ABRASION DAMAGE TO LIGHTGUIDE FIBER SURFACES

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ABSTRACT

The two general types of strength-reducing defects most commonly found in lightguide fibers are those due to chemical interactions of refractory particles with the hot glass and those due to mechanical damage resulting from indentation or scratching. In this paper we review some work on the mechanical damage in silica lightguide fibers and some of the approaches which have been taken to simplify the study of such damage. We are reminded that the indentation and scratching behavior of silica glass is not simple and that in all probability there are a variety of different types of flaws which may have different fatigue and aging behavior, and therefore result in different lifetimes. It is therefore suggested that either the actual proof test level flaws, or at least a wide variety of synthetic flaws, must be studied in order to develop sufficient background and experience to confidently predict fiber lifetimes.

INTRODUCTION

The suggestion that the high strength mode observed in silica fibers and silica lightguide fibers is essentially the intrinsic strength of this glass in a flaw-free state[1] is interesting and poses a number of scientific problems relating to the mechanisms of mechanical failure in such a material. Of more practical interest, yet still of much scientific interest, is the understanding of the rest of the strength distribution. In Figure 1 of our previous paper in this volume (Kurkjian, Critical Issues...), we illustrated qualitatively some of the history of the improvement of the long length strength distribution over the years. The interpretation of the highest strengths in terms of a Griffith or fracture mechanics model is somewhat unclear. The calculation of the length (c) of a sharp crack which has a strength of ~14 GPa (the inert strength of the ~5.5 GPa high strength mode) from :

$$\sigma = K_{IC}/Yc^{1/2}$$

gives a value 2 nm if K_{IC} (the fracture toughness)is taken as 0.8 MPam$^{1/2}$) and Y (a constant describing the crack shape) is taken as 1.24). While it is not unreasonable to think of a flaw in this material as a defective SiO_4 interstice[2], this is not expected to be 'sharp' and thus such a calculation is perhaps inappropriate. Another approach is to think of the surface as truly flaw-free and to suppose that some sort of crack must first be nucleated before fracture occurs. Equally as interesting however, is the question of the meaning of the apparently continuous strength (flaw) distribution. This extends from the narrow high strength mode to strength values which can easily be visualized as real cracks and can be observed in an SEM, say c ~ 0.25 to 1μm (σ ~ 700 - 1400 MPa). It is well known that the primary defects responsible for the lower strength mode are (1) refractory particles, and (2) mechanical damage. It is reasonable therefore, to suggest here that the strengths between, say 1.4 GPa and 5.5 GPa (at room temperature or ~ 3 and 14 GPa at liquid nitrogen temperature) are due not to extremely small sharp cracks, but are due to the

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residual stresses associated with the events responsible for the larger cracks. The paper in this volume entitled "Effect of Refractory Particles on the Strength of Optical Fibers" by D.J.Wissuchek deals with defects resulting from the chemical reaction of refractory particles with the surface of the hot glass fiber or preform. In the present paper we are concerned with different types of mechanical damage and the effect which they may have on fiber strength and lifetimes.

INDENTATION AND SCRATCHING

Occurrence of mechanical damage

There are several stages in the production and handling of lightguide fibers where mechanical damage can easily occur. While care is normally taken in the handling of preforms, their initial 'perfect' fire-polished surface may sustain some damage. Since the preform is then heated to ~2200°C, it might be expected that healing of any flaws in the preform would be complete, and for this reason it may be felt that damage incurred during the handling of the preform is not critical. However, the small amount of work which has been done in this area suggests that this is not necessarily true and that some remnants of these original preform flaws may remain on the final fiber. In fact, it has been shown that the earliest fibers which were drawn suffered from just such preform surface damage[3]. Such mechanical damage should be easily avoided. On the other hand, when fixturing fibers, i.e, for termination, splicing, etc., the protective polymer coating must be removed and in this process and/or in the subsequent handling of the bare fiber, surface damage is almost unavoidable. It has been shown that by careful chemical stripping with hot sulfuric acid, no damage is done to the fiber[4]. In the more normal mechanical stripping procedure, however, substantial damage is usually sustained. The characteristic damage[5] resulting from such a procedure is shown in Figure 1. Here again, the strength distribution due to this mechanical interaction is continuous, and it is unclear just how such minute sharp flaws could be produced by the scratching action of the stripping tool.

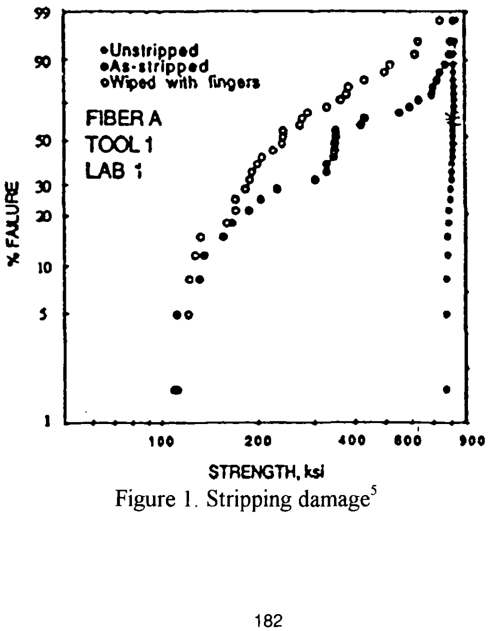

Figure 1. Stripping damage[5]

Indentations

Although indentation behavior in brittle materials has been studied extensively, the case of fused silica glass is somewhat special and is little understood. It was appreciated rather early that soda-lime silica ('normal'), (i.e., window and container glasses) and fused silica ('anomalous') glass responded differently to indentations. Arora et al[7] have summarized these differences in both cracking and flow behavior. Cook and Pharr[8] indicated that the 'normal' and 'anomalous' distinction was not so straight forward. They showed that radial cracks appeared on unloading the indenter, while in fused silica cone cracks were developed on loading. Dabbs et al[9] however, had earlier studied Vickers indents in high strength SLS rods and silica fibers. They found that in both glasses the primary cracks formed were radials. More recently, Kurkjian et al[10] and Lin, et al[11], by implication, showed that the development of cone cracks is a function of the 'surface condition' of the glass. At low temperature or humidity where slow crack growth is eliminated, or with as-drawn flat fiber, no cracks are formed with loads of 5 N or higher. At higher loads, the first cracks to develop are radial cracks

Pop-in

In a series of important studies starting with the work of Baikova, et al[12], it had been shown that low load indentations could be made without cracking. It was found that a discontinuity in strength vs. indentation load occurs when cracks appear; i.e., 'pop-in'. The load which causes this 'pop-in' depends both on time and environment.

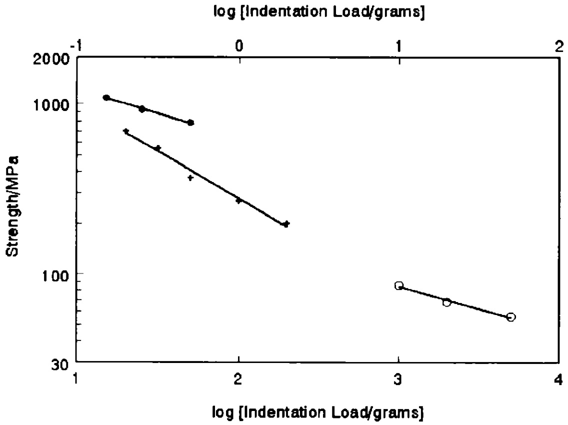

Fig. 2. Crack Pop-in. Circles are Vickers indents and correspond to the lower x-axis. Crosses are cube-corner indents and correspond to the upper x-axis.[13]

Figure 2 shows the results of such Vickers indentation and strength experiments. These were carried out in normal ambient, and the region in which there is a time dependence (the transition region) is not shown in detail. While there have been several attempts to produce a detailed model to explain this behavior, none has been completely successful. A more recent study sheds additional light on the problem[13], and is both interesting and of concern in predicting fiber lifetimes. This is illustrated by the crosses in Figure 2. These are the result of indentations with a cube corner rather than a Vickers indenter. This indenter was chosen to allow the development of cracks smaller than the 20 μm produced by a Vickers indenter. It can be seen that the loads necessary to produce a given strength are nearly two orders of magnitude lower with the cube corner indenter than with the Vickers. Cracks seem to have been produced in all of the tests with the cube corner indenter. The smallest of these is ~1 μm and results in a strength reduction to ~ 700 MPa. Such cracks are useful in the study of proof test level flaws since they are of the same size. However, the fact that strengths can be reduced to the normal proof test level by loading a sharp particle with a load of the order of grams is unexpected and the cause for special concern for lightguide processing and handling. In early studies with the cube corner indenter, Pharr et al[14] suggested that for a given contact area, the cube-corner indenter displaces three times more volume than does the Vickers indenter, and thus the cube-corner generates higher stresses for a given load. Whatever are the details of this pop-in behavior, it is clear that this effect might have an important influence on the apparent lifetime of damaged fibers. Subthreshold indentations may pop-in after proof-testing, and an unexpectedly large decrease in strength to substantially below the proof test level could result.

Residual stresses: glass composition, indents and scratches

As indicated earlier, the 'flow' under a Vickers indenter is different in different glasses. The resulting residual stresses (as characterized by the birefringence) are thus also expected to be different. While these residual stresses give rise to the development of cracks, they also contribute to the propagation of these cracks; that is, the residual stress contributes to the driving force on the crack system . When a tensile test is performed on a fiber containing an indentation- or scratch-induced crack, the total stress intensity factor is the sum of that due to the applied stress and the residual stress. The residual stress intensity factor (K_R) has been found to vary with the type of indentation involved. Thus for a 'point' indentation[5] (that typified by a Vickers indent), the residual stress intensity factor is proportional to $c^{3/2}$, while for a 'line' crack[15](such as that produced by a scratch), it is proportional to $c^{1/2}$. This leads to apparent fatigue parameters (n'):

$$n = 4/3n' - 2/3 \qquad \text{for point contact}$$

$$n = 2n' - 2 \qquad \text{for line contact}$$

Experimental studies[6,15] for these types of damage have been done on soda-lime-silica glass, and the results are summarized in Figure 3. Shown here are dynamic fatigue data for an as-indented Vickers indent and for a line scratch (dashed curve), plus data for an annealed Vickers indent. The corresponding values of n' or apparent n, are ~ 13, 10 and 17, respectively. These differences may at least partially explain the differences in the curves of Mould and Southwick[16] which were shown in Figure 5 in our first paper (Kurkjian, Critical Issues....). Since long fiber lengths are expected to have a variety of flaw types, a simple knowledge of the proof strength may be insufficient to define lifetimes. Another explanation for inconsistencies which are found in studies on abraded fiber, may be the interaction of flaws when their areal density is high[17].

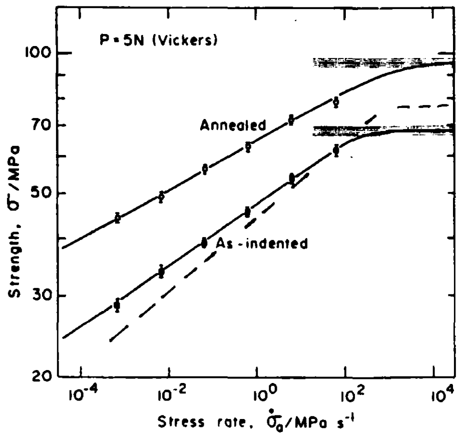

Figure 3. Differences in behavior of point[6] (circles) and line damage[15] (dashed curve).

SUMMARY

Mechanical damage to lightguide fibers can be developed at various stages in the production and use of these fibers. Damage severe enough to reduce the strength to the normal proof test levels (~700 MPa) can be the result of damage by a sharp, hard indenter with a very small loads of the order of grams or less. Thus it is very important that extreme care be taken in the handling of both fibers and preforms. Since mechanical damage can take a variety of forms which may then lead to different strength and fatigue characteristics, it is important that studies be carried out on defects having different origins. Thus a fundamental understanding of these flaws is needed for reliable lifetime predictions, or large quantities of individual fibers will need to be tested in order to develop statistical data in each case.

REFERENCES

1. C.R. Kurkjian and U.C. Paek, Appl. Phys. Lett. **42**, 251 (1983).
2. B. Proctor, Phys. Chem. Glasses. **31**, 78 (1990).
3. U.C. Paek and C.R. Kurkjian, unpublished work, 1983.
4. M.J. Matthewson. C.R. Kurkjian and J.R. Hamlin, J. Lightwave Tech., **15**, 490 (1997).
5. H.H. Yuce, P.L. Key and T.Wei, Proc. IWCS, 400 (1990).
6. D.B. Marshall and B.R.Lawn, J. Am. Ceram. Soc., **63**, 532 (1980)
7. A. Arora, D.B. Marshall, B.R. Lawn and M.V. Swain, J. Non-Cryst. Solids, **31**, 415 (1979).
8. R.F. Cook and G.M.Pharr, J. Am. Ceram. Soc., **73**, 787 (1990).
9. T.P.Dabbs, D.B. Marshall and B.R. Lawn, J. Am. Ceram. Soc, **68**, C224 (1980).
10. C.R. Kurkjian, G.K. Kammlott and M.M. Chaudhri, J. Am. Ceram. Soc. **78**, 737 (1995).
11. B. Lin and M.J. Matthewson, Philos. Magazine, **74**, p.1235 (1996).
12. L.G. Baikova, V.P Pukh and S.N. Talalakin, Sov. Phys., Solid State, **15**, 1437 (1974).
13. C.R. Kurkjian, O.S. Gebizlioglu and S.L. Semjonov, Proc. NFOEC, 63(1997).

14. G.M. Pharr, D.S. Harding and W.C.Oliver in Mechanical Properties and Deformation Behavior of Materials Having Ultra-fine Microstructures, ed. M. Nastasi, et. al., Kluwer Pub., the Netherlands, 449 (1993).
15. B.L Symonds, R.F.Cook and B.R.Lawn, J. Mat. Sci., **18**, 1306 (1983).
16. R.D.Mould and R.D.Southwick, J. Am. Ceram. Soc.**42**, 582 (1959).
17. M.V. Swain and J.T. Hagan, Eng. Fracture Mechanics, **10**, 299 (1978).