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**FOURTH PROGRESS REPORT OF
LIGHT ARMOR PROGRAM**

M. L. Wilkins

C. F. Cline

C. A. Honodel

June 4, 1969

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Lawrence Radiation Laboratory
UNIVERSITY OF CALIFORNIA
LIVERMORE

MASTER

UCRL-50694

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Key

Foreword

The work reported here is a continuation of the program initiated by the Advanced Research Projects Agency (ARPA) and is described in the first, second, and third progress reports.¹⁻³

The work of the following personnel of the Lawrence Radiation Laboratory is acknowledged. Hugh Heard performed the tri-axial compression tests. Richard Landingham made the cermet studies. William Gust did the equation-of-state measurements. Albert Holt did the work on graded armor. Charles Honodel, John O'Connor, and Nate Rawls performed the ballistic experiments. Dick Giroux was the programmer on the HEMP code, and T. Suyehiro worked on the special computer plot routines. Beb Luicanan and Katherine Spence were the computer assistants. Frank Wittmayer and Sam Digiallonardo did the microscopy studies.

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FOURTH PROGRESS REPORT OF LIGHT ARMOR PROGRAM

Abstract

The sequence of events that occurs when a sharp steel projectile strikes ceramic armor is examined using computer calculations in conjunction with experiments. Important physical properties of target materials that defeat penetration are identified. Experiments and calculations verify that the ballistic limit is very sensitive to the ceramic's tensile properties. The high stress load transmitted

through the ceramic to a backup plate results in the backup plate yielding directly under impact. Introduction of ductility into a ceramic promises to improve armor facing material. Fabrication of ceramics with ductility is discussed. Experimental equation-of-state data have been obtained for many current armor materials as well as for new armor material candidates.

Introduction

The light armor program has been aimed at understanding penetration mechanics in order to indicate the direction the material scientists should take to develop new materials. From the experiments and calculations that have been done previously,¹⁻³ the current positions are these:

- (1) The physical processes that occur during the penetration process are fairly well understood for impacts on hard and soft targets by blunt and sharp projectiles.
- (2) The important material parameters that operate during the penetration process have been identified.

The problem that remains now is to incorporate this information into a

material development program. There are two general directions that can be taken with the knowledge obtained in (1) and (2). One is to modify existing armor materials so as to have a better balance of physical properties needed to defeat penetration. The other is to fabricate new materials that may have intrinsically the desired properties. Both of these approaches are currently being undertaken at LRL.

A clearer picture of the motivation of the material development program can be given by summarizing some of the work that supports the two positions stated above for impacts of sharp projectiles on ceramic armor. (The projectiles are made from Allegheny steel 609, cylinder length 0.9 in, diameter 0.3 in. See Fig. 1 of Ref. 1.)

Physical Process in Penetration of Ceramic Armor

Figure 1 gives the energy-vs-time history obtained from a calculation of a sharp steel projectile that strikes an Al_2O_3 ceramic target supported by an aluminum backup plate. The impact velocity is 2800 ft/sec which is just below the ballistic limit for this target; other details are given in Ref. 3. The following sequence of events correspond to Fig. 1:

Fig. 1. Energy vs time for an impact at 2800 ft/sec of a sharp steel projectile with a target of 0.34-in. AD85 alumina backed by 0.25-in. aluminum.

(1) From 0 to 9 μsec , the projectile tip is being destroyed. During this period, the backup plate yields at the ceramic interface. Resulting tension in the ceramic as it follows the motion of the backup plate initiates an axial crack.

Figure 2 shows a calculation of the initiation of the axial crack at the ceramic backup-plate interface as well as the fracture conoid which starts at the

impact surface. The axial crack moves back toward the impact surface. The fracture conoid proceeds toward the backup plate and effectively limits the amount of ceramic that participates in the impact process.

Figure 3 shows framing camera views of the tip of a projectile being destroyed by an impact on Al_2O_3 . Figure 4 shows views of the interface of an Al_2O_3 target as seen through a glass backup plate during the impact of a sharp steel projectile. The breakout of the fracture conoid is seen. The last frame shows evidence of the axial crack which has been obscured by the grid lines.⁴

(2) From 9 to 15 μsec , the projectile is eroded by the ceramic which has become a rubble from coalescing cracks within the fracture conoid. Approximately 40 percent of the projectile mass and 40 percent of the initial projectile energy is carried off by eroded projectile material. Erosion takes place because the stress level on the projectile is greater than the material strength of the projectile. Projectile material yields and flows perpendicular to the impact and is then swept back as the projectile moves into the ceramic. The maximum decelerating force on the projectile is thus related to projectile yield strength. The erosion process continues until the prevailing stress level falls below the projectile's yield strength.

Figure 5 shows a calculation and a flash radiograph immediately after an initially sharp projectile has completed the erosion process. Both the calculation and the radiograph show that projectile material has flowed out in the radial

Fig. 2. Calculation of development of fracture and axial crack in alumina.

$t = 1 \mu\text{sec}$

$t = 9 \mu\text{sec}$

$t = 17 \mu\text{sec}$

$t = 33 \mu\text{sec}$

Fig. 3. Impact of a 30-caliber sharp projectile on Al_2O_3 ceramic. Impact velocity = 2800 ft/sec.

direction of the projectile and is being swept back by the ceramic. This material will eventually be stripped off the projectile.

(3) After 15 μsec , the erosion of the projectile stops and additional projectile

energy loss equals the target energy gained. Total energy in the target-projectile system is 60 percent of the original energy and is absorbed by the backup plate.

The important energy loss mechanism for a projectile that strikes a ceramic

Fig. 4. Experimental breakout of the fracture conoid in Al_2O_3 ceramic with a glass backup plate.

target is due to the loss of projectile mass during step (2) above. The rate of projectile energy loss shown in Fig. 1 is such that an increase of 250 ft/sec in the ballistic limit would occur if the time duration of step (2) could be extended by $2 \mu\text{sec}$.

Because the duration of the projectile erosion phase is related to the time of breakup of the ceramic, it is important to maintain ceramic integrity for as long as possible. This can be done in two ways:

HEMP
PL22Y.005L

Cycle
+2241.

Time
+16.27924

Flash radiograph

Fig. 5. Comparison of a calculation and an experiment at $16 \mu\text{sec}$ after impact of a sharp steel projectile with 0.34-in. AD85 alumina bonded to 0.25-in. aluminum. Impact velocity = 2800 ft/sec. (The flash radiograph of the cylinder appears to have a smaller diameter than the actual diameter because of x-ray penetration of the edges.)

Fig. 6. Section view of rigid ceramic struck by a sharp steel projectile at 2300 ft/sec.

(1) Synthesize a new backing material that will better support the ceramic in order to delay the onset of the ceramic tensile failure. The importance of this approach was investigated experimentally with a target of 0.34-in. thick AD-85 alumina supported by 1 in. of Starrett stock steel, which provided a nearly unyielding backup plate. There was no penetration of the ceramic when struck by the sharp

steel projectile at a velocity of 2300 ft/sec. The projectile was completely destroyed, but the ceramic still maintained its integrity (Fig. 6). This experiment suggests that ceramic armor would be even more effective if the backup plate could sustain higher loads.

(2) Synthesize a new ceramic which has improved tensile properties so as to delay the onset of the axial crack.

Important Parameters in Penetration Process

Experiments and calculations verify that the ballistic limit is very sensitive to the ceramic's tensile properties. The high stress load transmitted through the ceramic to the backup plate results in the backup plate yielding directly under the impact. The attendant motion of the ceramic as it follows the backup plate is slight, but due to the high elastic modulus of the ceramic, a large tensile stress is set up in the ceramic. The axial crack that results is the beginning of the ceramic breakup under the projectile. A ceramic material property that would allow the ceramic to follow the motion of the backup plate further before failure is ductility. Two computer calculations were made (Fig. 7) to translate an improvement in ceramic tensile properties into a time delay before fracture. These calculations show fracture patterns for different times when the ceramic is considered to fracture when the tension reaches 3 kbar and 8 kbar. The 3-kbar tension corresponds to about 0.1 percent strain, and the 8-kbar tension corresponds to about 0.3 percent strain. Therefore the calculations compare the fracture pattern as a function of time when one ceramic is capable of a greater elongation before fracture. It should be emphasized that it does not matter if the material properties are such that a high tensile stress can be reached before fracture occurs, or if the material deforms plastically at a lower stress magnitude but fractures at the same value of strain. The results will be the same because the important parameter is the magnitude of the strain when fracture occurs. Thus, the expression "improved tensile

properties" is used in the sense of an increase in strain before fracture from a tensile load.

Figure 7 shows that the accumulated damage to the ceramic is about the same at times $t = 4.15 \mu\text{sec}$ and $t = 5.87 \mu\text{sec}$ for the calculations with fracture at 3-kbar tension and fracture at 8-kbar tension, respectively. Thus, for these calculations, a 0.2 percent increase in strain before fracture results in a $1.7 \mu\text{sec}$ increase in the breakup time of the ceramics. When the ductile property of the ceramic is increased so that the time of breakup has been increased by $2 \mu\text{sec}$, the calculated increase in the ballistic limit is 250 ft/sec.

Note that there is not just one, but several important characteristics a target needs to defeat penetration. A ballistic impact on a ceramic creates large compressive stresses and subsequent tensile stresses. We have already discussed the importance of ductility or resistance to tensile failure. Adding ductility in a ceramic obviously would bring about an improvement if other parameters remained unchanged. But most ceramics have essentially no ductility. However, they are effective in armor because they present a high resistance to deformation from compressive loads.

We can identify several parameters associated with this characteristic with the aid of Fig. 8. Shown in this figure are Hugoniot curves for several ceramics and the reflected Hugoniot for steel. The intersection of the reflected Hugoniot curve for steel with a ceramic curve gives the stress state for a one-dimensional strain impact. The penetration process, of

Fig. 7. Effect of fracture strength on time for fracture (left column: fracture strength = 3-kbar tension; right column: fracture strength = 8-kbar tension) (time in μsec).

Fig. 8. One-dimensional strain impacts of steel on ceramics.

course, involves more than one space dimension. However, some of the important aspects of material properties can be identified with reference to a one-dimensional strain impact. The slope of the ceramic curves up to the Hugoniot elastic limit correspond to the elastic impedance, $\rho^0 C_\ell$. For a given projectile impact velocity, a higher elastic impedance implies a higher impact stress. The sharp point of a projectile is vulnerable to a high stress impact.

Another parameter implicit in these curves is the compressive yield strength Y^0 . The elastic impedance is a measure of the resistance to distortion of the material, while the yield strength is a measure of the amount of distortion that may occur before an elastic failure. It is not the elastic impedance at zero stress level that is important, but the impedance or resistance to distortion at the impact stress level. The compressive yield strength, Y^0 , is the material parameter that determines the stress level where the impedance of the material changes value.

A high compressive strength Y^0 when taken alone does not necessarily imply a good candidate for armor material. This point can be appreciated by referring to the curve for Si in Fig. 8. The yield strength for Si is $Y^0 = 0.09$ Mbar. This is higher than the yield strength for AD85, which is $Y^0 = 0.038$ Mbar. However, the ballistic limit for a target with Si is much lower than a target with AD85 (Table 1). The low ballistic limit for Si can be understood by noting that the impact stress level is much lower for Si than it is for AD85 (Fig. 8). Actually, high impedance and high yield strength are components of a lumped parameter which is incompressibility at the impact stress.

From this discussion, it is obvious that a high elastic impedance and a high yield strength are important parameters when they exist together in the material. For these characteristics to be maintained as long as possible during the impact process, the tensile properties of the ceramic become important, i.e., ultimately a tensile load is placed on the material and a tensile failure becomes the weakest link.

Table 1. Elastic properties of TiC cermets.

vol % (Ni, Mo)	Density ρ^0 (g/cc)	Longitudinal sound velocity C_ℓ (cm/ μ sec)	Shear velocity C_s (cm/ μ sec)	Bulk modulus K (Mbar)	Shear modulus μ (Mbar)	Elastic impedance $\rho^0 C_\ell$
0 ^a	4.88	0.95	0.58	2.61	1.32	4.63
8 ^b	5.56	0.93	0.57	2.41	1.81	5.17
10 ^b	5.63	0.91	0.56	2.39	1.77	5.12
14 ^b	5.81	0.89	0.54	2.39	1.69	5.17
30 ^b	6.25	0.86	0.50	2.56	1.54	5.38

^aKennametal.

^bFord Motor Company product.

Thus, the optimum ceramic should have a balance between elastic impedance, compressive yield strength, and the ability to withstand tensile stresses before fracture. These parameters are important in addition to low density. For a given target density, a low ceramic density permits a thicker ceramic facing, which in turn reduces the stress level on the backup plate,

other parameters being equal. It has already been described how the stress level on the backup plate sets the stage for subsequent events. Studies have indicated that microhardness correlates fairly well with the Hugoniot elastic limit. Hardness measurements are easy tests that can be useful in evaluating ceramic materials for application to armor.

Ductility in a Ceramic

Information obtained from the calculations described indicates that the ballistic limit should improve if a ceramic were able to sustain tensile stresses longer before an ultimate failure by fracture. It is natural to examine ballistic limit experiments with various ceramics with the idea that perhaps ductility had contributed to the ballistic limit. Of the experiments available, all the ballistic limits except BeO correlated with the material properties of incompressibility at the impact stress and low density. When tested at equal ceramic thickness, the ballistic limit for BeO is greater than the value for B_4C and is almost equal to the value for hot pressed Al_2O_3 .

Figure 8 shows that the Hugoniot curve for BeO lies below the curves for B_4C and hot pressed Al_2O_3 . It was suspected that BeO might possess better tensile properties than the other materials; this would compensate for the lower elastic impedance and compressive strength of BeO. Experiments that measure the shear strength of materials as a function of confining pressures showed that B_4C and hot pressed Al_2O_3 failed in a brittle manner at confining pressures up to 7.5 kbar (see

Appendix A). However, BeO yielded plastically and reached strains as high as 5 percent when the confining pressure was about 3 kbar and greater. The conclusion is that BeO has the desirable physical property of elongation before fracture as long as a confining pressure greater than 3 kbar is present. This is just the calculated stress range that is operating at the ceramic backup-plate interface during the first 20 μ sec after the impact. (See Fig. 5 of Ref. 3.)

Optical examinations of the BeO taken from high-pressure shear tests indicated that the material had indeed deformed in a plastic manner. Additional evidence of plastic flow was obtained from BeO specimens subjected to ballistic tests and examined by microscopy.

Flash x-ray experiments were done to compare the sequence of events for impacts of sharp steel projectiles on targets of equal thickness of AD85, B_4C and, BeO, all backed by aluminum. Figures 9 and 10 show results for impacts on AD85- and BeO-faced aluminum targets.

The interface between the ceramics and the aluminum is seen by the x-rays because a 5-mil-thick gold foil, 1/16 in. wide

Fig. 9. Flash radiographs of a sharp steel projectile striking 0.25-in. AD85 alumina bonded to 0.25-in. 6061-T6 aluminum. Projectile velocity = 2400 ft/sec.

$t = 2.0 \mu\text{sec}$

$t = 5.0 \mu\text{sec}$

$t = 10.0 \mu\text{sec}$

$t = 15.1 \mu\text{sec}$

$t = 20.0 \mu\text{sec}$

$t = 25.1 \mu\text{sec}$

$t = 30.0 \mu\text{sec}$

$t = 40.0 \mu\text{sec}$

Fig. 10. Flash radiographs of a sharp steel projectile striking 0.25-in. BeO bonded to 0.25-in. 6061-T6 aluminum. Projectile velocity = 2400 ft/sec.

Fig. 11. Flash radiographs of a sharp steel projectile striking targets of 0.25-in. B_4C , BeO , and AD85 each bonded to 0.25-in. 6061-T6 aluminum. Projectile impact velocity = 2400 ft/sec.

was laid along the interface at the impact center perpendicular to the x rays.³ Inspection of these figures at the same times shows that the projectile has not penetrated as far into the ceramic and is shorter (has undergone more erosion) for the impact on the BeO target compared to the AD85 target. At times $t = 30 \mu\text{sec}$ and $t = 40 \mu\text{sec}$ the gold foil reveals that the load on the aluminum backup plate is not as localized for the target with BeO as it is for the AD85 target.

Figure 11 compares B_4C , BeO , and AD85 targets struck by a sharp steel

projectile. More projectile erosion has occurred with the BeO target compared to the other two.

A series of experiments was performed using shadowgraph x rays to measure the residual mass and residual velocity for a sharp projectile penetrating AD85 and BeO . Results showed the most projectile energy is lost for an impact on the BeO .

These results mean that the capability for plastic flow, which delays the onset of tensile stresses, permits BeO to remain integral for a longer period of time.

Thus, BeO is able to present an incompressible barrier to the projectile for a longer time than other ceramics that fail

sooner when a tensile load occurs. For more data on the properties of BeO see Appendix B.

Introducing Ductility into a Ceramic

The impedance and yield strength of B_4C appear to be more than adequate for the application of this material to ceramic armor. For example, these parameters are considerably smaller in magnitude for AD85 than they are for B_4C , yet AD85 is still a good armor material. The weakness in B_4C , as well as most other ceramics, is failure by brittle fracture from a tensile load.

It might be possible to degrade the impedance and yield strength of B_4C in favor of improved ductility properties. Thus, a trade-off in impedance and strength for ductility is sought with an already low density ceramic. The values of elastic impedance and yield strength of AD85 provide a lower limit for the trade off of these parameters.

Preliminary work at LRL has considered the following systems:

TiC CERMETS

The indication that improved ductility might lead to a higher ballistic performance for ceramics has led us to the class of materials known as "cermets." The particular model system chosen for study was TiC-(Ni, Mo). This particular system has been the subject of extensive study by Humenik and coworkers⁵⁻⁸ at Ford Motor Company. A cooperative effort was set up with Humenik, and samples were obtained with metal contents ranging from 8 to 30 vol %. A control series of high

density, hot pressed TiC was obtained from Kennametal, Inc. Elastic properties of the various materials are tabulated in Table 1. Note that longitudinal sound and shear wave velocities decrease linearly with increasing metal content.

The TiC-(Ni, Mo) samples were in the form of 3-in. diameter disks, 1/4 in. thick. For ballistic testing, the disks were bonded to 6 by 6 by 1/4 in. 6061-T6 aluminum plates using a Scotchcast 221 adhesive.

Ballistic performances of the various samples are listed in Table 2. The LRL standard 30 caliber projectile was used in the tests. The TiC samples were examined by metallographic techniques to characterize the fracture behavior, and typical fracture features are shown in Figs. 12 and 13. We concluded that the fracture propagation is transgranular cleavage. It is not clear if the main fracture mode has changed due to metal additions, although the figures do show evidence of blunting of cracks at metal agglomerates. Figs. 12 and 13 show that in some cases, the crack front is stopped at the metal phase between TiC grains.

The metal phase of the TiC cermet has a density twice the value of the ceramic phase. Thus, the areal density of the target has been increased by the metal addition. However, the improved ballistic limit of the cermet (Table 2) is thought to be independent of the increased density due to the metal addition. In Table 2 a

Table 2. Ballistic performance of TiC cermets with 0.25 in. 6061-T6 aluminum backup plate.

vol % (Ni, Mo)	V_{BL} (1/4-in. TiC)
0	2500 $\pm$ 80
8	2960 $\pm$ 50
10	3050 $\pm$ 80
14	2970 $\pm$ 50
30	2810 $\pm$ 50

500 ft/sec improvement in ballistic limit is seen for a 10 vol % metal addition. This result is viewed as an encouraging test of the concept. The problem now is to develop a cermet with a low density ceramic in a low density metal.

B₄C CERMETS

The B₄C ceramic was the first choice for a light weight cermet; however, the problem is to find a low density metal that has the needed wetting behavior, and forms strong bonds with the ceramic.

Aluminum was tried initially, but proper bonding could not be obtained (see Appendix C). Further emphasis on a low-density metal was then waived so that a better understanding of the effect of bonding could be obtained. The Russian literature⁹ states that iron and copper wet B₄C. We turned our efforts to fabricating B₄C cermets with copper and iron. Samples were prepared by liquid-phase hot pressing (Appendix C).

We found that the wetting behavior of pure copper was not sufficient to permit the fabrication of structurally sound cermets, so this system was discarded. The addition of iron to B₄C showed some promise. But the complexity of the reaction

products and their interactions with the bonding process made Fe-B₄C a very poor system for studying the chemistry of bonding.

Recent bonding studies on systems of titanium with the compounds TiBe₁₂ and BeO have been very encouraging. These systems are also very desirable for light armor materials because of low density. Therefore, further work on the understanding of bonding is planned for these systems.

GRADED ARMOR

Since the cracks which eventually compromise the ceramic armor facing originate from the interface between the ceramic and the armor backing, it is natural to ask if the armor can be improved by eliminating the interface and thereby inhibiting crack formation. At least in principle, the interface between the ceramic facing and the metal backing can be eliminated by means of a transition zone which is ceramic and metal combined in proportions which change smoothly from 100 percent ceramic at the ceramic side of the zone to 100 percent metal at the metal side of the zone. We call the material in the transition zone a "graded cermet," and

20,000X

Metal phase

Cleavage cracks
in TiC grain

21,000X

25,000X

35,000X

Fig. 12. TiC + 30 vol % (Ni, Mo) ballistically tested. Negative replica Pt-Pd shadow.

we call the armor built this way a "graded armor."

To test the graded armor idea, an outside manufacturer was asked to construct, for ballistic testing, sample plates of alumina and aluminum using the

"plasma-spray" technique. The transition zone was plasma-sprayed onto a layer of ceramic produced by conventional means, then the metal backing was sprayed onto the transition zone. A smooth transition from ceramic to metal was not

9,500X

16,000X

24,000X

29,000X

Fig. 13. TiC + 30 vol % (Ni, Mo) ballistically tested. Negative replica Pt-Pd shadow.

attempted, instead a succession of small, but finite, composition changes were made.

The attempt to produce plates for ballistic testing was unsuccessful, probably because the original choice of materials

turned out to be a bad one. The large difference in the coefficient of thermal expansion and melting temperatures of the aluminum and alumina made it very difficult to bond successive layers in the transition zone. This caused warping and

cracking on the ceramic plate. The plasma-sprayed cermet was less than 90 percent of theoretical density and was unsuitable for armor applications. Finally, under ballistic loading, the plasma-sprayed aluminum behaved as a brittle, rather than ductile, material and it, too, had to be considered unsuitable for armor.

Failure to produce an alumina-aluminum graded armor suggested certain stringent requirements for the components of a graded armor. We now feel that these requirements may be met by a graded mixture of titanium and titanium dodecaboride (TiBe_{12}). An encouraging feature of this system is that the two materials make an excellent metallurgical bond.

Another promising system that is currently being evaluated at LRL is beryllium oxide-titanium.

NICKEL-WIRE-REINFORCED Al_2O_3

An attempt to gain macroscopic ductility in a ceramic was made by hot pressing a composite consisting of 20 vol % nickel wire (0.005-in. diameter by 1/4 in. long) in WESGO 300 Al_2O_3 powder. It was necessary to slurry the mixture in isopropyl alcohol in order to get a good distribution. The mixture was then vacuum-dried and hot pressed in a graphite die to a temperature of 1365°C at a pressure of 5000 psi. The maximum temperature was maintained for 30 min.

We knew the materials were dissimilar in thermal expansion properties, but felt that hot pressing might yield a dense structure. This optimism was not justified; final densities were about 85 percent of theoretical. This low density indicated that there was not a continuous ceramic phase in the material. This was borne

A. Impact side

B. Rear surface

Fig. 14. Nickel-wire-reinforced Al_2O_3 (20 vol % Ni in WESGO 300 Al_2O_3)

out in ballistic tests (see Fig. 14). The structure was not strong enough to break off the projectile point, which is a minimum requirement. At this point, it was decided to drop this particular system, but the concept should be reexamined with a system involving more compatible materials, i.e., BeO-Ti .

Experimental Ballistic Limits for Ceramics with an Aluminum Backup Plate

Experimental ballistic-limit studies with targets of AD85 alumina backed by aluminum³ showed that for a given backup-plate thickness, the ratio of the ballistic limit to the ceramic thickness was constant. This relation is true for AD85 thicknesses between 0.2 and 0.4 in. and corresponds to a large range of ballistic-limit velocities. The physical reason for this result is that the behavior of ceramic targets is strongly dependent on the support provided by the backup plate. And the behavior of the backup plate, in turn, depends on the stress transmitted to it

by the ceramic. Hence, the behavior of ceramic targets with backup plates is strongly dependent on the prevailing stress at the interface of the two materials. Now, the stress in the ceramic is directly proportional to the impact velocity since it is an elastic material. If it can be assumed that the stress in the ceramic falls off linearly with ceramic thickness, then the stress on the backup plate is proportional to the impact velocity and inversely proportional to the ceramic thickness.

We concluded from previous work that the ballistic limit is sensitive to the stress

Fig. 15. Experimental ballistic limits for B_4C bonded to aluminum. Straight-line fits of data pass through the origin.

Fig. 16. Experimental ballistic limits of ceramics bonded to 0.25-in. aluminum. Straight-line fits of data pass through the origin.

on the backup plate. With the above assumptions, the stress on the backup plate is the same when the impact velocity divided by the ceramic thickness is the same. Thus, for a given backup-plate thickness the ballistic-limit velocity, V_{BL} , divided by the ceramic thickness, Δ , should be a constant. The importance of this relation is that V_{BL}/Δ equals a constant is a line through the origin for plots of V_{BL} vs Δ . In the evaluation of new ceramics, it is desirable to make comparisons at the same ballistic limit velocity. With the

assumption that V_{BL}/Δ equals a constant, the experimental ballistic limit for only one ceramic thickness is necessary in order to determine the ceramic thickness required to defeat a given impact velocity.

The assumption that V_{BL}/Δ is a constant was checked experimentally for B_4C and BeO. Figure 15 shows V_{BL} vs Δ experimental data for B_4C with three different aluminum backup plate thicknesses. The data lie on lines through the origin. Figure 16 shows a comparison of V_{BL} vs Δ experimental data with a

Table 3. Ballistic limits for various ceramics bonded to 0.25-in. 6061-T6 aluminum.

Ceramic	Density ρ^0 (g/cc)	Δ^a (in.)	V_{BL}^b (ft/sec)	Δ_{2700}^c (in.)	$\rho^0 \Delta_{2700}^d$ (g/cm ²)	ϕ^e
B ₄ C	2.50	0.290	2700 ± 50	0.29	1.84	1.00
BeO	2.84	0.250	2500 ± 50	0.27	1.95	0.94
BeO + B	2.50	0.250	2320 ± 80	0.29	1.84	1.00
Coors AD85 Alumina	3.43	0.340	2850 ± 50	0.32	2.79	0.66
Coors AD999 alumina	3.96	0.250	2660 ± 80	0.25	2.51	0.73
WESGO 995 alumina	3.85	0.303	2870 ± 50	0.285	2.79	0.66
Diamonite alumina	3.72	0.340	3170 ± 50	0.29	2.74	0.67
Carborundum hot-pressed alumina	3.92	0.250	2550 ± 50	0.26	2.59	0.71
Sapphire	3.98	0.250	2470 ± 80	0.27	2.73	0.67
Si	2.33	0.250	1200 ± 50	0.56	3.31	0.56
SiC	3.09	0.250	2175 ± 75	0.31	2.43	0.76
SiN ₃	2.815	0.355	1450 ± 50	0.66	4.72	0.39
Quartzite	2.58	0.250	1250 ± 100	0.54	3.54	0.52
B	2.36	0.226	2100 ± 80	0.29	1.74	1.06
AlB ₁₂	2.53	0.250	2250 ± 80	0.30	1.93	0.95
Be ₂ B	2.03	0.250	2150 ± 50	0.31	1.60	1.15
TiB ₂	4.46	0.236	2270 ± 50	0.28	3.17	0.56
TiBe ₁₂	2.25	0.250	2250 ± 100	0.30	1.74	1.06
Pure TiC	4.88	0.25	2500 ± 80	0.27	3.25	0.55
TiC + (Ni,Mo) cermet	5.63	0.25	3050 ± 80	0.22	3.15	0.58
WC-(k-a)	15.24	0.25	3300 ± 200	0.20	7.7	0.24

^a Δ = ceramic thickness

^b V_{BL} = experimental ballistic limit

^c Δ_{2700} = extrapolated ceramic thickness for $V_{BL} = 2700$ ft/sec

^d $\rho^0 \Delta$ = ceramic areal density

^e $\phi = B_4C$ areal density/ceramic areal density

0.25-in. aluminum backup plate for AD 85 alumina, B_4C , and BeO . The relation, V_{BL}/Δ equals a constant, holds for these three ceramics.

Table 3 gives the experimental V_{BL} for several ceramics all backed by 0.25-in. aluminum. The fifth column of the table gives the ceramic thickness required to defeat an impact of 2700 ft/sec obtained from the assumption that V_{BL}/Δ equals a constant. The seventh column gives a figure of merit, ϕ , for the areal density of a given ceramic compared to the areal density of B_4C . The value of $\phi = 1.15$

shown for Be_2B implies a 13 percent weight saving for this material compared to B_4C to defeat the same ballistic threat. (See Appendix D for additional information on Be_2B and other beryllium compounds.)

Table 3 shows that low density ceramics are good light weight armor materials. Si appears to be an exception, but it is noted in Table 4 that the bulk modulus of this material is much less than the others; hence, it is much more compressible than the other materials, which is an undesirable characteristic.

Table 4. Elastic constants.

Material	Density ρ_0 (g/cm ³)	Bulk modulus K (Mbar)	Shear modulus μ (Mbar)	Sound speed C_ℓ (cm/ μ sec)	Hugoniot elastic limit (Mbar)	Yield strength σ_0 (Mbar)	Elastic impedance $\rho_0 C_\ell$
Coors AD85 alumina	3.43	1.54	0.83	0.88	0.06	0.038	3.02
Diamonite alumina	3.72	1.93	1.27	0.99	0.08	0.057	3.68
WESGO 995 alumina	3.81	2.14	1.41	1.03	0.084	0.060	3.92
Hot-pressed alumina	3.92	2.50	1.49	1.07	0.140	0.093	4.19
Boron carbide	2.50	1.96	1.85	1.33	0.150	0.126	3.33
Beryllium oxide	2.84	1.85	1.43	1.15	0.085	0.064	3.27
Be_2B	2.03	1.26	1.23	1.20	0.067	0.056	2.40
Al B_{12}	2.58	2.00	1.82	1.31	0.096 ^a	0.079 ^a	3.38
B	2.36	2.09	2.01	1.42	-	-	3.35
SiC ^b (KT Carborundum)	3.09	2.0	1.6	1.10	0.08	0.06	3.41
Ti Be_{12}	2.28	1.36	1.27	1.16	0.054	0.045	2.64
TiC ^c	5.63	2.39	1.77	0.91	-	-	5.12
Ti B_2	4.52	2.60	2.41	1.13	-	-	5.11
AD-999 Al_2O_3	3.95	2.56	1.58	1.09	-	-	4.30
Zr Be_{13} + 2 wt % $Zr_2 Be_{17}$	2.69	1.32	1.19	1.04	-	-	2.8
Sapphire (~58° to "C" axis)	3.98	2.36	1.79	1.09	-	-	4.35
Si $\langle 100 \rangle$	$\begin{Bmatrix} 2.33 \\ 2.33 \end{Bmatrix}$	$\begin{Bmatrix} 0.60 \\ 0.62 \end{Bmatrix}$	$\begin{Bmatrix} 0.80 \\ 0.76 \end{Bmatrix}$	$\begin{Bmatrix} 0.84 \\ 0.84 \end{Bmatrix}$	$\begin{Bmatrix} 0.090 \\ 0.080^a \end{Bmatrix}$	$\begin{Bmatrix} 0.086^a \\ 0.074^a \end{Bmatrix}$	$\begin{Bmatrix} 1.96 \\ 1.96 \end{Bmatrix}$

^aRef. 2.

^bResults are tentative due to inhomogeneous samples.

^cNi bonded.

$t = 0 \mu\text{sec}$

$t = 72 \mu\text{sec}$

$t = 16 \mu\text{sec}$

$t = 96 \mu\text{sec}$

$t = 32 \mu\text{sec}$

$t = 120 \mu\text{sec}$

$t = 48 \mu\text{sec}$

$t = 180 \mu\text{sec}$

Fig. 17. Framing camera views of the deflections of a 0.25-in. woven-roving backup plate. The 0.34-in. AD85 ceramic armor was struck by a sharp steel projectile at an impact velocity slightly above the ballistic limit.

Backup Plates

An optimum backup plate is made of a material that is stiff to a deflection, but not so stiff that a shear failure will occur at the periphery of the loading area. The plate must also be able to withstand a large deflection. Fiber glass backup plates are effective because they can undergo a large deflection, thus preventing high shear or punching stresses from occurring at the loading surface.

Figure 17 demonstrates the large excursion taken by a woven-roving fiber glass backup plate. In order to accomplish the large excursion the glass fibers must break away from the resin matrix. (See also Ref. 4.) It is thought that the finish on the glass fibers is the most important aspect of fiber glass composites used for armor because it is the

finish that influences the bond between the fibers and the resin. A study has begun to vary the bond between the glass and the resin by varying the viscosity of the finish material. The object is to find the optimum bonding strength compatible with the resistance to shear or resistance to punching failure from the applied load. Evaluation must be made by dynamic tests since the performance of fiber glass backup plates is very sensitive to loading rates. The study is also planned for Be wire partially coupled to an aluminum matrix where the aluminum is an alloy chosen for a balance of good elongation and shear strength properties. The point here is to check the understanding of a fiber glass-resin system using a metal system.

Material Physical Parameters

The measured parameters for several ceramics are summarized in Table 4. A

method for determining the strength of a material is given in Appendix E.

Conclusions

Computer calculations in conjunction with experiments have been used to gain understanding of ballistic penetration phenomena. The important target material parameters needed to defeat penetration by armor piercing projectiles have been identified. Ductility has been shown to be an important material parameter that is not present in most current ceramic armor materials. Weaknesses

in the failure process of a target have been pointed out. Direction has been given for future material development where a tradeoff in material parameters is used to obtain a better balance of the necessary parameters to defeat penetration. Experimental equation-of-state data have been obtained for many current armor materials as well as for new armor material candidates.

Recommendations

We recommend that emphasis be placed on material development for continued research in light armor. It is not necessary to do continued equation-of-state measurements and computer calculations for armor development. Nothing new is expected from continued work in this area. Cermet with not more than 10 vol % metal show promise for an improved ceramic facing. Work is necessary to understand the wetting and bonding of the metal to the ceramic phase.

We also recommend that specific cermet systems, Ti-TiBe₁₂ and Ti-BeO must be studied, and that the graded

armor concept be pursued once a satisfactory armor cermet has been fabricated.

Boron and beryllium represent the ultimate light-weight materials for armor application. Compounds of these elements should be synthesized and evaluated.

In fiber glass composites, a systematic study is necessary where the bond strength of the fibers and resin matrix is changed in a controlled way. When the optimum dynamic decoupling has been found for a given glass-resin system, a similar study should be made with glass fibers with different elongation characteristics.

Appendix A

Mechanical Properties of Armor Materials at High Pressure

The strength and ductility of most rock-forming minerals (silicates, carbonates, and oxides) as well as the brittle metals and semi-metals may be enormously enhanced by mechanically testing the material while it is confined by a high hydrostatic pressure.

When a brittle material is deformed at atmospheric confining pressure, the axial differential stress increases linearly with axial strain up to the ultimate strength (rupture strength) and then immediately falls to zero. The test sample fragments, and cohesion is lost across planes of both tensile and shear fracture. However, the shape of the stress-strain curve of the same brittle material, when tested under a confining pressure, is different. Here, due to the frictional sliding of broken fragments past one another with continued axial displacement, the differential stress does not fall to zero but decreases to a value near one-half to two-thirds of the rupture strength, and oscillates near that range until the test is terminated.

These fracture surfaces are observed to be uniquely oriented to the principal stress field, i.e., if we take σ_1 , σ_2 and σ_3 to be the three principal stresses ($\sigma_1 \geq \sigma_2 \geq \sigma_3$, tension = +), then the tensile fractures are always normal to σ_1 , and the shear fractures are always 20 to 45 degrees to σ_3 , containing σ_2 in their surfaces. Tensile fracture occurs at low mean pressures, shear at the intermediate ones, and complete ductility at high pressures. The tensile stress (σ_1) for fracture is independent

of mean pressure; the shear stress necessary to activate a shear fracture is strongly dependent on mean pressure, while true plastic flow is very nearly independent of pressure. Plastic flow at high pressures in the ordinarily brittle materials noted above involves crystallographic slip and/or twinning within individual grains of the minerals involved.

Limited test data available on ceramic materials, triaxially tested at high pressures, suggest that their behavior is similar to that of the brittle silicates and oxides noted above.

Triaxial compression tests have been performed on hot pressed Al_2O_3 , BeO , and B_4C material at confining pressures ($-\sigma_1$) ranging to 16 kbar at a strain rate of 10^{-4} /sec at 24°C. Numerical data are presented in Table A-1 (also see Fig. A-1) in the form of maximum shear stress before and after fracture (brittle), or maximum shear stress at 5 percent strain

Fig. A-1. Schematic of experimental stress configuration (see Table A-1).

Table A-1. Summary of triaxial compression tests.

Materials	$-\sigma_1^a$	Behavior	Virgin			After Fracture		
			$-\sigma_3^b$	$\frac{\sigma_1 - \sigma_3}{2}$	P^c	$-\sigma_3$	$\frac{\sigma_1 - \sigma_3}{2}$	P
BeO	0	Brittle	13.6	6.8	4.5	0	0	0
	1.0	Brittle	29.7	14.4	10.6	6.2	2.6	2.7
	2.0	Brittle	31.6	14.8	11.9	13.8	5.9	5.9
	2.0	Brittle	31.3	14.7	11.8	20.0	9.0	8.0
	3.5	Ductile ^d	36.3	16.4	14.4	-	-	-
	7.5	Ductile ^d	40.4	16.5	18.5	-	-	-
	10.0	Ductile ^d	44.2	17.1	21.4	-	-	-
Al ₂ O ₃	0	Brittle	13.3	6.7	4.4	0	0	0
	1.0	Brittle	53.3	26.2	18.4	14.2	6.6	5.4
	1.0	Brittle	43.9	21.5	15.3	11.0	5.0	4.3
	2.0	Brittle	52.7	25.4	18.9	19.4	8.7	7.8
	2.0	Brittle	46.2	22.6	16.7	17.6	7.8	7.2
	2.0	Brittle	60.8	29.4	21.6	13.4	5.7	5.8
	3.5	Brittle	55.1	25.8	20.7	32.9	14.7	13.3
	5.6	Brittle	58.3	26.4	23.1	36.0	15.2	15.7
	8.3	Brittle	71.4	31.6	29.3	43.7	17.7	20.1
	8.3	Brittle	69.7	30.7	28.7	49.5	20.6	22.0
	9.2	Brittle	62.2	26.5	26.8	33.4	12.1	17.2
	11.0	Brittle	72.6	30.8	31.5	49.8	19.4	23.9
	12.5	Brittle	71.5	29.5	32.2	36.7	12.2	20.6
B ₄ C	0	Brittle	13.0	6.5	4.3	0	0	0
	1.0	Brittle	54.2	26.6	18.7	7.3	3.1	3.1
	2.0	Brittle	62.6	30.3	22.2	20.1	9.1	8.0
	4.2	Brittle	67.3	31.6	25.2	27.4	11.6	11.9
	7.2	Brittle	81.7	37.3	32.0	35.6	14.2	16.7
	9.0	Brittle	70.2	30.6	29.3	39.0	15.0	19.0
	15.8	Brittle	91.8	38.0	41.1	54.2	19.2	28.6

Sign conventions: tensile stresses taken as positive

^a $\sigma_1 = \sigma_2 =$ confining stress^b $\sigma_3 =$ load stress

$$^c P = -\left(\frac{\sigma_1 + \sigma_2 + \sigma_3}{3}\right)$$

^dFracture only at large strain, σ_1 calculated at 5 percent strain

(ductile) vs both mean pressure and confining pressure. The Al_2O_3 and B_4C behave quite similarly; both are brittle over the entire pressure range investigated and both exhibit a 5 to 6 fold increase in shear strength as the confining pressure ranges from zero to the maximum value.

BeO possesses a transition from brittle fracture to ductile (intragranular) flow near 3-kbar confining pressure. The strength of this material was enhanced by about 2.5 fold in the brittle region as confining pressure increased; only a slight

strength increase was noted in the ductile flow region. It is expected that by increasing the deformation rate by 10^8 , strengths of Al_2O_3 and B_4C will increase less than 5 percent per decade in rate; in BeO this increase should be near 5 to 10 percent per decade in the ductile region. The brittle-ductile transition in BeO is to be expected at 5 to 7 kbar at the higher rate ($10^4/\text{sec}$).

In Fig. A-2, envelopes of shear strength as a function of confining pressure for Al_2O_3 are shown connecting the values of

Fig. A-2. Maximum shear stress before and after fracture vs confining pressure for Al_2O_3 .

ultimate fracture strength (top) and shear strength after fracture (bottom) from many tests at different values of confining pressure. Although there appears to be a rather large scatter in the data, there is no doubt that the shear strength of the virgin Al_2O_3 as well as the shear strength after fracture is increasing with confining pressure. We believe that a large portion of this scatter may be due to local small scale inhomogeneities in the starting material.

In Fig. A-3, similar data are presented for B_4C . Results here seem more

internally consistent. B_4C possesses a slightly higher strength than the Al_2O_3 but the mechanical properties of the two materials are quite similar.

The third material, BeO behaves quite differently. At confining pressures of 2 kbar or less, its stress-strain curve is that of a brittle material. But at 3.5-kbar confining pressure and higher, this material behaves like a ductile metal; it flows at nearly a constant stress difference until the load is removed. The strength does not, however, decrease during the change of mechanical behavior. The shear strength

Fig. A-3. Maximum shear stress before and after fracture vs confining pressure for B_4C .

Fig. A-4. Maximum shear stress before and after fracture vs confining pressure for BeO. (The stress shown in the ductile region is at 5 percent strain.)

of BeO is plotted in Fig. A-4 as a function of confining pressure. As both the virgin strength and strength after fracture are shown, direct comparisons may be made with Al_2O_3 and B_4C in Figs. A-2 and A-3. BeO is noted to be appreciably weaker, especially at the higher confining pressures.

Clearly, to predict the mechanical response of one material to the impact of a projectile of another, the physical and

especially the mechanical properties of both have to be delineated accurately. In extrapolating the mechanical response of brittle materials from tests at ordinary laboratory conditions to impact conditions, perhaps the most important parameter is the effect of a confining pressure (hydrostatic or isostatic). The effects of strain-rate temperature and loading path are of lesser importance.

Appendix B

BeO Studies

A research program on BeO was initiated with the premise that the material's ductility improved under high pressure. Triaxial compression tests on polycrystalline B_4C , BeO, and Al_2O_3 as a function of pressure were conducted (Appendix A). B_4C and Al_2O_3 behaved in a brittle fashion at all confining pressures, while BeO deformed in a ductile manner when the confining pressure exceeded 3.5 kbar. Metallographic examination of the samples after etching indicated that dislocation motion had indeed occurred. The etching behavior was not homogeneous, and indicated that only preferred orientations of grains might be involved, i.e., the deformation was not isotropic. See Figs. B-1 and B-2 for examples of the control and deformed sample.

The material used in initial investigations was obtained from Brush Beryllium Company and was their Thermalox grade. Samples had a density of 2.84 g/cc and an average grain size of 30 μ .

Samples of Brush Beryllium Company "UOX" grade BeO powder were then fabricated at LRL by two different techniques.

One involved induction hot pressing (Fig. B-3) at 1700°C and 5000 psi in graphite dies, the other involved isostatic pressing and sintering (1850°C in H_2). The two techniques used powder from the same lot. Ballistic performance and elastic properties of the three sets of samples are given in Table B-1. It shows that the hot pressed material is inferior. This behavior is because of the orientation of the UOX powder induced by the hot-pressing operation. X-ray intensity measurement of the various samples indicated a strong $\langle 10\bar{1}0 \rangle$ (prismatic) orientation for the hot-pressed plates, and also that the "c" axis is oriented parallel to the face and perpendicular to the impact. The consequence of this orientation is twofold in that: (1) the cleavage plane in BeO is the $\langle 10\bar{1}0 \rangle$ which could lower the fracture strength for this loading geometry, and (2) basal slip can not occur. Dislocation motion occurs on the basal plane. This plane should be oriented at 45 degrees to the impact direction so that the dislocations can carry the plastic strain.

Table B-1. Ballistic and elastic properties of BeO.

Type fabrication	ρ^0 (g/cc)	C_ℓ (cm/ μ sec)	C_s (cm/ μ sec)	$\rho^0 C_\ell$	V_{BL}^a (ft/sec)
Brush Thermalox	2.84	1.15	0.703	3.27	2500 $\pm$ 50
Hot-pressed UOX	3.00	1.15	0.673	3.45	2140 $\pm$ 50
Isostatic pressed and sintered	2.87	1.15	0.701	3.30	2460 $\pm$ 75

^aBallistic limits, V_{BL} , are for 0.25-in.-thick disks of BeO bonded to 0.25-in. 6061-T6 aluminum.

Current experiments are using hot-pressed plate samples which were cored out in three orientations: (1) perpendicular to the pressing direction, (2) parallel to the pressing direction, and (3) at 45 degrees to the pressing direction. These right circular cylinders will be characterized by pole figure

measurements and then deformed in tri-axial compression as a function of confining pressure to determine their strength properties. Furthermore, single crystals have been prepared in orientations suitable for the study of their mechanical properties under tri-axial compression.

1,000X

3,000X

4,000X

5,200X

8,000X

Fig. B-1. BeO (Brush Thermalox) control. As-received material. Etched 45 min in a solution of one drop hydrofluosilic acid to 1 cc H_2O at $\sim 90^\circ C$. Note low etch pit density (dislocation density) even though etch time was 45 min. Negative replica Pt-Pd shadow.

5,000X

5,000X

5,000X

5,200X

6,000X

10,000X

Fig. B-2. BeO (Brush Thermalox) deformed at 25°C in tri-axial compression with a confining pressure of 10.0 kbar = $\sim 10^{-5}$ /sec. Etched 14 min in a solution of one drop hydrofluosilic acid to 1 cc H₂O at $\sim 90^\circ$ C. Note high (greater than $10^8/\text{cm}^2$) dislocation density. Negative replica Pt-Pd shadow.

Fig. B-3. Induction hot press.

Appendix C

B₄C Cermets

VACUUM INFILTRATION OF ALUMINUM

Initial attempts to infiltrate isostatically pressed compacts of B₄C powder with aluminum at 1000°C in Al₂O₃ crucibles were unsuccessful. To improve wetting channels throughout the B₄C compact, the isostatic compacts were made of a mixture of B₄C and aluminum powders. The B₄C powder was a 7:1 blend of 20- to 30-μ diameter and 1- to 5-μ diameter B₄C, respectively, and the aluminum powder was a reagent grade -325 mesh powder. Aluminum bar stock (99.99 percent purity) was used to provide an excess of molten aluminum to cover the compact in the crucible.

Complete wetting of the B₄C powder was not observed until the oil diffusion pump on the infiltration chamber was replaced with a turbomolecular pump. The improved vacuum, from 10⁻⁴ to 10⁻⁵ mm Hg, allowed molten aluminum to fill all pores greater than 0.35-μ diameter.

Isostatically pressed Al-B₄C compacts (3-in. diameter by 5 in. long) of 35, 30, and 20 vol % aluminum have been infiltrated with aluminum by this technique. The aluminum content in the infiltrated bar does not continue to decrease with decreasing aluminum content in the isostatic compacts. The packing density of B₄C particles ultimately determines the lower limit of aluminum content.

Subsequent heat treatments on 3-in. diameter by 1/4-in.-thick disks cut from the infiltrated bars have reduced the aluminum content by forming AlB₂. Additional heat treatments are in progress to reduce

the aluminum content below 20 vol %. Ballistic results on as-infiltrated and heat-treated disks of Al-B₄C and Al-AlB₂-B₄C indicate that these disks behave similar to pure aluminum. The B₄C in the aluminum matrix must be bonded at the interfaces in order to provide the microstructure required for the erosion of the projectile.

LIQUID-PHASE HOT PRESSING OF B₄C WITH COPPER

Commercial B₄C powder was blended in a 7:1 weight ratio, 20- to 30-μ diameter to 1- to 5-μ diameter B₄C powder respectively, to improve the compaction density of the irregular shaped B₄C powder. Copper powder of -200 mesh and iron powder of -200 mesh were V-blended and ball milled with the blend of B₄C powder to get 30 vol % Cu-B₄C and 30 vol % Fe-B₄C blends for hot pressing. Powder from the above blends was loaded into an 0.81-in. diameter graphite die, resistance heated to the desired temperature and then pressed at 5000 psi. After 1 min at the desired temperature, the die was allowed to air cool, and the pellet was extracted for evaluation.

The density of the pellets listed in Tables C-1 and C-2 was determined from the measured dimensions of the pellet and its weight. The retained metal content in the pellet was determined by the difference of initial metal content and extruded metal.

The rapid decrease in copper retained in the hot-pressed pellet above 1450°C was attributed to the preferred extrusion of

Table C-1. Summary of data on hot pressed 30 vol % Cu-B₄C powder blend. Compacts (0.81-in. diam) were held at temperature for 1 min at 5000 psi in a graphite die.

Pellet number	Hot press temp. (°C)	Dimensional density (%)	Theoretical density ^a (%)	Vol % of Cu retained
5	1700	2.43	95	1
7	1400	3.30	83.0	22.4
8	1300	3.26	83.4	21.6
9	1200	3.23	80.4	22.5
10	1525	2.37	82.0	5.8
11	1450	3.39	90.6	18.9
12	1475	2.92	87.3	12.8
13	1425	3.27	87.3	19.2
15	1500	3.39	90.0	18.5
16	1450	2.9	79.3	17.7

^aBased on actual composition after hot pressing.

copper between die and punches rather than into the finer porosity in the B₄C powder (see Fig. C-1). The density varies with the amount of retained copper and the tolerance of the die assembly. One pressfit die assembly did retain 18.5 vol % copper, but the density of less than 95 percent was still too low. The only pellet that achieved sufficient density had extruded nearly all of the copper (<1% retained). This latter pellet indicates that relatively low-temperature consolidation of B₄C is possible with liquid-phase hot pressing. Insufficient wetting of the B₄C by copper was observed under these conditions to warrant further study of the Cu-B₄C system.

LIQUID-PHASE HOT PRESSING OF B₄C WITH IRON

The initial experiments in the Fe-B₄C system used 30 vol % iron. A rapid reaction rate of the metal phase with the B₄C occurred at 1375°C, thus forming FeB and Fe₃C. The result is a rapid decrease in the amount of free iron (see Fig. C-2). This is undesirable since the object of introducing the Fe was to provide a ductile metal phase around the ceramic. Table C-2 summarizes the initial hot pressing experiments.

Metallographic examinations of pellet No. 3 showed that two major phases were present (Fig. C-3). The grey phase was identified by its high microhardness as B₄C. The white phase was identified by

Table C-2. Summary of data on hot-pressed 30 vol % Fe-B₄C powder blends. Compacts (0.81-in. diam) were held at temperature for 1 min at 5000 psi in a graphite die.

Pellet number	Hot press temp. (°C)	Dimensional density (g/cc)	Theoretical density ^a (%)	Wt % of Fe retained
1	1450	2.46	97.8	0.6
2	1350	3.69	91.8	54.7
3	1375	3.18	93.1	38.6
4	1300	3.40	84.2	55.0
5	1200	3.33	81.1	57.1
6	1450	2.53	92.4	11.6
7 ^b	1100	2.76	67.4	56.5
8	1100	3.10	75.8	56.5
9	1250	3.39	82.5	57.1
10	1400	2.72	98.2	13.6
11 ^c	1350	1.77	65.5	0.8
12 ^c	1345	3.57	90.6	46.1

^aBased on actual composition after hot pressing.

^bPressed at 2500 psi instead of 5000 psi.

^cHeld at temperature for 3 min instead of 1 min.

x-ray diffraction analysis as a mixture of Fe, FeB, and Fe₃C.

A 3-in. diameter disk is required for ballistic evaluation of a ceramic face material. The existing resistance-heated hot press is limited to disk diameters less than 1-1/2 in.; therefore an induction-heated hot press was used to consolidate the 3-in. diameter disks. The use of an induction-heated hot press resulted in a longer time to reach the consolidating temperature. The additional time allowed undesirable reactions to occur between the iron and B₄C. Table C-3 summarizes the data.

No vaporization of iron was observed in any of the 3-in. diameter hot pressing operations. This retention of iron was attributed to the press-fit dies and slow heating rate (~55°C/min). The most significant feature observed was the difference between the disks from Mix D-4 and the other five mixes. Disks from Mix D-4 achieved higher densities than the other mixes because of compound formation. Further attempts to reach similar densities (greater than 4.00 g/cc) by going to higher hot-pressing temperatures (1388° C for Mix D-6-2) were unsuccessful.

Fig. C-1. Hot pressed pellets (0.81-in. diam) from 30 vol % Cu-B₄C powder blend. Resistance heated in graphite dies and held at temperature for 1 min at 5000 psi.

Evaluation of the fracture behavior on ballistically tested samples indicated a marked increase in transgranular fracture for the high-density disks as compared to the low-density disks, implying a stronger bond between the binder and ceramic phase. Metallographic examination of pieces of the high- and low-density disks after

ballistic testing confirmed this mode of fracture. The microstructure of the low density induction-heated disks (A of Fig. C-4) was similar to that observed for the resistance hot-pressed samples (Fig. C-3), while extensive reaction at the surfaces of the B₄C particles has taken place in the high-density disks (B

of Fig. C-4 and Fig. C-5). X-ray diffraction results revealed only the presence of FeB, B_4C , and trace quantities of graphite in these disks.

Emission spectrographic results on the original B_4C powder used in the D-4 mixes are listed in Table C-4. The increased reaction at the surface of B_4C particles

in Mix D-4 is attributed to the excessively high impurity content in the B_4C powder.

In summary, the unanticipated formation of the compound FeB resulted in a stronger bond between the particles. This increased the path length for fracture, which is a desirable effect, but there was no ductile metal phase present.

Fig. C-2. Hot pressed pellets (0.81-in. diam) from 30 vol % Fe- B_4C powder blend. Resistance heated in graphite dies and held at temperature for 1 min at 5000 psi.

Table C-3. Summary of data on hot-pressed 30 vol % Fe-B₄C powder blends. Compacts (3-in. diam) were pressed at 5000 psi.

Pellet number	Hot pressing temp. (°C)	Dimensional density (g/cc)	Hg porosimeter	
			density (g/cc)	Porosity (%)
Mix D-4-				
1	1325	3.56	-	-
2	1360	4.09	4.10	7.7
3	1365	broke in die	4.40	1.8
4	1350	4.13	4.20	4.8
Mix D-4a-				
1	1350	3.74	-	-
2	1357	3.74	-	-
3	1359	3.80	-	-
4	1352	3.83	3.84	5.2
Mix D-4b-				
1	1360	3.68	-	-
2	1362	3.69	-	-
3	1362	3.72	3.87	7.2
Mix D-4c-1	1360	3.69	-	-
Mix D-5-				
1	1352	3.65	-	-
2	1358	3.65	-	-
Mix D-6-				
1	1362	3.70	-	-
2	1388	3.72	-	-

Table C-4. Emission spectrographic results of B_4C powders used in the 30 vol % Fe- B_4C mixes.

Mix number B_4C size	D-4		D-4a, 4b, 4c	
	Fine	Coarse	Fine	Coarse
Fe	>5000	4000	200	5000
Al	800	>>1000	1000	>1000
Si	600	>1000	> 800	400
Ca	400	400	400	400
Cr	300	400	400	200
Zr	150	250	250	150
Ni	300	300	250	250
Ti	100	100	100	200
Cu	60	60	125	18
Pb	100	100	100	<20
Mn	70	45	4	500

Fig. C-3. Microphotograph (500X) of 30 vol % Fe- B_4C powder after hot pressing at 1375°C for 1 min in a graphite die.

A. Typical of low density (~ 3.7 g/cc) 3-in. diam pellets.

B. Typical of high density (~ 4.1 g/cc) 3-in. diam pellets.

Fig. C-4. Photomicrographs (200X) of 30 vol % Fe-B₄C powder consolidated by induction heated hot pressing at 5000 psi in graphite dies.

Fig. C-5. Photomicrograph (500X) of 30 vol % Fe-B₄C powder consolidated by induction heated hot pressing at 5000 psi in a graphite die.

Appendix D

Beryllium Compounds

BERYLLIUM BORIDES

A class of materials that boasts the desirable properties of low density, high hardness, and high sound speeds appears to be optimized in the borides and beryllides. The beryllium-boron system is therefore promising for developing new armor materials. A summary of the main beryllium-boron compounds is given in Table D-1. Table D-2 gives the elastic properties of several beryllium compounds.

The LRL program centered about diberyllium boride (Be_2B), which has these properties:

- (1) Low density
- (2) An adequate hardness which implies an adequate Hugoniot elastic limit
- (3) High sound velocities which implies incompressibility at least at the zero stress level.

The original powder was purchased from a vendor and the samples fabricated

Table D-1. Properties of beryllium boride.

Compound	Theoretical ρ^0 (g/cc)	Crystal structure	Melting point ($^{\circ}\text{C}$)	Hardness (HK_{100})
Be_4C	2.01	Tetragonal	~ 1160	600-900
Be_2B	1.89	Cubic	~ 1500	~ 1300
BeB_2	2.42	Hexagonal	> 1970	3180
BeB_6	2.35	Tetragonal	2020-2120	2580
BeB_{12}	2.42	Hexagonal	-	-

Table D-2. Elastic properties of beryllium compounds.

Compound	ρ^0 (g/cc)	C_ℓ (cm/ μsec)	C_s (cm/ μsec)	$\rho^0 C_\ell$	K (Mbar)	μ (Mbar)	Y^0 (Mbar)
BeO^a	2.84	1.15	0.703	3.27	1.85	1.43	0.064
TiBe_{12}	2.28	1.14	0.75	2.60	1.23	1.28	0.045
Be_2B	2.03^b	1.27	0.800	2.58	1.52	1.30	0.057
BeB_2	2.25^c	1.28	0.828	2.88	1.63	1.54	-

^aBrush-Thermalox Grade BeO.

^bDensity is higher than the theoretical density, Table D-1, because of a small amount of BeB_2 .

^cDensity is lower than the theoretical density, Table D-1, because of a small amount of Be_2B .

at LRL. The technique used was conventional hot pressing in graphite or graphite lines with BN dies. The pressing conditions were 1060° C for a half-hour at a pressure of 5400 psi. The samples were pressed in the form of 3-in. disks of varying thicknesses. These samples were evaluated for ballistic and equation-of-state properties. (See Tables 3 and 4.) Typical fractography pictures of pieces received after ballistic tests are shown

in Figs. D-1 and D-2. The salient features are cleavage and twinning. There is as yet no evidence of macroscopic ductility.

The obvious advantage of this material for armor is the low density which permits a greater thickness per equivalent areal density. The ballistic results shown in Table 3 show that this is the highest performing armor material currently available.

Bright field at 600X

Polarized light at 600X

Fig. D-1. Twinning in ballistically tested Be₂B.

Ti AND Zr BERYLLIDES

The only beryllide compositions currently under consideration are the compounds TiBe_{12} and ZrBe_{13} . Samples of

TiBe_{12} were purchased from a vendor and ballistically tested. A typical fracture replica is shown in Fig. D-3. The ballistic results are promising in that they are superior to B_4C (Table 3).

R12

3000X

R10

1500X

R11

3000X

Fig. D-2. Typical fractograph of ballistically tested Be_2B .

3000X

5000X

6000X

Fig. D-3. TiBe₁₂ impact fracture. Negative replica Pt-Pd shadow.

Appendix E

Strength of Materials Measurements

When a projectile of an elastic-plastic material strikes a rigid boundary, the stress at the impact end reaches the elastic limit and plastic flow occurs. A plastic wave front moves back into the projectile and decelerates the undeformed part of the projectile. The magnitude of the decelerating force applied to the still elastic portion of the projectile is related to the yield stress of the material. A strong material would be stopped sooner than a weak material, other factors being equal.

Calculations show that the final length of a blunt cylinder that impacts a rigid boundary is very sensitive to the strength of the cylinder material, Y^0 . In experiments we used 0.762-cm diameter, 2.347-cm length cylinders of 1090 steel, fired at a rigid boundary provided by a 0.34 in. Al_2O_3 ceramic backed by 1-in. hardened steel. A streak camera followed the projectile rear surface for some of the experiment.

Figure E-1 shows the results of streak camera records compared with

Fig. E-1. Position-time for the rear surface of cylinders that have struck a rigid boundary.

calculations. A yield strength of $Y^0 = 12$ kbar was obtained by adjusting the Y^0 in calculations until the calculated cylinder minimum length agreed with an experiment where the impact velocity was 0.0175 cm/ μ sec and the final length 2.138 cm. This value of Y^0 was then used to calculate the impact shown in Fig. E-1. The sensitivity of the final cylinder length to the yield strength Y^0 is seen by comparing the dashed curve with curve 1. The strength of the material is independent of the impact velocity for the impact velocity range investigated, 0.01 to 0.06 cm/ μ sec. The target failed for impacts above 0.06 cm/ μ sec; this accounts for the calculation and experiment difference shown in curve 3.

A second series of experiments used cylinders of 6061-T6 aluminum. The results correlated very well with a yield strength $Y^0 = 4.2$ kbar for the impact velocity range 0.005 to 0.05 cm/ μ sec. For impact velocities greater than 0.05 cm/ μ sec, the temperature increase due to the impact stress reduces the material strength, and the material starts to behave like a fluid. This technique, then, permits the material strength and the upper bound of the material strength to be determined.

The yield strength of 6061-T6 aluminum obtained from Hugoniot elastic-limit experiments and tension tests is $Y^0 = 3$ kbar. The higher value, $Y^0 = 4.2$ kbar, from these dynamic tests is not considered to be a strain rate effect. Previous calculations using dislocation theory have shown that even if the material had a rate dependent strength property it would not be apparent in this experimental geometry.¹ The increase in strength is thought to be due entirely to work hardening. This conclusion is substantiated by hardness

measurements made after impact that show about a 30 percent increase in hardness in the deformed section of a cylinder. Table E-1 compares the final cylinder lengths, computed with a work hardening model for the aluminum, with the experimental value. (Also see Fig. E-2.) Agreement is very good up to impact velocity of about 0.05 cm/ μ sec. It is estimated that at this velocity the temperature is about 200° C. Material strength probably will drop as the impact velocity and temperature increase.

Cylinder initial length: $L = 2.347$ cm

Cylinder initial diameter: $D = 0.762$ cm

Fig. E-2. Cylinder geometry referred to in Tables E-1 and E-2.

Figure E-3 compares the calculated final cylinder geometry with the recovered experimental cylinder for three impact velocities of Table E-1. The agreement of the cylinder contour was found to be better when work hardening was used in the calculation. These experiments can be used to obtain information on material strength including temperature and work-hardening effects.

A third series of experiments was done with pure copper cylinders. Table E-2 gives the results of calculations with a work-hardening model and the experimental

Table E-1. Impact of 6061-T6 aluminum cylinders on a rigid boundary. (See Fig. E-2 for cylinder geometry.)

Experiment	Impact velocity (cm/ μ sec)	Experimental cylinder final length (cm)	Calculated ^a cylinder final length (cm)
1	0.005	2.330	2.33
2	0.0169	2.159	2.16
3	0.0191	2.126	2.12
4	0.0262	1.981	1.95
5	0.0283	1.892	1.90
6	0.0306	1.839	1.84
7	0.0373	1.651	1.65
8	0.0478	1.318	1.35
9	0.0603	0.8687	0.96
10	0.0757	0.457	0.59

^aCalculations use a work hardening model for the yield strength:

$$Y = \left[0.003 + \left(\frac{\gamma}{0.05 + \gamma} \right) (0.0015) \right] \text{ Mbar}$$

where

$$\gamma = 2/3 \sqrt{(\epsilon_1^p - \epsilon_2^p)^2 + (\epsilon_2^p - \epsilon_3^p)^2 + (\epsilon_3^p - \epsilon_1^p)^2}$$

$\epsilon_1^p, \epsilon_2^p, \epsilon_3^p$ = principal components of the plastic strain.

Table E-2. Impact of pure copper cylinder on a rigid boundary. (See Fig. E-2 for cylinder geometry.)

Experiment	Impact velocity (cm/ μ sec)	Experimental cylinder final length (cm)	Calculated ^a cylinder final length (cm)
1	0.0089	2.10	2.10
2	0.0123	1.96	1.96
3	0.0153	1.83	1.80
4	0.0183	1.68	1.69
5	0.0204	1.57	1.60
6	0.0210	1.51	1.57

^aCalculations use a work hardening model for the yield strength:

$$Y^0 = \left[0.0012 + \left(\frac{\gamma}{0.2 + \gamma} \right) (0.005) \right] \text{ Mbar}$$

γ is defined in Table E-1.

Fig. E-3. Calculated and experimental final shapes of 6061-T6 aluminum cylinders after an impact on a rigid boundary. Top photo: original cylinder; center photo: final shape; bottom photo: calculated shape.

values for the cylinder final lengths. Figure E-4 shows the calculated and experimental geometries.

Microphotographs* of material structure are shown in Figs. E-5 and E-6 for an aluminum cylinder and a copper cylinder.

One impact experiment was done for a pure beryllium cylinder. Figure E-7 shows microphotographs of the recovered cylinder. Note that fracture has occurred in the impact end. More experiments are required to obtain a reliable strength value.

*All specimens were ground from coarse to 3/0 silicon carbide paper and then polished several hours with 1-micron diamond paste using a Syntron vibratory table covered with a nylon microcloth. The final finish was given with Linde B alumina, also on nylon microcloth using a Syntron vibratory table.

Grain boundaries were delineated on the aluminum specimen by cathodically etching. A dichromate chemical-etching solution consisting of 100 ml H₂O, 8 ml H₂SO₄, 2 g K₂Cr₂O₇, and 4 ml NaCl was used on the copper specimen, while only polarized light was used on the beryllium specimen to reveal the microstructure.

Fig. E-4. Calculated and experimental final shapes of pure copper cylinders after impact on a rigid boundary. Top photo: original cylinder, center photo: final shape; bottom photo: calculated shape.

Aluminum

8X

a. Macrostructure of 6061 - T6 aluminum cylinder.

Impact velocity: 0.0262 cm/ μ sec.

Area A above

Area B above

(100X cathodically etched)

b. Microstructures of areas A and B.

Fig. E-5. Microphotographs of sectioned aluminum cylinder.

Copper

10X

a. Macrostructure of pure copper cylinder
Impact velocity: 0.0123 cm/ μ sec.

Area A above

Area B above

Area C above

Area D above

100X - dichromate etchant

b. Microstructures of areas A, B, C, and D.

Fig. E-6. Microphotographs of sectioned copper cylinder.

Beryllium

Bright field, 8X

Polarized light, 8X

a. Macrostructures of pure beryllium cylinder
Impact velocity: $0.0190 \text{ cm}/\mu\text{sec}$

Area A above

Area B above

Area C above

Area E above

Area D above

b. Microstructures of areas A, B, C, D, and E (polarized light, 100X)

Fig. E-7. Microphotographs of sectioned beryllium cylinder.

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