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ROCKET ENGINE OPERATIONS - NUCLEAR

REPORT NO. RN-S-0021

TO

AEC-NASA SPACE NUCLEAR PROPULSION OFFICE

FRAGMENTATION RESEARCH PROGRAM (U)

NERVA PROGRAM

CONTRACT SNP-1

MAY 1964

MASTER

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15 May 1964

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Subject: AGC Report N. RN-S-0021, Fragmentation
Research Program

Reference: (a) AGC ltr. 765:963 dated 5-8-64,
NERVA Reports under Subtask 1.6

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Enclosure (1) is forwarded in partial fulfillment of Subtask 1.6, Contract SNP-1.

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ABSTRACT

This report describes the basic mechanisms of graphite fracture resulting from destruction of the NERVA engine reactor core by explosive means. It was determined that the prime core fragmenting mechanism is mechanical interaction between neighboring core elements and engine structure produced by the explosive pressure waves. Explosive shock is effective in reducing the core material to dust sized particles in a zone very near the explosive detonation centers.

It was determined that the anisotropic nature of the reactor design, its structural components, and the core material (on a microcrystalline basis) strongly impede the capability of completely pulverizing the core material in spite of the application of large quantities of explosive energy.

It is suspected that application of an explosive destruct system during engine operation or while the engine is under elevated temperature will increase the formation of plastically deformed core material and consequently reduce the degree of fragmentation produced. All of the data and analyses on which these conclusions are based were derived from simulated core material tested at ambient temperatures.

W. D. Stinnett
REON Technical Systems Manager

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I. INTRODUCTION

This special report is submitted to AEC-NASA Space Nuclear Propulsion Office (SNPO) to describe the basic mechanisms of graphite fracture resulting from the destruction of the NERVA engine by explosive means. This information has been applied to the concurrent development programs for the Anticriticality Destruct System (Sub-Subtask 1.6.4) and Post-Operative Destruct System (Sub-Subtask 1.6.5) to provide a better understanding of the physical processes surrounding the particular destruct application.

The objective of the Fragmentation Research Program has been to investigate the fracture characteristics of graphite in order to determine the limitations in the use of explosive energy to fragment the reactor core materials of the NERVA engine. These data will also be correlated to the materials used to simulate the core for test purposes.

The publishing of this report completes Milestone 5 of Task Item 1.6 under SNPO Contract SNP-1 Mod 8, Contract Year 1963.

The work was accomplished by the Aerojet-General Corporation, Ordnance Research Division under the provisions of Sub-Subtask 1.6.1 during Contract Year 1963.

The preliminary technical evaluation of candidate anticriticality destruct systems is reported in the REON Report No. RN-S-0020 "Special Summary Report for Milestone 19," Contract Year 1963. The preliminary design of the selected ejectable anticriticality destruct system is reported in the REON Report No. RN-S-0022 "Special Summary Report for Milestone 25," Contract Year 1963. The evaluation of candidate Post-operative destruct systems is reported in the REON Report No. RN-S-0033 "Special Summary Report for Milestone 30," Contract Year 1963.

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II. SUMMARY

The objective of the Fragmentation Research Program was to study the basic mechanisms of graphite fracture so that limitations in the use of explosive energy to fragment the reactor core of the NERVA engine can be established. Because of the wide variety of parameters involved in a study of this type and since simulative materials were used exclusively, only a limited amount of pertinent data were evolved. The significant results are summarized below in two broad and overlapping areas of investigation: The Fracture Mechanisms Study and the Material Correlation Study. The data and conclusions presented are by no means complete nor of desirable accuracy. However, this information is of great value in assisting the understanding of the basic dynamic fracture processes involved in the destruction of the NERVA engine. Proper application of this data will support the development of an Anticriticality Destruct System and a Post-Operative Destruct System to provide safe nuclear reactor flight programs in the future.

The fracture mechanisms study indicated that the prime core fragmenting mechanism is mechanical interaction between neighboring core elements and engine structures that is produced by the explosive pressure waves. Explosive shock, originally believed to be responsible for the greater fraction of dust production, is limited in its action to the core material in a zone very near the explosive detonation centers. Feasible explosive Post-Operative Destruct System (PODS) designs do not permit distribution of the explosive centers into a large number of point sources within the core mass, although increasing the distribution was found to improve fragment production. Consequently, a large portion of the core material escapes the direct explosive shock action and some fraction of the core escapes the subsequent mechanical pulverization. These latter factors tend to produce a small quantity of fragments exceeding the destruct size limitation even when increased quantities of explosives are used. On the other hand, it has not been possible through testing to fracture graphite without producing some dust, even from simple blows with a hammer.

It was determined that the anisotropic or uni-directional nature of the reactor design, its structural components, and the core material (on a micro-crystalline basis) strongly impede the capability for completely pulverizing

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the core material. Under controlled conditions of extremely high explosive shock pressures, accompanied by repeated intense mechanical interaction, it was determined that sliver shaped (large l/d) core fragments exceeding the fragment size limitation of $1/32$ -inch could not be eliminated regardless of the element hole geometry, material properties, constraint conditions, and direction of destruct action. This condition is more evident in the B-4 element design than in the B-1 or solid round rod configurations primarily because of the thin web dimensions found in the B-4 geometry. Material receiving an impulse typically fractures in a manner influenced by some characteristic dimension. A characteristic dimension of 0.075-inch is present in the B-4 design derived from an average of the maximum and minimum longitudinal breakage surface lengths, measured on the diamond shaped cross-sectional coolant hole pattern. This minimum characteristic dimension, which is greater than twice the desired $1/32$ -inch maximum size, tends to influence production of a range of fragment sizes in excess of this limiting value. Characteristic grain structure dimensions would then influence fragment sizes below that induced by the larger geometric dimension effects.

Fuel element geometry has more effect upon the fragmentation size than grain structure or other physical properties. However, if an exact mass fraction relationship between these effects is required, then almost all material properties must be considered. Grain structure, grain size, grain orientation, primary direction of attack and loading strain rate all figure prominatly in the damage mechanisms for graphite which, although outwardly appearing to be extremely brittle, have remarkable elastic and plastic like properties.

From a large number of related tests in which the core element geometry (B-4, B-1, solid rod), basic material properties, and strain rates applied (explosive, mechanical) were varied over wide ranges of energy levels and confinement conditions, it was found that lateral loading (perpendicular to an element) causes much greater fragmentation and shock than longitudinal loading (parallel to an element). This factor suggests lateral explosive attack upon the reactor core material as being the most promising. Zones of high pressure and high temperature are formed in the core mass by the explosives. Some of

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these zones exist for an extended duration and compact or plastically deform the core material into large dense pieces. The deformed graphite has been observed many times among the fragments recovered from a wide range of element and destruct concept geometries tests (including the Mark D type), in the fuel depleted loaded material from Test APG-2, and in both full and small scale tests.

Application of PODS during engine operation or while the engine is under elevated temperature conditions will tend to increase the possibility of creating localized zones of plastically deformed graphite material and consequently reduce the degree of fragmentation produced. The behavior of deformed graphite under re-entry conditions has not as yet been determined. The strength of graphite increases with temperature rise up to about 2500°C, but graphite is still considered a very brittle material up to at least 1800°C.

The increased strength of graphite at high temperature will significantly reduce the efficiency of fragmentating the core, however the inclusion of fuel material should enhance the fragmentation over that which resulted from the tests conducted against unfueled graphite.

Graphite's brittleness characteristic at the lower temperatures make it highly sensitive to stress concentrations such as holes, edges, and intrained foreign particles (i.e., fuel).

High temperature engine structure under PODS attack will provide early release of confinement and probably cause a reduction in the effectiveness of the core pulverization through a premature loss of this confinement.

The presence of heavy fuel particles in the lighter graphite matrix is believed to create momentary stress concentrations and provide numerous well distributed sources of fracture. It is expected that the new coated fuel elements will aid the fragmentation because of better fuel dispersion control which should reduce the tendency to generate random fragment sizes.

The material correlation study indicated that graphite is a high incident energy absorbing material under a wide variety of conditions. The study also indicated that graphite is an extremely complex structural material whose physical nature is difficult to predict under the conditions involved in the

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destruct processes. It was determined from the test results that the unloaded B-4 simulant and the depleted fuel loaded B-4 simulant fractured more easily, produced smaller average fragment sizes, and transmitted less incident energy under all types of loading and restraint conditions than the crude 3/4-inch diameter solid graphite rods.

The extruded solid round rods transmit lateral explosive shock loads 2.5 times better than the extruded unloaded hexagonal rods having the B-4 hole pattern. The round solid rods transmit lateral explosive shock loads 1.33 times better than hexagonal rods containing the B-1 hole pattern. These values indicate a gross measure of the relative lateral attenuating properties of the various design geometries. They also represent an inverse measure of the fragmentation performance relationships brought about primarily by the geometric differences.

The B-4 hole and hexagonal element geometry machined into bars of extruded graphite material is noticeably weaker under impulses than the extruded unloaded B-4 simulant of apparently similar extrusion process quality supplied from government sources. This suggests that the fabrication process greatly influences fragmentation properties. The major source of this effect is the tough "skin" condition produced by the extrusion process which is more prevalent in the case of the B-4 design because of the larger surface area to volume ratio as compared to either the B-1 or solid rod configuration.

Graphite having larger grain structure will attenuate incident energy better than the finer grained graphite material, and will also produce more data scatter. This is true for all types of incident loading and strain-rates up to direct high brisance explosive contact. The large grained graphite provides increased void volume. The fine grained graphite is typically the reactor grade type. The fragmentation properties of the finer grained material is influenced more by the extrusion process than the large grained material because of increased grain and void alignment, and "skin" toughening.

A study of the characteristics and application of graphite is presented in Appendix A. Because of the lack of information concerning the effect of

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explosive impact upon the mechanical properties of graphite, a number of "static" theories have been postulated by experiments. The "weakest link" theory is the one most generally accepted as describing the brittle-fracture of a system of microcracks and internal stresses caused by the fabrication process.

A discussion is presented for each of the significant series of experiments conducted during the program. Results are presented substantiating in part the conclusions presented herein.

A useful research byproduct of the development of time-of-arrival and pressure instrumentation for NERVA reactor core fragmentation tests was the determination of the Hugoniot for AGSX grade graphite. The Hugoniot and shock propagation velocity equations were found to be as follows, as derived from test data up to approximately 140 Kbar.

$$V = 2.263 + 0.811 u$$

$$P = 47.5 u + 6.5 u^2$$

The data obtained thus far have established general trends which must be evaluated further if proper engineering utility is to be achieved. It is not known whether the use of actual reactor core material under ambient destruct conditions will influence the validity of test results and conclusions reached through use of simulative material. However, it is felt at this time that whatever relative effects exist, they are probably of secondary importance. The most significant relationship of unknown character is that existing during high reactor temperature destruct.

The greatest emphasis for future study should be placed on the determination of the influence of temperature on the explosive fragmentation characteristics of actual core material.

The recommendations for future studies, then, are fairly clear. First, it is imperative that an absolute relationship be established with regard to the arbitrary reference standards defined by the previous work, and upon which all variations of system and component parameters were tested and compared.

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This can be accomplished, essentially, by carefully repeating a limited number of the basic mechanical and explosive tests using the new coated fuel elements with known fabrication history, and selected to be within acceptable production tolerances. These results will permit a measure of absolute performance instead of only relative performance established thus far for all destruct system research and development tests.

Secondly, a limited number of mechanical and explosive destruct tests must be conducted at reactor operating temperatures to determine whether a significant tendency to produce massive plastic deformation of the reactor core material instead of the desirable pulverization observed under ambient destruct conditions.

From these tests and the results of all Anticriticality and Post-operative Destruct System development work plus the results of the Radiation Damage to Explosives program, the basic operational limitations of the NERVA Engine Destruct System can be firmly established.

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III. TECHNICAL DISCUSSION

A. GENERAL APPROACH

The primary objective of the Fragmentation Research Program was to define the basic mechanisms by which graphite is fragmented. In compliance with this objective, investigation of all phases of the phenomenon of graphite fracture was required so that a general correlation of graphite core element configurations, destruct system geometry, material properties, and other related variables could be made with regard to the characteristic fragment sizes produced by a given input shock-energy level. From this, engineering data could be evolved leading to the successful development of a NERVA engine flight safety system based upon explosive fragmentation of the reactor core material.

An extensive literature survey was made initially to determine the state-of-the-art of high strain rate fracture of graphite material. The results of this survey (Appendices A and B) provided evidence of the lack of useful technical information that could be applied to the problem of reducing the reactor core to dust sized particles no larger than 1/32-inch diameter. From this result, it was clear that two simultaneous efforts were required. Consequently, the Basic Fragmentation Research Program was divided into two basic areas of investigation: the Fracture Mechanisms Study, and the Material Correlation Study. Although these two areas of study were conducted for entirely different purposes, the investigation of the two were often combined into a single test with overlapping objectives. In general, the first study concerned basic experiments in a number of related areas which would support the understanding of the mechanisms of fracture independent of the energy source and the second concerned the development of correlative factors which would assist in predicting the fracture characteristics of fuel loaded core elements under specific explosive destruct conditions extending beyond the present testing capability and technical knowledge.

Because of the wide variety of uncontrolled parameters involved in the correlation study, only one technique was available for use. It consisted of the establishing an arbitrary standard of reference to which variations of

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system and component parameters were tested and compared. In this way, the primary variables could be identified and simple relationships established. Secondary variables are more difficult to identify and relate since uncontrolled combinations of secondary variables generally produce vague results. Complete parametric testing of the many variables for all combinations was determined to be beyond the scope of the program. Use of the "Latin Squares" pre-planned testing approach might have been helpful, but the large number of variables, the lack of exact analytical expressions and the unfortunate lack of simulative reactor material made this procedure unwieldy. A source of confusion in conducting a crude comparison study is the possibility of wide data scatter producing erroneous trends when only limited data points are available. Because neither expended NERVA engines nor engine components were available, it was necessary to conduct the Explosive Fragmentation Program using simulations of the engine hardware. The choice of simulants was based wherever possible upon dynamic response characteristics calculated to be similar. Accurate simulation of the core elements, however, was not possible because graphite rods that approached the NERVA B4 configuration and intergranular structure were not available. The only material available, in the quantities required, and within cost limitations was solid, round commercial graphite electrode stock (National Carbon Type AGSX). A rod diameter of 0.75 inch was chosen so that the void-volume of the solid rods approximated the void-volume of the coolant holes in the NERVA B4 configuration. The dynamic response of the solid rod simulant, however, was found to differ grossly from that of the unfueled NERVA B4 material; therefore, as a part of further study it remains essential to investigate the characteristic of this anticipated variation with respect to fueled NERVA B4 material.

With this absolute correlation data, it would be possible to calculate not only the end fragmentation results from the destruct of a "theoretical model" by a candidate destruct system, but also, it would permit the quantitative analytical interpretation and transaction of the actual full scale test results using inexpensive core material simulants, into terms of the fragment-size-distribution results to be anticipated if actual material had been used.

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B. SPECIFIC INVESTIGATIONS

Early in the investigation, it was postulated (based upon available information and limited supporting tests) that impulsive shock loading was a primary fracture mechanism. Therefore, it was decided that the influence of shock loading would be the first area of investigation. This initial postulation was later supported by the results of the Explosive Fragmentation Program conducted during Contract Year 1963.

Answers to the following questions were sought:

1. What impulsive shock-loading energy-levels are required to produce the characteristic 1/32-inch fragment size?
2. What are the factors governing the transmission of impulses to adjoining graphite members?
3. What are the shock attenuating or scattering effects produced by the exterior geometry of the fuel elements?
4. What is the attenuation or shock wave dissipation characteristics of the coolant holes in the fuel elements?
5. Is it possible to transmit sufficient energy through the core to adequately fragment the graphite elements furthest from the explosive charges?
6. Is fracture of these "extreme distance" modules a function of "impulsive shock," or a function of the crushing action (plane pressure wave) imparted by adjoining modules?
7. Is the transmission of "impulsive shock" greatly effected by temperature extremes or radiation environments?
8. What is the effect of fragmentation produced by the inclusion of metallic fuel particles in the graphite matrix?
9. What is the relationship between the fragment-size-distribution and the incidental shock pressure?
10. What is the effect on fragmentation produced by confinement of the reactor core bundle?

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It was anticipated that the answers to these questions would provide, among other things, answers to the following questions which regard the function of the PODS Mark D concept:

1. Is fulfillment of the PODS fragment-size objective possible?
2. What are the projected useful operational limits of the system (spatial distribution extremes for the explosives, maximum insertion angle for projectiles, functional temperature extremes, etc.)?
3. What is the optimum mode of utilization of available explosive energy?
4. What is the Hugoniot equation-of-state for the fracture of graphite?

Careful examination of these questions will reveal that they are all primarily concerned with the phenomena of "impulse shock loading," and therefore the solution would seem to be somewhat straight-forward if measurement of applied or resultant shock loads could be correlated with the fragmentation produced (i.e., measure the applied shock and measure the resultant fragments) over a wide range of energy levels.

In answer to these questions, the basic approach was simplified as follows.

1. To determine the shockwave impedance of massive graphite.
2. To determine the shockwave impedance and scattering between adjacent core elements of B4 configuration fuel elements.
3. To determine the shockwave impedance produced by the interior coolant holes.
4. To determine the relationship between shock energy input and the resultant fragment-size-distribution.

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C. GRAPHITE APPLICATION CONSIDERATIONS

From the intensive literature survey accomplished at the outset of the Fragmentation Research Program, a number of essential factors were noted which describe the peculiar characteristics inherent in graphite. Most of these significant facts are summarized below, as applied to the pulverization of the NERVA engine core material, rather than to the manufacture and physical properties of graphite of various forms. Appendix A is presented as a comprehensive survey of the state-of-the-art of graphite manufacture knowledge. Appendix B is presented to describe the generally accepted "static" theories concerning the effects of explosive impact upon the mechanical properties and response of graphite.

1. Structural Properties

Graphite is a material with many unique characteristics. It is a high incident energy absorbing material, and consequently, it is difficult to pulverize large quantities into dust especially when the energy source is not uniformly distributed within the graphite mass. The material has a complex nature which defies extrapolation to new and more critical environmental and operational conditions. The physical behavior of the various forms of graphite is well known at ambient temperatures and low strain-rates. However, data concerning the plastic response of graphite to extremely high strain-rates at elevated temperature are a necessary pre-requisite to the application of explosives to the NERVA engine core destruct task. Data are needed to describe the fracture characteristics of graphite over the entire thermal range extending from cryogenic to reactor-operating temperatures. To develop meaningful correlations between the various forms of graphite under consideration, from the commercial AGSX round rods, the unloaded GFE geometric simulant and to the coated fuel loaded material, knowledge of the combined thermal and fracture properties is very important. Material strength, element geometry, grain structure, and incident environment, all contribute to adequate correlation, but the relative importance of each is not known at this time. The element geometry is expected to be the variable having the greatest effect and grain structure, environment, and strength variables falling into secondary categories.

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2. Fabrication Effects

The effects of some forms of fabrication upon the fracture characteristics of graphite core elements has been noted. It was determined by numerous small tests that the B-4 hole geometry machined into extruded graphite bars is weaker under strong impulse loading than similar grade graphite extruded in B-4 hole geometry. Noticeable fracture performance differences probably exist as a result of the manufacturer's batch tolerances in grain size and alignment, void volume, and strength. The extremely hard and tough "skin" effect that the extrusion process produces on the longitudinal surfaces of the graphite element might be more significant than the hole pattern. The material is known to exhibit increased strength properties from this "skin" effect from the denser and better aligned grain structure.

It is the basic geometric differences, though, that affect the degree to which the skin influence is noticed. For instance, in the 19 hole B-4 design, the extrusion process produces significant geometry and skin effects compared to that of the 7 hole B-1 design, while the skin effects of the solid rod simulant are hardly noticeable. The per unit length, longitudinal surface area to bulk volume ratios vary from 17.6 for the B-4 to 13.2 for the B-1, to 5.33 for the solid rod. In other words, the B-4 design provides a well distributed toughening effect because of the thin web design which makes the B-4 type practically all skin effect. Yet even this extra toughness does not compensate for the basic geometric induced fragmentation weakness inherent to the B-4 design.

It was determined by a number of simple tests and by APG-2* that both the GFE unloaded B-4 simulant and the GFE depleted fuel loaded B-4 simulant fractured more easily, produced smaller average fragment sizes, and transmitted less incident energy under all types of loading conditions than did the solid round graphite rods. The conditions under which this overall observation was made was neither fully controlled nor directly correlatable. However, sufficient repetitive evidence was obtained in many tests over a period of 20 months to permit this statement to be made. Another overall

*Refer to REON No. RNS-0033

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observation is that the larger grained graphite is more attenuant to incident energy than the finer grained graphite material and also produced considerably more data scatter. The scatter effect was felt to be caused primarily by the greater batch tolerances used in the larger grained graphite manufacturing process and also the wider range in basic grain sizes used in the mix process.

Reactor grade graphite is typically fine grained and has a higher bulk density than the commercial types used for such things as electrodes. The finer grained graphite is effected more by the extrusion than the large grained material and possesses increased grain and void alignment and "skin" toughening characteristics.

3. Fuel Loading Effects

From the cursory observations made to date, no apparent differences exist as to the presence of heavy fuel particles in the graphite elements and the unloaded state. It would appear that the presence of tiny well distributed dense masses would assist the fracture process at the matrix-fuel interfaces, especially under strong shock conditions. If this is true, then more dust production is possible for the fuel in a loaded state. However, if insignificant differences exist, then future FSSE tests can be conducted entirely with the lower cost GFE unloaded B-4 simulant.

Since no material was available with which to perform tests, the influence of the new encapsulated fuel type or of the fission product status on the localized fracture mechanisms of the B-4 design is unknown. It is expected that the new encapsulated fabrication process will aid the fragmentation performance by increasing dust production because of the better fuel dispersion and graphite particle size control which should reduce the tendency toward random fragment size production in the smaller sizes. Also, any reduction in basic element strength of grain structure damage as a result of fission product buildup or chemical change during engine operation will produce better fragmentation performance for low to medium C/M ratio Mark D PODS which depend to a great degree upon repetitious mechanical impact type dust generating mechanisms.

4. Temperature Effects

Conclusive evidence has been continuously observed through tests which indicates the extreme influence that temperature and pressure have upon the detailed graphite fracture mechanisms. Under the extreme conditions involving the application of explosives to this destruct task, intense pressure and temperature zones are created within the core which compact or plastically deform a significant portion of the normally fragile graphite material into large dense pieces instead of pulverizing the material into dust. This condition has been observed in the solid round rod, B-1 (round and hexagonal element) and B-4 geometrics, in most destruct concept geometries (including Mark D and especially noted in the Mark G concept), and in both full and small scale tests. It has also been observed in the case of the depleted fuel loaded B-4 elements in test APG-2.

The temperature effects upon the physical properties of graphite and the related effects upon the fragmentation properties are not known at present. However, it is feared that any tendency toward high temperature conditions within the core regardless of the source (i.e., engine under power) will reduce the explosive fragmentation effectiveness. This is a result of the apparent capability of graphite material to absorb strong impulsive loads (at high strain rates) under extremely high temperature conditions by exhibiting plastic behavior. This characteristic behavior would tend to increase the possibility of creating localized zones of plastically deformed elements instead of the desirable dust.

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D. TEST PROGRAM

The test program was planned to study all pertinent aspects of the impulsive fracture of unfueled reactor grade graphite. Basically, five related areas were covered in the test program:

- Shock Wave Propagation Studies
- Shock Pressure vs Fragmentation Tests
- Shock Wave Instrumentation Tests
- Equation-of-State Tests
- Fracture Mechanism Tests.

A discussion of the specific details of each test series are presented in the following sections. Because of the dual function of many tests, the analytical discussion and summarization of the test results as applied to either the fracture mechanisms or material correlation phases is presented in the conclusions section.

At the outset of the testing program the inadequacy of existing "standard" test instrumentation became apparent. It was necessary as the program progressed to develop both new instrumentation and techniques by which required data could be obtained. Established measuring devices such as high-speed photography, piezoelectric crystal transducers, and ionization pin switches (commonly used in ordnance tests) produced unusable test data. It also became necessary to develop techniques to color code the graphite components, used in some tests, so that the origin of specific fragments could be determined at the conclusion of the test. Extreme difficulty in obtaining reproducible test data from small scale tests was experienced. For this reason, a great deal of effort was spent in the area of instrumentation techniques and improvement in the testing procedure.

1. Shock Wave Propagation Study Tests

This series (SPS tests) consisted of a total of 28 tests that were conducted in an attempt to photographically determine the behavior of explosive shockwaves within individual elements and within bundles of elements

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representing the reactor fuel rod geometry. Four basic test setups were used in this series: singular 0.75 inch diameter Plexiglas rods (Figure 1); multiple 0.75 inch diameter Plexiglas rod bundles representing a single 90° section of the full scale simulated reactor core (Figure 2); singular rectangular Plexiglas bars with hole patterns representing "scaled-up" B4 configuration element hole geometries (Figure 3), and radially confined bundles of simulated B4 configuration rods (Figure 4) - (machined from solid round stock).

This test series was conducted in support of both the Fracture Mechanisms and Material Correlation Studies. Plexiglas targets were used to facilitate photographing the propagation of the explosive shock wave within the elements and thereby provide a pictorial sequence to indicate the shock wave behavior when it encounters the shock wave disturbing geometrical interfaces presented in the B4 design and the solid round simulated core materials. Graphite rods were later used in an attempt to corroborate derived results. The test series was interrupted (with the close of Contract Year 1963) before studies of hexagonal B4 configuration Plexiglas modules could be performed. It was anticipated that correlation of shock wave scatter and decay across bundles of Plexiglas B4 modules, as opposed to bundles of solid round Plexiglas rods, would indicate the shock energy level transmitted to various points within the actual core bundle. With this knowledge of the shock energy required to produce a particular characteristic fragment-size-distribution (as derived from the "Disc" tests, Section III.D.2), the correlation between fragment-size-distributions of B-4 and solid round simulants could then be calculated. These data would have permitted the calculation of hypothetical B4 fragment-size-distributions based upon actual full scale test results using solid round simulated core elements.

The test series was to serve the added function of investigating a simplified method of applying a variable and known input shock pressure into graphite samples for use in the energy input V_s MMD test series (see Section III.D.2). The technique employed was to photograph the explosively induced shock wave as it propagates the length of a solid round Plexiglas rod. Measurement of the shock wave velocity could then be made from the photographs. Through use

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List of Material

- ① Detonator: X-98AA
- ② Tetryl pellets: 0.75 in. in dia $\times$ 0.75 in. long (2 each).
- ③ Blast shield: Plexiglas, 0.06 in. thick $\times$ 3.00 in. long $\times$ 3.00 in. wide.
- ④ Rod: Plexiglas, 0.75 in. in dia. $\times$ 0.75 in. to 4.125 in. in length.
- ⑤ Piezoelectric crystal assembly.

Figure 1
Piezoelectric Crystal Tube
Test Setup

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Figure 2

Test Setup: Test SPS-9

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- (1) Plexiglas: 3.00 in. thick x 6.00 in. wide x 12.00 in. long
- (2) Explosive: Composition B, loaded in a 6.250-in. -OD x 0.375-in. -thick wall x 14.00-in. -long tube. Explosive charge length: 7.5 in. (Note: For clarity, tube is not shown - charge is centered in tube.)

Figure 3

Test Setup and Materials:
Tests SPS 7, 8, 11 and 12

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Figure 4
Test MC-96 - Tube with
NERVA Core

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of the following equation for Plexiglas (Reference 1, also Section III.D.4), the shock pressure was calculated.

$$P = 5.51V^2 - 14.03V \quad (1)$$

Illustration of a typical Beckman-Whitley continuous writing streak camera photograph of shock wave propagation is presented in Figure 5. A planned modification of this concept was to place Plexiglas rods on either side of a graphite sample (in various configurations) and, by measuring the input and exit shock pressure as seen in the Plexiglas, determining the capability of graphite to absorb incidental shock energy. Use of this basic system was also planned for piezoelectric crystal transducer calibration.

A total of 22 shock wave propagation tests were conducted using singular 0.75 inch diameter Plexiglas rods. The least-squares-fit method was used to reduce the time-vs-distance film data for a specific test to shock velocity-vs-distance for that test. The shock velocity-vs-distance data for all the tests (combined data) were then used to determine the best-fit curve through the points by the least-squares-fit analysis. The best-fit curve, and the combined data, are presented in Figure 6. With the shock velocity-vs-distance curve and the shock pressure equation for Plexiglas, the curve of shock pressure-vs-distance was calculated and is presented in Figure 7. This curve can be used to determine the shock pressure at any point from the explosive/Plexiglas interface.

Four tests were conducted using the multi-perforated rectangular Plexiglas slab targets (illustrated in Figure 3) for the purpose of visually studying the dispersion of shock waves encountering these types of interfaces (semi-simulated B4 configuration hole pattern). Typical photo-sequences appear in Figures 9 through 11 which illustrate the rapid attenuation of the shock wave (within 2 to 3 inches from the explosive/graphite interface) as it encounters the hole patterns. Additional tests would be meaningful using the actual B-4 element hole pattern (in order to obtain quantitative attenuation factors).

Two tests were performed using bundles of solid round Plexiglas rods representing single quadrants of the simulated engine core structure. The

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Figure 5
Photographic Streak Used to
Determine Shock Pressure

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Figure 6
Plexiglas Analysis
Shockwave Propagation

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Figure 7

Shock Pressure vs Distance
in 0.75-in. Dia Plexiglas Rod

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Figure 8

Shock Pressure vs MMD for Graphite Explosive
Fragmentation Tests

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Figure 10
Test Results: Test SPS-12

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A. Time-of-Arrival Test.

B. Reset Generator.

Figure 11

Test Results: Test SPS-11

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test setup is illustrated in Figure 2. Radial shock wave propagation and fracture plane patterns were to be studied using photographs obtained with a Beckman-Whitley framing camera. These tests were unsuccessful because of the fire flash given off by the explosive which obscured the event.

Eight shock propagation tests (SV tests) were conducted on confined bundles of simulated B4 graphite elements using mechanical pin switches to measure shock times of arrival. See Section III.D.3 for a description of the use of pin switches. The first tests included only one row of rods around a 1.5 inch diameter x 6 inch long composition-B charge. Later, tests were conducted with two rows of rods and future tests were to have increased the number of rows to 14. This is the approximate maximum distance that a core element would be located from the casing of a projectile during reactor destruction. Figure 4 illustrates a typical test setup.

Because each flow hole was expected to impede the shock wave, each row of holes parallel to the explosive surface contained a pin switch. This was in addition to the pin switches at rod-to-rod interfaces parallel to the explosive.

Because of the limited number of tests that were conducted, much of the effort went into development of the required instrumentation and, therefore the data was incomplete. The shock transit time between switches was erratic and unreliable because of the poor performance of the pin switches. It was planned to change to piezoelectric crystals for use as time of arrival gages, but these were not developed in time (see Section III.D.3 for a discussion of crystals as time of arrival gages). Consequently, the test results were more qualitative than quantitative.

2. Shock Pressure vs Fragmentation Tests

The objective of this test series (D tests) was to determine the fragment-size distribution as a function of input shock pressure for graphite. It was anticipated that future tests would determine the pressure necessary to produce the 1/32 inch maximum fragment size required of the destruct system. The

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pressure distribution data, from the full scale reactor destruct tests, combined with this information would have indicated those areas of the reactor core not subjected to great enough shock pressure for proper fragmentation.

A simplified test setup was used, consisting of a 0.75 inch diameter x 0.5 inch thick sample of commercial grade AGSX graphite explosively shocked by two stacked tetryl pellets of equal size. Plexiglas discs 0.75 inch diameter and of various thicknesses were inserted between the explosive and the graphite to attenuate the shock wave. The result was controlled input shock pressure for the graphite. The Plexiglas thicknesses used in the test were 0.0626 inch, 0.1875 inch, and 0.525 inch.

Initial tests were conducted in air and the fragment recovery result was poor. All subsequent tests were conducted under water in a 12 inch diameter x 24 inch deep steel tank. This test setup yielded nearly 100% recovery of fragments. The fragments were then graded according to sizes using various mesh sieves, and after careful washing and drying they were weighed and the resultant mass median diameter (mmd) was calculated for each test. For any sample, the fragment sizes of 50% of the mass will be larger than the mmd.

Knowing the thickness of the Plexiglas attenuating disc, the incident shock wave velocity (V_1) and pressure (P_1) at the Plexiglas/graphite interface were read from Figures 6 and 7. These were substituted into the following equation to determine the incident particle velocity (u_1):

$$P = 10 \rho V u \quad (2)$$

where

P = shock pressure, kilobars

ρ = material density, gm/cc

V = shock propagation velocity, mm/ μ sec

u = particle velocity, mm/ μ sec

After the shock wave partially reflects from the interface, the pressure and particle velocities are the same on both sides of the interface. The desired

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point (P_D, u_D) is the intersection of the reflected Plexiglas Hugoniot and the graphite Hugoniot:

$$P = 30.1 (2u_1 - u_D) + 25.4 (2u_1 - u_D)^2 \quad (\text{Plexiglas}) \quad (3)$$

$$P = 47.5 u_D + 6.5 u_D^2 \quad (\text{graphite}) \quad (4)$$

Equations (3) and (4) were combined with u_1 found above to produce u_D , which was substituted back into Equation (4) to find the shock pressure input (P_D) for the graphite sample.

The data from 14 tests were plotted (Figure 8) to show the relationship between input shock pressure and the resulting mass median diameter (mmd) of the fragments. The curve shows the increasing difficulty of producing smaller size fragments from higher shock pressures. More tests are needed at higher input shock pressures to confirm this trend. If this is true, then the explosive charge required in an operational system intended to achieve the 1/32 inch maximum fragment size may be very large. Additional tests are necessary to determine the energy required to produce given fragment sizes. The graphite disc in future tests could be sandwiched between two Plexiglas rods and then explosively shocked. Streak cameras could record the shock propagation velocity in the two Plexiglas rods. The shock energy balances on both sides of the graphite sample could be used to determine the energy absorbed by the graphite. The resulting fragments could again be graded according to size. Such tests could be inexpensive and fast so that much data could be easily gathered regarding fracture energy of graphite.

3. Shock Wave Instrumentation Tests

Instrumentation of sub-scale and full-scale simulated engine (FSSE) destruct system tests was planned primarily for support of the Post Operation Destruct System (PODS) evaluation tests. The objective was to determine how the explosive shock wave propagated through the reactor core, how the core collapsed and fractured, and the time delay between shock arrival and reactor collapse. Knowledge of the time of arrival (TOA) of a shock wave and

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The attendant shock pressure (stress) was necessary in order to understand the phenomena occurring within the reactor core undergoing destruction, and for determining areas which require greater loading effect for proper fragmentation. Although most of the testing combined the two shock measurements (TOA and pressure), the development of the instrumentation to record these two measurements was at times conducted separately because of the different problems that were encountered.

Initially, the measurement of shock TOA was to be obtained using mechanical pin switches with the pressure calculated from these data and the equation of state of the core material (graphite). In this system, a voltage potential was applied across the pin switch and was discharged when the switch was closed, either by being mechanically crushed or by ionization as a shock wave passed it. When closed, the switch applied the potential to an oscilloscope and appeared as a pip on the horizontal sweep. By careful coding of the pin switches and using a raster attachment on the oscilloscope, the time of shock wave arrival could be determined. Several attempts were made to measure the propagation of a shock wave through a solid graphite rod using pin switches equally spaced along the rod, but with little success. The problem was that the shock pressure attenuated through the graphite, after a short distance from the explosive, to a level insufficient to mechanically close the pin switch. The reliability was poor and the variation between switches was about 50%. A brief attempt was made to increase the sensitivity of the switches but was unsuccessful.

Since knowledge of shock pressure as well as TOA data was desired from the FSSE tests and conventional pressure transducers were unacceptable, it was decided to try to measure the pressure directly at the same time that times of arrival were recorded. In the quest for a TOA system that was both sensitive and reliable at low shock pressures, several alternate systems were considered. The most promising system involved the use of piezoelectric crystals. A piezoelectric material, when stressed either mechanically or by a shock, produces electrical energy proportional to the applied stress (pressure).

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The original system conceived was expected to produce reliable data and, with additional instrumentation it was expected the shock impulse (pressure x time) could be obtained at the location of the piezoelectric crystal.

Several experiments (TAS tests) were conducted to determine the feasibility of using piezoelectric crystals as TOA gages and pressure transducers. The test setup consisted of either a Plexiglas or graphite rod with a 0.25 inch diameter x 0.1 inch thick PZT-5 (Clevite Corp.) ceramic crystal attached to one end. The other end was explosively shocked by a tetryl pellet and the crystal output was recorded on an oscilloscope via a reset generator (a device that produces one pulse when sufficient input voltage is applied). Plexiglas was used because of the abundance of data available from other tests. The test results indicated the method was highly reliable and sensitive to low shock pressures (about 4×10^5 picocoulombs output per kbar of pressure). The system jitter, the time between pressure input and signal output, was less than 1 μ sec. However, the crystal output inexplicably did not follow the expected pressure vs time history of an explosive shock wave. The conclusion was that, if properly calibrated, piezoelectric crystals could be used to determine shock TOA and pressure.

Once feasibility was proven, test efforts were directed toward improving instrumentation techniques and obtaining the required pressure calibration curves for the crystals. These tests were necessary because, although piezoelectric crystals have been used for many years as pressure transducers, they were never before used at the pressures under consideration (5×10^4 to 1.5×10^6 psi). Several tests were conducted with two types of Clevite Corporation crystals: 0.25 inch diameter x 0.1 inch thick discs of PZT-5 and 0.125 inch O.D. x 0.01 inch wall x 0.5 inch long tubes of PZT-4. The test setups for the two types of crystals were similar. Figure 1 illustrates the general arrangement of equipment for the tests. A Beckman and Whitley continuous writing streak camera was used to record shock velocity in the Plexiglas rods.

The crystal assembly for the disc tests consisted of a piece of Plexiglas with a central hole into which was placed a piece of gold foil and

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the crystal was pressure cemented in the hole on top of the foil. Gold wires were secured to the foil and to the crystal on the side opposite the foil. The two wires were connected to the oscilloscope.

The tubular crystals were made by placing a wire in the center of the crystal and filling the passage with solder. A second wire was soldered around the circumference of one end, and the assembly was completed by cementing the crystal into a piece of Plexiglas.

The results indicated the use of Plexiglas rods was a bad choice. In the Plexiglas was a forerunning elastic wave front (precursor wave) connected to a slower plastic wave form by a transition region. At high pressures the two waves were very close together and difficult to distinguish on the streak camera film. Deformation of the Plexiglas under high pressure created additional difficulties. Hence, it was decided to try water as the calibration medium for future tests.

The data from the disc crystal tests were rather disappointing. Although the crystal response was excellent and output was good, the recording system was unable to record the data. The oscilloscope response and screen persistence were too low. Further, the disc type of crystal was found to have two disadvantages. First, the polarity of the output voltage was dependent upon the direction the shock wave was moving; one face was positive and the other was negative. Secondly, the output was proportional to the angle at which the shock wave struck the crystal face. Here, an orientation error of $\pm 6^\circ$ (with the normal to the crystal face) would introduce an error of $\pm 10\%$ in peak pressure data.

Fewer tests were conducted with the tubular crystals so their application was not fully investigated. The major test result was the ringing (natural frequency oscillation) of the crystals. The result was a high frequency oscillation superimposed on the shock wave form. This was thought to be caused by crystal orientation. The shock struck the tubes radially during all tests. The tubular crystal did not have the disadvantages of the disc crystal because, with a shock wave striking perpendicular to the axis, the output was not dependent on the direction of the wave. Further, the tubular crystals were smaller and could fit into smaller places during FSSE tests.

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At this point, the tubular crystal appeared to be more adaptable to reactor core instrumentation and therefore the disc crystal was temporarily excluded from developmental experiments. A new series of tests was conducted with tubular PZT-5 crystals of the same size as in previous tests. The test setup was also similar to Figure 1 except the calibration medium was usually water instead of Plexiglas. The crystal preparation included soldering wire leads inside and outside the crystal and filling the inside with epoxy. The crystals were explosively shocked both radially and longitudinally. The previously discovered high frequency ringing was again present but was found to be less of a problem for longitudinal shocking than for radial shocking of the crystal. When the crystal was shocked longitudinally, it was excited to natural frequency ringing, but the amplitude (voltage) of the oscillation was not as high as that obtained when the crystal was shocked radially. The difficulty of shocking the tubular crystal longitudinally was that the crystal length was greater than the shock wave thickness and, therefore, the pressure recorded was the average value of the shock wave stressing the crystal. To record pressures accurately, the length of the crystal should have been small compared to the shock wave thickness. The results of a study published in the Journal of Applied Physics (Reference 2) indicated that PZT-5 crystals cannot be used as pressure transducers above approximately 20 kbar unless the pressure range is known in advance. The reason is that, above this pressure the crystal output drops sharply, then rises and then drops again as pressure increases. Consequently, a given crystal output can be indicative of several input pressures. Hence, it was concluded that crystals made from PZT-4 and PZT-5 materials could not be used as pressure transducers for FSSE tests, and that tubular crystals would have to be much shorter in order to record accurate pressures.

Several TOA tests were conducted to improve techniques and increase the system reliability. The test setup is illustrated in Figure 12 along with the recording circuit. Tubular PZT-5 crystals, of the size used previously, were placed 1 inch apart in round holes along the 0.75 inch diameter graphite rod so that the shock wave struck them radially. The immediate results

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Figure 12
Time-of-Arrival Test System

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showed the system was highly reliable if the crystals were cemented into the holes. This is a logical conclusion since considerable energy and shock pressure are lost at a material-air-crystal interface. The cementing gave an added advantage of ensuring that the crystals did not shift locations prior to testing.

Other TOA tests were conducted with 1.5 inch square graphite bars (Figure 13) but a recording difficulty arose. The system did not function at high shock velocities. Apparently, the time duration of the crystal output was inadequate to trigger the reset generator, even though the voltage was high. This problem was solved with a circuit utilizing germanium diodes (Figure 14), to maintain the applied voltage on the grid of the reset generator long enough for proper triggering.

An unexpected result of these tests was the discovery of a precursor wave in the graphite bar similar to that found earlier in the Plexiglas rods. This was noticed when the crystal output was connected directly to the oscilloscope in addition to the usual connection via the reset generator. The phenomenon was further investigated by placing a steel plate between the explosive and the graphite sample, simulating a projectile casing. The low amplitude precursor wave was still present. The phenomenon became more pronounced as the explosive-to-crystal distance was increased, but the cause was not isolated.

Two more TOA tests were conducted to determine the shock attenuation rate, and to verify estimated times-of-arrival of an explosively initiated shock in solid graphite. A large block of graphite, instrumented with 12 PZT-5 tubular crystals located up to 12 inches from the explosive, was shocked by a simulated 155 mm projectile (Figure 15). The output of each crystal, via a reset generator, was displayed on a raster oscilloscope. The shock attenuated uniformly over the first 4 inch of travel and the crystals provided excellent TOA data (Table I). However, beyond 4 inches from the projectile the data were inconsistent and erratic. This was attributed to precursor disturbance. Some precursor voltage measurements were beyond the capability of the oscilloscope (50 v) and apparently, the reset generator was triggered by the precursor wave before the shock wave arrived at the crystal.

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- ① Gas deflector: graphite, 0.375 in. thick x 6.00 in. square
- ② Block: graphite, 1.50 in. square x 5.00 in. long
- ③ Tubular Piezoelectric Crystals (four)
- ④ Explosive: Composition C-4, 1.50 in. square x 3.00 in. long

Figure 13
Test Setup and Materials:
Test TAS-58

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- ① Detonator: No. 8 EBC
- ② Line wave generator
- ③ Explosive; Composition C-4, 1.50 in. square x 3.00 in. long
- ④ Gas deflector: graphite, 4.00 in. square x 0.25 in. thick

- ⑤ Graphite block; 1.50 in. square x 3.00 in. long
- ⑥ Tubular Piezoelectric Crystals (two)
- ⑦ Cable: RG 58/U
- ⑧ Diode

Figure 14
Germanium Diode Circuit Schematic

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Figure 15

Test Setup: Piezoelectric Crystal Instrumentation
System Development Tests

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TABLE I

TEST RESULTS: TEST TAS-64

<u>Distance from Explosive/Steel Interface</u>		<u>Time*</u>	<u>Average Velocity**</u>
<u>(mm)</u>	<u>(in.)</u>	<u>(μsec)</u>	<u>(mm/μsec)</u>
34.9	1.375	17.0	2.05
60.3	2.375	37.0	1.27
85.7	3.375	57.3	1.25
111.1	4.375	73.5	1.57

*Zero time measured at explosive/steel interface.

**Measured between adjacent instrumentation points.

Although the PZT-5 crystals served as good TOA gages they made poor pressure transducers because of the multiple pressures possible at one crystal output level. On the other hand, quartz crystals have been studied extensively up to 300 kbar pressures and no phenomenon similar to this has been reported. Because of the better response of quartz crystals, a few preliminary, qualitative tests were conducted to determine applicability to FSSE instrumentation. A thin disc of x-cut quartz, 0.5 inch diameter x 0.1 inch thick, was attached to one end of a graphite specimen while the other end was explosively shocked with a plane wave generator. The crystal output, as displayed on an oscilloscope (Figure 16) was excellent but exact pressures could not be calculated because of the lack of an equation of state for graphite.

Although 70 tests had been conducted during this part of the program, more testing was required to obtain pressure calibration curves for the quartz crystals. The program was terminated prior to complete development of shock wave instrumentation equipment and techniques.

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Figure 16
Crystal Output in Preliminary
Test of Quartz-Crystal Pressure
Transducer

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4. Equation of State Tests

The use of piezoelectric crystals to measure shock wave time of arrival and pressure in the NERVA reactor fragmentation tests is discussed in the previous section. In order to use the crystal data it was necessary to account for the interface between the graphite and the crystal in shock energy balances. This required the Hugoniot equation of state for shock transmission, or simply Hugoniot of graphite. The equations of state for the piezoelectric materials were already available from other research.

The Hugoniot of a material is the locus of states which can be reached only by a shock transition from an initial ambient state and involves shock pressure (stress) and particle velocity. The Hugoniot and shock propagation velocity equations determined during this test series for commercial graphite grade AGSX (S.G. = 1.75 gm/cc) were:

$$V = 2.263 + 0.811 u \quad (5)$$

$$P = 47.5 u + 6.5 u^2 \quad (6)$$

where

V = shock propagation velocity, mm/ μ sec

P = shock pressure, kilobars (kbar)

u = material particle velocity, mm/ μ sec

The test data from which these equations were derived was limited to approximately 140 kbar. Therefore, any higher shock pressures must be treated as extrapolations, even though equation (6) is mathematically proper above this pressure.

The 63 experiments (ES tests) conducted for the purpose of determining equations (5) and (6) consisted of a small graphite specimen and a shock device. A few of the tests used a gas gun which propelled a 2.5 inch diameter aluminum projectile against a 0.75 inch diameter x 5 mm thick graphite disc to apply shock. The resulting shock pressure in the graphite was below 10 kbar. The majority of the tests used explosive shocking to achieve higher

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pressures in the graphite. This test setup is illustrated in Figure 17. The shock was produced using a baratol plane wave generator. Buffer plates placed between the explosive and the graphite were used to attenuate the shock pressure. By varying the buffer material and the thickness, the test pressure at the buffer plate/graphite interface was controlled. It was necessary to use a combination of materials in a buffer plate sandwich to achieve certain desired attenuations. The buffer plate materials were aluminum 2024, Plexiglas, lead, and copper. All graphite specimens were X-rayed before testing to detect manufacturing voids. In this sense, the specimens were probably more homogeneous than actual reactor fuel rods, but it was necessary for these experiments using small selected samples. The graphite was shocked both parallel and perpendicular to the grain.

The time of arrival of the shock wave was detected by mechanical pin switches on both faces of the buffer plate and a piezoelectric crystal on the back face of the graphite disc. The tests also doubled as pressure calibrations for the PZT-5 (Clevite Corp.) and quartz crystals, in conjunction with shock wave instrumentation (Section 3). The arrival of the shock wave was recorded on a rasteroscilloscope. For sandwiched buffer plates, only the layer immediately prior to the graphite was instrumented with pin switches.

The technique used to calculate the graphite equation of state (Hugoniot) is commonly called the Impedance Matching Method (Reference 3). The Hugoniots and velocity equations of the buffer plate materials are as listed in Table II.

TABLE II

HUGONIOTS AND VELOCITY EQUATIONS OF BUFFER PLATE MATERIALS

<u>Material</u>	<u>Velocity Equation (3)</u>	<u>Hugoniot (4)</u>
Aluminum 2024	$V = 5.37 + 1.34 u$	$P = 150.1 u + 37.5 u^2$
Plexiglas	$V = 2.646 + 2.153 u$	$P = 30.1 u + 25.4 u^2$
Lead	$V = 2.066 + 1.517 u$	$P = 234.3 u + 172.1 u^2$
Copper	$V = 3.972 + 1.478 u$	$P = 353.5 u + 131.5 u^2$

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Figure 17

Test Setup: Explosive Shock Generator

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Just prior to striking the buffer plate/graphite interface, the shocked buffer plate material is in state (P_1, u_1) in Figure 18. Part of the incident shock is transmitted across the interface into the graphite and part is reflected back into the buffer plate as a rarefaction. The reflected Hugoniot for the buffer plate is nearly (and is assumed to be exactly) a mirror image of its direct Hugoniot about prereflection state (P_1, u_1) and is a function of $(2u_1 - u_D)$. Just after the original shock strikes the interface and is partially reflected, the pressure and particle velocity are the same in both materials. This shocked state is the intersection of the reflected buffer Hugoniot and the graphite Hugoniot, point D in Figure 18. This is the desired point on the graphite Hugoniot.

The average velocities of the shock waves through the buffer plate and graphite sample were calculated from the oscilloscope time information. Using the buffer plate velocity equation, the particle velocity u_1 was obtained. Since the shock pressure in the graphite can be expressed as

$$P_G = 10 V_G u_G \quad (7)$$

and both materials are at the same pressure, the common particle velocity (u_D) was calculated by equating equation (6), with $2u_1 - u_D$ substituted for u , to equation (7). The u_D ($\neq u_G$) was substituted into equation (7) and P_G was determined.

Each test yielded one point on the graphite Hugoniot and velocity equation plots. The extreme points were thrown out and a best-fit curve was drawn through the remaining data. Equations (5) and (6) give these two curves.

5. Fracture Mechanisms Test

The Fracture Mechanisms Test series was directed toward the analysis of the physical response of graphite to applied impulsive loads. The analyses investigated the combined effects of a wide range of primary variables throughout to affect fracture, such as graphite configuration, input energy level, strain rate, shock transmission between adjacent members, and environmental (sample) temperature. The results of this test series were to contribute

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Figure 18
Impedance Matching Method

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directly to the Fracture Mechanisms Study, and indirectly to the Material Correlation Study, by defining the effects produced by these prime variables so that their relationship to graphite fracture could be determined.

The Fracture Mechanisms test series was divided into several related but distinct areas of investigation:

- (1) Static strength tests.
- (2) Dropweight mechanical impact tests.
- (3) Impulsive loading (explosive) tests.
- (4) Microscopic and density analysis of resultant fragments.

In order to maintain the significance of each test series, each series is presented separately.

a. Static Strength Tests

A series of static strength tests was conducted to determine the characteristics of the specific configurations and types of graphite employed in the fracture mechanisms tests, supplemental to the data derived from the literature search. These tests consisted of investigations of compressive, transverse, and tensile strength at high, low, and ambient temperatures to define the stress-strain relationships for the various sample materials being tested. The results of this series are discussed in Section III.D. so that the proper relationship to the data derived from the literature survey might be maintained.

b. Dropweight Mechanical Impact Tests

The original intent of the dropweight impact tests was to determine the minimum energy required to fracture various types and configurations of graphite rods for use in the Material Correlation Study. Initial test samples were generally individual round or hexagonal rods, both with and without axial holes to simulate the coolant passages of the B⁴ configuration.

Data derived from the Shockwave Propagation Test Series (Section III.D.1) indicated that the shockwave, previously considered a primary

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fracture mechanism, was attenuated very rapidly (within 2 to 3 inches from the explosive/graphite interface) and that the majority of the rod breakup was probably caused by simple mechanical crushing, as the individual rods were forced against one another and the reactor pressure vessel. The dropweight impact tests, therefore, took on new meaning, and efforts were intensified in this area to determine the parameters surrounding fracture by mechanical impact for use in the Fracture Mechanisms Study.

Individual rods and clusters of rods were subjected to mechanical impact using a conventional test apparatus (Figure 19) with a 2 K gram dropweight. Because of the influence of the method of deploying the explosives in the Mark D and Mark J (internal burster charges) post-operation destruct system concepts, the effects of transverse loading were of primary interest during the investigation. Several types of data were derived from the test series. Individual rods were impacted at various energy levels, both with and without confinement, to determine the minimum energy required to produce fracture under varying conditions. Loads were increased to determine the characteristic fragment shapes at high energy inputs. High-speed motion pictures were taken of impacted rods and rod clusters to determine both the fracture pattern within individual rods and the load or impulse transmission characteristics between elements.

Figures 20 and 21 illustrate the typical test setups that were used and the direction of the applied force. The results of these tests in terms of the fracture pattern produced and the minimum energy input required for fracture is illustrated in Figure 22 and the results are tabulated in Table III. High-speed photo sequences of fracture propagation are presented in Figures 23 through 27 and are typical of those from which the fracture diagrams were derived.

In an effort to determine the general effects produced by greater energy inputs, a short series of tests were conducted in which the energy-level was increased by 3 to 5 times the accepted minimum. An attempt was made to determine the fragment-size-distribution and the new surfaces produced for the various types of specimens as a function of impact energy.

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Figure 19
Dropweight Fixture
for Mechanical Fracture Tests

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MC-46
MC-47
MC-61
MC-62

MC-48

MC-50
MC-59
MC-60

MC-52

MC-53
MC-57

MC-51
MC-54
MC-57
MC-58

Figure 20
Test Setup and Direction
of Force: Mechanical Impact Tests

Figure 21
Hammer and Anvils Used
Through 8-Test Fracture Series

Figure 22

Test Results: Mechanical Fracture Tests

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TABLE III

MECHANICAL IMPACT TEST SERIES

Test No. And Date	Test Specimens Configuration (all specimens 2 in. long)*	Minimum Energy to Fracture (Ergs)	Comments**
MC-46 2/6/63	0.75 in. dia. extruded AGSX Grade Graphite Rod	204×10^6	See Conclusions
MC-47 2/6/63	0.75 in. hex (across flats) machined from HLM Grade Graphite extruded bar. Grain parallel to longitudi- nal axis.	204×10^6	See Conclusions
MC-48 2/13/63	Same as Test MC-46 (Specimen at 200°C)	66.5×10^6	See Conclusions
MC-50 2/13/63	Same as Test MC-47	166.4×10^6	See Conclusions
MC-51 2/13/63	Same as Test MC-47	- -	Fracture occurred similar to that in Test MC-37 (previously reported). The purpose of this test was to ob- serve the fracture with a high speed damera. There was no great differ- ence between the hex rods and the round rods used in Test MC-37.
MC-52 2/14/63	Same as Test MC-47	166.4×10^6	See Conclusions
MC-53 2/13/63	Same as Test MC-47 with B-1 hole pattern	13.65×10^6	See Conclusions
MC-54	Same as Test MC-47 with B-1 hole pattern	27.4×10^6	See Conclusions
MC-57 2/14/63	Same as Test MC-47 with B-4 hole pattern	11.6×10^6	See Conclusions
MC-58 2/14/63	Same as Test MC-57	- -	The proximity of the holes to the edge caused the rods to fracture only at the corners. The observed energy to fracture was not consid- ered valid.
MC-59 2/14/63	Same as Test MC-53	18.6×10^6	See Conclusions
MC-60 2/14/63	Same as Test MC-57	12.4×10^6	See Conclusions
MC-61 2/14/63	Same as Test MC-53	117×10^6	See Conclusions
MC-62 2/14/63	Same as Test MC-57	114×10^6	See Conclusions

* Specimen arrangements are shown in Figure D- .

** Fracture results are shown in Figure D- .

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Figure 23
Fracture Propagation;
Test MC-34

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Figure 24
Fracture Propagation;
Test MC-35

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Figure 25
Fracture Propagation;
Test MC-35

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Figure 26
Fracture Propagation;
Tests MC-39 and MC-40

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Figure 27
Fracture Propagation;
Test MC-44

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These efforts were, however, ineffective as the data scatter was excessive. The resulting fragments do, however, indicate the order-of-magnitude of the relationship of fracture to energy-input. The results of increased energy-level impact tests are tabulated in Table IV and photographs of resulting fragments are illustrated in Figures 28 through 33.

A short series of tests was conducted in which the graphite samples were heated prior to impact testing. Because an adequate inert-atmosphere furnace with impact test adaption capability was not available, it was necessary to conduct high temperature impact tests by heating the samples in an electric furnace and, when the sample stabilized at the prescribed temperature, removing the sample from the furnace and placing it in the impact apparatus. Samples were tested at temperatures ranging from ambient to 200°-1000°C. Since atmosphere and temperature control could not be maintained during the transport of the sample from furnace to test equipment, the impact data derived from this test series was too erratic to be of any quantitative value. It can be generally stated, however, that impact strength increased significantly with increased temperatures. Strength increases of the order of magnitude of 2 were indicated. Additional high temperature testing was withheld pending the fabrication of a controlled inert atmosphere drop test furnace. This equipment was only partially complete when the program was terminated.

A final test series was conducted to determine the degree of data scatter that might result from dropweight impact tests. The 900 test specimens were 2 inch lengths of 0.75 inch diameter solid round extruded AGSX graphite. Twenty specimens were impacted at each of 45 energy levels from 1.5- to 8.0-j (1 joule = 10^7 ergs). The percentage of specimens fractured at each energy level has been plotted against energy level and is presented in Figure 34. The scatter diagram obtained by plotting the fracture percentage of the first ten specimens at 0.25-j increments did not differ significantly from that of Figure 34. Further reduction in the number of data points made it quite difficult to define a curve. The previously determined minimum fracture energy of 5.35-j (Table IV, Test MC-71) for this type of specimen was based upon eight tests and, as seen in the scatter plot, falls near the 90 percent point. The

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TABLE IV

TEST DATA: TESTS MC-71 - MC-80

Test No.	Test Specimen	Direction of Impact	Minimum Fracture Energy (joules)	Energy Input for Increased Energy-Level Tests (joules)
MC-71	AGSX grade graphite 0.75-in.-dia x 2-in.-long, extruded	Radial	5.35	7.47, 9.96, 14.94
MC-72	Same as MC-71	Axial	16.43	
MC-73	HLM grade graphite, B-4 configuration without hole pattern, 2-in.-long	Radial	7.47	
MC-74	Same as MC-73	Axial	6.97	
MC-75	HLM grade graphite, B-4 configuration, 2-in.-long	Radial	0.78	14.95
MC-76	Same as MC-75	Axial	5.97	
MC-77	AGSX grade graphite, 0.75-in.-dia x 2-in.-long. Machined from 1.25-in.-dia stock	Radial	3.98	
MC-78	Same as MC-77	Axial	10.95	
MC-79	HPC grade graphite, 0.75-in.-dia x 2-in.-long. Machined from 1.5-in.-dia stock	Radial	4.48	
MC-80	Same as MC-79	Axial	11.45	

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Figure 28
Mechanical Impact Results:
Test MC-71

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Figure 29
Mechanical Impact Results:
Test MC-72

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Figure 30
Mechanical Impact Results.
Test MC-73

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Figure 31
Mechanical Impact Results:
Test MC-74

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0.775j

0.775j

14.95j

Figure 32
Mechanical Impact Results:
Test MC-75

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5. 98j

5. 98j

Figure 33
Mechanical Impact Results:
Test MC-76

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Figure 34
Scatter Diagram of Impact
Energy vs Percentage of
Fracture in Graphite

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minimum-energy values for multiple element tests should be regarded as questionable because the number of tests conducted to obtain these values does not make adequate allowance for the extreme data scatter.

c. Impulsive Loading (Explosive) Tests

Early attempts to analyze the response of graphite to impulsive loading by photographing the fracture propagation within individual explosively shock-loaded rods was largely unsuccessful. This was primarily because the camera view was obscured by the detonation products. A low level of test activity was maintained to develop rod photography techniques.

At the time of the tests, shockwaves of known intensity could not be introduced into the specimens or to measure induced shock pressures, and therefore, evaluation of initial crude tests was somewhat qualitative in nature. It was decided that until such time as shock pressures could be measured, the cause-and-effect relationships of explosive shock-loading would be investigated. Parallel effort was at this time being expended on the development of shockwave measurement techniques. It was anticipated that a more quantitative analysis of explosive loading might be conducted upon the successful completion of the instrumentation development program.

Preliminary fracture mechanisms and material correlation tests utilized a cylindrical explosive charge centrally placed in a cylindrical graphite test module to study the type of fragmentation obtained from various core element simulants made of commercial grade graphite. The envelope of the test module was a 6 inch diameter x 6 inch long right cylinder. An end view of a typical module is illustrated in Figure 35. The graphite configurations tested were: solid cylinders, 0.75 inch diameter extruded rods, machined solid hexagonal (0.75 inch across flats) rods, machined hexagonal (0.75 inch across flats) rods with a drilled Kiwi B-1 hole pattern (7 hole), machined Kiwi B-4 (19 hole) configuration, and depleted fuel loaded B-4 configuration rods. In all cases, the grain was parallel to the long axis of the configuration and the grades of graphite used were all (with the exception of the actual B-4 elements) comparable with respect to density, grain size, and strength.

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CROSS-SECTIONAL VIEW

- 1 Low-carbon steel tube, 6 in ID x 0.75-in.-thick wall x 6 in. long
- 2 AGSX grade graphite rod, 0.75 in. in dia x 6 in. long
 - Twenty-one in outer ring
 - Fifteen in center ring
 - Nine in inner ring
- 3 Explosive Charge:
 - Cast Composition B, 1.50 in. in dia x 6 in. long; 296 gm

Figure 35

Explosive Fragmentation Test Module

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Preliminary tests were conducted to size the explosive charge in order to produce a broad range of fragment sizes and types of fracture specimens for further study, including microscopic examination. A 0.435 inch diameter x 6 inch long cast Composition-B charge (approximately 25.4 gm, or a graphite/explosive mass ratio of ~150:1) produced a good distribution of fragment sizes and was sufficiently small so that enough fragments (25-75 percent of original graphite mass) remained in the test area to make a crude overall quantitative comparison of the fragment-size-distribution in addition to the qualitative comparison of the type of fracture produced.

Additional testing-method-development tests were conducted to determine the effect of end confinement on the test module fragmentation.

The test results indicated an approximate 25% increase in breakup between supporting the target on a steel plate to suspending the target 4 ft above the ground. This elevation appeared to be the limit of significant ground effect for the subscale charges. Because the results of this test series, subsequent tests were conducted by suspending the target above the ground.

The next series of tests was designed to analyze the effects produced by module confinement upon the fragmentation characteristics of a variety of types of core elements. In order to define these characteristics, the various types of available graphite configurations previously mentioned (e.g., solid cylinder, and 0.75 inch diameter solid rods) were placed within mild steel cylindrical tubes of various wall thicknesses, and fragmented by a centralized burster charge of the established size. Wall thicknesses of 0.060, 0.250, 0.500 and 0.750 inch were used. In addition, unconfined modules were tested by taping bundles of rods together to form the cylindrical configuration. In order to identify the origin of particular fragments from within the module, a technique of color coding the rods by sections was used with limited success. Circumferential bands of rods were painted contrasting colors and effectively formed an inner and outer ring as viewed from either end of the module. By this method, the effects of shockwave fracture adjacent to the explosive charge could be directly compared to the fracture produced by crushing the outer ring of rods through the explosively generated pressure wave.

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The effect of radial confinement differed according to the graphite configuration. The solid cylinder and the solid hexagonal rods fragmented in much the same manner when confined (tests MC-31 and MC-38). The graphite was forced outward against the 0.75 inch thick confining cylinder walls but was poorly fragmented except for an approximate 1.75 inch diameter axial zone through the center of the module, and spalling at either end. The hexagonal rods remaining in the cylinder were generally broken into 2 to 4 pieces with the fractures occurring across the long axis. The portion of the solid graphite cylinder remaining in the confining tube was essentially unbroken. By comparison, in test MC-10, an unconfined solid graphite cylinder was fragmented into numerous pieces, the larger fragments being of an approximate 15cc screen size. An unconfined hexagonal rod array was not tested.

An unconfined module of 0.75 inch diameter solid rods (test MC-12), color coded in two areas, was fragmented such that rods from the center area were fractured along the longitudinal axis and rods from the outer area were sheared across the axis. A comparative test (MC-32) with the rod bundle confined in a 0.75 inch thick wall steel tube produced primarily longitudinal fractures in the outer rod ring. The effect of reducing the confinement to a 0.060 inch thick wall (test MC-30) was that the longitudinal fracture of the outer rod ring was not as predominant. Approximately half of the outer rod fragments were produced by shearing across the axis. In both tests (MC-39 and MC-32) the rods in the center portion of the module were primarily fractured along the longitudinal axis.

Tests MC-49 and MC-56 were directly comparable to test MC-32 and were identical except that the test modules were comprised of hexagonal rods with a Kiwi B-1 hole pattern and machined Kiwi B-4 configuration elements respectively. Fragment recovery was 82% for MC-49 and 76% for MC-56, making a close quantitative comparison of fragment distributions somewhat questionable. The larger fragments from the modified B-1 elements consisted of intact cross-sections caused by shearing of the elements across the longitudinal axis, whereas the larger B-4 element fragments had been longitudinally fractured in addition to cross-axis shearing.

A comparative fragment-size-distribution tabulation for tests MC-26, MC-32, MC-38, MC-49 and MC-56 appears in Table V.

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TABLE V

MATERIAL CORRELATION EXPLOSIVE FRAGMENTATION TESTS

<u>Test</u>	<u>Core Material</u>	<u>Core Geometry</u>	<u>Screened Maximum Particle Size (mm vs % Recovered)</u>							<u>Unre- covered</u>
			<u>Duct</u>	<u>>0.5</u>	<u>>1.0</u>	<u>>3.4</u>	<u>>8.0</u>	<u>>16.0</u>	<u>>32.0</u>	
MC-26	HLM Grade Graphite	1 Pc Round 6-in.-dia x 6 in. long No hole pattern	0.5	1.0	1.5	2.0	2.0	3.0	80.0	10
MC-32	AGSX Grade Graphite (Extruded)	46 pc Round 0.75 in. dia x 6 in. long No hole pattern	4.0	2.0	8.0	11.0	24.0	24.0	18.0	9
MC-38	HLM Grade Graphite (machined from rectangular bar)	54 pc Hexagonal 0.75 in. (across flats) x 5 in. long No hole pattern	0.5	0.5	3.0	7.0	15.0	24	25	25
MC-49	HLM Grade Graphite (machined from rectangular bar)	54 pc Hexagonal 0.75 in. (across flats) x 5 in. long Kiwi B-1 hole pattern	4.0	2.0	13.0	20.0	25.0	8.0	9.0	19
MC-56	HLM Grade Graphite	54 pc Same as Test Mc-49	5.0	5.0	21.0	29.0	8.0	6.0	2.0	24

Because of the inadequate fragment recovery, a close quantitative analysis of fragment-size-distributions as a result of confinement effects was very difficult, and the identification and examination of the more subtle variations was virtually impossible. For this reason, a simple and effective recovery technique was developed which produced consistent recoveries of 95 to 98% of the resultant fragments.

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The recovery technique consisted of suspending the test module within a large tank of water that is violently aerated by compressed air escaping from a multi-perforated tube coiled within the tank. The air bubbles effectively negated the explosive shock and pressure waves, and thereby permitted the use of a smaller tank than would otherwise be required. The entire test module is placed within a polyethylene bag to prevent water from entering the rod array. After detonating the charge, the water is drained from the tank, during which time, the fragments are captured within a sieving unit which forms a part of the drainage system. A schematic illustration of the recovery system is presented in Figure 36.

A series of seven tests was conducted using the water tank recovery system and modules of the type used in previous tests. In the first test (MC-86) a 24 gm cylindrical explosive charge was placed in the center of three concentric rings of 0.75 inch diameter graphite rods confined in a 0.75 inch thick walled steel cylinder. Each ring of rods was color coded to aid in post-test fragment analysis. The target configuration is illustrated in Figure 35. Test MC-86, the first of the series, used 0.75 inch diameter solid rods and a 24 gm Composition B burster charge, as used previously. The resultant fragment-size-distribution is tabulated in Table VI, and is illustrated as a curve in Figure 37. As can be seen, more than 70 percent of the fragments were retained on a sieve with a 0.433 inch opening. One test (MC-90), using an explosive charge intermediate in size to those of tests MC-86 and MC-91 (54 gms), and conducted to gain additional information on the influence of charge size on fragmentation, but the results did not appear reasonable enough to report without duplicating the test for verification. This verification test was cancelled upon termination of the Contract Year 1963. In an effort to further investigate the relationship between increased explosives to reduction of fragment size, the following tests (MC-91) and some subsequent tests were made. These tests used a 132 gm booster charge for a graphite/explosive mass ratio of approximately 20:1, which more closely represents the ratio in the full scale target.

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Figure 36
Fragment-Recovery Technique Schematic

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Figure 37
Overall Fragment-Size Distribution:
Tests MC-86, MC-91, MC-92, and MC-93

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TABLE VI

FRAGMENT-SIZE DISTRIBUTION: Test MC-86

Fragment Size*		Mass (gm)	Percent of Original Mass**	Cumulative Percent
(in.)	(mm)			
				100.07
1.25	32.00	127.0	3.50	96.56
0.625	16.00	2248.5	62.00	34.56
0.312	7.93	597.0	16.50	18.06
0.132	3.33	456.0	12.60	5.46
0.0394	1.00	140.0	3.89	1.57
0.0197	0.50	31.45	0.86	0.69
less than 0.0197	less than 0.50	25.36	0.69	-

* For sizes 1.25 in. and greater, the size listed represents the length of a rod fragment with the 0.75-in.-dia either completely or partially intact. Smaller sizes represent the opening size of the sieve on which the fragment was retained.

** Discrepancy in total percentage imposed by inherent post-test weighing error.

Test MC-91 was identical to MC-86 with the exception of the explosive charge size. By comparison to the 70 percent of test MC-86, MC-91 produced 52 percent of fragments larger than the 0.433 inch screen size.

In Table VII all fragment sizes greater than 0.75 inch represent nearly whole rods (diameters intact) with the fragment size given as the length. As shown by the data in Table VII and Figure 37, many of the rods in test MC-86 were broken into 1 to 3 inch lengths. The larger fragments from test MC-86 showed little surface damage and little smearing of the color-coding paint. In test MC-91, however, the large fragments either had numerous surface cracks in planes approximately perpendicular to the longitudinal axis or had

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Test	Fragment Size ¹	(in.) (mm)	6.00 152.4	4.50 114.3	3.50 88.9	2.25 57.15	1.25 32	0.433 11.1	0.312 7.93	0.312 3.327	0.055 1.397	0.028 0.701	0.014 0.351	0.010 0.246	0.007 0.175	less than 0.007 less than 0.175
MC-91	Mass (gm)	-	-	-	-	188.5	1752.0	381.5	645.7	361.6	140.0	2.0	83.5	28.5	16.5	
	% of original mass	-	-	-	-	5.4	49.7	10.8	18.3	10.2	4.0	0.1	2.4	0.8	0.5	
	Cumulative % ²	-	-	-	102.2	96.8	47.1	36.3	18.0	7.8	3.8	3.7	1.3	0.5	-	
Inner Ring	Mass (gm)	-	-	-	-	66.0	422.9	65.0	58.6	-	-	-	-	-	-	
	% of original mass	-	-	-	-	9.7	63.0	9.55	8.63	-	-	-	-	-	-	
	Cumulative % ⁴	-	-	-	100.0	90.3	27.3	18.7	10.1	-	-	-	-	-	-	
Middle Ring	Mass (gm)	-	-	-	-	122.6	767.7	89.5	88.8	-	-	-	-	-	-	
	% of original mass	-	-	-	-	10.85	67.6	7.9	7.84	-	-	-	-	-	-	
	Cumulative % ⁴	-	-	-	100.0	89.1	21.5	13.6	5.7	-	-	-	-	-	-	
Outer Ring	Mass (gm)	-	-	-	-	-	559.0	252.4	343.7	-	-	-	-	-	-	
	% of original mass	-	-	-	-	-	35.3	15.92	21.7	-	-	-	-	-	-	
	Cumulative % ⁴	-	-	-	-	100.0	64.7	48.8	27.1	-	-	-	-	-	-	
MC-92	Mass (gm)	-	815.3	347.6	702.4	439	358.1	138.3	221.7	172.1	99.4	71.8	33.0	39.1	128.4	
	% of original mass	-	22.3	9.5	19.2	12.0	9.7	3.76	6.03	4.68	2.7	1.95	0.9	1.06	3.50	
	Cumulative % ³	100.0	77.7	68.2	49.0	37.0	27.3	23.5	17.5	12.8	10.1	8.1	7.2	6.2	2.7	
MC-93	Mass (gm)	-	76.8	309.9	613.6	848.3	576.0	153.8	311.0	202.7	101.4	64.8	25.4	23.1	80.0	
	% of original mass	-	2.16	8.72	17.3	23.9	16.2	4.33	8.75	5.7	2.86	1.83	0.71	0.65	2.25	
	Cumulative % ³	100.0	97.8	89.1	71.8	47.9	31.7	27.4	18.6	12.9	10.0	8.2	7.5	6.8	4.5	
Inner Ring	Mass (gm)	-	-	-	-	57.5	96.2	27.3	13.1	-	-	-	-	-	-	
	% of original mass	-	-	-	-	8.38	14.20	3.98	1.91	-	-	-	-	-	-	
	Cumulative % ⁵	-	-	-	-	-	-	-	-	-	-	-	-	-	-	
Middle Ring	Mass (gm)	-	-	77.9	173.0	223.8	280.9	26.5	21.3	-	-	-	-	-	-	
	% of original mass	-	-	6.83	15.15	19.60	24.60	2.32	1.87	-	-	-	-	-	-	
	Cumulative % ⁵	-	-	-	-	-	-	-	-	-	-	-	-	-	-	
Outer Ring	Mass (gm)	-	76.8	232.0	440.6	506.5	89.7	6.8	13.0	-	-	-	-	-	-	
	% of original mass	-	4.79	14.50	27.55	31.65	5.6	0.43	0.81	-	-	-	-	-	-	
	Cumulative % ⁵	-	-	-	-	-	-	-	-	-	-	-	-	-	-	

1. For sizes 1.25 in. and greater, the size represents the length of a rod fragment with the 0.75-in.-dia either partially or completely intact. Smaller sizes represent the opening of the sieve on which the fragment was retained.
2. The 2.2% gain in material is probably due to inaccuracies in the original mass determination of the material. This obviously could not be rechecked after the test was conducted; the error is small enough that the test results are not appreciably effected.
3. In these tests, the 0.175-mm screen continually clogged and was removed after draining of the test tank was partially completed; consequently, it has been assumed that the unrecovered material was less than 0.246 mm and the material unbalance appears under the 0.175-mm size in the cumulative percent tabulation.
4. In this test, all the rod fragments were identified by color through the 0.312-in. screen size. The largest size was therefore taken as 100% in the cumulative percent tabulation and the cumulative percent was calculated back from this point; this makes the 0.132-in. size cumulative percent value larger than it actually is because it includes sizes less than 0.132 in.
5. A large enough percentage of the rod fragments were unidentified by color that a cumulative percent tabulation for this test was considered to be misleading.

TABLE VII

FRAGMENT-SIZE DISTRIBUTION: TESTS MC-91, MC-92, AND MC-93

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undergone large plastic deformation, depending on their position in the module. Paint from the rods was smeared onto rods in the next ring out, indicating that they had been given a strong outward push. The fragments of test MC-91 were separated by color code reference, and the results are illustrated in Figure 38. The fragment-size-distribution difference found in test MC-91 was not noted in test MC-86. Figure 38 indicates that, in test MC-91, the rods in the center ring were fragmented less than in the other areas. It would be expected that the rods furthest from the explosive would be the least fragmented, but a visual examination of the fragments and the steel cylinder indicated that the outer rods were broken up more because they were forced against the relatively rigid steel cylinder, as evidenced by the indentations on the inner wall of the cylinder. In addition, the larger fragments from the outer ring appeared to have more surface cracks than those from the inner ring. Apparently the breakup of the outer ring of rods reduced the deceleration of the center ring to the point where a noticeable difference in fragmentation occurred. The increased fragmentation effect of the radial confinement was also noted in test MC-93.

The inner rod ring in test MC-91 did not undergo as extensive a breakup, relative to the outer rings, as might have been expected. However, a visual examination provided a probable explanation; the larger fragments had undergone large plastic deformation rather than fracture. The large plastic deformation may be caused by the fact that the rods near the explosive "feel" the strong shockwave as a hydrostatic compression. Static tests by Bridgman (Reference 4) and Von Karman (Reference 5) have shown that normally brittle materials will plastically deform under a pressure from all sides. A literature search was initiated to determine if the same phenomenon can be caused by the shockwave process. The "Locking Solids" Theory was forthcoming from the survey and has been presented in Appendix A as a possible explanation of the phenomenon.

In test MC-92 the test module of Figure 35 was modified by placing 0.75 inch thick confinement plates over the ends of the steel cylinder and joining them with four 0.375 inch diameter bolts. A 0.50 inch diameter hole was made in the center of one end plate to permit access for explosive

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Figure 38
Fragment-Size Distribution of Individual
Rod Rings: Test MC-91

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initiation. The end confinement caused a marked decrease in fragmentation, as compared with test MC-91 (Figure 37) which indicates that over 60 percent of the rods remained in lengths greater than 1.25 inch. Examination of the fragments indicated that numerous surface cracks had been formed parallel to the longitudinal axis, and that the original rod surface had eroded away, probably caused by the action of the confined detonation products. The loss of the paint color coding made it impossible to determine the original location of the fragments. As noted in the inner ring fragments of test MC-91, plastic deformation of the rods was evident, but in this test it was found in almost all of the rods, indicating that possibly the confined detonation products created a hydrostatic pressure effect.

In test MC-93, the basic test module of test MC-91 was modified by removing the steel radial-confinement cylinder and wrapping the rod assembly with one layer of masking tape to retain the cylindrical shape. Figure 38 illustrates that the fragmentation was markedly decreased from that of test MC-91. Loss of the color coding paint was significant enough that it was difficult to draw any definite conclusions from the fragment-size-distributions of individual rings. Only 30 percent of the inner ring, 70 percent of the center ring, and 85 percent of the outer ring were identifiable. Examination of the identifiable fragments did show that the large fragments from the outer rod ring had many more cracks on the surface in contact with the center ring of rods (inner surface identifiable by point smears) than on the outer surface. The increased fragmentation effect afforded by the radial confinement cylinder is evidenced by the fact that 98.5 percent (83.9 percent of the original mass) of the identifiable outer ring fragments were larger than 0.433 inch; whereas in test MC-91 approximately 30 percent of the outer ring fragments were in the same size range. The results of this test support the hypothesis that the shockwave effects are minor beyond the first or second ring of rods and that the impact of one rod upon another is the major factor causing fragmentation in the NERVA reactor core during PODS operation.

Test MC-94 was conducted in an attempt to determine the fragmentation effects produced by extreme increases in explosives. The test module was identical to that used in test MC-91 except that an explosive charge of

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296 gms of Composition B was used. The test module is illustrated in Figure 35. The explosive charge was too large for the existing underwater fragment recovery setup and, consequently, a large amount of water, apparently containing fragments, was thrown from the tank by the explosion. The fragment recovery was only 85 percent, whereas the previous minimum fragment recovery was approximately 95 percent.

Two fragment-size-distribution curves were obtained from the test results; one by assuming the unrecovered material to be of dust dimensions, and the other by assuming that the unrecovered fragments were greater than 0.433 inch screen size. These fragment-size distributions are illustrated in Figure 39. Fragments from the inner ring had undergone more extreme plastic deformation than those in the same ring in test MC-91. One of the most deformed fragments had its original 0.75 inch diameter circular cross-section deformed to an approximate ellipse with a 0.875 inch major diameter and a 0.500 inch minor diameter.

In test MC-96, the 0.75 inch thick steel confining cylinder was filled with extruded, loaded fuel depleted core elements of the B-4 configuration (Figure 4). A 132 gm centralized explosive charge was used. The test was basically the same as test MC-91 except that the solid, round-rods were replaced by the B-4 elements. The comparative fragment-size-distribution curves for the two tests are presented in Figure 40 and the results of test MC-96 are tabulated in Table VIII.

In addition to the marked difference in fragment-size distribution, the extensive plastic deformation noted in tests MC-91 and MC-94 was evident only in fragments retained on the 11.1 mm and 7.93 mm screens; this represents only 3% of the total fragment mass and is a sizable reduction in amount of plastic deformation. Two fragments were selected from each of these two screen sizes and their densities were determined. The fragments were large enough that parts of the hole pattern were retained (holes closed). Their fragment densities are listed in Table IX.

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Figure 39
Fragment-Size Distribution Comparison

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Figure 40
Fragment-Size Distribution Comparison

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TABLE VIII

FRAGMENT-SIZE DISTRIBUTION; TEST MC-96

<u>Screen Size (mm)</u>	<u>Fragment Mass Retained On Screen (gm)</u>	<u>Percent of Original Mass</u>	<u>Cumulative Percent</u>
			100.00
11.100	12.7	0.35	99.65
7.930	96.5	2.67	96.98
3.327	1318.4	36.61	60.37
1.397	1179.1	32.74	27.63
0.701	304.7	8.46	19.17
0.351	170.0	4.71	14.46
0.246	73.0	2.02	12.44
0.175	68.5	1.90	10.54*
Dust	141.8	3.94	

*6.6% unrecovered material was assumed to be dust.

TABLE IX

DENSITIES OF SCREENED FRAGMENTS

<u>Screen Size (mm)</u>	<u>Mass (gm)</u>	<u>Volume (cc)</u>	<u>Calculated Density (gm/cc)</u>	<u>Average Density (gm/cc)</u>
11.10	1.468	0.75	1.96	1.955
11.10	1.461	0.75	1.95	
7.93	0.995	0.50	1.99	1.850
7.93	1.023	0.60	1.71	

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For comparison, density measurements were made on 3 short lengths (~1-in.) of unfragmented elements, and on 3 "slivers" (inter-cooling hole web) about 2 inch long sawed from unfragmented elements. The densities of these samples are listed in Table X.

TABLE X
DENSITIES OF SAMPLES

<u>Description</u>	<u>Mass (gm)</u>	<u>Volume (cc)</u>	<u>Calculated Density (gm/cc)</u>	<u>Average Density (gm/cc)</u>
NERVA B-4 Cross-Section	8.542	4.50	1.90	1.870
NERVA B-4 Cross-Section	7.385	4.10	1.80	
NERVA B-4 Cross-Section	7.656	4.00	1.91	
"Sliver"	0.271	0.20	1.36	1.969
"Sliver"	0.467	0.25	1.87	
"Sliver"	0.446	0.24	1.86	

Five fragments each were then taken from the 3.327 mm and 1.397 mm screens and their densities were determined. The 3.327 mm screen-size fragments were in the form of "slivers" approximately 0.25 in. long and 1.397 mm screen-size fragments were more nearly spherical with an approximate diameter of 0.13 in. Their densities are shown in Table XI.

A comparison of the average densities indicates that a slight density increase occurred in the plastically deformed pieces, and that the 3.327 mm screen size, and smaller, fragments were virtually unchanged in density. To more concisely relate the actual fragment dimensions to screen size, an average fragment size was calculated for the fragments retained on the 3.327 mm and 1.397 mm screens. Samples were taken from each of the two screen-size fragments, and the "average" fragment was determined in terms of a sphere using

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densities of 1.7 gm/cc and 1.9 gm/cc. The calculated spherical diameter was compared to the average opening of the screen on which the fragments were trapped and the next larger screen that they had passed through; the average "spherical" fragment diameter was quite close to the average screen size.

TABLE XI

DENSITIES OF SCREENED FRAGMENTS

Screen Size (mm)	Mass (gm)	Volume (cc)	Calculated Density (gm/cc)	Average Density (gm/cc)
3.327	0.3890	0.2000	1.94	
3.327	0.3840	0.2100	1.83	
3.327	0.3060	0.1700	1.80	1.875
3.327	0.2890	0.1500	1.93	
3.327	0.2400	0.1400	1.72	
1.397	0.0646	0.0460	1.40	
1.397	0.0317	0.0216	1.47	
1.397	0.0790	0.0453	1.74	1.650
1.397	0.0548	0.0295	1.86	1.79
1.397	0.0325	0.0183	1.78	

Extensive fragment analysis of test MC-96 by zones was not possible because the color-coding paint did not remain on the fragments. The extruded surfaces of the NERVA B-4 elements are non-porous; consequently the paint will not adhere securely. In test MC-98, a 7/16-in.-diameter explosive charge was used in a test module similar to that of Figure 41, and the paint remained on the fragments. The fragment analysis was not completed at the end of this reporting period, however, and the results will be presented in a later report.

A preliminary test of the quartz-crystal pressure transducer in a graphite test specimen, shocked with an explosive planewave generator,

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Figure 41
Test MC-96 - Tube With NERVA Core

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produced a crystal output as shown in Figure 16. The wave shape is similar to that obtained in other materials (by other investigators), and is theoretically correct. The peak pressure is determined by finding the area under the positive portion of the crystal output curve and multiplying by appropriate constants. The impedance-mismatch constant for graphite to quartz cannot be determined until the equation-of-state testing and analysis have been completed.

Photographic instrumentation of multiple element breakup under impact loading is becoming increasingly successful; portions of a Dynafax camera sequence at 26,000 frames per second (Figures 28 through 33) indicate that element breakup occurs randomly in the module (dark areas indicate breakage of white-painted element ends). In future tests, the 1-inch-thick steel plate (3 inches wide x 3 inches long) between the 3-inch-diameter x 15-inch-long steel ram and the rods (Figures 28 through 33) will be removed in order to decrease shockwave attenuation prior to the beginning of the graphite core element module. This decrease in attenuation should enable further breakup to occur before the event is obscured by explosive detonation products and the breakup should progress more from the top to the bottom of the module than is indicated in Figures 28 through 33.

A preliminary microscopic examination of some selected graphite specimens was conducted. Prior to polishing, these graphite specimens were impregnated with epoxy resin (at 2000 psig) to prevent breakdown of the intergranular wall during mounting and polishing. The samples were selected on the basis of their unusual deformation and apparent structural modification. The mounted specimens were examined under a microscope at 200X magnification. Photographs were taken of these areas for detailed examination. These photographs are presented in Figures 42 through 44 and the tabulated test data are presented in Table XII.

Figure 42 illustrates the photomicrograph of a standard sample of AGSX graphite. By comparison, Figure 43 illustrates a specimen of AGSX graphite that, by exterior visual analysis, apparently had undergone severe plastic deformation. The photograph illustrates that the normally present voids are almost non-existent, and even in the areas where voids are present, they are small.

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Standard
AGSX Sample

Figure 42
Sample Microphotograph of
Standard AGSX Graphite

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Void Area
of Dense
Sample

Figure 43
Radially Compressed
Core Sample (Microphoto)

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Dense Area
of PO-45

Void Area
of PO-45

Figure 44
Microphotographs of Core
Samples from Test PO-45

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TABLE XII

RESULTS OF MICROSCOPIC EXAMINATION

Test	Sample Description	Area Investigated	Comment
PO-36	Deformed and broken	Dense area	Minute pores or holes, large areas filled with epoxy, less porosity than ATJ and HLM graphite.
		Void area	Large holes, very little epoxy filling holes, comparable with ATJ graphite.
PO-41	Sample dent	General structure	Small porosity, large grain size, no epoxy fill, much less porosity than ATJ graphite.
PO-41	Badly deformed sample	Dense area	Minute pores, small areas filled with epoxy, same grain size as ATJ graphite.
		Void area	Large porosity, small areas filled with epoxy, same grain as ATJ graphite.
PO-45	Sample squeezed and deformed	Dense area	Minute pores, small areas of epoxy, large grain size, very dense.
		Void area	Large pores, large area of epoxy, large grain size, much less dense.
PO-45	Sample broken at 45°, examined this facet	Dense area	Small pores, large area filled with epoxy, grain size same as ATJ graphite.
		Void area	Large holes, minute pores of epoxy, small grain size.
R-1	Deformed piece	General structure	Similar amount of medium size pores and epoxy-filled area, grain size about same as ATJ graphite.
R-3	Sample had splintered along hole (KIWI core sample)	Dense area	Large holes filled with epoxy, similar to porosity of ATJ graphite.
		Void area	Large empty holes and small holes filled with epoxy, structure similar to ATJ graphite.
PI-A	One side deformed	Dense area	Series of large holes forming a crack, large area of epoxy fill.
		Void area	Very many areas of epoxy fill, small pores.
PI-A	No damage to sample	General structure	Large grain, large holes, similar to ATJ graphite.
PI-A	Splintered along hole (KIWI core sample)	General structure	Very large grain, much more dense than ATJ graphite.
*	Sample broken and deformed	Dense area	Fine grain size, many small pores filled with epoxy.
		Void area	Medium sized pores, large grain size, similar to ATJ graphite.
*	Sample broken and deformed	General structure	Large grain size, small pores, medium pores filled with epoxy.
*	Sample broken and deformed	Dense area	Large grained graphite, very small holes, stringers of epoxy, more dense than ATJ graphite.
		Void area	Large grains, large holes, very little epoxy.
*	Sample broken and deformed	Dense area	Fine-grained graphite, very small holes, epoxy-filled stringer.
		Void area	Small holes and small epoxy-filled holes, fine grain, more dense than ATJ graphite.

* These samples were taken from Area 1-C at Chino Hills. In tests at this area, the samples were associated with uranium.

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Figure 44 is a microphoto from PODS test PO-45 (Mark G concept), and illustrates that an interesting structural change occurred. The dense areas of the graphite had been compressed so that the voids (grey lines) are uni-directional. The reasoning for this is not clear, but the material must have been compressed or worked in such a way that the void areas were compressed and elongated. Of additional interest is that this type of densification has been accomplished previously, but only at high temperatures and with special procedures and materials. At ambient conditions, graphite will normally crush and crumble as pressure is applied. Close examination of the sample did not indicate that the graphite had broken into fine pieces and had been reformed. If this transpired, the sample would have no strength and would easily crumble.

Three standard types of graphite samples, along with samples from test PO-45, were tested for compressive strength to determine the effects of densification and void displacement. The standard structures tested were: a 0.75 inch diameter AGSX solid rod, a Kiwi B-1 configuration rod, and a 0.750 inch O.D. by 15/32 inch I.D. AGSX graphite tube. The samples from test PO-45 were graphite rods which were bent, deformed, and compressed into unusual shapes. (Figure 45). The samples were machined to a diameter such that they were no longer bent (0.600 inch diameter) to allow this compression. The results of the compressive strength tests appear in Table XIII. The results indicate that the samples retained about half the compressive strength of the standard AGSX rod, even though the samples appeared to be stronger when broken by hand.

Photographic examination of multiple element breakup under impact loading became increasingly more successful during the latter portion of the program during Contract Year 1963. Portions of a Dynafax camera sequence at 26,000 frames per second (Figure 46) indicate that element breakup occurs randomly in the module (dark areas indicate breakage of white-painted element ends). Additional tests were planned in which it was anticipated that the optical resolution would be improved by reducing the quantity of shock wave attenuating materials. This would permit the shockwave to proceed into the target array prior to the camera's view being obscured by the detonation products. The program for Contract Year 1963 terminated prior to conducting a test of this type.

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Figure 45

Deformed Rods - Test PO-45

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TABLE XIII

COMPRESSIVE STRENGTH TEST RESULTS

<u>Material</u>	<u>Specimen</u>	<u>Ultimate Compressive Strength (psi)</u>	<u>Compressive Modulus (psi)</u>
Type AGSX Graphite Rod (0.75-in. dia)	1	7,500	150,300
	2	7,000	140,000
Kiwi B-1 Graphite Rods (Type AGSN)	1	4,300	54,900
	2	4,100	49,900
	3	4,100	-
Type AGSX Graphite Tube (3/4-in. OD x 15/32-in. ID)	1	5,500	190,100
	2	5,400	170,700
Type AGSX Graphite from Test PO-41 (turned to 0.60-in. dia)	1	4,300	54,900
	2	4,100	49,400

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(a)

Reference Time: 0

(b)

Reference Time: 1.33 msec

(c)

Reference Time: 2.99 msec

(d)

Reference Time: 3.65 msec

(e)

Reference Time: 4.65 msec

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Figure 46
High Speed Photographs of Multiple-Element Breakup
Under Impact Loading

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APPENDIX A

MANUFACTURE OF GRAPHITE

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APPENDIX A

MANUFACTURE OF GRAPHITE

I. GRAPHITE CHARACTERISTICS

A. ARTIFICIAL GRAPHITE PRODUCTION

The term "graphite," strictly applied, describes a particular form of crystalline carbon structure. Natural graphite is found in deposits; however, artificial graphite, as required for the NERVA engine reactor, is developed as a by-product in such processes as the solidification of carbon-iron, or by the high temperature treatment of carbonaceous materials. Commercially produced, artificial graphites are not completely graphitic throughout. The reasons for the imperfections are several: actual voids may exist; the carbon may not be completely converted to graphite; impurities may exist in the carbonaceous materials used in the manufacturing process; and the graphite crystals themselves may contain substructures and microscopic faults. Many of the defects, or departures, from the perfect graphitic structure result from the rather unique methods employed in the manufacture of artificial graphites. Briefly, in the standard process for the production of graphites, other than pyrolytic graphites, the particles of starting material, usually petroleum coke, are bonded together by pitch or some other binder, this green mix is then carbonized by heating, and, subsequently, is subjected to prolonged heating about 2500°C or higher to cause graphitization of the carbon.

A large number of variables exist in such a sequence, including the nature and composition of the starting powders, the composition of the binder, and the pressures and temperatures involved. Table A-I is reproduced from an Armour Research Foundation report (Reference A-1) listing variables present in formulating, mixing, fabricating, and heat treating the carbon base material from which a graphite shape emerges.

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TABLE A-I

INDEPENDENT VARIABLES IN SYNTHETIC GRAPHITE MANUFACTURE*

1	Aggregate	Size, shape, and distribution
2		Purity (includes ash, interplanar compounds, etc.)
3		Shrinkage (to carbonization temperature)
4		Wetted area
5		Graphitizability
6	Binder	Coke residue value
7		Composition
8		Viscosity - temperature characteristics
9		Graphitizability
10	Mixing	Proportion of binder to aggregate
11		Condition of mixing
12	Fabricating	Temperature
13		Pressure
14		Additives (lubricants, solvents, surfactants)
15	Carbonizing	Temperature
16		Pressure
17		Time
18		Atmosphere
19	Graphitizing	Temperature
20		Pressure
21		Time
22		Atmosphere
23		Specimen Dimensions

*Reference A-1

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Manufacturer-controlled variations, introduced at various stages in the production sequence, can result in materials which exhibit relatively large differences in their physical, mechanical, and chemical properties. Artificial graphite is therefore not a specific material, but a family of materials, each essentially pure carbon but varying in such things as grain size; orientation of crystallites; apparent density; the number, size, and shape of the pore spaces; and the degree of graphitization.

Graphite is unique among materials in that it resembles both metals and ceramics in some of its properties (Reference A-2). It has high thermal conductivity, is an electronic conductor, and is black when finely divided and in these respects it is like a metal. However, it lacks ductility at moderate temperatures and is porous, and in these respects it behaves like some ceramics. Moreover, its manufacture resembles that of ceramics more than that of metals since it cannot be melted and then cast, rolled, or forged.

To understand the unique properties and characteristics of graphite, it is necessary that some knowledge be acquired concerning both its microstructure and its macrostructure. The microstructure mainly depends on the types of materials used as the filler and the binder and the interactions between them. The macrostructure results largely from the selection and sequence of the processes used in the manufacture of graphite (Reference A-3).

1. Crystalline Structure of Graphite

Crystalline graphite was one of the first structures to be examined by X-rays, and has been the subject of many investigations by various workers over the past 35 years. The basis of the present knowledge of the atomic structure of graphite was given by Bernal (Reference A-4) in 1924, who showed that the carbon atoms lay in almost perfect planar sheets in a hexagonal arrangement similar to the benzene ring configuration. The sheets are stacked together in such a manner that the vacant centers in any sheet are always overlaid by atoms in the next sheet. Stacking the layers so that such a relationship exists can be accomplished in two ways. In one case, alternate sheets are identically disposed; i.e., by fixing one layer and an axis normal

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to it as a reference, designated the c-axis, and then placing the third layer similarly with respect to the c-axis as is the first layer. Such a stacking sequence leads to the normal graphite crystal structure - a hexagonal lattice in which the c-axis is equal to twice the interlayer spacing. In the second case, the same flat hexagonal networks are found, but the third layer bears the same relationship to the second as the second does to the first. Such a layer arrangement results in a rhombohedral crystal lattice. Usually, five to ten percent of the layers in a graphite crystal are stacked in this second manner (Reference A-5).

The atoms in the hexagonal network within the layer planes are held together by covalent bonds with the C-C bond length being $1.415 \text{ \AA}$ (Reference A-6). Because the bonding electrons are all used in forming C-C bonds within the carbon layers, the layers themselves are held together only by weak Van der Waals forces. The bond strengths within the layer planes are approximately 150 kcal/gm atom, whereas the interlayer binding energy is estimated to be from 1.3 kcal/gm atom (Reference A-7) to 4 kcal/gm atom (Reference A-8). The distance between the layers along the c-axis is $3.35 \text{ \AA}$. Thus, the c-axis unit cell dimension is $6.7 \text{ \AA}$ in the hexagonal crystal structure since, in that structure, the c-axis is twice the distance of the interlayer spacing.

The described sequences of layer stacking, either ABAB (hexagonal) or ABCABC (rhombohedral), are found only in a perfect single crystal of graphite. Usually, however, even microscopic crystals are not perfect. The Van der Waals forces which bind the neighboring layers are so weak that successive layers of atoms may be parallel, but slightly rotated, in the horizontal plane with respect to one another so that an imperfect stack of layers is built up. Such a structure is termed turbostratic, the term being derived from the two words: stratic, meaning the layers are stratified or stacked, and turbo, meaning that the alternate layers are slightly rotated or mismatched. In graphite used for graphite-moderated nuclear reactors, the average error in layer stacking exists at intervals of about six layers (Reference A-5).

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The weak Van der Waals attraction between the planes of carbon atoms in a graphite crystal is responsible for the high compressibility and the relatively high thermal expansion in the c-direction (Reference A-9). Single-crystal graphite exhibits anisotropy in most of its physical and mechanical properties. The large anisotropy of single crystals is illustrated by the ratios of the values of thermal expansion and thermal conductivity measured parallel and perpendicular to the a-axis. Calculated values of linear thermal expansion coefficients from data obtained by Nelson and Riley (Reference A-10) gives a ratio of about 19 for c-axis expansion to a-axis expansion at room temperature. The thermal conductivity at room temperature measured parallel to the a-direction (i.e., in the plane of the layers of carbon atoms) is about 5 times the value measured in the c-direction (Reference A-11). This anisotropy is reflected in polycrystalline, artificial graphite, but usually not to the same degree as in single crystals.

2. The Manufacture of Polycrystalline Graphite

Graphite can be manufactured from almost any organic material that leaves a high carbon residue when heated. The major source of organic material used in commercial graphites is a refinery by-product, petroleum coke. The variations present in coke materials necessitate selecting a particular type in order that the desired properties in the finished graphite are realized. The purity of the coke material is, of course, of prime importance.

For reactor graphite, cokes are selected that have reasonably low values for the impurities, boron, vanadium, and titanium, although these are less important with modern purification techniques. Analyses of a number of cokes actually used in the production of graphite yielded the following typical values of these three elements in parts-per-million of coke: boron, 0.4; vanadium, 70; and titanium, 10 (Reference A-2).

Cokes have been classified as "soft" or "hard," depending on the extent of graphitization attained when the cokes are heated to the temperatures used in the graphitization process. The soft cokes are those that are relatively easy to graphitize, while some of the hard cokes show hardly any graphitization regardless of the temperature to which they are subjected. The domains of carbon

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atoms that will become graphite crystallites after subsequent heat treatment are fixed in the coking operation by the polymerization of hydrocarbons into large polycyclic aromatic molecules.

The raw coke as received from the refinery is still a hydrocarbon, and is only about 95% carbon. The first step in the preparation of raw coke is a heat treatment to a temperature of 1000° to 1300°C. This heat treatment, or calcination, accomplishes two purposes. First, there is a considerable amount of volatile material driven out of the coke including such gases as hydrogen, carbon dioxide, carbon monoxide, methane, and water vapor. The volatiles are removed from the edges of the larger aromatic arrays in the coke during calcination and, in the process, carbon bonds are broken. These bonds either reform and cause further polymerization and cross-linking (Reference A-12), or the peripheral carbon valences are restored and are available for rapid growth of aromatic ring structures into graphite crystallites at later stages in the manufacturing process (Reference A-1). Although it is not one of the purposes of calcination, one of the results of the evolution of the volatile materials is to substantially contribute to the porosity of the coke. Second, calcination is accomplished to effect an appreciable volume shrinkage of the raw coke material. Approximately 80% of the total shrinkage the raw coke will undergo in the manufacture of graphite occurs during calcination (Reference A-1). If raw coke is mixed with a pitch binder prior to calcination, the shrinkage of the filler coke during subsequent baking may be sufficiently severe to break the pitch-coke bonds and result in an extremely weak or a cracked structure. The rapid evolution of excessive volatile constituents of an uncalcinated pitch-coke binder mixture during the early stages of baking can also contribute to a weak carbon body. Thus, it is important that the raw coke undergo calcination in order that volume changes can be controlled in subsequent processes, allowing the manufacture of large blocks of graphite.

After calcination of the coke, it is crushed and screened, yielding various sizes of particles ranging in diameter from a fraction of a millimeter to a centimeter or larger. The material finer than about 0.5 mm is further crushed into a powder which consists of grains of coke with a

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continuous distribution of diameters from 2 to 500 μ . The individual grains of coke produced in this grinding process are irregular in shape. Cokes from various sources will fracture in different ways and the resultant shapes can range from sphericity to splintery or needlelike. The differences in particle shapes arise from the variations in the degree of the alignment of the rudimentary crystals in the raw coke. The orientation of the aromatic molecules in a soft coke is such that the planes of the benzene rings are parallel to the cellular walls. Using X-ray, small-angle-scattering techniques, Akamutu and Kuroda (Reference A-13) concluded that thin plate-like crystallites are present in raw petroleum coke. When the coke is crushed, it fractures in a manner such that the longer particle axis is parallel to these plates and the resulting particle shape is splintery. The elongated particle shape causes the anisotropic properties of the crystal to be reflected in the properties of the graphite body.

A blend is made from the particles of coke, and the proportion of fraction of various grain sizes is varied, within limits, to control the properties of the end product. The proportion of the very fine, to larger sized particles has an effect not only on the strength of the graphite, but also on its degree of anisotropy. With regard to the effect of grain size on physical properties, two generalizations may be made. The strength of the finished product increases as the grain size is reduced, and the anisotropy is increased.

The next step in the manufacture of graphite is mixing the blend of coke particles with a binder. A good binder for the dry aggregate of coke particles must meet three requirements. First, it must be a thermoplastic material, i.e., one which is solid at room temperature and fluid at higher temperatures. Second, it must have a high carbon content and a major portion of the carbon must be deposited as a bonding coke during the baking process. Third, it should have as high a specific gravity as possible, which will permit the deposition of the greatest number of carbon atoms in and around the filler coke particles. Of all the materials which are commonly available and economically feasible for use in large production quantities, coal tar pitch best meets the requirements. This pitch is obtained as a residue of coal tar distillation.

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The fact that coal tar pitch is thermoplastic material permits thorough mixing of the filler and binder materials during the forming process and allows storage and handling of the formed articles at room temperature. The viscosity of pitch changes rapidly with temperature. Values of the viscosity of a typical binder at various temperatures are as follows (Reference A-2): 13,000 cp at 130°C, 2200 cp at 150°C, and 540 cp at 170°C. Mixing of the pitch with the coke blend is accomplished at a temperature of approximately 165°C, at which temperature the pitch is quite fluid.

The "coking value" of the pitch is very important because it is the quantity of infusible carbon remaining as a bond between the filler coke particles. It is measured by slowly heating the weighed pitch sample in a nonoxidizing atmosphere to a temperature of 1000°C. The residue, which is expressed as a percentage of the original sample weight, is defined as the coking value. This property of a pitch varies with the rate of heating and the environment during the test. Riesz and Susman (Reference A-14) investigated 50 materials for use as potential binder materials. Only five of these had a coke residue value over 50%. Of these five materials, three were coal tar pitches and two were phenol-formaldehydes. A liquid furfuryl alcohol polymer had a coke residue value of 49.1%. The highest value obtained in this group was 52.5% for a coal tar pitch. Other coke residue values for coal tar pitches range from 55% (Reference A-15) to 65% (Reference A-2).

The normal proportions of filler and binder used in the production of commercial graphites is in the order of 75% of calcinated petroleum coke and 25% coal tar pitch binder (Reference A-14). Hutcheon and Price (Reference A-16) conducted an investigation into the variation of the physical properties of graphite bodies resulting from a change in the percent of binder content. On the basis of apparent density, bend strength, air permeability, electrical resistivity and thermal expansion, it was determined that the optimum binder content was approximately 23% for all properties. Wiggs (Reference A-17), using a blend of petroleum coke particles with a size distribution given in a paper at the 1955 Geneva Conference (Reference A-2) for a nuclear graphite, varied the binder content over the range from 20 gm of pitch binder to 100 gm of coke up to 50 gm of binder to 100 gm of coke.

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This variation in binder content was made in 5 gm increments. Measurements of the baked mixings of coke and binder indicated that the greatest transverse strength was achieved by using 35 gm of pitch binder with 100 gm of coke. The pitch proportion by weight in this mix was then approximately 26%.

The mixture of the binder and filler is then formed into the desired shape. The forming is accomplished with the mix at a temperature sufficiently high, usually about 165°C, so that the binder is fluid. The forming method is by either an extrusion or molding technique. A lubricating oil is often added to the mix when extrusion is used as the forming method. The oil which is insoluble in the coal tar pitch, has no coking value and, therefore, does not remain as carbon in the final product. Its purpose is merely to reduce the friction between the surface of the extension die and the mix.

If an extrusion process is employed, the coke particles tend to line up with their long axis parallel to the direction of the extrusion. In molding, the long axis is perpendicular to the direction of applied pressure. The final graphite shape retains the same pattern of grain orientation. The terms "with-the-grain" and "against-the-grain" are used in connection with graphite bodies. The with-the-grain direction is parallel to the direction of extrusion in extruded graphite and perpendicular to the molding pressure in molded graphite. The against-the-grain direction is perpendicular to the with-the-grain direction in each case. The anisotropy of the graphite crystals within the grains is carried over into the final graphite body because of the preferred orientation of the grains. Thus, there is a difference in the values of properties measured with-the-grain and against-the-grain. The thermal and electrical conductivities, the strength, and Young's modulus are higher with-the-grain, and the thermal expansion is higher against-the-grain.

After the plastic mixture of filler and binder is formed, the green forms are cooled to harden the pitch binder and render them less plastic for handling purposes. When cool, the green forms may be stored without any distortion occurring in their shape. The formed green bodies at

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this point consists of a calcinated petroleum coke filler, bonded with a coal tar pitch which softens at a temperature of approximately 100°C . The next process in the manufacture of graphite is baking of the green bodies. The purpose of the baking processes is to convert the pitch from a thermoplastic material into an infusible solid. In this process, the green bodies must be supported to prevent slumping or distortion as they are heated through the temperature range in which the pitch is liquid. The final baking temperature is between 800° and 1000°C . The pitch binder evolves large quantities of hydrocarbon gases in the temperature range of 300° to 500°C and becomes rigid at approximately 450°C . In the course of baking, approximately one-third of the pitch binder is driven off as volatiles. Because the large amount of volatiles given off must escape through a relatively impermeable mass, the heating rate must be slow so that the evolution of the gases is not so rapid that the structure of the body is disrupted.

Above approximately 450°C , a fairly large shrinkage occurs as the binder hardens, and gives rise to internal stresses which are sometimes great enough to crack the carbon body. As the volatile hydrocarbons are being evolved and subsequent shrinkage occurs, polymerization and cross-linking proceeds within the binder and between the binder and filler materials. If there is an excess of binder material, there will be no locking process between the coke particles and the body will distort and blister as the volatiles evolve during the baking. A deficiency of binder will result in a weakened bond between particles. This effect relates back to the previous discussion as to the optimum binder content of the mix which is to be formed.

The baked carbon body is very brittle and hard. The body is also susceptible to thermal shock because of its low thermal conductivity, high coefficient of thermal expansion, and high elastic modulus.

After the baking cycle is completed, the carbon bars are either graphitized or undergo an impregnation operation prior to graphitization. The porosity of baked carbon is approximately 25%. The purpose of the impregnation operation is to reduce the porosity and thereby increase the density of the resulting graphite. In some cases the impregnation is deferred until

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graphitization has been accomplished. Normally, the impregnant used is a coal tar pitch. However, this pitch differs in composition from the pitch used as a binder in that some of the heavier fractions usually found in a binder pitch are missing. Other materials (References A-18 through A-21) have been investigated for use in impregnation operations among which are an aldehyde/aniline mixture, polyacrylonitrile, and sugar. Whatever organic material is used as an impregnant, it must be one which will deposit secondary carbon in the voids or pores during pyrolysis in a rebaking or graphitization operation.

Pitch impregnation is accomplished in a large autoclave containing the baked carbon rod. The carbon rod is preheated to approximately 250°C prior to insertion into the autoclave such that its temperature is well above the softening point of pitch. This assures that penetration of the pitch is not retarded because of its being chilled by the carbon. With the heated carbon body in the autoclave, the pressure is reduced to about 1/10 atmosphere to remove some of the air from the pores in the carbon. Hot pitch is then pumped into the autoclave completely immersing the carbon. A pressure of approximately 100 psi is applied and maintained for a sufficient length of time to ensure impregnation to the center of the rod. The pressure is then released, the unused pitch is pumped out, and the excess pitch on the rod is allowed to drain off.

Graphitization is the next and final operation for those graphites which are not required to be extremely pure. This process is performed in a large Acheson electric furnace, which is capable of providing temperatures up to 3000°C while maintaining a nonoxidizing atmosphere. The impregnated carbon bars are laid cross-wise on top of one another in the furnace to form stacks of carbon. Granular coke is placed around the rods to fill the intervening spaces between the stacks. The coke forms the resistor element of the furnace and most of the resistance heating is generated in it, especially during the terminal stages of graphitization. Power is supplied at rates of 400 to 600 kw (Reference A-2) depending on the size of the furnace, the type of graphite to be produced, and the heating cycle position. The duration of the heating cycle and the temperatures attained are also dependent upon the furnace size and the type as well as the amount of graphite to be produced. In general, a peak temperature between 2600° and 3000°C is attained. A period of three to

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four days are required to heat the stock, and ten days to two weeks are required to cool the charge to a temperature at which the graphite can be exposed to the air.

Chemical analysis of graphite shows that extensive purification occurs during the graphitization. However, to obtain graphite of the highest purity, the graphite bars are bathed in a purifying gas containing chlorine and fluorine while they are electrically heated to approximately 2500°C. This process removes many impurities having high neutron absorption cross-sections such as the rare earths, vanadium, and boron. In addition, it generally reduces the ash content to less than 20 ppm.

The graphitization process not only eliminates most of the impurities in the carbon body, but it also promotes growth and ordering of the graphite-like crystallites. Large-scale ordering of the crystallites into three-dimensional graphite crystals is not accomplished until the higher temperatures of the graphitization process are attained. Prior to the formation of the graphite crystals, several microstructural changes have occurred to the aromatic molecules in the soft petroleum coke.

During the coking process, flow patterns are formed within the coke particles and relatively large aromatic plate-like molecules are aligned parallel to the flow. Upon solidification, the raw coke has a textured and partly aligned structure. The structure does not appreciably change during calcination of the coke, although some porosity develops as a result of the evolution of volatile hydrocarbons. Mrozowski (Reference 9) describes the structure of a calcinated coke as follows: "At temperatures around and somewhat above 800°C, these substances having gradually lost most of the atoms other than carbon by evolution of gases, become calcinated or baked carbons. At this stage, these carbons are composed of small turbostratic crystallites of approximately 30 Å in diameter, built of disordered stacks of parallel equidistant planes of condensed benzene rings, and of the disorganized carbon phase forming a matrix which surrounds these crystallites." With increase of heat treatment temperature, the crystallite growth continues, but proceeds rather slowly at first. According to Mizushima (Reference A-22) the layer

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diameter increases from 20 Å to about 40 Å in the temperature range from 1000° to 1500°C. However, at a temperature of about 1800°C, the crystallite growth suddenly accelerates (Reference A-23). The most probable mechanism explaining the rapid crystallite growth is the displacement and rearrangement of crystallites relatively to each other. If two neighboring crystallites are aligned so that their basal planes are nearly parallel to each other, they can be easily rearranged so as to have their c-axis completely parallel. Such a rearrangement does not involve mass transfer or the breaking of C-C bonds. Thus, two nearly parallel crystallites can coalesce to form a much larger crystallite. However, even though the layer planes become two-dimensionally ordered, i.e., the plane of carbon atoms in a hexagonal arrangement is quite large and the stack of parallel planes is fairly deep, there is little ordering of the planes with respect to one another and no strictly graphitic structures have been formed at this stage.

At the higher, graphitization temperatures minor crystal growth occurs. However, the major effect of the heat treatment is one of three-dimensional ordering of the stacked parallel planes into an ABAB stacking sequence to form a graphite crystalline structure.

From the foregoing discussion of the crystal structure of graphite and the manufacturing processes involved in its production, a picture of a commercial polycrystalline graphite can be formulated. It is composed of approximately 80 to 85% filler coke and 15 to 20% binder pitch. The particles, or grains, of graphitized coke contain relatively large areas of crystallites which may be highly aligned. The crystallites themselves are made up of carbon atoms arranged in the characteristic and periodic manner of a crystal structure. The normal crystalline lattice structure in graphite is hexagonal, although some rhombohedral crystalline structures are present. The crystallite may contain structural defects such as atomic vacancies, interstitial atoms, dislocations, impurities, and turning. There is also some microporosity at the crystallite boundaries resulting from imperfect alignment of the crystallites with respect to one another. Lock and Austin (Reference A-24) have calculated that the average spacing between adjoining crystallites is from 3 to 8 Å.

Additional porosity is present in the coke particles resulting from the evolution of gases during the calcination process. This is a more coarse porosity than that between adjacent crystallites within the coke particle.

The filler material is a blend of coke particles and coke powder. The particles in a typical nuclear-grade graphite range in size from 0.015 inch to 0.03 inch. The filler coke material is surrounded by and, to some extent, infused with a binder. During the mixing and forming process, the binder material, usually a coal tar pitch, wets the particles and is subsequently carbonized during the baking and graphitizing processes. The carbonized binder material forms bridges between the filler particles, holding them together. There is considerable macroporosity in the resultant body because of the evolution of gases from the binder material and the shrinkage during the baking process. This porosity can be substantially reduced by successive impregnation and graphitization cycles. However, the remaining porosity may still be as large as 20 to 25% of the total volume of the graphite.

The extent of the porosity in a graphite body is evident from a measurement of its density. The theoretical density of single crystal graphite is 2.26 gm/cc. The density of a normal nuclear-grade graphite with its associated micro- and macroporosity is only 1.7 gm/cc. Therefore, polycrystalline artificial graphite has a complex structure and can hardly be considered to be a homogeneous material either on a micro- or macro-structural level.

3. Other Types of Artificial Graphite

All of the previous discussion has dealt with what has been termed polycrystalline artificial graphites. Most of the commercially produced graphite is of the general type which has been described. However, there are other types of graphite produced commercially which also are polycrystalline artificial graphites, but are considerably different both in their method of manufacture and in their structure. One of these types, pyrolytic graphite, will be briefly discussed since it is sometimes present in nuclear fuel elements as a coating for fuel-material particles.

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Pyrolytic graphite is formed by the decomposition of a carbonaceous gas on a heated surface. The reaction has been accomplished over a range of temperatures from 1400° to over 2000°C. The resultant material deposited on the heated surface varies not only with the temperature of the deposition surface, but is also affected by the pressure, gas flow rate, geometry of the deposition surface, and the starting raw material (Reference A-25). The process of deposition may be made continuous by maintaining a constant flow rate and gas composition. However, careful control of all the process variables is necessary. As the deposit grows, it acts as a layer of low thermal conductivity between the exposed surface and the original deposition surface and this, by upsetting the thermal distribution, adds to the difficulty in maintaining constant deposition conditions.

The carbonaceous gases usually used in the production of pyrolytic graphite may be either methane, benzene, propane, or a purified natural gas. The deposition temperature of commercially produced pyrolytic graphite is usually 2000°C or higher. The deposits of carbon atoms at these elevated temperatures are highly oriented crystallographically, but some turbostratic stacking of layer planes occurs. The turbostratic nature of the deposit is greatly reduced on subsequent heat treatment at temperatures above the deposition temperatures.

The density of the deposited material is a function of both the deposition temperature and initial gas pressure. Brown and Watt (Reference A-26), using benzene, propane, and methane, varied the gas pressure over a temperature range from 1600° to 2100°C to observe the effects on the rate of deposition and density of the deposits. It was determined that an increase in pressure at a given deposition temperature tends to increase the density of the deposit. The variation of the density of deposited carbon with temperature of deposition appeared to be largely independent of the hydrocarbon gas that is used. The density was minimum at a deposition temperature of about 1700°C and increased as the temperature increased. At a deposition temperature of 2100°C, deposits with densities of 2.22 gm/cc were obtained using methane and benzene and a density of 2.23 was obtained using propane. These densities

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are greater than 98% of the theoretical density of 2.26 gm/cc for ideally crystalline graphite.

X-ray examination of pyrolytic graphite shows that the deposited carbon is oriented so that the hexagonal layers of atoms lie parallel to each other and to the surface of the original substrate. Three-dimensional ordering does not occur to any great extent for deposition temperatures less than 2000°C. Microscopic examination of a cross-section normal to the surface of a deposit reveals the presence of a fibrillar structure. The fiber axis coincides with the direction of the growth of the deposit. The individual fibers are approximately normal to the surface of the plate-like deposit. The fibers do not lie parallel to one another, however, but are slightly divergent and arranged in conical bundles. The larger cones are initiated at the bottom surface of the deposit and may have an apex angle of 40° to 50° at their origin (Reference A-27). The divergence of the fibers diminishes suddenly in the early stages of growth and the fibers are almost parallel at the end of the cones. Within a conical bundle, a number of smaller cones are initiated at random points and these are apparently the cones whose ends are visible on the upper surface.

When deposited pyrolytic graphite is subjected to a heat treatment at a temperature above the deposition temperature, a transformation takes place which results in a decrease in the turbostratic structure, an increase in the preferred orientation, permanent growth parallel to the basal planes and shrinkage perpendicular to these planes, and an elimination of the growth cone structure (Reference A-28). The dimensional changes which occur on annealing are a result of the increase in three-dimensional ordering (which results in a decrease in the distance between planes) and the straightening out of wrinkles in the AB-plane caused by the with-conical-grain growth (Reference A-29). Annealing effects begin to be evident at a heat treatment temperature of about 2600°C and continue up to a temperature of at least 3250°C (Reference A-30). Ubbelohde, et al (Reference A-31), annealed pyrolytic graphite with simultaneous hot-pressing at 2800°C and found that the effects on its physical properties were comparable to or exceeded those of conventional annealing at 3500°C.

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Pyrolytic graphite is very anisotropic as a result of its high degree of preferred orientation. This material has been produced in which the electrical resistivity in the c-direction is 5000 times that in the a-direction and thermal conductivity in the a-direction 250 times that in the c-direction (Reference A-31).

Although not strictly a different type of graphite, some mention will be made of the methods of producing fueled graphite elements. The green carbon mix is not appreciably different than that used in the production of commercial polycrystalline graphite. However, the means by which the addition of the fuel material is accomplished can vary. Three techniques for the preparation of fueled graphite have been reported. The first method of preparation is by impregnation, the second method is by encapsulation, and the third method is by admixture. Impregnation of graphite is possible because of its relatively high porosity and is accomplished by impregnating with a uranium or a thorium bearing solution. One of the problems associated with the impregnation of graphite is the nonuniformity in the uranium concentration (Reference A-32). A high concentration of the uranium or thorium presents the problem of a possible hot spot forming during reactor operation. Also, the fission product release is reported to be many orders of magnitude greater than admixed elements. An encapsulated fuel element consists of an unfueled shell enclosing or encapsulating a fueled core. Such elements have suffered considerable impairment of physical properties after irradiation (Reference A-33). By using the admixture technique it is possible to maintain good control of the uranium distribution. Fuel elements of this type have shown very small dimensional changes after irradiation.

At the present time, three materials are being utilized primarily in the admixture technique of production of fueled graphite. These are uranium oxide, uranium carbide, and uranium carbide spheres which have been coated with pyrolytic graphite.

Fueled graphite using any one of these materials can be formed by either an extrusion or a molding process. It is reported (Reference A-33) that the coefficient of variation of uranium dispersion can be held

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to 3.7% in molded fuel elements 1/2 inch diameter which contains a dispersion of 180 to 250 mm diameter UC_2 coated with pyrolytic graphite. Unclad uranium carbide particles are formed in a graphite matrix by dispersing UO_2 particles, which are subsequently converted to the carbide.

Fueled graphite elements with higher densities and greater strengths can be produced by adding the pyrolytic coated UC_2 particles to the green carbon mix instead of the UO_2 . This is because that in converting to the carbide from the oxide, carbon atoms are robbed from the binder material. Also, this mixing in the clad UC_2 particles allows for impregnation of the element to achieve higher density. Impregnation can not be carried out when unclad UC_2 is present because of the reaction with the evolving material of the impregnating pitch during baking.

4. Strength of Graphite

In addition to its good nuclear properties, graphite has other properties which make it well suited for use as a material in high temperature nuclear reactors. The relatively high thermal conductivity of graphite, coupled with its low thermal expansion makes it insensitive to thermal shock. The strength of graphite increases with increase in temperature up to about $2500^{\circ}C$ (Reference A-34). At this temperature graphite's strength is about twice its room temperature strength, but above this temperature the strength begins to fall off rapidly.

There is an extensive amount of literature available on the physical and mechanical properties of graphite. However, this discussion will be restricted primarily to its strength properties. Other properties will be discussed only to the extent that they relate to, or affect the strength characteristics of, graphite.

In making a determination of the strength of a material such as graphite, many problems are encountered. The methods employed in the manufacture of graphite results in the production of a two-phase system (binder and filler) which is porous. Even when the utmost care is exercised in the selection of raw materials and close control is maintained over the

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manufacturing processes, variations still exist in the finished production. This variation can exist between two production batches, between two pieces from the same batch, or even in different portions of a single piece (Reference A-34). The variation resulting from location in a single graphite body is more pronounced in the large graphite pieces. Therefore, in a series of tests on graphite specimens under supposedly identical testing conditions and environment, it is difficult to obtain a high degree of reproducibility in results because of the nature of the test material itself. This difficulty is increased since graphite is considered to be a brittle material up to temperatures approaching 1800°C. The fact that brittle materials are highly sensitive to stress concentration makes the design of sample holders and jigs difficult. This same sensitivity to stress concentrations must be considered in determining an acceptable geometrical configuration for test specimens. Also, because of the extreme range of temperatures over which testing must be conducted, modifications and changes must be made in the materials used in testing apparatus and in testing techniques. All of the foregoing is merely a prelude to the statement that test results of the strength of graphite from different investigators can show considerable disparity. The degree of scatter in test data can also vary appreciably when the same type and grade of commercial graphite is tested at different laboratories, depending on the type of test and the care exercised in conducting the test.

An indication of the degree of scatter which results from the testing of graphite can be seen in the variation of strength values obtained in conjunction with a study conducted in this program of the fracture mechanism of graphite. Samples of 3/4-inch-diameter rods were tested in compression at room temperature to failure. The test specimens were 1-1/4 inch long and were cut from rods of AGSX graphite, 36 inch long. The ends of the test specimens were machined flat to within 0.002 inch and perpendicular to the axis of the rod length. The first group of specimens were selected at random from a larger group which had been machined from an unknown number of different rods. In an effort to reduce variation, the specimens comprising a second group were machined from only three rods. Although it would have been better to use specimens from only one rod, it was necessary to use three rods in order that a

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sufficiently large number of specimens could be obtained. Both groups were tested on the same machine and at a cross-head speed of 0.05 inch/min. The first group had an average compressive strength of 7200 psi with a standard deviation of 747 psi. The difference between the highest and lowest strength values obtained in this group was 2480 psi. The second group tested had an average strength of 7780 psi with a standard deviation of 317 psi. The extremes in the high and low strength values of the second group was 1210 psi. The 95% confidence range for the average strength was ± 152 psi. The fact that much less scatter occurred in the tests conducted on the second group may have merely been fortuitous since selection of the rods from which these specimens were made was based on limited information. At any rate, a great deal of scatter can and does occur. McIntosh, et al (Reference A-35), has reported the results of room temperature tensile strength determinations on some British reactor-grade graphites. They found the average strength to be 1620 psi tensile with the 95% confidence range of the average to be ± 260 psi.

Because of the large scatter obtained in individual test, a relatively large number of specimens must be tested when a determination of the strength of graphite is to be obtained. From these tests, an average value is obtained. Determination of these average values reveal that even though variations exist in the individual test values, certain general factors affect the strength properties of graphites. Among these are grain size, method of forming, density, and the starting raw materials. Graphite bodies also exhibit differences in strength when measured with-the-grain as opposed to against-the-grain. The effect of grain size is illustrated by data reported by Currie, et al (Reference A-2), for the elastic modulus and flexural strengths of graphites of varying grain size.

The coarse grained stock from which the values in Table A-II were obtained is not entirely comparable with the other stocks in that it was not given a pitch impregnation and, consequently, has a lower density. However, the point illustrated is that the strength increases as the grain size is reduced, and the ratio of the properties measured parallel and perpendicular to the grain is increased. The ratio of the mechanical properties

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TABLE A-II

EFFECT OF GRAIN SIZE

	Fine Grain Stock	Medium Grain Stock	Coarse Grain Stock
Elastic Modulus (10^6 lb/in. ²)			
Parallel to grain	1.84	1.47	0.89
Perpendicular to grain	0.78	0.76	0.61
Flexural Strength (lb/in. ²)			
Parallel to grain	4520	2220	1280
Perpendicular to grain	3000	1630	990

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when measured with-the-grain and against-the-grain for the same grain size graphite is greater for extruded stock than it is in molded stock. This again indicates the greater anisotropy for the extruded material.

A most unusual property of graphite is its increase of strength at high temperatures. Malstrom, Keen and Green (Reference A-34) reported extensive tensile tests indicating that the strength of graphite increases up to a temperature of 2500°C. At this temperature, the strength for short-time loads is from 50% to more than 100% higher than the room temperature values. The same investigators also reported that the elastic modulus increases with temperature and that a 2000°C is about 40% greater than at room temperature. Similar results of the change in strength with temperature have been reported by several other investigators (References A-36 through A-43). Wagner, et al (Reference A-36), measured the compressive strength of H4LM and uranium-loaded graphite over the temperature range 20° to 3000°C and concluded the strength increased from room temperature to a maximum at some temperature above 2000°C. Above the 2000°C maximum, strength dropped off sharply. The average compressive strength value of H4LM measured parallel to the grain increased from a room temperature value of about 4800 psi to a maximum of 8500 psi. This maximum strength occurred at a temperature of approximately 2600°C. Tensile tests were also performed on these materials and it was found that a maximum strength was attained at a temperature between 2000°C and 2500°C, and was approximately twice the room temperature value. Martens, et al (Reference A-39), conducted tensile tests on commercial graphites at room temperature, 4000°F, 4500°F, and 5000°F, and at strain rates ranging from 5.2×10^{-5} to 1×10^{-2} sec⁻¹. They noted a two-fold increase in the tensile strength at a temperature of 4500°F. At 5000°F the ductility of one lot of specimens decreased from 40% to 1% elongation as the strain rate was increased from 5×10^{-5} to 1×10^{-2} sec⁻¹. At this temperature, the strength appeared to increase slightly with increasing strain rate although the scatter in strength values was quite large.

A strain rate effect has also been observed on the strength at room temperature (Reference A-40). A series of tests measuring the transverse bend strength at increasing jaw speeds from 0.0002 inch/min to 20 inch/min resulted in an increase in bend strength from about 8200 psi to 9300 psi.

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Green, et al (Reference A-36), made strength determinations at 2500°C on unimpregnated and impregnated H4LM as well as uranium-loaded graphite, furnished by LASL. The results were substantially the same as those obtained by Wagner, et al (Reference A-41), who tested similar materials. Strain measurements were recorded during the tensile tests and stress-strain curves were obtained. The stress-strain curves of the H4LM were in several instances S-shaped. However, in other cases the curvature was always in the same direction, being concave downward. This behavior of the material was investigated to see if hysteresis effects would occur in repeated-loading tests. Specimens of the unimpregnated H4LM were loaded at 2400°C to a point short of failure, unloaded to a low stress level, and then reloaded to failure. Large hysteresis effects were observed during the initial loading and unloading, with the specimens exhibiting essentially elastic but nonlinear behavior to fracture on the second loading.

The molded graphite, furnished by LASL, contained uranium in the form of UO_2 in the following concentrations: 0, 125, 250, and 350 mg/cc. The breaking strength values of the loaded graphite indicated an increase with temperature, similar to that of the pure grade, over the range from ambient to 2400°C. The graphite containing 125 mg/cc of uranium exhibited the highest strength at the elevated temperatures. This confirmed the result obtained by Wagner, et al (Reference A-41). Because of the scatter in the breaking strength data, no ranking by strength was possible among the other three grades. The stress-strain curves indicated that the pure-grade material (no UO_2) provided the least plasticity at 2000°C, failing at an ultimate elongation of about 0.1%. At 2400°C, the ultimate elongation among the specimens of the grade that were tested ranged from 0.45% to 2.0%.

Because of the S-shaped stress-strain curves obtained on the H4LM specimens, the Young's modulus of this material was determined by dynamic means. The values obtained show the thermal dependence noted by other investigators. The modulus increased from ambient temperature up to approximately 2000° to 2200°C where it reached a maximum of approximately 1-1/2 times the room temperature value.

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The investigations cited, as well as many others which determined the mechanical properties of graphite and the changes in these properties as a function of temperature, point out the uniqueness of graphite as a high temperature material. All of the investigations show an increase in the tensile and compressive strength of graphite as the test temperature is increased to a temperature above 2000°C. Depending upon the type of graphite, the temperature at which maximum strength occurs is between 2000° and 2600°C, with the maximum strength approximately twice the room temperature value. Young's modulus values exhibit the same trend although the maximum generally occurs at a slightly lower temperature than the temperature at which the maximum strength occurs. At temperatures above those at which these maxima occur, both the strength and modulus values decrease. At low strain rates, creep effects become predominant at these higher temperatures. Prior to a discussion concerning the theories which have been proposed to explain the mechanisms of failure in graphite and its high temperature strength properties, a discussion of the effects of irradiation on the mechanical properties of graphite will be presented.

5. Effects of Neutron Irradiation on the Mechanical Properties of Graphite

Exposure of graphite to neutron irradiation results in significant changes in its physical and mechanical properties under certain conditions. The degree to which these properties are affected is very sensitive to the temperature at which the graphite is irradiated. The effects are a result of crystal lattice defects (interstitial atoms and vacant lattice sites) which are produced by displacement of atoms by neutron collisions.

Neutron irradiation of graphite at a temperature of 30°C results in an increase in mechanical strength and hardness while the electrical and thermal conductivities decrease. Other effects which are noted are a change in the gross dimensions of a graphite shape and a storing of potential energy within the graphite. A decrease in magnitude is noted in all of these effects at high irradiation temperature.

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Changes in the physical properties of graphite irradiated at room temperature are such that a stronger, harder, and more brittle material results. Woods, et al (Reference A-44), reported at the 1955 Geneva Conference that the compressive strength of a sample of KC graphite, cut parallel to the extrusion axis, increased by a factor of almost three when exposed to an integrated flux of 5×10^9 nvt at 30°C . The compressive strength decreased at higher exposures and a value approximately 2-1/4 times the unirradiated value was attained at an exposure of 4×10^{20} nvt and remained at this level to the highest exposure data reported of 10×10^{20} nvt. Meyer and Bourdeau (Reference A-45) have reported that the compressive strength of a British made graphite increased from 6204 lb/inch^2 to $10,000 \text{ lb/inch}^2$ after irradiation. The corresponding strains at these strength levels were 0.022 inch/inch and 0.008 inch/inch respectively. The irradiation dose for this sample was 1.8×10^{20} nvt at a temperature of 40° to 50°C . Under similar irradiation conditions AGHT graphite measured $10,000 \text{ lb/inch}^2$ and 0.0035 inch/inch compared to pre-irradiation values of 4400 lb/inch^2 and 0.028 inch/inch.

Neutron irradiation of graphite at room temperature also results in a significant increase in Young's modulus. At the 1955 Geneva Conference, Woods, et al (Reference A-44), reported a threefold increase occurred in the Young's modulus of KC graphite with an exposure of 1×10^{20} nvt (about 175 mwd/AT). At the same conference, Kinchin (Reference A-46), reported an increase of 2-1/2 times the pre-irradiated value in the Young's modulus of a graphite subjected to an exposure of 5×10^{19} nvt.

The changes produced in the physical and mechanical properties as a result of neutron irradiation can, in many cases, be almost completely annealed out of a graphite at sufficiently high annealing temperatures. Kinchin (Reference A-46) has investigated the effect of one hour anneals at varying annealing temperatures on graphite which was irradiated initially at 30°C . The irradiated graphite had a Young's modulus 2-1/2 times that of the pre-irradiated material. After a one-hour anneal at 250°C , the modulus had decreased to a value which was only 1-3/4 that of the pre-irradiated value. The modulus values of irradiated samples subjected to one-hour anneals continued

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to decrease as the annealing temperature was increased. After a one-hour anneal at 1500°C, the recovery was almost complete and the modulus of the irradiated sample was only slightly more than that prior to irradiation. The effects of annealing irradiated CSF graphite in regards to a reduction in its compressive strength have also been reported (Reference A-44). The graphite previously had an exposure of 1.14×10^{20} nvt and the unannealed irradiated material had a compressive strength of about 13,900 lb/inch². An anneal for 16 hours at 215°C reduced the compressive strength to about 8500 lb/inch². A 16-hour anneal at 375°C resulted in a compressive strength of about 6500 lb and a 4-hour anneal at 520°C resulted in a compressive strength of slightly over 6000 lb/inch².

Graphite which has been damaged by reactor irradiation at about 30°C can be annealed by a second irradiation at a higher temperature. The degree to which the damage is annealed by this second irradiation is appreciably more than that which would be accomplished by a thermal anneal at the same temperature. This phenomenon is called irradiation annealing. Nightingale, et al (Reference A-47), reported at the 1957 Geneva Conference that reactor radiation at 335°C is effective in partially removing damage which in some cases is approached only by thermal annealing to 1300°C.

The changes in the mechanical properties vary considerably at different temperatures of irradiation. Investigations by Simmons (Reference A-48) have shown the relationship of the fractional change in Young's modulus of graphite rods as a function of irradiation at various temperatures. For exposures of 1×10^{20} nvt at temperatures of 25, 80, 180, and 350°C, the corresponding fractional increases in Young's modulus were 1.7, 1.5, 0.7, and 0.2.

The changes which occur in the mechanical properties of graphite as a result of neutron irradiation are dependent not only upon the type of graphite, but also upon the total exposure and the thermal history both during and after irradiation.

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6. Theories Concerning the Mechanical Strength and the Mechanisms of Failure in Graphite

As a part of the Fragmentation Research program it was proposed that a study be made concerning the mechanisms of fracture in graphite and such a study was initiated. Samples of AGSX graphite were fractured by various means (compression, impact, etc.) and the fracture surfaces examined in an effort to determine if a particular mode of fracture could be associated with a given type of stress application. All of the failures were of a brittle nature and produced a similar brittle fracture surface. No variations, other than those naturally occurring in a material such as graphite, were observed by microstructural examination of the fractured specimens. It is felt that a thorough study of the fracture mechanisms of graphite would require facilities beyond the scope of this phase of the program. Such a study would require that specimens be prepared in the laboratory so that extremely close control could be maintained in their production and variations in the specimens held to a minimum. Also, any information obtained concerning the fracture mechanisms operating as a result of static loading or loading at low strain rates would then have to be interpreted and related to fracture because of explosive shock and impact. Work has been done on the effect of shock and impact on the mechanical properties of materials but such information is available primarily concerning metals. A literature search for this type of information concerning graphite revealed a very limited number of investigations have been carried out in this field and they were not concerned with the fracture mechanisms involved.

Because of the lack of information concerning the effect of explosive impact on the mechanical properties of graphite, some of the theories developed and proposed as a result of investigations carried out using static or low rates of loading will be presented. It is hoped that additional information on explosive impact phenomena will be forthcoming as a result of continuing experimentation on this program and that some of the static theories can be either modified or extended to aid in understanding the behavior of graphite under such conditions.

Graphite is a brittle material at temperatures below about 1800°C and loading to failure below this temperature produces a brittle fracture

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surface. Various general theories have been advanced to describe the fracture phenomenon in brittle materials. Some of the theories which have been proposed (Reference A-49) are: (1) the weakest-link theory, (2) the parallel-chain theory, and (3) the strength-summation and combination theory. These theories are all based on the premise that the variations in the strength of brittle material specimens are related to the flaw concept. The flaw concept explains the reason for the statistical nature of the brittle-fracture phenomenon, but does not describe the actual manner in which the strength is affected by the flaws. The theories listed above deal with the effect of a flaw or the distribution of a group of flaws on the strength properties. Of these theories, the weakest-link theory is the one most generally accepted as describing the brittle-fracture phenomenon.

In the weakest-link theory, each element of the material is regarded as being similar to a link in a chain. When the weakest link or, in terms of flaws, when that element which contains the most severe flaw is stressed to its capacity, the link will break and the entire body or chain will subsequently fail. Mason (Reference A-50) has postulated that the weakest-link theory is applicable in the failure of graphite and that the weak-links which control the strength are the microcracks which exist in the material. As pointed out by Mrozowski (Reference A-51), formation of these microcracks are a result of the heat treatment during the graphitization process and subsequent cooling. After graphitization, the crystallites in the cooling graphite shrink mainly in the c-direction and lead to the formation of a complicated pattern of internal stresses. In the absence of plastic flow, these stresses are not relieved and may become so great that they produce internal cracks within the microcrystallites. Cracks are also present in the binder material (Reference A-52) because of the large shrinkage which occurs during baking.

As opposed to some other investigators, Mrozowski has proposed (References A-9 through A-51) that the initial failure of graphite occurs in the coked binder material. Fractures can occur either across the bridge between the particles or along the surface of the particle. The particles which are coated with the binder usually have been calcinated to temperatures in excess of 1000°C and therefore have undergone very little additional shrinkage during the baking

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of the binder material. However, the binder shrinks considerably during baking and can decrease its volume only by means of opening cracks. This cracking results in a weakening of the binder as well as developing frozen-in stresses. Mrozowski maintains that when mechanical failure occurs, the break goes through the binder material because this is the path of minimum cross-section and because the material is weakest there. The frozen-in stresses in graphite, as previously stated, are present not only in the binder material, but also in the filler material. Mrozowski attributes the increase in the strength of graphite at elevated temperatures as being caused by the release of these frozen-in stresses (Reference A-9).

Martens, et al (Reference A-53), have proposed a similar explanation for the temperature dependence on the strength of a graphite although they do not propose failure in the binder material as does Mrozowski. Hove (Reference A-54) has suggested that the increase in Young's modulus with increasing temperature is related to the thermal expansion of the crystallites into the void spaces. Such expansion, it is postulated, tends to lock the crystallites together more tightly and an increase in the modulus results. He does state that additional effects must be present, however, such as the existence of locked in stresses. This proposal, that the increase in modulus is related to the expansion of the crystallites, is based on the similarity of the effect which neutron irradiation has on the crystallites of graphite. Both thermal expansion and neutron irradiation produce an increase in the c-direction of a graphite crystal. It is known that irradiation of graphite results in an increase in the Young's modulus. Hence, the proposed explanation of the increase in modulus of unirradiated graphite as being caused by crystallite expansion.

Simmons (References A-48 through A-58) states that the change in the elastic modulus of irradiated graphite is not likely to be caused by expansion of the crystallite into pore spaces because the change occurs at a dose which is too small to affect the thermal expansion. As a result of irradiation experiments on a crystal of Tranvacore graphite, Simmons suggested that the change in Young's modulus is a result of a change in the shear constant corresponding to the layer planes. His interpretation implies that when artificial graphite is subjected to a load, the deformation is determined by the resistance to shear of the crystallites.

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Other investigators have discounted the increase in modulus as a result of irradiation as being caused by expansion of the crystallites into the voids. Davidson and Losty (Reference A-56) state that an integrated flux of 0.5×10^{20} nvt causes a threefold increase in the modulus and about a 0.7% change in the lattice spacing. A similar expansion of 0.7% can be achieved by a temperature rise of approximately 250°C, but is accompanied by a relatively insignificant change in the modulus.

Losty and various co-workers (References A-56 through A-60) have proposed that a shear mechanism within the crystallite is the strength determining factor in graphite. Such a mechanism can account for the increase in modulus because of irradiation. In graphite, the layer planes, if not constrained, slide over one another with relative ease, as evidenced by the well known lubricating properties of graphite. It is believed that neutron bombardment results in the displacement of carbon atoms from positions within the layer planes. These atoms subsequently come to rest between the planes. The introduction of such an interstitial atom would serve to act as a grit and inhibit the motion of the layer plane. Such an effect on the elastic properties as a result of irradiation can also explain how the increase in modulus saturates with higher doses. Prolonged irradiation will cause the single interstitials to combine into clusters which will lead to an increase in the spacing between the layer planes. When this occurs, the introduction of additional single interstitial atoms will have no effect.

An explanation for the temperature dependence of the modulus is also given by these investigators. Measurements were made of Young's modulus and the modulus of rigidity on a variety of specimens of different crystallite structure and crystal perfection. It was found that the modulus values increased with test temperature for all those specimens which were regarded as having a well developed graphite structure, while the specimens made from non-graphitizing carbons lacked a marked temperature coefficient of elasticity. These results were rationalized on the basis of a shear mechanism by assuming the increased test temperature enhances the vibrations of the carbon atoms perpendicular to the layer planes. Such increased vibrations could then inhibit the movement of the planes relative to each other. This process will be pronounced with the more

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perfect crystallites, but will probably be unimportant in those materials where the motion of the layer planes are already restricted because of gross imperfections.

Losty and Orchard (Reference A-57) conducted an investigation in an effort to obtain some correlation between the strength and modulus of graphite. A set of flexural test specimens, consisting of both impregnated and unimpregnated rods, were prepared. The impregnated specimens had been impregnated from one to five times. The specimens were subjected to the same heat treatment so that the crystal structure development was similar, but the varying number of impregnations produced specimens of different densities. Flexural tests were performed to determine the Young's modulus, flexural strength, breaking strain, and ultimate strain energy per unit volume. The Young's modulus values of the test specimens were found to be directly related to the number of impregnations received by the specimens, i.e., the more impregnations, the larger the modulus. The strength of the rods varied in a similar way, but the breaking strain remained fairly constant for all specimens tested. The ultimate strain energy rose rapidly during the first two impregnations and then tended to saturate. The apparent density changed in this manner also.

The next series of tests was conducted on specimens having the same density but a variation in the degree of crystallinity. It was found that generally, the breaking strain rose progressively as the crystallinity increased. However, the ultimate strain energy per unit volume remained constant within experimental error. On the basis of this investigation, previous work that had been done, and the known behavior of graphite when subjected to irradiation, the following model was proposed as the method of deformation of graphite to the point of failure. When a stress is applied to a polycrystalline granular graphite, it is accommodated by shear deformation within the grains. These shears are accompanied by slips, but the accumulation of slip is not the ultimate cause of failure. When sufficient strain energy is present, within the grain, to perform the work necessary to propagate a crack across the misoriented grain (or sub-grain) boundaries, failure occurs. Taking into account the effects of impregnation of graphite, they concluded that although impregnation increases

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the total breaking strain, the strain energy required to produce shear within the grains is a constant, and the portion of the strain exceeding that experienced by unimpregnated graphite is accommodated by the impregnating material.

It was previously pointed out that the current theories of the failure mechanisms of graphite were formulated largely on the basis of experimentation conducted under static conditions or at low strain rates. At this time, very little is known concerning the behavior of graphite when subjected to explosive impact. It is known that impulsive loads with their high pressures and rapid rates of loading may markedly alter the mechanical properties of a material (Reference A-61). Thus, changes in the mechanical properties of a material may cause the behavior of an impulsively loaded body to be quite different from that of a conventionally loaded body of the same material. Recent work performed by the Air Force Special Weapons Center (Reference A-62) on distended materials has provided invaluable information on shock propagation and Hugoniot's equation-of-state of commercial graphite. The results of these tests are listed in Table A-III and are shown graphically in Figures A-1, A-2, and A-3. These results will be compared with data collected from the gas gun experiments.

Some approximate theories for shock propagation in distended material were also presented in the referenced report. Simplified equations of state of the type termed "locking solids" were assumed in these theories. Two different models, were considered. While these models do not produce exact results, the simplification will permit approximation of the behavior of the distended material.

The first, and the simplest, case considers the distended material collapsing under negligibly small pressure. Thereafter, the density of the compressed material is assumed to be either independent of pressure or perfectly rigid in compression. Therefore, the material behaves as a solid under dynamic loads. This model is termed the "simple-rigid" locking-solid.

The second case considered is more applicable to graphite. In this theoretical model, the distended material is assumed to withstand a finite pressure before collapse of pore space; this model is termed the "elastic-rigid" locking solid.

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TABLE A-III

SUMMARY OF EXPERIMENTAL DATA OF GRAPHITE EQUATION OF STATE

<u>Density (g/cc)</u>	<u>Acoustic Velocity (mm/μsec)</u>	<u>Shock Velocity (mm/μsec)</u>	<u>Particle Velocity (mm/μsec)</u>	<u>Shock Pressure (kb)</u>	<u>Specific Volume (cm³/g)</u>	<u>Test Method*</u>
1.93	-	4.38	0.99	83.9	0.400	a
1.93	-	3.81	0.76	55.4	0.417	a
1.93	-	2.46	0.24	11.2	0.468	b
1.93	-	2.06	0.08	3.1	0.498	b
1.93	-	3.74	0.76	54.0	0.413	b
1.93	-	3.27	0.45	28.0	0.447	b
1.93	-	2.87	0.30	16.4	0.464	b
1.93	-	2.45	0.27	12.6	0.461	b
1.57	1.64	3.18	1.13	57.6	0.406	a
1.57	1.64	2.55	0.79	3.13	0.445	a
1.57	1.64	1.28	0.45	8.9	0.418	b
1.57	1.64	0.66	0.30	3.1	0.452	b
1.05	2.67	2.28	1.26	30.2	0.425	a
1.05	2.67	1.60	0.83	14.0	0.455	a
1.05	2.67	1.94	1.17	23.8	0.378 ^c	b
1.05	2.67	1.70	0.75	13.4	0.532	b
1.05	2.67	1.43	0.56	8.4	0.579	b

*Legend

- a. Driver Plate Method.
- b. X-ray Method.
- c. Result is questionable.

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Figure A-1. Experimental Data for Foams:

- (a) Shock Velocity, Particle-Velocity Plot
- (b) Pressure, Particle-Velocity Plot (Hugoniot Curve)

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Figure A-2. Experimental Data for Foams:

- (a) Shock Velocity, Particle-Velocity Plot
- (b) Pressure, Particle-Velocity Plot (Hugoniot Curve)

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Figure A-3. Experimental Data for Foams:

- (a) Shock Velocity, Particle-Velocity Plot
- (b) Pressure, Particle-Velocity Plot (Hugoniot Curve)

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Inclusion of a finite yield strength into theory shows that for a range of pressures, a double-wave system exists. The precursor shock front travels with acoustic velocity and has a stress amplitude equal to the yield strength. (This yield strength is the dynamic yield point under uniaxial stress.) Following the elastic wave is a slower plastic shock which collapses the pore space. This plastic shock has a stress amplitude greater than the elastic wave and a wave velocity less than sonic.

The proof of these theories is presented in Appendix B. A closed-form solution which completely describes the propagation of the double-shock system in the distended material and predicts the attenuation of a shock in a distended material due to internal losses is also presented in the appendix.

The tests conducted with commercial graphite have shown that the theories of locking solids can be used to predict the behavior of graphite, although the predictions are only fair. Modification to the elastic-rigid locking solid theory could provide better predictions, but the theory would become more complex. As a simple first approximation, this theory is adequate.

APPENDIX B
THEORY OF "LOCKING SOLIDS"

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APPENDIX B

THEORY OF "LOCKING SOLIDS"*

I. INTRODUCTION

In order to predict the effectiveness of a given countermeasure** in reducing the peak pressure delivered to a structure, a theoretical description of the propagation of a pressure pulse in the countermeasure is essential. In this section an approximate theory is presented which is intended to serve that purpose. This theory is not expected to give exact results as might be obtained from high-speed machine calculations in which more details of the flow are considered. Nevertheless, it should be approximately valid. Further, its relative simplicity allows the effects of changes in parameters to be easily evaluated.

Essentially two problems are to be considered. These fall into the general class of wave propagation in "locking solids," for which some theoretical work already exists (References B-1 and B-2). To the writer's knowledge the specific problems considered here, however, have not previously been treated.

Locking solids in general are materials in which irreversible changes in density occur under pressure. Porous, or distended, materials are the major representatives, since the pore space of the material collapses at some critical pressure determined by the elastic yield strength.

For both problems*** the collapsed, or locked material, is considered to be perfectly rigid in comparison to the distended material. This is then dropped

* This theory was extracted from the Air Force Special Weapons Center Report AFSWC TDR-62-22, Experimental Testing of Shock Attenuating Materials, March 1962, p. 21 through 44 (SECRET). The text that was extracted and is presented herein is unclassified.

** In this report a countermeasure is the distended material under examination.

*** The first problem considered is that of shock propagation in a simple locking solid in which the elastic yield strength is considered to be negligible. In the second problem the theory is extended to include effects which result from significant yield strength.

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and the collapsed material is assumed to have compressibility characteristic of the solid material.

In these problems it is also assumed that an impulse is delivered to the free-boundary of the distended solid in a time very short in comparison to other times of interest. Hence, it is assumed for convenience that the impulse is delivered instantaneously.

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II. SIMPLE-RIGID LOCKING SOLID

We consider the problem of the propagation of a shock wave in a distended material for which the critical pressure for locking, or collapse of pore space, is negligible. The form of equation of state is illustrated in Figure B-1. The initial, distended specific volume is designated as V_0 . Under very small pressure the specific volume decreases to V_1 , and thereafter remains constant.

We assume that an impulse, I_0 , is delivered to the free boundary, $x = 0$, of a semi-infinite slab of such a material at time, $t = 0$.

We wish to describe the propagation of the shock pulse produced in the distended material and, also, to ascertain the effects of placing a solid boundary at some position, $x = x_1 > 0$, in the flow field. An $x - t$ diagram is illustrated as Figure B-2.

At $t = 0$, an impulse, I_0 , is delivered instantaneously to the free boundary of the distended material, Material "A." This impulse propagates as a shock pulse of velocity U_1 . The shock changes the state of the material from that designated by subscripts "0" to that designated by subscripts "1." P denotes pressure, V specific volume, and u mass velocity. The coordinates of the shock front are x_s, t_s ; the coordinates of the rear free-surface are x_b, t_b .

At time $t = t_1$, the shock reflects at a boundary with a solid material, Material "B," the states of which are designated by prime symbols. At this point the assumption of rigidity of the locked solid is dropped. The reflected shock in Material "A," of velocity U_2 , transforms the state to that denoted by subscripts "2." Finally, the reflected shock strikes the rear free-boundary of the locked solid, where it reflects as a rarefaction fan. The particle velocity behind this fan is denoted by u_3 .

The problem is considered only to the point of calculating the reflected state, P_2, u_2 , and the mass velocity, u_3 . These quantities are sufficient only to predict the peak pressure and the total momentum delivered to the solid boundary. The subsequent behavior of the shock in the solid can be deduced from standard methods if desired.

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Pressure-Volume Curve for Simple-Rigid Locking Solid

Figure B-2

Time-Distance Diagram Illustrating Shock Propagation
in Simple-Rigid Locking Solid, and Reflection at a Solid Boundary

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From the Rankine-Hugoniot jump conditions we have

$$\frac{V_1}{V_0} = 1 - \frac{u_1}{U_1} \quad (1)$$

and

$$P_1 = \rho_0 U_1 u_1 \quad (2)$$

where P_0 is assumed zero, and $\rho_0 = V_0^{-1}$.

Because of the assumption that the locked material is rigid, the mass velocity, u_1 , and the specific volume, V_1 , are constant throughout the compressed region bounded by $x_s(t)$ and $x_b(t)$. Conservation of momentum requires, then

$$\begin{aligned} I_0 &= \rho_1 u_1 (x_s - x_b) \\ &= \rho_0 u_1 x_s \end{aligned} \quad (3)$$

but, from Equation (1),

$$\begin{aligned} u_1 &= U_1 \left(1 - \frac{V_1}{V_0} \right) \\ &= \dot{x}_s \left(1 - \frac{V_1}{V_0} \right) \end{aligned} \quad (4)$$

where the dot signifies differentiation with respect to time. Substituting Equation (4) into Equation (3),

$$I_0 = \rho_0 \dot{x}_s \left(1 - \frac{V_1}{V_0} \right) x_s$$

This equation can be integrated to give

$$x_s^2 = \left(\frac{2I_0 V_0^2}{V_0 - V_1} \right) t_s \quad (5)$$

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Equation (5), thus, describes the shock path.

The rate of decay of mass velocity, u_1 , is given by Equation (3)

$$u_1 = \frac{I_o V_o}{x_s} \quad (6)$$

or, substituting for x_s from Equation (5)

$$\begin{aligned} u_1 &= \frac{I_o V_o}{\left[\left(\frac{2 I_o V_o^2}{V_o - V_1} \right) t_s \right]^{1/2}} \\ &= \left[\frac{I_o (V_o - V_1)}{2 t_s} \right]^{1/2} \end{aligned} \quad (7)$$

The expression for x_b is easily found to be

$$x_b = \frac{V_o - V_1}{V_o} x_s$$

The decay in pressure, P_1 , is found by combining Equations (1), (2), and (6) or (7)

$$P_1 = \frac{\rho_o u_1^2}{1 - \frac{V_1}{V_o}} = \frac{u_1^2}{V_o - V_1} \quad (8)$$

$$= \frac{I_o^2 V_o^2}{x_s^2 (V_o - V_1)} \quad (9)$$

$$= \frac{I_o}{2 t_s} \quad (10)$$

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Denoting the total mass of shocked (locked) material by μ , we have

$$\mu = \rho_o x_s$$

so that Equation (9) can be written

$$P_1 = \frac{I_o^2}{\mu^2 (V_o - V_1)} \quad (11)$$

We now consider the reflection at the solid boundary between Material "A" and Material "B." At this point ($x = x_1$, $t = t_1$), the assumption that the locked countermeasure is perfectly rigid is dropped, and Material "A" is henceforth assumed to have an equation of state typical of the solid material from which the distended countermeasure is fabricated. The assumption of perfect rigidity of the locked countermeasure has been satisfactory up to this point because its rigidity has been, in effect, compared to that of the distended countermeasure. Upon reflection at the solid boundary, however, the rigidity of the locked countermeasure is effectively compared to that of Material "B," and the assumption of perfect rigidity is no longer adequate.

The states produced in both materials upon reflection at x_1 , t_1 , are found from the Hugoniot jump conditions. We have

$$\frac{V_2}{V_1} = 1 - \frac{u_2 - u_1}{U_2 - u_1} \quad (12)$$

and

$$P_2 - P_1 = \rho_1 (U_2 - u_1) (u_2 - u_1) \quad (13)$$

for the state in Material "A."

For Material "B," we have

$$\frac{V'_2}{V} = 1 - \frac{u_2}{U'_2} \quad (14)$$

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and

$$P_2 = \rho_o' U_2' u_2 \quad (15)$$

In Equations (14) and (15) the boundary condition requiring continuity of pressure and particle velocity across the boundary has been invoked.

Eliminating u_2 from Equations (13) and (15), we get

$$P_2 = \frac{\rho_o' U_2' (\rho_1 u_1 U_2^m + P_1)}{\rho_1 U_2^m + \rho_o' U_2'} \quad (16)$$

where the shock velocity with respect to the material ahead of the shock in Material "A" is denoted by U_2^m . That is,

$$U_2^m = - (U_2 - u_1)$$

Substituting for P_1 and u_1 in Equations (16) from Equations (6) and (11)

$$P_2 = \frac{\rho_o' U_2' \left[\frac{\rho_1 U_2^m I_o}{\mu} + \frac{I_o^2}{\mu^2 (V_o - V_1)} \right]}{\rho_1 U_2^m + \rho_o' U_2'} \quad (17)$$

or

$$P_2 = \frac{\rho_o' U_2' I_o}{\mu} \left[\frac{\rho_1 U_2^m + \frac{I_o}{\mu (V_o - V_1)}}{\rho_1 U_2^m + \rho_o' U_2'} \right] \quad (17a)$$

In general, the velocities U_2' and U_2^m in Equation (17) will be dependent on the pressure P_2 . For most solid materials, however, the variation in shock velocity with pressure is small so that Equation (17) can be used for estimation purposes. An explicit solution is quite feasible if desired.

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Equation (17) can be rewritten in more concise form by letting the ratio $\rho_o' U_2' / \rho_1 U_2^m = R$; and recalling from Equations (2) and (8) that

$$\frac{I_o}{\mu (V_o - V_1)} = \rho_o U_1$$

We can then write Equation (17) as

$$P_2 = \frac{\rho_o' U_2' I_o}{\mu} \left[\frac{1 + \frac{\rho_o U_1}{\rho_1 U_2^m}}{1 + R} \right] \quad (18)$$

since, in all cases of interest

$$\rho_o U_1 \ll \rho_1 U_2^m$$

It is evident that, for fixed $\rho_o' U_2'$, P_2 decreases with increasing R . From the standpoint of minimizing P_2 , therefore, it is desirable that the locked countermeasure have as low a shock impedance as possible. If, as is discussed below, momentum transfer is also to be minimized, then this places certain restrictions on the value of $\rho_1 U_2^m$.

In addition to the pressure, P_2 , the mass velocity, u_3 , produced as the reflected shock, U_2 , reflects from the free boundary, x_{b1} , of Material "A" is of interest. Before estimating u_3 , however, we need to find the value of u_2 . Substituting for P_2 in Equation (15) from Equation (17)

$$u_2 = \frac{I_o}{\mu} \left[\frac{\rho_1 U_2^m + \frac{I_o}{\mu (V_o - V_1)}}{\rho_1 U_2^m + \rho_o' U_2'} \right] \quad (19)$$

The change in mass velocity across the rarefaction generated by reflection of U_2 at the rear free boundary is found from the Riemann integral, viz.

$$u_3 - u_2 = \int_{V_2}^{V_1} \frac{dP}{\rho c}$$

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where ρ and c are the values of density and sound speed passed through in crossing the rarefaction fan. We can estimate u_3 conveniently by assuming the value ρc of the locked material to be independent of pressure. This is equivalent to assuming the shock speed, U_2^m , to be independent of pressure. It is also equivalent to the assumption that the Hugoniot and (adiabat) of the locked material is a straight line of negative, finite slope in the pressure-volume plane.

We have then, with this assumption

$$u_3 - u_2 = \frac{-P_2}{\rho_2 c_2} \quad (20)$$

It is easily shown from the jump conditions and the definition of sound velocity that

$$\rho_2 c_2 = \rho_1 U_2^m$$

Hence, Equation (20) is

$$u_3 - u_2 = \frac{-P_2}{\rho_1 U_2^m}$$

Upon substituting for P_2 and u_2 from Equations (17) and (19), we find

$$u_3 = \frac{I_o}{\mu} \left[\frac{\rho_o U_2^m + \frac{I_o}{\mu (V_o - V_1)}}{\rho_o U_2^m + \rho_o' U_2'} \right] \left[1 \frac{\rho_o' U_2'}{\rho_1 U_2^m} \right] \quad (21)$$

Alternatively, we can evaluate the quantity

$$\mu (u_1 - u_3) = I_1$$

where I_1 is approximately the total momentum delivered to Material "B." We get

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$$I_1 = I_0 \left\{ \left[\frac{\rho_1 U_2^m + \frac{I_0}{\mu (V_0 - V_1)}}{\rho_1 U_2^m + \rho'_0 U'_2} \right] \left[\frac{\rho'_0 U'_2}{\rho_1 U_2^m} - 1 \right] + 1 \right\}$$

or, denoting the ratio $\rho'_0 U'_2 / \rho_1 U_2^m$ by R , and the quantity $I_0 / [\rho_1 U_2^m \mu (V_0 - V_1)]$ by K , we can write,

$$I_1 = I_0 \left[(1 + K) \left(\frac{R - 1}{R + 1} \right) + 1 \right] \quad (22)$$

From Equation (22) it is evident that for

$$R > 1 ; I_1 > I_0$$

$$R = 1 ; I_1 = I_0$$

$$R < 1 ; I_1 < I_0$$

In the latter case, although $I_1 < I_0$ for the first reflection, subsequent reflections not considered in the theory would necessarily increase I_1 to equal I_0 . Otherwise, momentum would not be conserved.

Consideration of Equations (18) and (22) shows that the minimum pressure is delivered to Material "B," within the restriction that the momentum delivered not exceed I_0 , when Material "A" in the locked condition has the same shock impedance as Material "B."

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III. ELASTIC-RIGID LOCKING SOLID

We now consider a problem similar to that of the previous section, but no longer assume that the critical pressure at which the pore volume of the countermeasure material collapsed is negligible. We therefore assume an equation of state of the form shown in Figure B-3.

Figure B-3
Pressure-Volume Curve
for Elastic-Rigid Locking Solid

As in the previous problem the initial, distended specific volume is denoted by V_0 . Below a certain critical pressure, P_e , the material behaves as an isotropic elastic solid. Above a critical elastic wave amplitude, designated by P_e , V_e , the pore space collapses irreversibly and the material locks at volume V_1 . Thereafter, the locked material behaves as a perfectly rigid solid in comparison to the rigidity of the distended material. As in the previous problem this assumption is dropped at a time when all of the countermeasure material has been locked. Thus, in effect, when the rigidity of the locked

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material is no longer compared to that of distended material, the locked material is assumed to have a finite compressibility characteristic of the solid material from which the countermeasure is fabricated.

The problem is considered in two parts; first we treat the propagation of a pressure pulse in a semi-infinite half-space of the distended material. Secondly, the effect of inserting a solid boundary at some point in the flow field is treated. Our ultimate goal, as before, is to determine the peak pressure and total momentum delivered to such a solid boundary.

The first phase of the problem, viz., the propagation of a pressure pulse in a semi-infinite half space of distended material, is illustrated qualitatively in the x, t plane in Figure B-4.

Figure B-4

Time-Distance Diagram Illustrating
Shock Propagation in Elastic-Rigid Locking Solid

As before, an impulse, I_0 , is delivered to the free boundary, $x = 0$, in a negligibly short time at time, $t = 0$. Initially this impulse propagates as a

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single shock front whose coordinates are denoted by x_s , t_s . The state behind this front is denoted by P_1 , u_1 , V_1 . The coordinates of the rear free boundary are specified by x_b , t_b .

As the pressure at the shock front drops, it eventually reaches the value P_c (Figure B-4). For shock pressures between P_c and P_e a single front is unstable and an elastic front of pressure, P_e , splits off and precedes the main shock. The coordinates at which this instability first occurs are designated by t_c , x_{sc} , and x_{bc} .

After time t_c , the pulse thus consists of two shock fronts. The first travels with constant velocity, U_e . Between this front and the second front the state is constant and is denoted by subscripts "e." The second shock decays as it propagates for two reasons. First, in order that momentum be conserved, the total momentum in the second shock decreases as the momentum in the first wave increases. Secondly, the total mass in the second wave, over which its momentum is distributed, increases with time. Hence, since the mass velocity is constant over the entire locked region ($x_s - x_b$), this mass velocity must decrease with time.

At some point, denoted by x_s^* , t^* (Figure B-4), the mass velocity behind the second shock is just equal to the mass velocity in the elastic wave. At this point the pressure difference across the second front also reduces to zero. A density change still exists across the boundary, $x_s(t)$, however, as the material behind the second front is locked.

For times greater than t^* the total mass between the boundary $x_s(t)$ and the rear free-boundary, $x_b(t)$, remains constant. In order for momentum to be conserved, a rarefaction fan emanates from the boundary $x_s(t)$. The first characteristic of this fan is centered at x^* , t^* . This fan propagates into the elastically shocked region, reducing the pressure and mass velocity in that region.

With this qualitative description we now turn to quantitative calculation of the propagation of such a pulse.

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The critical pressure, P_c , above which a single shock exists is determined by the intersection of the straight line forming an extension of the P-V curve in the elastic region with the vertical line, $V = V_1$ (Figure B-3).

We can write, therefore,

$$P_c = \frac{P_e (V_o - V_1)}{V_o - V_e} \quad (23)$$

Until such time as the pressure reduces to P_e , a single shock front exists and the analysis of the preceding section is valid. Thus, we have

$$P_1 = \frac{I_o^2 V_o^2}{x_s^2 (V_o - V_1)} = \frac{I_o}{2 t_s} = \frac{I_o^2}{\mu^2 (V_o - V_1)} \quad (t < t_c) \quad (24)$$

from Equations (8) and (11) of Section II.

Also, from Equation (6)

$$u_1 = \frac{I_o V_o}{x_s} = \frac{I_o}{\mu} ; \quad t < t_c$$

The coordinates at which the elastic precursor splits away from the main shock are then

$$t_c = \frac{I_o}{2 P_c} = \frac{I_o (V_o - V_e)}{2 P_e (V_o - V_1)} \quad (25)$$

and

$$x_{sc} = \frac{I_o V_o}{P_c (V_o - V_1)} = \frac{V_o I_o}{V_o - V_1} \sqrt{\frac{V_o - V_e}{P_e}} \quad (26)$$

$$x_{bc} = I_o \sqrt{\frac{V_o - V_e}{P_e}} \quad (27)$$

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Further,

$$\mu_c = \rho_o x_{sc} = \frac{I_o}{V_o - V_1} \sqrt{\frac{V_o - V_e}{P_e}} \quad (28)$$

For times greater than t_c , we require two sets of jump conditions, valid across each shock front, respectively. For the elastic wave we have

$$u_e = \frac{U_e (V_o - V_e)}{V_o} \quad (29)$$

and

$$P_e = \rho_o U_e u_e$$

For the second shock we have

$$u_1 - u_e = \frac{(U_1 - u_e) (V_e - V_1)}{V_e} \quad (31)$$

and

$$P_1 - P_e = \rho_e (U_1 - u_e) (u_1 - u_e) \quad (32)$$

Combining Equations (29) and (30), we can write

$$P_e = \frac{u_e^2}{V_o - V_e} = \frac{(V_o - V_e)}{V_o^2} U_e^2 \quad (33)$$

Similarly, combining Equations (31) and (32),

$$P_1 - P_e = \frac{(u_1 - u_e)^2}{V_e - V_1} = \frac{(V_e - V_1) (U_1 - u_e)^2}{V_e^2} \quad (34)$$

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With Equations (29) to (34) we can expression Equations (25) to (28) for convenience as

$$t_c = \frac{I_o V_o^2}{2 (V_o - V_l) U_e^2} \quad (35)$$

$$x_{sc} = \frac{I_o V_o^2}{U_e (V_o - V_l)} \quad (36)$$

$$x_{bc} = \frac{I_o V_o}{U_c} \quad (37)$$

$$\mu_c = \frac{I_o V_o}{U_e (V_o - V_l)} \quad (38)$$

For times greater than t_c , conservation of momentum requires

$$I_o = \rho_e u_e (x_e - x_s) + \rho_l u_l (x_s - x_b) \quad (39)$$

Moreover,

$$x_e = U_e (t - t_c) + x_{sc} \quad (t > t_c) \quad (40)$$

$$x_s = \int_{t_c}^t U_l dt + x_{sc} \quad (t > t_c) \quad (41)$$

$$x_b = \int_{t_c}^t u_l dt + x_{bc} \quad (t > t_c) \quad (42)$$

For convenience, we introduce a change in variables as follows. Let

$$x_e - x_{sc} = \eta_e$$

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$$x_s - x_{sc} = \eta_s \quad (43)$$

$$x_b - x_{bc} = \eta_b$$

$$t - t_c = \tau$$

In terms of these variables, Equations (40) to (42) are

$$\eta_e = U_e \tau \quad (44)$$

$$\eta_s = \int_0^\tau U_1 d\tau \quad (45)$$

$$\eta_b = \int_0^\tau u_1 d\tau \quad (46)$$

Equation (39) can then be written

$$I_o = \rho_e u_e (\eta_e - \eta_s) + \rho_1 u_1 (\eta_s - \eta_b + x_{sc} - x_{bc}) \quad (47)$$

Substituting for U_1 in Equation (45) from Equation (31),

$$\eta_s = \int_0^\tau \left[u_e + \frac{V_e (u_1 - u_2)}{V_e - V_1} \right] d\tau$$

which reduces to

$$\eta_s = \frac{1}{V_e - V_1} \left[V_e \int_0^\tau u_1 d\tau - V_1 u_e \tau \right]$$

Or, by substituting from Equation (46),

$$\eta_s = \frac{1}{V_e - V_1} \left[V_e \eta_b - V_1 u_e \tau \right] \quad (48)$$

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Substituting in Equation (47) for η_s , and letting $u_1 = \dot{\eta}_b$, where the dot denotes differentiation with respect to time,

$$\begin{aligned} I_o &= \rho_e u_e \left[U_e \tau - \frac{1}{V_e - V_1} (V_e \eta_b - V_1 u_e \tau) \right] \\ &+ \rho_1 \dot{\eta}_b \left[\frac{1}{V_e - V_1} (V_e \eta_b - V_1 u_e \tau) - \eta_b + \frac{I_o V_o V_1}{U_e (V_o - V_1)} \right] \end{aligned} \quad (49)$$

After some simplification, Equation (48) becomes

$$(V_e - V_1) I_o = \dot{\eta}_b [\eta_b - u_e \tau + \mu_c (V_e - V_1)] - u_e \eta_b + u_e^2 \left(\frac{V_o - V_1}{V_o - V_e} \right) \tau \quad (50)$$

This equation is the differential equation describing the path of the rear free boundary.

Consider now the derivative

$$\begin{aligned} \frac{1}{2} \frac{d}{dt} \eta_b - u_e + \mu_c (V_e - V_1)^2 &= \eta_b - u_e + \mu_c (V_e - V_1) \dot{\eta}_b - u_e \\ &= \dot{\eta}_b \left[\eta_b - u_e \tau + \mu_c (V_e - V_1) \right] - u_e \eta_b \\ &+ u_e^2 - u_e \mu_c (V_e - V_1) \end{aligned}$$

The first two terms of this derivative are the same as the first two terms on the right hand side of Equation (50). Hence, Equation (50) can be written

$$(V_e - V_1) I_o = \frac{1}{2} \frac{d}{d\tau} \eta_b - u_e + \mu_c (V_e - V_1)^2 + u_e \mu_c (V_e - V_1) + u_e^2 \left(\frac{V_e - V_1}{V_o - V_e} \right) \tau$$

or

$$I_o = \frac{1}{2 (V_e - V_1)} \frac{d}{d\tau} [\eta_b - u_e \tau + \mu_c (V_e - V_1)]^2 + u_e \mu_c + \frac{u_e^2 \tau}{V_o - V_e}$$

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This equation integrates immediately by separation of variables to give

$$[\eta_b - u_e \tau + \mu_c (V_e - V_1)]^2 = 2 (V_e - V_1) \left[(I_o - u_e \mu_c) - \frac{u_e^2 \tau^2}{2(V_o - V_e)} \right] + C \quad (51)$$

The constant of integration is evaluated by observing that at $\tau = 0$, $\eta_b = 0$.

Hence,

$$C = [\mu_c (V_e - V_1)]^2$$

Equation (51) can therefore be written

$$(\eta_b - u_e \tau)^2 + 2 \mu_c (V_e - V_1) (\eta_b - u_e \tau) = 2 (V_e - V_1) \left[(I_o - u_e \mu_c) \tau - \frac{u_e^2 \tau^2}{2(V_o - V_e)} \right]$$

Solving for $(\eta_b - u_e \tau)$ we have, after some simplification,

$$\eta_b - u_e \tau = (V_e - V_1) \left[-\mu_c + \sqrt{\mu_c^2 + \frac{2 I_o \tau}{V_o - V_1} - \frac{P_e \tau^2}{V_e - V_1}} \right] \quad (52)$$

Equation (52) is thus the solution for the path of the rear free boundary for times greater than t_c .

The position of the second shock front follows immediately from Equation (48). In order to find the solution for the mass velocity, u_1 , we differentiate Equation (52). This gives

$$u_1 = \frac{dx_b}{dt} = \frac{d\eta_b}{d\tau} = u_e + \frac{(V_e - V_1) \frac{I_o}{V_o - V_1} - \frac{P_e \tau}{V_e - V_1}}{\sqrt{u_c^2 + \frac{2 I_o \tau}{V_o - V_1} - \frac{P_e \tau^2}{V_e - V_1}}} \quad (53)$$

The pressure P_1 is then determined from Equation (34).

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Evidently the mass velocity u_1 will be equal to u_e when the numerator of the second term of Equation (53) is zero. Denoting this time by τ^* , we have

$$\tau^* = \frac{I_o (V_e - V_1)}{P_e (V_o - V_1)} \quad (54)$$

At this time no difference in mass velocity exists across the boundary x_s , and the speed of the second shock with respect to the material ahead of the shock has dropped to zero. Consequently, no pressure difference exists. The amplitude of the second shock at this time therefore has decayed to zero. However, a change in density exists across x_s , since the material to the left of the boundary has been locked. Up to this time the mass of locked material has increased with time. After time τ^* , however, the mass of locked material remains constant.

Values of other parameters at time τ^* are written below for convenience.

$$\eta_e^* = U_e \tau^* = \frac{U_e I_o (V_e - V_1)}{P_e (V_o - V_1)}$$

$$\eta_b^* = u_c \tau^* + (V_e - V_1) \left[-\mu_c + \sqrt{\mu_c^2 + \frac{I_o^2 (V_e - V_1)}{P_e (V_o - V_1)^2}} \right]$$

$$\eta_s^* = \frac{1}{V_e - V_1} V_e \eta_b^* - V_1 u_e \tau^*$$

$$\mu^* = \rho_1 (x_s^* - x_b^*)$$

For times greater than t^* , the jump conditions, Equations (31) and (32) no longer apply across the boundary, x_s . At this time a rarefaction wave begins to issue from the boundary, x_s . As this wave propagates it slows down the mass velocity and reduces the pressure in the elastic wave. At the same

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time the mass of locked material, μ^* , also slows down in order to preserve continuity of pressure and particle velocity across the boundary. The rate at which the locked material slows down is found from conservation of momentum. Hence, we can write

$$\frac{dI_s}{dt} = - \frac{dI_e}{dt}$$

where I_s is the momentum of the locked material and I_e the momentum in the elastic wave.

Referring to Figure B-4, and recalling that the mass of locked material, μ^* , remains constant for $t > t^*$, we have

$$\mu^* \dot{u}_1 = - \frac{d}{dt} \int_{x_s}^{x_e} \rho u dx \quad (55)$$

where ρ and u are the local density and mass velocity in the elastic wave.

We now make the simplifying assumption that the slope of each characteristic of the rarefaction fan is the same, and is equal to that of the elastic shock front. This is equivalent to neglecting the particle velocity of the elastic wave in comparison to the sound speed.

The rarefaction travels into material in a constant state; hence, it is a simple wave. The characteristics are therefore straight lines along which all the flow parameters are constant. Moreover, they are all parallel by virtue of the assumption stated in the preceding paragraph.

Under these conditions the total momentum between the elastic shock front and any characteristic curve of the rarefaction fan is constant with time. Referring to Figure B-4, we see that the momentum lost by the locked material in a time increment, dt , is just equal to that gained by the elastic wave in the region bounded by the curve, $x_s(t)$, and the characteristic issuing from x_s , whose path is denoted by $x_r(t)$. This region is shaded in Figure B-4.

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Equating the momentum lost by μ^* to the momentum gained by the elastic wave, we have

$$\mu^* \dot{u}_1 = - u_1 \rho (\dot{x}_r - \dot{x}_s) \quad (56)$$

However,

$$\dot{x}_r = U_e$$

and

$$\dot{x}_s = u_1$$

Moreover, Equations (29) and (30) also apply to the rarefaction since the sound speed has been assumed constant and equal to the elastic shock speed.

Hence

$$\rho = \frac{\rho_o U_e}{U_e - u_1}$$

Substituting these last three relations into Equation (56), there results

$$\begin{aligned} \mu^* \dot{u}_1 &= \frac{u_1 \rho_o U_e}{U_e - u_1} (U_e - u_1) \\ &= - u_1 \rho_o U_e \end{aligned} \quad (57)$$

Equation (57) is thus the differential equation for the mass velocity in the collapsed material as a function of time for $t > t^*$.

Separating variables in Equation (57), and integrating, we have

$$\log \frac{u_1}{u_e} = - \frac{\rho_o U_e}{\mu^*} (t - t^*)$$

or

$$u_1 = u_e \exp \left[- \frac{\rho_o U_e}{\mu^*} (t - t^*) \right] \quad (58)$$

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In order to find the trajectory of the boundary, x_s , Equation (58) is integrated once again with respect to time. We get

$$x_s = \frac{u_e \mu^*}{\rho_o U_e} \exp \left[- \frac{\rho_o U_o}{\mu^*} (t - t^*) \right] + A \quad (59)$$

where "A" is a constant of integration.

The constant of integration can be evaluated at $t = t^*$. This gives

$$x_s^* = - \frac{u_e \mu^*}{\rho_o U_e} + A \quad (60)$$

Substituting for A from Equation (60) into Equation (59) we obtain

$$x_s = x_s^* + \frac{u_e \mu^*}{\rho_o U_e} \left[1 - \exp \left(- \frac{\rho_o U_e}{\mu^*} (t - t^*) \right) \right] \quad (61)$$

From Equation (61) it is clear that the ultimate depth of penetration of the collapsed material in laboratory coordinates is

$$x_{s(max)} = x_s^* + \frac{u_e \mu^*}{\rho_o U_e} \quad (62)$$

At this point we have obtained a complete solution for the propagation of the pulse in the distended material. The trajectories of both waves for times less than or greater than t^* are known. Further, profiles of the entire pulse at any time are easily found graphically by drawing straight characteristic curves beyond (x^*, t^*) , along which all flow parameters are constant.

Having obtained a complete solution for the propagation of the pulse in the distended material, we now turn to consideration of the effect of a solid boundary at some point in the flow field.

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Firstly, the reflection of the elastic shock at a solid boundary causes an increase in pressure at the boundary over that of the elastic shock. If the solid material is assumed to be rigid in comparison to the distended material, we can compute the reflected and transmitted pressure, P_2 , from the jump conditions for the reflected shock. These are

$$u_2 - u_e = \frac{(U_2 - u_2)(V_e - V_1)}{V_e} \quad (63)$$

and

$$P_2 - P_e = \rho_e (U_2 - u_e)(u_2 - u_e)$$

Setting $u_2 = 0$, and combining these two relations,

$$P_2 = P_e + \frac{u_e^2}{V_e - V_1}$$

Substituting for P_e from Equation (33)

$$P_2 = P_e \left[1 + \frac{V_0 - V_e}{V_e - V_1} \right] \quad (64)$$

From Equation (64) it is evident that for moderate distensions, e.g., distensions of two or more, the second term on the right is a small fraction of the first term. Hence, the transmitted and reflected pressure, P_2 , is nearly equal to the incident pressure, P_e .

The velocity of the reflected shock (in laboratory coordinates) in the distended material is found from Equation (63)

$$U_2 = - \frac{u_e V_1}{V_e - V_1} \quad (65)$$

It is clear from Equation (65) that, again for moderate distensions ($V_e/V_1 \sim 2$ or more), the velocity of the reflected shock is of the same order of magnitude

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as the mass velocity in the elastic wave. Since V_e is always greater than V_1 , the sign of U_2 is always negative, indicating that the wave always travels toward $-x$.

As this reflected wave travels toward the rear boundary it eventually interacts with the boundary x_s . This interaction may occur either prior to or later than time t^* . If the interaction is later than t^* , the reflected wave will first interact with the rarefaction fan; otherwise it will not. Consequently, we have two cases to consider.

We first consider the case of the interaction occurring at a time less than t^* . This collision takes place between two regions of material both in the locked condition, traveling with different mass velocities. Hence, the collision produces shock waves in each region. Since all of the distended material has been locked at this time the assumption of perfect rigidity of the locked material is henceforth dropped. The pressure, P_3 , produced by the collision is computed as before from the jump conditions. We have

$$P_3 - P_2 = \rho_1 (U_3 - u_2) (u_3 - u_2)$$

where subscripts 3 refer to the state behind the reflected shock. Letting $u_2 = 0$, then $u_3 = (1/2)u_1$ from symmetry. Hence

$$P_3 - P_2 = \frac{1}{2} \rho_1 U_3 u_1 \quad (66)$$

The velocity U_3 depends on the properties of the locked material and, in general, also depends on P_3 .

Equation (66) gives the peak pressure in the shock waves produced by the interaction. If the locked material has the same shock impedance as the solid, Material "B," then P_3 will also be the peak pressure delivered to Material "B."

Just as in Section II, it can be shown that minimum peak pressure is delivered to Material "B," without increase of impulse, when the shock impedances of Materials "B" and "A," in the locked condition, are the same.

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The solution for the case in which the reflected shock, U_2 , interacts first with the rarefaction fan before colliding with the boundary, x_s , is much more complicated. At this writing the solution has not been obtained.

As shown in the previous AFSWC work, it is questionable whether the theories of the strength and mechanisms of failure are applicable to graphite which has been subjected to impulsive loading.

To illustrate the large changes in the behavior of graphite under impulsive loading as opposed to conventional loading, consider the deformation which occurs in each case prior to fracture. Graphite is not a perfectly elastic material and will, in fact, attain a permanent set at room temperature. However, graphite is considered to be a brittle material and loading to failure by conventional means at test temperatures less than about 2000°C results in a brittle fracture. The specimen deformation occurring with such a failure is barely observable, if observable at all. Tests have been carried out in this program at ambient temperature on 3/4-inch diameter rods of graphite which were exposed to explosive impact. These rods have undergone very extensive deformation which, it would appear, is a result of plastic flow. It is not known what temperatures are attained during such tests, but it is certainly less than the temperature at which plastic flow occurs during conventional testing. There has also been an appreciable increase in the density of the deformed material, indicating a reduction in the void volume has occurred. The manufacturer's data sheet on the graphite used in these tests lists the density as 1.67 gm/cc. The measured density of various deformed pieces of this material has ranged from about 1.8 gm/cc to as high as 1.9 gm/cc.

There are marked differences between the fracture phenomena associated with impulsive loads and those associated with static loads. The effects of impulsive loading have been studied primarily on materials which are of a more homogeneous and non-porous nature than graphite (Reference B-3). Many things not now known must be determined in order to understand the behavior of graphite under explosive impact even in a qualitative manner. Among these are the factors which control the partition of energy among plastic flow and fracture since both are energy absorbing processes. An extensive amount of investigation and experimentation will be required before an understanding of the fracture mechanisms is attained.

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