Transparent Glasses and Ceramics
ABSTRACT

A major challenge in achieving a physics-based computational capability for designing glass and ceramic armor is to develop accurate constitutive equations and fragment flow models that are usable in continuum codes. We describe a model that uses microdamage and fragment flow constitutive data, shows how the model links to continuum models, and compares computational results with glass penetration tests.

BACKGROUND AND PROGRAM GOALS

Traditional mechanical constitutive relations for glass targets undergoing microscopic damage need additional physics and scale design of transparent armor. The goal of the microdamage model is to relate material failure on the microstructural level to continuum behavior, and to give guidance to continuum models that are used in designers.

Generation of thin targets of both brittle and ductile materials occur by the formation of a region of yielded, flowing material at the penetration target interface. The flow of this material allows penetration to occur. For brittle materials like glass and ceramics, the yielded material is observed to consist of fine fragments in a thin region near the interface zone (1995).

Our goal is to construct a microdamage model that describes the microdamage evolution, i.e. the nucleation, growth, and coalescence of microcracks in the HIC, and subsequent granular flow of the damaged material out of the path of the advancing penetrator. Our microdamage model is empirical, and is based on observations and data from experiments designed to measure microdamage evolution and fragment flow.

In this paper we describe a proposal initial framework for a microdamage, incorporate available data, discuss preliminary conclusions of predictions with observations, and discuss future program opportunities.

MECHANICAL APPROACH

Mechanically based mechanical constitutive relations have been successfully developed during the past several decades to relate material failure in metals and composites to the underlying micromechanics processes, thereby helping to select appropriate continuum models and macroscale parameters. The key to this approach has been the development of experiments for determination of “strength and grain-size fragmentation” (MRA/RAO) tests in a relevant volume element (587) for the evolution of size distributions of microcracks and cracks and their coalescence to form fragments, as well as the subsequent movement of these fragments.

MRA/RAO experiments are designed to measure key properties. As summarized in the 2004 handbook by Eames, Hartman, and Forooz, glass presents several challenges, as follows.

1. Flow stress: Whereas most brittle materials contain internal flaws that can serve as microcracks, the flow stress of a high-strength glass has primarily random surface flaws. This means that in an uniaxial stress field, plate impact experiments, for example, the microcracks are not localized in a region adjacent to the impact surface.
2. High pressure: Glasses do not exhibit a distinct deviatoric stress limit in plate impact experiments, partly because of a contact damage contribution of the fragment at low pressures.
3. Interlayer structure: At high pressures, brittle glasses become ductile. The molecular structure of glass allows identification without heating at pressures exceeding 600 MPa. At such pressures, evidence of
yield may vanish, and the response may be difficult to distinguish between elastic and hydrostatic.

Prior experiments have emphasized measurement of very low and short period loading tests using load, displacement, and load or strain-controlled tests. These measurements include stress or strain as a function of time. The techniques used include thermal expansion, strain measurements, and load or strain-controlled tests. However, more recently, experiments specifically designed to yield microstructural resolution for the evolution of microscopic damage upon loading to the present state we focus on these types of experiments that can potentially provide such information.

1. Blast impact (impact strain) experiments. These experiments simulate the loading conditions under which the structure could experience the impact shock waves.
2. Fracture propagation of quasi-brittle materials. These experiments simulate the loading conditions near the onset of fracture at propagation rates of several hundred times.
3. Quasi-brittle material property tests, including compression-tension tests of prisms. These experiments provide basic properties of the material.

To guide planning and interpretation of the above experiments, we used a conceptual model (CMM), which will serve as an initial framework in the modeling as we obtain more micro-structural resolution data.

CONCEPTUAL MICROSTRUCTURAL MODEL

The CMM is based on modifications of the MACRO-BOND (MB) concept of quasi-brittle behavior in brittle materials. The micromechanical theory is assumed to be totally elastic, with elastic fractures affecting the failure of inter-granular interfaces, and is treated by analogy to multi-scale plasticity models based on multi-scale plasticity models for multi-scale plasticity models. The focus is on the movement of lines of failure between the fracture, called microcrack localization (MCL), on a finite number of slip planes. The fracture is inherently size-dependent because of finite crack propagation and growth rates and fragment breakage. Other factors, like stress-strain mechanics (SSM) in order to govern the microcrack propagation, and the “fracture toughness” is a property closely associated with the material’s toughness.

Interactions of microstructural entities suggest that the fracture is characterized by a needle-like array of discontinuous fractures in which there is initiation, growth, blockage of fragmented blocks, and growth rates of fragmented blocks. These features contribute to the microcrack propagation, and the “fracture toughness” is a property that easily changes, which also illustrates the expected importance of confining boundaries.

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![Confinement principally encourages shear-enhanced compaction but not bulking](image)

Fracture

MD: the hole extends into the plane of the Figure to form a dislocation-like object.

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[Figure 1. Schematic picture of confinement]
\[ \Psi_{BM} = \Psi_{BM}^{c} \Psi_{BM}^{f}(m_{v}^{1} r > m_{v}^{2}) \]  
where \( m \) is the maximum shear stress. \( \Psi_{BM}^{c} \) is the pressure, \( \Psi_{BM}^{f} \) is a critical fragment size, \( m_{v}^{1} r > m_{v}^{2} \) is the corresponding critical flow stress, \( r_{BM} \) is this plane strain fracture toughness, and the \( \Psi_{BM}^{c} \) allows local stress enhancement over the remote stresses. Figure 1 shows that we identify the fragment cross-sections as square. An equivalent circle with the same area would thus have a radius \( r_{BM} = \sqrt{\frac{4A}{\pi}} \).  

Eq. (1) is clearly an oversimplified relation to be improved as more data are obtained.\(^{11}\) For example, the dynamic friction on most surfaces and a combination of Mode II and Mode III toughnesses would be more appropriate than \( \Psi_{BM}^{c} \) which is used here as a simple measure of brittleness, and in shear mode.\(^{12}\) We expect that a more detailed model, such as that of Simmons and Work, may eventually be needed.\(^{13}\)  

As the applied stress increases, comminution breaks down the larger fragments until either sufficient embrittlement has occurred to allow flow of the remaining fragments, or comminution ceases because the flow zones are self-limited.\(^{14}\) Specifically, we fit our data to an initial Poisson's fragment size distribution

\[ P_{BM}(R) \propto \frac{1}{R^{4/3}} \]  
where \( P_{BM}(R) \) is the number of fragments per unit area of a cross-section with size greater than \( R \), \( N \) is the total number of fragments per unit area in a cross-section, and \( R_{BM} \) is the initial average size of the fragments.\(^{14}\) The fragment density function is

\[ dN/dR = \frac{N}{4/3 \pi R^{4/3}} \]  
We assume that the initial distribution has an upper limit, a largest fragment, \( R_{BM} \). We also assume that the initial size (POD) size distribution undergoes the fragment size distribution, with the larger holes being associated with the larger fragments, and the average hole size is equal to the average fragment size. Integrating the hole area \( \psi_{BM}^{h} \) with a density function \( \psi_{BM}^{h} \) from \( R = R_{B} \) to \( R_{BM} \), gives the total initial porosity

\[ \psi_{BM}^{h} = \int_{R_{B}}^{R_{BM}} \rho_{BM} r^{2} (1 - \psi_{BM}) \]  
where \( \rho_{BM} \) is the total number of holes per unit area of a cross-section, \( m_r \) is the fraction of \( R_{BM} \) that specifies the width of the holet, i.e., \( R_{BM} \) is the maximum size of the holet, and

\[ \psi_{BM} = \frac{R_{B}^{2} \rho_{BM}}{2 \pi (1 - \psi_{BM})} \]  
where \( R_{B} = 100 \rho_{BM} \).\(^{14}\)  

For a given \( R_{B} \), say \( R_{B} = R_{B_{max}} \), the available porosity is the total porosity minus the integral from 0 to \( R_{B} \), and the rate of embrittlement to total porosity is

\[ \psi_{BM}^{h} = \int_{R_{B}}^{R_{BM}} \rho_{BM} r^{2} (1 - \psi_{BM}) \]  
where \( \rho_{BM} = \rho_{BM}^{c} / 2 \pi \) and \( r = \rho_{BM}^{2/3} \).\(^{14}\)  

For example, Figure 2 shows that when \( m_{BM} = 100 \) and \( m_{BM} = 200 \), \( \psi_{BM}^{h} = 0.25 \). That is, when the largest fragment in the distribution has been reduced to \( m_{BM} = 200 \), 25% of the original fragments and associated holes have been impressed.\(^{14}\)  

Under the high confinement provided by the impact interface of a unit size plate fragment, we note that the mass of a and impact, the holes will tend to be driven into this containing boundary, resulting in comparison of the fragments can move into the RSDL, but not out. If the confinement is maintained, the subsequent response must be elastic. This postulated behavior is shown schematically in Figure 2.
The experimental results also describe the process by which macroscopic cracks and holes (voids) are produced in originally void-free glass under compression and shear. "Wing cracks" are a candidate for causing brittle fractures and delamination under compression and shear (see Figure 2). In general, a weak surface flaw can extend progressively as a Mode II shear crack, or form and grow to the crack plane and propagate at a wing (expanding) crack. In the latter case, microcracks (MCDs) are produced in the ligament, causing delamination. The extensive literature on this subject is reviewed in the 2004 book by Rappe, Bouamrane, and Loubatier. Wing cracks were observed in glass plates under compression in 1983 by Daus and Bouamrane. Subsequent work by Wannasirrattana and Hopia, Hopia and Wannasirrattana, Mose and Gupta, and Wannasirrattana and Guzel among others, described complex behavior under different stress states. MCDs¶ preferentially propagate with an edge impact technique on a number of brittle and fragile materials, and found that the impulse energy depended on whether there was a mechanism for stress softening (e.g., adhesive heating) sufficient to stabilize a propagating shear crack.

Thus, wing cracks appear to be a possible source for deriving microcracks (MCDs) in previously void-free glass in plane impact tests. However, the brittle/delicate transitions discussed above may suppress their formation, in which case we would expect a more progressive layer of damage to the impact target material. To resolve this and other issues, we need more data.

**SUMMARY OF EXPERIMENTAL RESULTS**

Uniaxial strain plate impact response:

We begin by examining the response of Soda-Lime Glass (SLG) samples loaded by uniaxial strain impacts in experiments reported by Stothers and Gupta, and Alexander et al. We assume that
cracks (using modes or shear cracks), originate from a size distribution of flaws of size \( f \) on the impacted surface. We also assume that the distribution is of the form \( \mu g(\mu) \). To relate the flaw size \( f \) to the fragility \( q \), we draw on the "fracture" concept of the HUBERT model for brittle materials, which defines a parameter \( s0 \). For very brittle materials, \( s0 \) can be between 10 and 20. The HUBERT model assumes that each fragment contains a flaw of size \( f = 10 s0 \). Thus, \( W_0 = s0 \).

For a ductile slip plane, Eq. (4.1) with \( \phi \) set equal to the minimum friction, describes the Fig. 2 yield function in \( \mu, \eta \) space. When an elastic bond path crosses the surface, the model produces a basis of non-linear elastic or the plasticity referred to the impact surface. The shear stress drop at constant pressure to the "yield" shear \( \eta_{\text{y}} \).

To describe the non-linear flow of the "yielded" material as a measure that ensures stability and uniqueness, an analysis due to Whitman and Bue (1970) is applied, and a "stress-relaxing yield" relation used to describe the total strain rate on a slip plane as the sum of the plastic strain rate and the non-elastic strain rate:

\[
\frac{\partial \sigma_{\text{pl}}}{\partial t} = [42(\frac{\mu}{\eta_{\text{y}}})] \frac{1}{\rho} \frac{\partial \eta_{\text{y}}}{\partial t} + \frac{\partial \sigma_{\text{pl}}}{\partial t}
\]

where the non-elastic strain rate is given by

\[
\frac{\partial \eta_{\text{pl}}}{\partial t} = \frac{1}{\rho} \frac{\partial \sigma}{\partial \mu} \left[ \frac{\partial \mu}{\partial \mu} \right]^{\frac{1}{2}} \left[ \frac{\partial \eta_{\text{y}}}{\partial \mu} \right]^{\frac{1}{2}} \frac{1}{\mu} \frac{\partial \mu}{\partial t}
\]

and where \( \sigma \) the shear modulus, \( \rho \) is the density function, \( \eta \) is the maximum shear stress (\( \eta_{\text{y}} \)), \( \mu \) the intermediate friction coefficient, \( \mu \) is the plastic tip (plastic) penalty associated with the \( \omega \) plane, \( \eta \) is the average shear (\( \eta_{\text{y}} \)) height equal to the average fracture shear, and \( \psi \) is a dimensionless parameter that characterizes the average BHT width. The quantity \( \phi = \psi \eta_{\text{y}} \) can be considered a new parameter that specifies, for a given stress state on a given slip plane, both the penalty applied by the BHT and the BHT width. \( \eta_{\text{y}} \) has significant leverage on the predicted behavior, and it will vary in this parameter study.

In Table 1, we list measured and assumed material properties for the Sandstone 1 and 2, and compare them with those for BHC, a ceramic for which a clear two-phase response has been recognized. The glass and BHC data were previously discussed, since both sets of experiments were well-documented, and give an opportunity to study the effects of different morphologies, fracture toughness, and toughness.

**Table 1. Properties for Sandstone 1 and 2**

<table>
<thead>
<tr>
<th>Property</th>
<th>Sandstone 1</th>
<th>Sandstone 2</th>
</tr>
</thead>
<tbody>
<tr>
<td>Density (g/cc)</td>
<td>2.64</td>
<td>2.61</td>
</tr>
<tr>
<td>( C_0 ) (kN/m)</td>
<td>5.76</td>
<td>5.09</td>
</tr>
<tr>
<td>( C_1 ) (kN/m)</td>
<td>3.41</td>
<td>2.67</td>
</tr>
<tr>
<td>( C_{\text{eff}} ) (kN/m)</td>
<td>4.19</td>
<td>3.68</td>
</tr>
<tr>
<td>( m ) (Monomial index)</td>
<td>2.34</td>
<td>2.67</td>
</tr>
<tr>
<td>( \omega ) (fracture parameter)</td>
<td>0.17 ( (\phi) = 0.05 )</td>
<td>0.17 ( (\phi) = 0.05 )</td>
</tr>
<tr>
<td>E</td>
<td>17.1 ( (\phi) = 0.05 )</td>
<td>17.1 ( (\phi) = 0.05 )</td>
</tr>
<tr>
<td>( H_{\text{R}} ) (GPa)</td>
<td>29.0 ( (\phi) = 0.05 )</td>
<td>29.0 ( (\phi) = 0.05 )</td>
</tr>
<tr>
<td>Fracture toughness</td>
<td>3.8</td>
<td>3.8</td>
</tr>
</tbody>
</table>

**BHC**

<table>
<thead>
<tr>
<th>Property</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Surface</td>
<td>16.15</td>
</tr>
</tbody>
</table>

Assumed material properties: \( \mu, \phi, \) and \( \eta_{\text{y}} = \phi \theta \), \( \eta_{\text{y}} = \phi \theta \), \( \phi = 0.3 \), \( \mu = 0.1 \), \( \phi = 0.05 \), \( \theta = 0.05 \), and \( \phi = 0.1 \).
The reported ambient pressure elastic moduli measured by ultrasound are inconsistent with the reported values and have not been considered. The measured values are given in the parentheses.

We chose three measurements of the ambient pressure disturbances in the plane, but we will show an example analysis assuming the ambient mechanical properties listed in Table 1, including setting $\beta = 1$ and $\beta = 0.5$. The measured values are given in the parentheses.

To construct the relations to those for the longitudinal stress, $\sigma$, we use the ambient strain conditions:

\[ \sigma = \frac{\rho \cdot \omega}{(1 - \nu)^2} \cdot \left( (1 - 2\nu) \cdot (\sigma_1 - \sigma_2) + \nu \cdot (\sigma_1 + \sigma_2) \right) \]

where $\rho$ is the longitudinal wave speed, $\omega$ is the angular frequency, and $\nu$ is Poisson's ratio.

Operating on Eq (6) with $\Delta \sigma$, combining with the equations for constraint of mass and momentum, and using the above ambient strain relations connecting $\sigma$ and $\nu$, via Poisson's ratio $\nu$, leads to

\[ \Delta \sigma = \frac{(\rho \cdot \omega)^2}{A} \cdot \left( (1 - \nu)^2 \right) \cdot \left( (1 - 2\nu) \cdot (\sigma_1 - \sigma_2) + \nu \cdot (\sigma_1 + \sigma_2) \right) \]

where $A$ is the cross-sectional area.

Using the phase values of Table 1 yields values of $\Delta \sigma$ ranging from 100 to 1000, depending on the choice of $A$.

Eqs. (8) and (9) fulfill a stability criterion due to Whitman [26], $\Delta \sigma > 0$, the same as in Eq (2) for $\sigma(\omega)$ in well-pressed and stable (stabilized). However, the true value of the strain is not known, so we will approximate the response by setting $\beta = 0.5$ equal to the 50% value.

For the above measurements, a 39.3% strain performed by Simmons and Gupta and Alexander et al. in which 59.5% longer than the measured elastic impact stress of 4 to 6 GPa in the impact surface. Both measured bending stress of 4 to 6 GPa of about 0.2% for the largest stresses of 8 to 16 GPa. Simmons and Gupta measured a 1 mm strain in the impact region. For example, at 4 GPa, they measured a slightly curved longitudinal wave phase of 8 GPa, followed by a peak stress of about 4 GPa. This peak stress is approximately equal to the stress of about 1.5 GPa, but with about 0.3% drop to about 1 GPa, only to rise again at about 2.5 GPa to a new plateau of about 1.5 GPa at about 3 GPa. This observed two wave structure was interpreted by Simmons and Gupta as a two dependent area of stress followed by a permanent increase of the wave structure, consistent with the CEBP model. In contrast, Alexander's experiment of this kind revealed no wave structure.

Simmons and Gupta developed an all-pass elastic moduli strength model that correlated well with the above experiments, but similar features as the CEBP model, and can therefore serve as a test of the CEBP model's validity.

This is, in summary, the CEBP model's predictions with the above data, we set $\beta = 1$, and observe the results in Table 1 to ensure a value of 4 GPa for the DBL. The values in Table 1 for $\sigma(\omega)$ given a material strain loading path of $\sigma = 0.08$GPa and Eq (1) gives a yield curve with a radius of $0.3$. 

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C/2, corresponding to $E (\text{103}) = 1 \text{ GPa}$, as desired. The Chent model calculates a difference from pseudopressure obtained by analogy to heat flow calculations, where calculations of heat flow from a hot slab, maintained at constant temperature, suddenly plunged in contact with a cold material, showed that the pseudopressure of propagation of one-fourth the hot slab temperature, was $3.8\sigma$, where $\sigma$ is the diffusivity. Taking the diffusivity to be equal to $6.7\times 10^{-6}$, and setting the pseudopressure equal to the diffusion field yields

$$H (\text{Diffusion}) = 1.7 C T (\text{m})$$

(3.1)

So far, the discussion has concerned individual slip planes, but a further consequence of the postulated "reinforcing yield condition" is that many slip planes would become active simultaneously. In Figure 1, for example, the vertical slip planes would start to slide. The fragment then accelerates, schematically shown as sequence, would become "wheels", and the material would become more like a liquid. The effective coefficient of friction might decrease as the particles begin to roll, but if the confinement is maintained, once the $6.8\sigma$, have flowed into the constraining fractures interfaces to compact the material, the subsequent relaxing from the reference state $E (\text{103}) = 1.4 \text{ GPa}$ ($\sigma (\text{m}) = 0.3$) would be that of an elastic liquid, and the approach wave speed would be the bulk wave speed.

Since the Chent model has many adjustable parameters, a wide variety of responses can be predicted. We performed a preliminary parameter study by varying the value of $6.8\sigma$, over the range shown in Table 1. Figure 4 shows the two extremes. Figure 4 shows the minimum and maximum temperature errors at $4.6 \text{ GPa}$. The heat output prediction is obtained with the large-temperature $T (\text{500} \text{ K})$, which forced the deficiency factor wave to travel at almost the elastic wave speed. The slower bulk wave speed for the elastic liquid delayed the arrival of the reflection wave to about 1.5 $\mu$s, in rough agreement with the Shinoh and Gupta results for the longitudinal stress. Moreover, the observed regain of strength is not predicted by the Chent model (although it is possible to imagine that the time particles compact and therefore become an effective solid again).

By choosing the small value of $T (\text{30} \text{ K})$, we can delay the following plate to agree with the experiment. Figure 4 shows that the delay simply reduces the first wave in less amplitude, resulting in a good correlation.

![Figure 4. Comparison of Chent model trends with simplified Shinoh & Gupta data. Measurement is from impactsurface.](image-url)
At higher impact stresses, the data suggest that the glass becomes plastic, and undergoes densification at a plane of failure, and no longer behaves like consideration of elastic strains. Structure models predict that the material will be somewhat elastic at the impact interface, and related collectively, we expect that both these effects and the overall behavior should remain elastic, as demonstrated schematically in Figure 2. Elastic unloading was at a rate determined as fast as 1000 m/s, at least 10^7 m/s.

In summary, these are enough adjustable parameters in the model to allow enough small changes in the plate impact velocities of the plate impact simulations, but we need additional stress-strain data to further constrain our model parameters. A proposed "soft response" plate impact experiment will be described later.

Partial penetration experiments provide valuable data under either limiting conditions, and are discussed next.

Partial penetration of armor targets

As described in a companion paper in this present conference, we have performed experiments in which hard small and large structural targets were struck by thin, hardened steel, and heat treated glass targets at impact velocities ranging from 300 to 5000 m/s. The steel, with a diameter of 3.8 cm and long rods, remained elastic, and impacted at partial penetration. The hardened targets showed a HED region around the penetration that consisted of partially melted material with a highly dense boundary. The HED width at the nose of the plate in all cases was less than 2 cm. The diameter of the target ranged from about twice the radius of steel at 3000 m/s impact velocity to about 4 times the radius of the steel at 5000 m/s impact velocity.

The microstructure showed distinct spatial components which are driven into the target from the nose geometry at the target surface, and impacted with less spatially relative large relative to form a fragmented rod emerging from the area with fragmented area some distance ahead of the plastic penetration or sub area sized fragment close to the boundary with the HED, within which the fragment size were less than 50 microns. The more intense areas remain in the region and produce a small porosity (600 m/s). This, similar to the potential role played by using steel, the plate impact case, a mechanism exists to produce both fragment and porosity (600 m/s) in a region ahead of an advancing interface with the impervious penetrator or target plate.

The Chariot model then suggests the following penetration criterion. When the actual compressive stress at a given location (PMCA ahead of the penetrator) is the critical value for the largest fragment in the PMCA (Fig. 2), these fragments and holes are generated, and non-classical flow begins. This flow initially consists of "channels" of small fragments, varying in size, into the target. The larger pressure near the penetrator move cause compression and generation of micron sized particles similar to those in the plate impact case. The pressure is absorbed by the phase transition, resulting in a compacted HED. When the HED material begins from the penetrator penetration, the plastic penetration begins. However, the fiber stress also exists with the penetrator material to become a metal, and the temperatures are high enough to remove some of the glass, as observed.

Preliminary measurements of the fragment size distribution in the HED case as the penetrator's nose roughly for a Poisson distribution with L_{0} in Eq. (2) equal to about 20 microns inside the HED boundary and 60 microns outside the boundary. If we now assume that L_{0} = 0.1 and L_{1} = 200 microns, Eq. (1) gives a critical driving stress of 3.6 GPa for the outer boundary of the HED. This appears to correlate well with supporting hydrocode calculations of an impact velocity of 450 m/s. Furthermore, the hydrocode calculations give a driving stress of 3.6 GPa for the outer boundary of the HED, which is consistent with the experimental data and the penetrator mass reduction. In this scenario, the main role of the larger fragment in the HED, as well as the fragments outside the HED, is to provide the pressure that partially absorbs the
Thus, the C3008 composites are perhaps encouraging, but we again need microdamage evaluation studies described next.

**PLATE IMPACT PROGRAM**

**Partial penetration tests:**

The partial penetration tests fulfill the basic requirements for measuring microdamage evolution: variable load amplitudes and durations, well-confined specimens allowing microscopic examination of the fractured material, and recovered broken material for property testing. A prototype test program is described in a companion paper in this conference.

**Quantitative property issues:**

Tests are needed to obtain basic material properties, especially for the polymerized material for input to both continuum and mesomechanical models. Pressure shock measurements on polymerized material recovered from the partial penetration test are underway. These tests include microstructural observations of the material before and after granular flow. We plan similar observations of damaged material from extended pressure test testing performed at SAE. Since interaction of the polymerized material with the penetrator nose and sides seemed important in the partial penetration test, we plan to also examine the material that adheres to the nose surfaces in the SAE experiments. We will compare the evolution of fragmentation size and shape for the two cases.

**Blast and residual plate impact tests:**

To help interpret prior plate impact data, we need tests that allow us to monitor the evolution of the microdamage. In prior work on brittle metallic fracture and fragmentation in Anechoic tests, for example, we were able to produce different damage levels in target "packs" and therefore perform iterative calculations with RAM/SA/AA models until we could correlate with the measured damage distributions. To follow the same procedure for compressive shock loads, we need a scheme to nondestructively test the target specimen for microstructural examination. A possible design is illustrated in Figure 5, which shows the following features based on our earlier work:

- **Flyer plate is 50 mm diam x 5 mm thick.**
- **Glass试 about a 2 μs pulse with a possible 2-wave structure.**

![Figure 5. Plate impact test design](image-url)
The samples are bonded to pre-fabricated cylindrical glass plates ("pucks").

Impact and fragmentation mechanisms are improved-reinforcing aluminum alloys.

The sample plate is surrounded by a glass collar whose purpose is to eliminate converging
unloading waves, and maintain uniaxial strain in the sample for the duration of the load pulse.

Impact velocities and transmitted force histories are recorded.

Increasing impact velocity leads to lower increasing degrees of damage.

A standard "tag angle" is used for all experiments of the target package.

The glass sample is characterized pre and posttest for fragment size, mass size, and fragment
growth density distributions.

CONCLUSIONS

Current experimental methods (Choi et al.) show some promise, but results from micromechanics evolution
data to complete. An approach similar to the work of T.F. will be replaced by an approach that
combines static and dynamic fractographic and acoustic emission, and sophisticated acoustic
growth and correlation models for correlating better with micromechanics evolution data. Some key questions are:

How is cavity initiation in impacted glass targets from inertial heating? Are using exactly the
thermal mechanism? The proposed experiments of Figure 3 should help us answer that and other
questions.

What growth to the plastic extension of fragment sizes inside the CTO? The picture of "strain burst"-
induced fragmentation followed by plastic networking under confinement and compression will
not doubt be modified by data from the proposed additional posttest penetration tests and plate impact
tests.

What are the three properties of the material in the CTO? Data from the extreme gasdynamic property
tests will be valuable in this area.

In general, the forthcoming detailed microscopic damage evolution data should help us replace
uncertain elements of our conceptional assumptions with empirically determined relations.

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