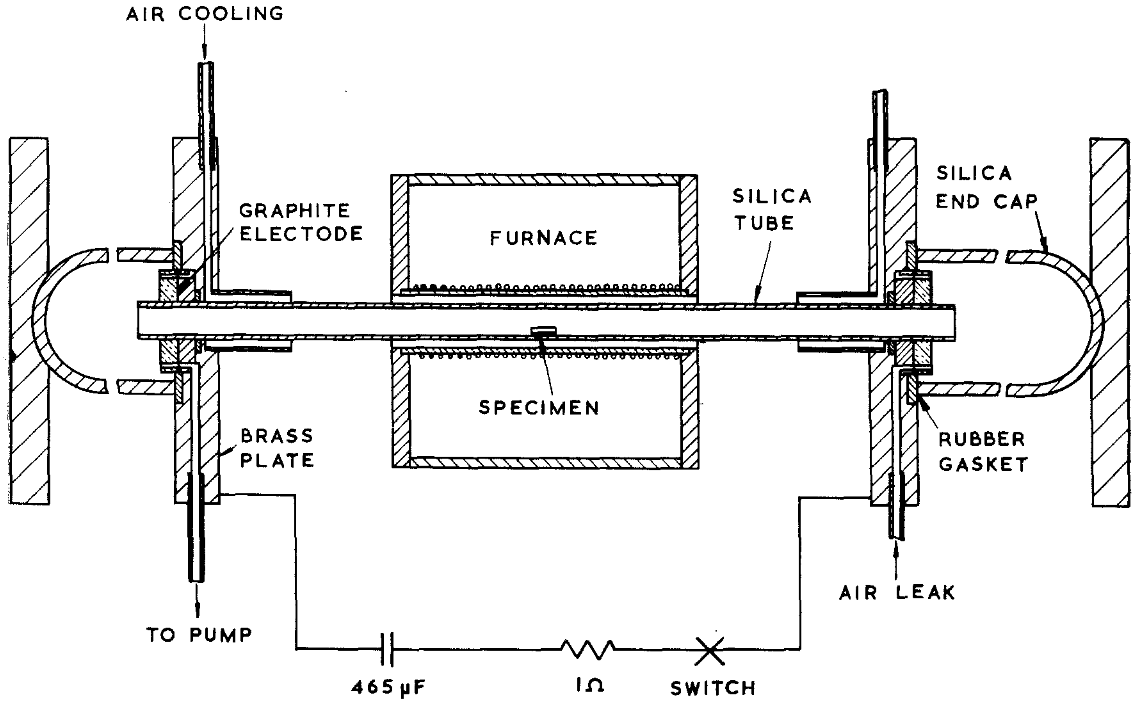
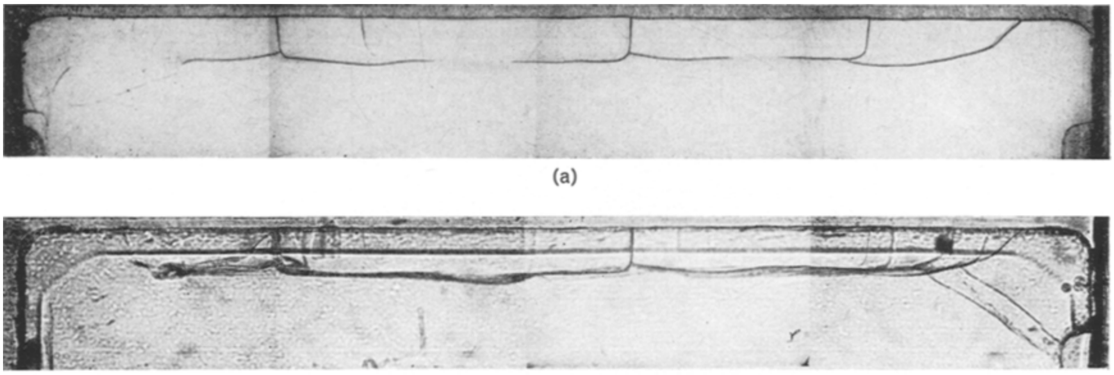


Figure 1 Scale diagram of the linear gas discharge apparatus.


(a)

(b)

Figure 2 Composite section of soda glass after exposure to a discharge at 7 kV, ambient temperature 350° C: (a) reflected light; (b) trasmitted light. (×28)

in the tube, it was estimated that peak surface temperature rises lay in the range 700 to 3500° C (if cooling by evaporation is neglected in this instance for simplicity) corresponding to the range of capacitor voltage from 2.5 to 7 kV. Owing to the similarity of the thermal constants, a given discharge was expected to produce a similar temperature distribution in depth and time in the three glasses, although evaporation would modify this at high heat-pulse intensity. Moreover, owing to the similarity of the power waveforms, the temporal distributions should be similar to those given by the short discharge tube. As described elsewhere [2], evaporation at 7 kV would produce high transient pressures of several atmospheres during the pulse.

3. Experimental

3.1. Single Exposures

A section through a soda-glass specimen after exposure is shown in fig. 2. This section illustrates the main features of the effects of a single discharge on glass. The form of the cracking, i.e. flaking, is typical for soda and borosilicate glasses. The section in transmitted light shows optical effects attributed previously [1] to a viscously deformed surface layer of glass, of high fictive temperature, and about half the flake thickness, and to a much thinner layer, close to the surface, in which evaporation has occurred. At high levels of discharge intensity, this latter surface layer in soda glass contained bubbles of ~30 μm diameter at a density of up to 3000/cm^2. Few, if any, bubbles appeared in borosilicate glass, and none in fused silica.

Just above the threshold discharge intensity which caused cracking, no flakes were formed; instead, a linked pattern of craze cracks appeared which penetrated into the glass approximately perpendicular to the surface. In fused silica, no flaking ever occurred; cracks, when produced, being of this type, i.e. perpendicular to the surface.

The general effect of ambient temperature is indicated by the diagrams of fig. 3. The diagrams were constructed from observations made on about a hundred specimens of each glass, each given a single exposure at a specific capacitor voltage and ambient temperature. A similar diagram for fused silica would contain only the regions A and B, i.e. perpendicular cracking only, in an area bounded approximately by lines at 6 kV and 400° C. A region unique to soda glass, region D, occurs inside the flaking region C and marks conditions under which exceptionally large flakes were formed, sometimes of the order of the specimen dimensions in size. These large flakes usually peeled off the specimen *in situ*. Plots of flake dimension as a function of capacitor voltage, given in fig. 4, also illustrate this effect. Flake dimensions for borosilicate glass were always larger than those

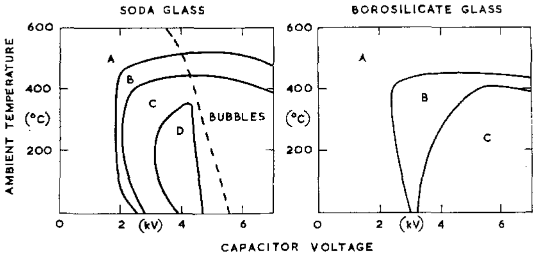


Figure 3 Diagrams indicating the cracking conditions in terms of ambient temperature and capacitor voltage: A, no cracking; B, perpendicular cracking; C, flaking; D, peeling.

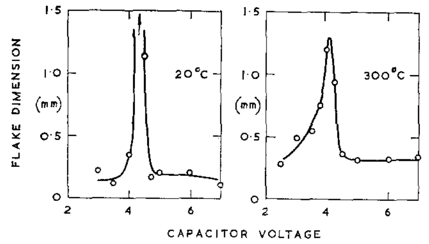


Figure 4 Flake dimension as a function of capacitor voltage for soda glass.

for soda glass under given conditions, with the exception of the conditions enclosed by region D. In general, again with the exception of region D, flake dimensions increased with ambient temperature. The variation of flake depth with ambient temperature is illustrated by the plots given in fig. 5; these diagrams also indicate the ambient-temperature dependence of the "viscous layer" depth. Crack and flake depths were always greater than the viscous layer depth except under peeling conditions, when the two were similar.

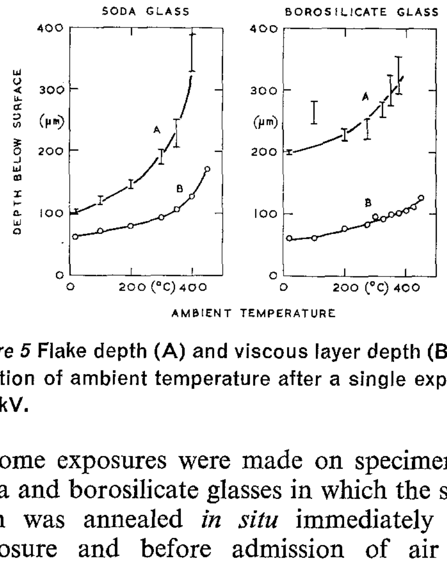


Figure 5 Flake depth (A) and viscous layer depth (B) as a function of ambient temperature after a single exposure at 7 kV.

Some exposures were made on specimens of soda and borosilicate glasses in which the specimen was annealed *in situ* immediately after exposure and before admission of air (the annealing conditions were 5 min at 550° C for soda glass and 5 min at 600° C for borosilicate glass). These indicated that, if crack diagrams of the type given in fig. 3 were to be drawn from data obtained in this way, the regions B and C would shrink and move to the right, i.e. the boundary in each case would move to higher capacitor voltage and lower ambient temperature.

In the case of borosilicate glass, this change was very marked, particularly as regards flaking, which only occurred at high voltage/low ambient temperature if annealing was given within a few minutes of the discharge. Thus the initiation of cracking and also of flaking after exposure was dependent on one or more of the three factors, time, temperature, and admission of air. Region D in soda glass, however, was not influenced by the heating schedule after exposure, thus indicating that the peeling phenomena occurred at low pressure soon after exposure.

The influence of time and admission of air on crack initiation and propagation was noted on silica and borosilicate glass specimens. Thus silica exposed near threshold required several hours contact with air for cracks to appear (but their initiation could be speeded up by scratching the surface). Such cracks travelled slowly across the surface sometimes taking several hours to propagate. Borosilicate glass, exposed near the threshold for flaking, acquired a pattern of perpendicular craze cracking on removal from the discharge tube which developed, during the next few minutes, into a flake system of the same pattern.

3.2. The Elastic Energy Dissipated to Form the Crack Pattern

A quantity, γ, the elastic energy released in forming unit area of new surface by cracking, may be determined from an estimate of the elastic energy and a measurement of the new surface area. As discussed previously [1], both systems of cracking, i.e. perpendicular cracking and flaking, release most of the elastic energy (*c.* 80%), although the exact proportion depends on the depth of cracking. If U_0 is the elastic energy introduced into the specimen per unit area of specimen surface, two approximate expressions for γ, containing crack dimensions, may be written:

$$\gamma_p \simeq 0.2\, U_0\, (s/d) \quad \text{(perpendicular cracking)}$$

$$\gamma_f \simeq \frac{0.4\, U_0}{1 + (2d/s)} \quad \text{(flaking)} \qquad (2)$$

where d is the depth of cracking below the surface (or flake thickness), and s is the mean flake dimension or distance between perpendicular cracks ($s = 2/l$ where l is the total length of crack intercepting unit area of the surface). The elastic energy U_0 is given by

$$U_0 = \frac{a\, E\, \alpha^2\, (\Delta T_v)^2}{(1 - \nu)} \qquad (3)$$

where a is the depth of the viscous flow layer. Values of $T_V = 750$ and 850° C were estimated for the soda and borosilicate glasses respectively [1]. Equation 3 assumes that cracking does not start until cooling is complete, i.e. until ambient temperature is reached again.

Measurements of s and d were made for both glasses under flaking conditions, i.e. in the regions C in fig. 3. The application of equation 2 was made to determine γ_f for these conditions. For borosilicate glass, values in the range 1×10^4 to 2×10^4 erg/cm^2 were determined. Similar values were obtained for soda glass on the left-hand side of region C, but on the right-hand side the values increased to about 4×10^4 erg/cm^2.

The experimental evidence tends to indicate that flaking is usually a relatively slow process following an initial, faster, perpendicular crack system. Since, as already mentioned in connexion with equation 2, most of the elastic energy can be released thereby, the residue available for the propagation of horizontal cracks to form the flakes is small. Thus the important release of elastic energy may usually occur by an initial propagation of perpendicular cracking, and it is therefore of interest to determine γ_p using the same measurements of s and d. The values so obtained lie in the range 5×10^4 to 10×10^4 erg/cm^2 for soda glass and in the range 5×10^4 to 7×10^4 erg/cm^2 for borosilicate glass.

Peeling, as occurs in region D, seems to be a noteworthy exception to this mode of crack propagation, since it is clear that the faster crack propagation is dominantly parallel to the surface. The values of γ in this region lie in the range 3×10^4 to 5×10^4 erg/cm^2. A possible requirement for peeling to occur may be the initiation of cracking during cooling, so that the initial perpendicular crack should barely penetrate into the compressive region beneath the tensile, viscous flow layer, but instead turn parallel to the surface. It is conceivable that heat pulses of such an intensity as to cause peeling may neither change the nature of the surface nor the presence and nature of crack-initiating flaws, as do more intense pulses, which, as discussed below, increase the resistance to crack initiation at the surface.

From the maximum ambient temperature for which peeling occurred (350° C), it is estimated from equation 1 that the tensile stress in the glass surface, after the heat pulse, would be, in the absence of cracking, about 5×10^4 lb/in.^{2} (1 lb/in.^{2} $= 7 \times 10^{-4}$ kg/mm^2). This might be regarded as defining a threshold for peeling, i.e. that a tensile stress approaching this magnitude may be required to propagate a surface microcrack while cooling is still in progress and the substrate still in compression. In borosilicate glass, the tensile stress cannot exceed about 4×10^4 lb/in.^{2} at an ambient of room temperature, and this may be the reason why peeling was not observed in this glass.

3.3. The Influence of the Heat Pulse on Strength

In previous work [1], it was shown that perpendicular cracking reduced the strength of glass specimens in accordance with Griffith's theory. Flaking, however, tended to have less effect. With the present apparatus and technique of annealing immediately after the heat pulse, it was possible, by choosing suitable operating points on the diagrams of fig. 3, to introduce perpendicular cracks of depths up to 300 μm, giving strengths as low as 2000 lb/in^2.

On the other hand, operation in regions A at high voltage, followed by annealing, caused the room-temperature strength of the soda and borosilicate glasses to be increased above values obtainable by annealing alone. As discussed previously [1, 2], very intense pulses evaporate the surface layers, causing the nature of the surface to be changed both physically and chemically by the loss of the more volatile constituents of the glass. As is well known, the strength is sensitive to the physical and chemical condition of the surface (see, for example, Ernsberger [5]), and evaporation coupled with removal of the tensile stress is thought to be responsible for the increase in strength.

The as-received soda-glass rod gave bend-strength values lying in the range 2×10^4 to 2.9×10^4 lb/in.^{2}; annealing at 525° C *in vacuo* produced no significant change. Specimens exposed to a single discharge at 7 kV at ambient temperatures of 500 to 550° C gave strengths spreading over the wide range 2×10^4 to 7×10^4 lb/in^2. This variation in results was thought to be associated with non-uniform effects on the surface of the glass brought about by patchy contamination. The presence of visible quantities of organic contamination at lower ambient temperatures had been observed to minimise crack damage in the area covered, and it was important to clean specimens well just before exposure. The mechanism of this protective process no doubt depends on the reduction of heat transfer to the glass by the decomposing and evaporating contamination. To reduce the possibility of this process occurring, it was felt desirable to expose the specimen to more than one discharge, thus exploiting the surface cleaning action of the heat pulses.

Another factor which might have influenced the strength of exposed soda glass was the presence of the many small bubbles produced near the surface. Increasing the number of exposures tended to reduce the number of bubbles, and so in this respect multiple exposures would again be beneficial.

3.4. The Increase in Strength Produced by Three Discharges

Three consecutive exposures spaced at 1 min intervals were chosen rather arbitrarily for further investigation of the increased strength effect. An ambient temperature of 525° C was fixed for soda glass, but a number of tests was made in which the duration of the annealing after exposure and the cooling schedule before admitting air were varied. For example, six specimens gained strengths lying in the range 4.9×10^4 to 7.6×10^4 lb/in.^{2} by a treatment consisting of 12 min at 525° C, followed by three exposures at 7 kV, and then a further 10 min at 525° C, followed by cooling to 400° C (about 10 min), before admitting air and removing the hot specimen. Variations on this procedure produced no significant improvement, although strengths as high as 9.4×10^4 lb/in.^{2} were occasionally achieved.

A factor which was thought to have a bearing on the scatter in the strength measurements was lack of concentricity of the specimen with the discharge tube. It proved difficult experimentally to ensure concentric geometry and, in many cases, as indicated by crack patterns on strength specimens exposed excentrically at low ambient temperature, differences in the heat-pulse intensity on diametrically opposite sides would occur.

A number of tests was made to investigate the durability of the strength increase. A standardised strengthening treatment, as described in the example above, was employed to prepare specimens which were then subjected to the tests. Abrading with emery paper to produce scratches to depths of the order of 10 μm brought down the strength into the as-received range. Handling and storing under atmospheric conditions for periods up to three days caused no reduction in strength; immersion in water for three days, however, brought the strength down into the initial range.

A series of similar experiments was conducted with borosilicate glass. A successful strengthening treatment consisted of 10 min at 600° C, followed by three exposures as before, then a further 10 min at 600° C, followed by cooling to 250° C (about 20 min), before admitting air and removing the specimen. Six specimens so-treated developed strengths in the range 3.5×10^4 to 9.5×10^4 lb/in.^{2}, whereas as-received specimens and control specimens, undergoing the same thermal treatment without discharges, gave strengths in the range 1.4×10^4 to 2.7×10^4 lb/in^2. Neither handling nor prolonged immersion in water caused any significant reduction in strength. No abrasion tests were carried out.

In all, a total of ninety-one specimens, largely soda glass, were broken to obtain these results.

3.5. Surface Effects in Multiple Discharges

Changes in the surface of soda glass were visible under the microscope after a single exposure at high voltage. In addition to the formation of bubbles, an undulatory structure developed. After three exposures, the surface had the appearance reproduced in fig. 6, which shows the surface of strengthened soda glass. As the number of discharges increased, the undulatory structure coarsened. After a hundred discharges, relatively large mounds of nearly 1 mm diameter, of different composition to the substrate, were

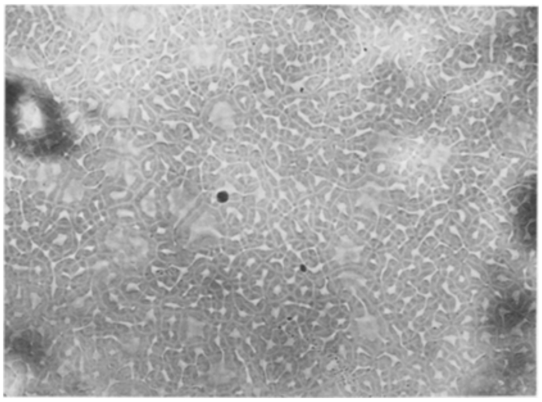


Figure 6 The surface of strengthened soda glass (three exposures at 7 kV, 525° C); transmitted light. (×50)

formed. A section through one of these is shown in fig. 7. The material of the mound had a refractive index of 1.620 compared with 1.518 for the substrate soda glass.

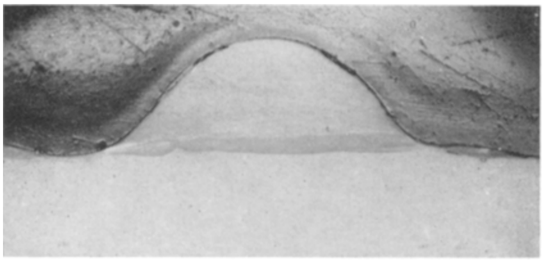


Figure 7 Soda-glass section showing second-phase surface projection after one hundred exposures at 7 kV, 525° C; reflected light. (×50)

Similar, but less prominent, changes occurred on the surface of borosilicate glass, although few, if any, bubbles appeared. Bubbles were only observed on borosilicate glass if exposed under conditions which caused cracking to take place; the bubbles were then formed in the vicinity of the cracks when partly healed by later exposures. The surface of borosilicate glass after many exposures also developed a second-phase structure, an example of which is given in fig. 8. Unlike soda glass, the second-phase regions did not project appreciably beyond the surface but penetrated with lens-like shape into the substrate to a depth of $\sim 10\ \mu m$. The refractive index of these regions was 1.579 compared with 1.474 for the substrate borosilicate glass.

It is considered that evaporation and condensation processes must be responsible for these various effects; none of them occurred on fused silica, which presented a featureless surface independent of the number of discharges.

The nature of the cracking was such that the single-exposure diagrams of fig. 3 were roughly still applicable after a hundred discharges, except that detached areas of bubbled and melted material formed on soda glass at low ambient temperature and medium voltage. The cracking behaviour of fused silica did not appear to change with increasing exposures.

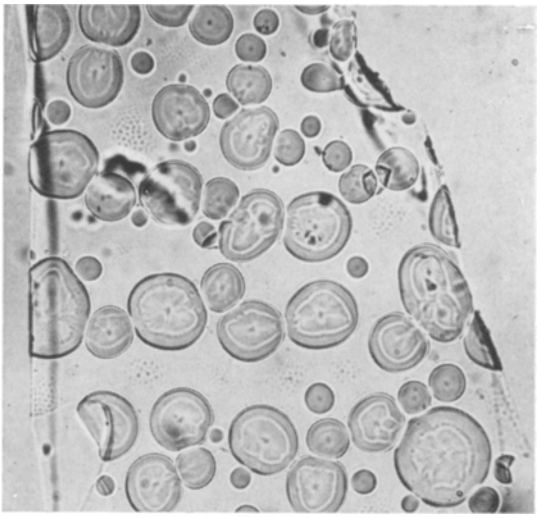


Figure 8 Surface fragment of borosilicate glass (one hundred exposures at 7 kV, 400° C) mounted in liquid of refractive index 1.474 showing second-phase regions; transmitted light. (×130)

4. Discussion

To sum up: a heat pulse of the type described causes cracking in the surface of glass, the nature of which depends on ambient temperature, on heat pulse intensity, and on coefficient of expansion. At temperatures above about 500° C, no cracking is induced in any glass, either because stress relief by annealing occurs (soda glass), or because the induced stress is too small (fused silica and probably borosilicate glass). Under many conditions of heat-pulse intensity and ambient temperature, an initial crack pattern appears to release most of the induced elastic energy with an expenditure of 10^4 to 10^5 erg/cm^2 of new surface. This initial crack pattern appears to consist of a network of cracks perpendicular to the glass surface, and this is frequently followed by relatively slow cracking, parallel to the surface, to form a flake system; this second stage dissipates much less energy per unit area of new surface.

A notable exception to this behaviour occurs in soda glass (i.e. at high coefficient of expansion) under a limited range of heat-pulse intensity and ambient temperature conditions, when peeling of the surface takes place. This is thought to occur when a surface flaw, present initially, is propagated by high tensile stresses during cooling of the heated surface layers. In this case, the moving perpendicular crack enters the compressive region in the substrate and is immediately turned parallel to the surface, to travel near the boundary of the tensile and compressive layers. Upper limits to the range of operating conditions for this process are set by:

(i) a high ambient temperature that gives too small an induced tensile stress to propagate a surface flaw in the time available before cooling is complete (i.e. when the substrate compressive stress is negligible); and (ii) a heat pulse sufficiently intense to modify the physical nature of the surface (and hence flaws) by evaporation. Moreover, peeling does not occur if the coefficient of expansion is too small to generate the high stress required to propagate a surface flaw during the cooling stage of the surface temperature transient, nor again if the heat-pulse duration is too small for sufficient elastic energy to be induced to drive a fast crack parallel to the surface.

Where cracking occurs in two stages, consisting of an initial, fast stage (perpendicular cracking) followed by a later, slow stage (flaking), then a quickly applied annealing treatment can preserve the perpendicular system, and low strengths are obtained. Near the borderline for conditions giving cracking, crack initiation and propagation may be delayed and slow, as observed in fused silica, and may be sensitive to contact with the atmosphere.

Thus the general behaviour of the observed cracking is understandable, qualitatively at least, in terms of the known fracture characteristics of glass as discussed by Shand [6].

The novel aspect of pulsed heating, however, is the change in physical and chemical structure of the surface which is brought about by evaporation. The possibility of increasing the strength of glass by changing the chemical composition of the surface in this way was at first verified in tests on soda glass. Here, the possibility of achieving a surface layer of increased silica content of low coefficient of expansion offered the prospect of achieving a surface compressive stress of magnitude in the range 0 to 60 000 lb/in.^{2}, depending on the extent of the differential evaporation. The experimental realisation of increased strength of the appropriate order at first seemed to support this view, but subsequent experiments with borosilicate glass seemed to indicate that the physical change in surface structure, brought about by evaporation, may be the more important strengthening process. Owing to its low coefficient of expansion, the observed strength increases in borosilicate glass cannot be entirely attributed to the presence of a compressive surface layer, and another mechanism, viz. modification of stress-raising surface flaws by evaporation, must be invoked. It is to be noted that this mechanism has already been invoked to account for an upper limit on peeling effects in soda glass.

It would appear, from the data obtained and the interpretation given here, that it should be possible to operate glass discharge tubes under pulsed conditions without failure, provided (i) the ambient temperature is high enough to avoid cracking (*c.* 500° C) and (ii) high stresses of non-thermal origin are avoided. If high stresses (e.g. arising from high transient pressure caused by evaporation) exist only during the middle of the pulse when the inner surface of the glass is in a viscous condition, then tube failure will be governed only by the stress which the outer surface can support without fracture.

The phase separation effects which develop with increasing number of exposures do not seem to invalidate these conclusions. However, should the glass be cooled from its ambient temperature of about 500° C, stresses would appear which might cause cracking; these would arise from the differing coefficients of expansion of the phases. These phase separation effects, thought to occur by evaporation mechanisms, are not understood.

In relation to machining operations on glass [1, 2], the present work gives some insight into the nature of surface damage; again the parameters ambient temperature, coefficient of expansion, heat-pulse duration, etc. may be chosen to minimise damage.

Pulsed heating of glass might be applied in other fabrication processes to gain advantages over processes involving steady or continuous heating. Examples which might be practical are the glazing of ceramics and enamelling of metals. Techniques of this type would have the advantage that the process might be completed without too much heating of the substrate material in bulk. Instead of having a single coefficient of expansion, as in the case of the glasses studied here, the situation would involve two different coefficients, that of the substrate and that of the coating; their difference would seem to be the important parameter governing the process. By controlling the conditions of ambient temperature and pulse length, in a manner based on the experience described here for glass alone, successful composite structures might be achieved.

5. Summary

Three glasses, soda glass, borosilicate glass, and

fused silica, covering a wide range in properties and coefficients of thermal expansion, were subjected to heat pulses generated in a gas discharge tube under conditions of controlled ambient temperature in the range 20 to 600° C. The power in the tube decayed exponentially with time-constant 400 μsec, and up to $\sim$15 joule/cm^2 were dissipated on the specimen surface. Thermal shock effects related to heat-pulse intensity and ambient temperature occurred in a surface layer of the order of 10^{-2} cm thickness. The elastic energy, released by the cracking, per unit area of new surface reached values in the range 10^4 to 10^5 erg/cm^2 when the cracks were propagated rapidly. Three modes of cracking were distinguished: (i) perpendicular cracking in which the crack system was perpendicular to the exposed surface; (ii) flaking in which an initial perpendicular system was supplemented by a later and slower process of cracking parallel to the surface; (iii) peeling in which large flakes formed and detached themselves before a perpendicular system could be established. The coefficient of expansion, the ambient temperature, and the heat-pulse duration and intensity were the important parameters influencing the cracking; annealing treatment following the heat pulse could preserve perpendicular cracking under some conditions to give strengths as low as 2000 lb/in.^{2} (1 lb/in.^{2} = 7×10^{-4} kg/mm^2). At ambient temperatures of about 500° C or more, no cracking occurred, and intense heat pulses which changed the physical and chemical nature of the surface by evaporation produced strengths of *c.* 70 000 lb/in^2.

These effects were interpreted in terms of the stress distribution previously predicted [1] and the known fracture characteristics of glass as summarised and discussed by Shand [6]. Their relevance to gas discharge tube failure and to thermal milling and other heat-pulse processing techniques was outlined. In multiple intense exposures, phase separation effects occurred that were thought to depend on evaporation mechanisms.

Acknowledgements

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