Dynamic high-strain-rate mechanical behavior of microstructurally biased two-phase TiB$_2$+Al$_2$O$_3$ ceramics

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The dynamic high-strain-rate behavior of two-phase TiB$_2$+Al$_2$O$_3$ ceramics with biased microstructures was investigated in this study. The microstructural bias includes differences in phase (grain) size and phase distribution such that in one case a continuous (interconnected) TiB$_2$ network surrounds the Al$_2$O$_3$ phase (qualitatively termed “T@A”) and in another case the TiB$_2$ and Al$_2$O$_3$ phases are interdispersed and uniformly intertwined with each other (qualitatively termed “TinA”). Quantitative microscopy was used to characterize the phase size and the integral curvature which is taken as a measure of TiB$_2$ phase connectivity around Al$_2$O$_3$. Dynamic compression and tension (spall) properties were measured using plate impact experiments. The measurements used piezoelectric polyvinylidene fluoride stress gauges to obtain the loading profile and to determine the Hugoniot elastic limit. In addition, velocity interferometry system for any reflector interferometry was used to obtain the spall signal and the tensile dynamic strengths of the materials. Experimental results reveal that while the $\sigma_{\text{HEL}}$ and the compressive strengths of TiB$_2$+Al$_2$O$_3$ ceramics are dependent on the average grain (phase) size, the tensile (spall) strength scales with the TiB$_2$-phase connectivity. This result suggests that the interconnected TiB$_2$/Al$_2$O$_3$ microstructural morphology provides a stronger impediment to failure in tension compared with the morphology with simply interdispersed phases. © 2002 American Institute of Physics. [DOI: 10.1063/1.1429770]

I. INTRODUCTION

The dynamic, high-strain-rate deformation behavior of materials (and consequently their ballistic performance) is dramatically influenced by their microstructural characteristics, including phase size, phase morphology, composition and texture. Such a trend has been revealed not only for steels used as heavy armor, but also for ceramics considered for lightweight armor applications. In addition to their low density, ceramics exhibit superior hardness and high compressive strength that enables erosion and “interface defeat” of projectiles. This mechanism makes ceramics highly desirable for use as armor materials. Over the last 30 years a number of studies on ballistic performance and dynamic behavior have been performed on various ceramics (e.g., AlN, Al$_2$O$_3$, B$_4$C, SiC, TiB$_2$, WC, ZrO$_2$). These studies have suggested that the desirable characteristics of ceramics that are beneficial for the defeat of projectiles include the combination of high yield strength or hardness, high tensile spall strength, high fracture toughness, high Poisson’s ratio, and high coefficient of friction. The properties typically measured to characterize the dynamic behavior of ceramics are the Hugoniot elastic limit ($\sigma_{\text{HEL}}$) and spall strength. In general, the spall strength is lower than the elastic limit. Materials such as TiB$_2$ have a spall strength that decreases with increasing impact stress and becomes negligible at the $\sigma_{\text{HEL}}$. In contrast, Al$_2$O$_3$ maintains its original spall strength even at impact stress levels that exceed the $\sigma_{\text{HEL}}$.

It has been demonstrated that SiC and TiB$_2$ exhibit the most desirable ballistic properties based on results of experiments performed to determine the transition between interface defeat and penetration involving W projectiles. The impact velocity for dwell/penetration transition for SiC and TiB$_2$ is higher than that for B$_4$C, in spite of the higher yield strength of the latter. This trend is consistent with the predictions from the dwell/penetration transition model which accounts for damage mechanism based on the extension of mode-I wing cracks. The predictions illustrate that SiC and TiB$_2$ ceramics have the ability to suppress the formation of wing cracks due to plastic relaxation of preexisting flaws, while B$_4$C shows a brittle behavior dominated by growth of wing cracks. Brittle fracture is also expected to dominate the behavior of Al$_2$O$_3$. Measurements of shock wave profiles have also illustrated the ability of SiC and TiB$_2$ ceramics to undergo deformation-induced hardening, which may give rise to improved ballistic properties.

Most dynamic behavior studies performed to date have focused on single-phase monolithic ceramics, although ce-
amics with glassy (impurity) phases, e.g., AD85 Al₂O₃, have also been investigated. In such ceramics, the presence of intergranular oxide glass has been shown to significantly lower the tensile (spall) strength, and, therefore, the fracture resistance. Dynamic behavior of ductile metal-matrix composites, e.g., Al alloys consisting of embedded ceramic particles, has been studied to a limited extent. It has been shown that both the dynamic yield strength and the spall strength are reduced, in comparison to strength increases observed in particle-reinforced composite materials under quasistatic loading conditions. The composite structure acts as a mechanical energy trap due to scattering of waves from incoherent boundaries and interfaces between the matrix and reinforcement phases having dissimilar shock impedance. The high-strain-rate mechanical behavior of ceramic-ceramic composites shows a strain rate dependence significantly different from that observed in metal-matrix composites. Dynamic behavior of ceramic composites has been limited only to the study of two-phase Al₂O₃ + TiB₂ ceramics (70/30 mass ratio). Past work on these two-phase ceramics has revealed an 80% increase in compressive strength with increasing strain rate (3.5 GPa at 10⁻⁴ s⁻¹ – 5.8 GPa at 10⁴ s⁻¹) as shown in Fig. 1(a). In addition, as illustrated in Fig. 1(b), the two-phase Al₂O₃ + TiB₂ ceramics have static and dynamic mechanical properties superior to their monolithic constituents. The Al₂O₃ + TiB₂ ceramics have also shown better penetration resistance than monolithic Al₂O₃, and the system in which TiB₂ is an interconnected phase surrounding Al₂O₃ has been shown to exhibit a superior ballistic performance compared with the system in which the two phases are simply uniformly interdispersed. Micro-mechanical simulations have also demonstrated the effect of microstructural bias on failure resistance. However, the influence of microstructural bias on the fundamental dynamic properties of these ceramics has not been fully established.

The objective of the present work is to characterize the high-strain-rate deformation and damage response of four types of microstructurally biased, two-phase TiB₂ + Al₂O₃ ceramics. The microstructural bias of the four ceramics falls into two morphological categories. The first category involves a continuous (interconnected) TiB₂ network that surrounds Al₂O₃ (qualitatively termed “T@A”) and the second category involves TiB₂ and Al₂O₃ phases that are interdispersed and uniformly intertwined with each other (qualitatively termed “TinA”). Normal plate impact experiments were used to measure the σHEL and the spall strength of the materials. The measured responses were then correlated with the microstructure morphological characteristics and the phase size scales.

II. PREPARATION AND CHARACTERIZATION OF THE TWO-PHASE CERAMICS

The two-phase TiB₂ + Al₂O₃ ceramics were fabricated as ~80 mm diameter by 20-mm-thick disks by hot-pressing powders. The powders were produced through either self-propagating high-temperature synthesis (SHS) reactions between powder precursors (3TiO₂ + 3B₂O₃ + 10Al = 3TiB₂ + Al₂O₃; 33:67 weight ratio), or through manual mixing (MM) of the constituents (nominally, 30 wt% TiB₂ + 70 wt% Al₂O₃) which are obtained using conventional powder processing techniques. Following the SHS synthesis or manual mixing, the powders were also ball milled to generate the microstructural bias. Details of the processing approaches used for fabricating these two-phase TiB₂ + Al₂O₃ ceramics are described elsewhere. Quantitative metallography was performed to determine the phase size (as measured by the linear intercept length) and phase connectivity (based on measurement of integral curvature). The connectivity of a phase in a multiphase material can be related to a quantitative measurement of curvature, k, defined as the angle, dθ, of a finite segment of an arc divided by the length of the arc dλ, or

\[ k = d\theta/d\lambda. \]  

Through a division of line integrals, the average curvature, \( k_{\text{ave}} \), can be calculated by dividing the angle of arc \( \theta_A^{\text{net}} \) (in phase A) by total length of arc, \( L_A \). Hence,

\[ k_{\text{ave}} = \frac{\theta_A^{\text{net}}}{L_A} = \frac{\left(\theta_A^+ - \theta_A^-\right)}{L_A}, \]

FIG. 1. (a) Strain-rate dependence of compressive strength for Al₂O₃ + TiB₂ (solid line) in contrast to that for Al₂O₃ (dashed line) (see Ref. 23). (b) A comparison of the dynamic compressive strengths of two-phase TiB₂ + Al₂O₃ with those of the monolithic constituents (see Ref. 23).

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where, $\theta_A^+$ is angle for a convex arc and $\theta_A^-$ is angle for a concave arc in phase A. If the phases are considered as closed as in the case of our two-phase ceramic, then the angle of the arc is $2\pi$. Hence the above equation can be written as

$$k_{ave} = (2\pi)N_{A}^{\text{net}} / L_A = 2\pi(N_A^+ - N_A^-) / L_A. \quad (3)$$

Table I. Measured average values of integral curvature, and sizes of constituent phases.

<table>
<thead>
<tr>
<th>Microstructure type</th>
<th>Average integral curvature ($\mu$m$^{-1}$)</th>
<th>TiB$_2$ phase size ($\mu$m)</th>
<th>Al$_2$O$_3$ phase size ($\mu$m)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>$-0.316 \pm 0.022$</td>
<td>7.0</td>
<td>10.4</td>
</tr>
<tr>
<td>B</td>
<td>$-0.476 \pm 0.046$</td>
<td>6.2</td>
<td>9.1</td>
</tr>
<tr>
<td>C</td>
<td>$-0.074 \pm 0.028$</td>
<td>8.7</td>
<td>25.1</td>
</tr>
<tr>
<td>D</td>
<td>$-0.375 \pm 0.031$</td>
<td>7.9</td>
<td>12.3</td>
</tr>
</tbody>
</table>

where, $N_{A}^{\text{net}}$, $N_{A}^{+}$, and $N_{A}^{-}$ are the net, positive, and negative numbers of loops of phase A. The length of the arc, $L_A$, of a closed loop is the perimeter and can be related to the mean linear intercept length $L$ using $L_A = (\pi/2)(1/L)$. The average integral curvature of phase A can then be expressed as

$$k_{ave} = 4N_{A}^{\text{net}}L. \quad (4)$$

In the present work, the average integral curvature values were determined using Eq. (4). The mean linear intercept length $L$ and the number of net loops of "A" (or TiB$_2$) phase around the "B" (or Al$_2$O$_3$) phase were determined using IMAGEPRO-PLUS, a commercial software package (product of Media Cybernetics). To account for anisotropy effects, the analysis was performed on at least ten micrographs of both planar and cross-sectional surfaces. However, the differences due to anisotropy were found to be in the range of the standard deviation.$^{29}$

The impact experiments used an 80-mm-diam, single-stage gas gun. Measurements of the $\sigma_{HE}$ and the shock wave speeds under dynamic compression were obtained from stress profiles recorded using polyvinylidene flouride (PVDF) stress gauges. As shown in Fig. 2(a), a TiB$_2$ flyer plate backed by an air gap and mounted at the head of an aluminum projectile, was used to impact the target assembly consisting of $\sim$3-mm-thick ceramic sample backed by a TiB$_2$ backer plate. Impact experiments were conducted at a nominal velocity of $\sim0.750$ km/s ($\sim15$ GPa nominal impact stress) to ensure shock loading of samples under similar conditions. The impact velocity was measured using arrival time pins. One PVDF gauge package was placed at the impact surface and another was placed between the sample and TiB$_2$. 

FIG. 2. Schematic illustrations of the setups used for measurements of (a) stress profiles with PVDF gauges and (b) free-surface velocity traces with VISAR interferometry.

FIG. 3. Optical micrographs of two-phase ceramics (TiB$_2$ white and Al$_2$O$_3$ dark phase). (a) Sample A, SHS, T@A microstructure, (b) sample B, SHS, TinA microstructure, (c) sample C, MM, T@A microstructure, and (d) sample D, MM, TinA microstructure.

FIG. 4. Plot comparing the average integral curvature values of the TiB$_2$ + Al$_2$O$_3$ ceramic samples of four different types of microstructures. Note that the TiB$_2$ phase connectivity is greater in sample A than in sample B, and likewise in sample C than in sample D.

Table I. Measured average values of integral curvature, and sizes of constituent phases.
backer plate. The experimental setup was designed such that a planar-parallel shock wave propagates through the target and the input and propagated stress profiles are measured with little interference from radial reflected waves. The PVDF gauges were connected to current viewing resistors to allow current versus time data acquisition with a 1 GHz frequency oscilloscope. The current versus time data were then numerically integrated to yield stress versus time profiles for the impact and propagated wave gauges. The arrival time at the respective gauges was taken as the travel time through the sample thickness, from which the shock wave speed was determined.

Figure 2(b) shows the experimental setup, illustrating the target sample (without any backer) being impacted by a projectile consisting of a 4140 steel (or silicon carbide) flyer plate and an aluminum sabot. An air gap exists between the flyer plate and the sabot to allow full unloading required for the spall strength measurements. All sample surfaces were lapped for flatness and parallelism. The ceramic targets were polished with 5 μm diamond paste to ensure reflectivity required by the velocity interferometer system for any reflector (VISAR) beam. The time-resolved longitudinal motion of the sample free surface was measured in the form of interference fringes with a Valyn VISAR and recorded on a digital oscilloscope. The interference fringes were then converted to time-resolved history of particle velocity.

III. RESULTS AND DISCUSSIONS

A. Microstructural characterization

Optical micrographs of the four types of microstructurally biased samples studied are shown in Figs. 3(a)–3(d). The micrographs correspond to sample A made by the SHS process with TiB₂ surrounding Al₂O₃ (T@A); sample B also made by SHS with intermixed TiB₂ and Al₂O₃ (TinA) phases; sample C made by MM with TiB₂ surrounding Al₂O₃ (T@A); and sample D made by MM with intermixed TiB₂ and Al₂O₃ (TinA) phases. It should be noted that the samples do not reveal 100% microstructural bias, i.e., while the micrographs of samples A and C generally illustrate a nearly continuous (interconnected) TiB₂ phase surrounding Al₂O₃, regions where the converse is true are also present. Likewise, samples B and D show not only an intermixed structure but also the continuous phase microstructures.

Table I lists the values of average integral curvature and the average size of TiB₂ and Al₂O₃ phases. In general, the average sizes of TiB₂ and Al₂O₃ phases in both SHS samples (A and B) are smaller than those in the manually mixed samples (C and D), with sample C showing the largest size for both constituents. Sample C also represents the microstructure with the highest value of connectivity for TiB₂, as illustrated in Fig. 4. All samples, in fact, show a negative curvature, since, within a given area of measurement, the number of loops around smaller particles is greater than that surrounding the larger particles. This skews the average integral curvature towards negative values. Nevertheless, the measured values provide a quantitative measure of the microstructural bias on the basis of phase connectivity, and illustrate that the connectivity of TiB₂ phase is greater in sample A than in sample B, and likewise in sample C than in sample D. The grain size of the respective phases also shows a similar trend, with sample A being coarser than sample B, and sample C being coarser than sample D.

B. Characterization of elastic properties

The elastic properties of the ceramics were characterized using an Ultralan Laboratories ultrasonic test apparatus. The setup allows determination of the longitudinal and shear wave velocities from which the elastic moduli are obtained using densities measured on individual samples by the Archimedean method. Table II lists the results of elastic wave velocity measurements and corresponding Young’s and shear moduli. It can be seen that the moduli of the various ceramic samples are relatively similar except in the case of sample B, which has ~4.4% porosity.

![Figure 5](image-url)

**FIG. 5.** Typical input and backer gauges stress profiles (sample B) obtained from Expt. 9916.

<table>
<thead>
<tr>
<th>Sample type</th>
<th>Density (g/cm³; %TD)</th>
<th>Long. vel (km/s)</th>
<th>Shear vel. (km/s)</th>
<th>Young’s mod (GPa)</th>
<th>Shear mod (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>4.095; 99.4%</td>
<td>11.14</td>
<td>6.477</td>
<td>505.8</td>
<td>171.8</td>
</tr>
<tr>
<td>B</td>
<td>3.993; 95.6%</td>
<td>10.336</td>
<td>6.145</td>
<td>420.8</td>
<td>148.7</td>
</tr>
<tr>
<td>C</td>
<td>4.108; 99.7%</td>
<td>10.796</td>
<td>6.527</td>
<td>478.8</td>
<td>175.0</td>
</tr>
<tr>
<td>D</td>
<td>4.083; 99.1%</td>
<td>10.773</td>
<td>6.523</td>
<td>473.9</td>
<td>173.7</td>
</tr>
<tr>
<td>TiB₂</td>
<td>4.509</td>
<td>10.790</td>
<td>7.43</td>
<td>524.9</td>
<td>248.9</td>
</tr>
<tr>
<td>Al₂O₃</td>
<td>3.55</td>
<td>9.280</td>
<td>5.47</td>
<td>306.2</td>
<td>106.4</td>
</tr>
</tbody>
</table>

C. Measurements of dynamic high-strain-rate mechanical behavior

Stress wave profiles obtained using PVDF stress gauges were used to determine the mechanical behavior under dynamic compression. VISAR interferometry was used to obtain free-surface velocity profiles to determine the tensile spall strength.

1. Dynamic compression behavior

Measurements of the Hugoniot elastic limit \( \sigma_{\text{HEL}} \) were obtained from normal planar impact experiments conducted at an impact velocity of 750 m/s. Figure 5 shows examples of stress histories recorded by the “input” and “backer” PVDF stress gauges from experiment 9916, for the sample B ceramic. Table III lists the sample thickness, density, impact velocity measured using shorting pins, wave speed measured by considering the times of travel through the sample thickness as recorded by input and backer gauges, the \( \sigma_{\text{HEL}} \), and the yield stress in simple tension \( \sigma_{\text{YS}} \) calculated from the \( \sigma_{\text{HEL}} \).

The wave speed in the material, which is a function of sample density and microstructure in addition to loading conditions, is measured to be similar for all four samples within the range of experimental scatter. The Hugoniot elastic limit \( \sigma_{\text{HEL}} \), the axial stress at which a solid loaded under conditions of uniaxial strain begins to exhibit plastic deformation, is observed to be a strong function of the microstructure including phase size. The \( \sigma_{\text{HEL}} \) was determined by considering one half the value between the peak and trough of the elastic precursor wave [Fig. 5(b)], and one half the difference was considered as the range. The \( \sigma_{\text{HEL}} \) is found to be the lowest for sample B (4.4 ± 1.2 GPa), due to its high level of porosity (~4%) and highest for sample D (8.5 ± 4.5 GPa) (with a large standard deviation). The \( \sigma_{\text{HEL}} \) values reported previously by Grady \( ^5,6 \) are 9–18 GPa for TiB\(_2\), and ~6.7 for Al\(_2\)O\(_3\). While \( \sigma_{\text{HEL}} \) identifies the limit of elastic response under dynamic (shock) uniaxial strain loading, it is also common to assume a Von Misses-type yield condition which asserts that plastic flow initiates when the second deviatoric stress invariant attains a critical value. Through this formalism, the yield stress in simple tension has been shown to be related to \( \sigma_{\text{HEL}} \) through \( \sigma_{\text{YS}} = 2 \left( C_2 / C_1 \right) \times \sigma_{\text{HEL}} \).\(^6\) The yield stress in simple tension, \( \sigma_{\text{YS}} \) for the various microstructurally biased samples follows the same trend as the Hugoniot elastic limit, with sample D showing the highest and sample B the lowest yield strength.

2. Tensile (spall) strength

Spall experiments were performed (with VISAR interferometry) on samples of microstructure A at impact velocities corresponding to input stresses above and below the \( \sigma_{\text{HEL}} \), and on samples of microstructures ‘C’ and ‘D’ under elastic loading conditions to ensure that the compression-induced damage does not influence the tensile response. Figure 6 shows the free-surface velocity traces obtained on samples of microstructure A at 237, 495, and 758 m/s, using a 2.66-mm-thick AISI 4140 steel flyer plate. Arrows indicate spall signal.

![Fig. 6. Free-surface velocity traces obtained on samples of Microstructure A at 237(a), 495(b), and 758(c) m/s, using a 2.66-mm-thick AISI 4140 steel flyer plate. Arrows indicate spall signal.](image)
The two-phase TiB$_2$ + Al$_2$O$_3$ ceramic maintains non-negligible spall strength even at input stresses exceeding the Hugoniot elastic limit (6.2 ± 3.4 GPa).

Tensile spall experiments were also performed on samples of microstructures C and D, at an input stress of ∼3.8 GPa. Figure 7 shows the free velocity traces for these samples. A spall strength of 0.311 GPa for sample C and 0.222 GPa for sample D was observed. A comparison of spall test results between sample C and sample D shows that the latter sample with dispersed microstructure has a lower spall strength, while the microstructure with interconnected TiB$_2$ has a higher spall strength, even though sample D has a smaller phase size. The measured high value of spall strength of the two-phase ceramic is similar to the published spall strength even at input stresses exceeding the Hugoniot elastic limit (0.33 GPa) but lower than that of Al$_2$O$_3$ (0.45 GPa).

**IV. DISCUSSION AND SUMMARY OF RESULTS**

The two-phase TiB$_2$ + Al$_2$O$_3$ ceramics, made either by the SHS or mechanical milling methods, reveal differences in microstructure which qualitatively show TiB$_2$ as a continuous (interconnected) phase surrounding Al$_2$O$_3$ (T@A), or TiB$_2$ and Al$_2$O$_3$ intermixed with each other (TiA). Quantitative microscopy analysis based on the measurement of the integral curvature showed that the samples investigated do not exhibit 100% microstructural bias. However, the overall trend of the influence of microstructural bias emerging from the results of experiments performed to date illustrates that the dynamic yield strength and the σ$_{HEL}$ are more dominantly dependent on the phase size. Sample C prepared by manual mixing and having the largest phase (grain) size shows the lowest values in contrast to the other samples of similar (~99%) density. In contrast, the tensile spall strength appears to scale with the continuity of the TiB$_2$ phase. Sample C, which has the most interconnected TiB$_2$ phase, has the highest value of the tensile spall strength.

The results therefore, illustrate that while the Hugoniot elastic limit and the dynamic compressive yield strength of Al$_2$O$_3$ + TiB$_2$ are dependent on the average grain (phase) size, the tensile spall strength scales with the TiB$_2$-phase connectivity, and less so with the average phase size. It is possible that the interconnected phase morphology is more effective in impeding initiation and progression of fracture under tensile conditions. If so, TiB$_2$ as an interconnected phase has the ability to suppress and relax the cracks formed in Al$_2$O$_3$. Alternatively, it is also possible that in the two-phase ceramic with the microstructure containing dispersed Al$_2$O$_3$ and TiB$_2$ phases, the mechanical energy trapping due to scattering of waves from incoherent boundaries and interfaces results in the lowering of the tensile (spall) strength. Further work is currently in progress to more clearly delineate the effects of interconnected versus dispersed TiB$_2$ phase, and to eventually fabricate two-phase Al$_2$O$_3$ + TiB$_2$ ceramics with the microstructural bias that yields the most optimal dynamic properties.

**ACKNOWLEDGMENTS**

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**TABLE IV. Loading conditions and summary of results of spall experiments.**

<table>
<thead>
<tr>
<th>Sample No.</th>
<th>Flyer type and material</th>
<th>Flyer &amp; target th. (mm)</th>
<th>Target density (%TMD)</th>
<th>Impact velocity (m/s)</th>
<th>Input stress (GPa)</th>
<th>Spall strength (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A-0005</td>
<td>4140 steel</td>
<td>2.66/16.22</td>
<td>98.8%</td>
<td>495</td>
<td>7.9</td>
<td>0.320</td>
</tr>
<tr>
<td>A-0007</td>
<td>4140 steel</td>
<td>2.66/16.20</td>
<td>98.8%</td>
<td>237</td>
<td>3.7</td>
<td>0.320</td>
</tr>
<tr>
<td>A-0008</td>
<td>4140 steel</td>
<td>2.66/7.99</td>
<td>98.8%</td>
<td>758</td>
<td>11.8</td>
<td>0.160</td>
</tr>
<tr>
<td>C-9925</td>
<td>SiC</td>
<td>4.30/7.54</td>
<td>99.7%</td>
<td>244</td>
<td>3.8</td>
<td>0.311</td>
</tr>
<tr>
<td>D-9926</td>
<td>SiC</td>
<td>4.77/7.10</td>
<td>99.1%</td>
<td>239</td>
<td>3.8</td>
<td>0.222</td>
</tr>
</tbody>
</table>

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**FIG. 7.** Free-surface velocity traces obtained for samples C and D, at similar loading conditions.
14 F. I. Grace, in Ref. 11, pp. 421–428.
26 M. Zhou and J. Zhai, presented at the American Ceramic Society 102nd Annual Meeting & Exposition, St. Louis, MO, April 30–May 3, 2000.