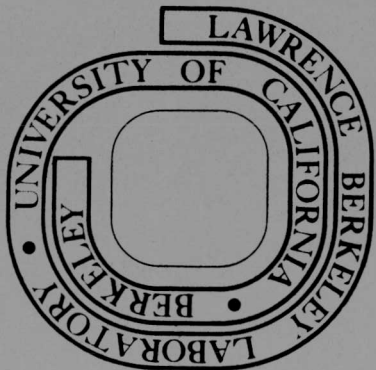


MORPHOLOGICAL CONTROL OF TITANIUM CARBIDE IN A
STEEL MATRIX

Federico Reinel
(M. S. thesis)

December, 1974

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TABLE OF CONTENTS

Abstract v

1. Introduction 1

2. Experimental Approach 2

 A. Materials Used 2

 B. Procedure Used 2

 C. Apparatus Used 5

 D. Variables Investigated 5

3. Experimental Results 6

4. Discussion and Conclusions 8

5. Summary 13

6. Suggested Areas of Further Research 14

Acknowledgements 15

Bibliography 16

Appendix I 18

Appendix II 19

Appendix III 24

Appendix IV 28

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Morphological Control of Titanium Carbide in a Steel Matrix

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ABSTRACT

One of the major methods of manufacturing steel bonded carbides by Powder Metallurgy Techniques have been investigated. The variables affecting the morphology of the microstructure have been observed and experimentally determined. Methods of achieving theoretical density as well as 35% porosity have been reviewed. Mercury porosimetry tests have been carried out to determine porosity and average porous size.

1. Introduction

Hard wear resistant materials with high strength properties at elevated temperatures have been subject of considerable study during the last twenty years.¹⁻⁴ By using powder metallurgy techniques, hard phases consisting of carbides, borides and nitrides have been able to be dispersed in a matrix which is commonly denominated as the binder. The reason for doing this is that the hard phase would provide the material with its wear resistant properties, while the binder would serve the dual purpose of locking these hard phases or grains in place and at the same time give the material some of their own properties and characteristics. Some of these properties are heat treatability, machinability, and corrosion resistant. It is within this context that the area of steel bonded carbides has been reviewed and subjected to further study.

Steel bonded carbides are a type of material in which a hard phase is distributed or dispersed in a steel matrix. The carbide phase is usually formed by the refractory carbides, some of them being titanium, tantalum, vanadium tungsten carbides and/or a combination of the above. The steel matrix may be almost any type of steel. For instance it has been possible to use work-hardenable types of steel, corrosion resistant steel, tempered resistant steels and many other classes.

It is within this context that the selection of a steel containing iron, chromium and molybdenum was considered. The criteria behind the choice being that the chromium would give corrosion resistant properties while the molybdenum would aid in the wetting abilities of the iron with respect to the titanium carbide grains.⁵

2. Experimental Approach

A. Materials Used

During the course of the investigation a steel matrix prepared from iron, chromium, carbon and molybdenum powders was used as a binder. The carbide phase consisted of titanium carbide. Steel 410L pre-alloyed powder was also considered and used as a binder for the titanium carbide powders. The powders used were in the submicron range, except for the steel 410L pre-alloyed powders which were -325 Mesh and the chromium powder (325 Mesh.).

In particular, the composition of the iron, chromium, molybdenum steel is as follows: Fe: 83.22 percent by weight; Cr: 12 percent; Mo: 0.75 percent; carbon: 0.03 percent. The titanium carbide consisted of: titanium carbide: 99.8 percent, free carbon: 0.1 percent, oxygen: 0.1 percent: The volume percent of the binder varied between 80 - 20 percent, accordingly the volume fraction of the carbide varied between 20 - 80 percent. Most of the studies were carried out under 65 percent. Vol steel - 35 percent Vol TiC carbide. Other volume fractions used were: 60 percent Vol steel - 40 percent Vol TiC, 58 percent Vol steel - 42 percent Vol titanium carbide.

B. Procedure Used

The procedure used in preparing the mixed alloy was as follows:

1. The elemental powders and the titanium carbide were weighed in the desired proportions and placed in a drum mill. The dimensions of the mill are given in Appendix II. One percent by weight of paraffin wax also added to the powders in order to improve their green strength

and at the same time to act as the die lubricant. The powders were ball milled in hexane. Careful consideration was given to the fact of not overloading the drum mill in order to obtain good mixing and tumbling action of the powders. In particular, 60.67 grs of powder was milled each time. This way, the powders and the grinding balls occupied 1/3 of the mill by volume. The amount of hexane added was 275 ml. This filled the mill completely with hexane. The powders were ball milled for twenty-four hours.

2. The hexane was decanted from the mill as much as possible. The amount of hexane decanted was 200 ml. This left the powders with 75 ml. of hexane and 0.1654 grs. of paraffin wax (0.27% by weight).

3. The powders were placed in a vacuum drying oven, and dried under a vacuum of 30 inches of mercury at room temperature for an hour.

4. Two grams of powder were weighed and compacted between pressures of 15000 psi and 70000 psi. The overall dimensions of the compacted powders varied but in general they were as follows: Diameter = 3/8 inches, height = 5/16 inches.

5. The material was then sintered in a reduced helium atmosphere of 5 inches of mercury. The procedure used in bringing the sample to the sintering temperatures was as follows:

i) The sample, after being compacted at the pressures above mentioned, was placed in an Alumina-Oxide crucible. Careful consideration was given to the placement of the sample in the crucible. The sample was always placed in the middle of the crucible. In order to standardize the position of the thermocouple inside the crucible, with respect to the specimen, a tip made of alumina-oxide was used. This tip went through a hole in the top of the sintering furnace. The thermocouple was placed

inside this tip. See Fig. 3, App II for further details of this arrangement.

ii) At room temperature, a vacuum of 4×10^{-5} mm of mercury was pumped then the sintering furnace was backfilled with helium to a reduced atmosphere of 5 inches of mercury. The procedure was repeated once more.

iii) In a reduced helium atmosphere of 5 inches of mercury the history was as follows: Enough power was supplied to the heating element in order to obtain a temperature of 630°C . The specimen was allowed to remain at this temperature for 10 minutes. Then the power was increased and a temperature of 1265°C was obtained. The sample was allowed to remain at this temperature for 10 minutes. The temperature was then increased to 1300°C and the specimen allowed to remain at this temperature for 10 minutes. Then, the temperature was increased 30°C each time and the specimen allowed to remain at the corresponding temperature for four minutes. The final sintering temperature and preferred time were 1450°C and 2 hours respectively. After sintering had taken place, the power was turned off and the temperature dropped from 1450°C to 800°C in ≈ 3 minutes.

6. The use of naphthalene, in conjunction with 410L steel prealloyed powders, as a means of producing porosity was also studied. The procedure used in milling the 410L steel prealloyed powders and the titanium carbide, as well as the heating history follows the one mentioned in the previous paragraphs. The only difference in this case being that no paraffin wax was used and that the naphthalene was added after milling the powders. This was done as follows: A solution consisting of 11.3 grs. of naphthalene and 150 ml. of hexane was prepared. Then 10 grs. of 410L prealloyed powder and titanium carbide were mixed with 6.50 ml. of the

above solution. The overall amount of naphthalene was five percent by weight of the mixed alloy. The powders were dried under the same conditions previously mentioned.

7. Mercury porosimetry tests carried out in order to determine porous size.

8. Etching of the sintered samples for the purpose of metallographic observation was carried out with a solution consisting of 30 prs. oxalic acid and 300 ml. of H_2O .

C. Apparatus Used

The milling was carried out in a drum mill. See App II, Fig. 2, for dimensions. The mill was made of 304L stainless steel and the grinding balls were made of high carbon chromium alloy, AISI 52100. (Rc hardness - 60-66).

A "Abar 90" high temperature resistance furnace was employed for the work described in this report. See Fig. 2, Appendix II, for detailed drawing of the hot zone.

D. Variables Investigated

The purpose of the investigation was to determine what kind of variables had an effect in the morphology of the titanium carbide phase dispersed in the steel matrix. In particular, it was desired to obtain round dispersed grains of titanium carbide in the steel binder. The following variables were investigated and their effect experimentally determined: sintering time, sintering temperature, milling conditions and milling time, volume percent of the steel matrix and titanium carbide, effect of molybdenum on the morphology of the titanium carbide grains, use

of naphthalene as a means of producing porosity, use of 410L steel as a binder for the titanium carbide.

3. Experimental Results

It has been found out that sintering time and sintering temperature are of paramount importance in the morphological control of titanium carbide dispersed in a steel matrix. The effects of these two variables are visible not only in obtaining round dispersed titanium carbide grains but also in the porosity of the final product. For instance, when sintering takes place, at a temperature of 1422°C and 40 minutes in a reduced helium atmosphere of 5 in. of Hg, the titanium carbide grains do not show a round configuration. There is also a sign of some conglomeration of the titanium carbide. However, when the sintering temperature is increased to 1460°C and the sintering time increased to two hours, round titanium carbide grains appear and a dispersed phase is visible. But the difference is not only in the morphology of the dispersed phase but also in the porosity as can be judged by comparing two micrographs 1 and 3^{App. III} both at a magnification of 50X of the above mentioned experiments.

It has been reported that milling conditions have an effect on the morphology of dispersed phases with respect to a binder.⁶⁻⁷ Experiments were carried out in order to determine the effects of milling time and type of milling action encountered. In particular, two batches of powder consisting of 65 percent steel and 35 percent titanium carbide were ball milled in the drum milled previously mentioned. One of them was milled for five hours and the other for 24 hours. By comparing Figs. 5 and 4 it can be said that there is no difference in the morphology of the titanium carbide phase dispersed in the steel matrix. Another experiment was

carried out in order to determine if the type of milling action had some effect on the microstructure of the titanium carbide phase. In particular, a batch of powder consisting of 65 percent by volume of steel and 35 percent by volume of titanium carbide was milled in a planetary mill^{See App. II, Fig. 4} for five hours. Then a sample was prepared and sintered at the same temperature and time as the two previously mentioned experiments. No difference in the morphology of the dispersed phase was observed.

A series of experiments were carried out in order to determine the effect of volume percent of titanium carbide in obtaining a dispersed phase. Earlier experiments had shown that by varying the volume percent of titanium carbide in the matrix a continuous network of titanium carbide would appear. However, these results were erroneous because sintering had taken place at a low temperature and short period of time. Once the optimum sintering conditions had been determined the experiments were repeated and it was found out that the volume percent of titanium carbide did not have an effect on the morphology of the dispersed phase. In particular, a batch of powder consisting of 58 percent by volume of steel and 42 percent of volume of titanium carbide was sintered at 1469°C for 2 hours. There was no continuous network of titanium carbide present. See Fig. 6, Appendix III for additional details.

The use of prealloyed powders as a potential binder for the titanium carbide was also investigated. A series of experiments were performed in order to determine the optimum sintering conditions. It was possible to obtain round dispersed grains of titanium carbide. No porosimetry tests were carried out with these samples. See Figures 8 and 9 for details in the microstructure and sintering conditions.

The use of naphthalene as a means of producing porosity was also investigated. In particular, samples consisting of 410L prealloyed powders and titanium carbide were compacted at 10000 psi and sintered at 1402°C for 1/2, 1, and 2 hours. Mercury porosimetry tests were carried out on these samples. The results are shown in Figs. 1-3 of Appendix IV. It can be seen that the level of porosity is in the order of 35 percent by volume. The average porous size is in the order of 4 microns.

Mercury porosimetry tests were also carried out on samples made out of elemental powders. In particular, a set of samples were sintered at 1410°C and 35 minutes and compacted at pressures of 10000 psi, 15000 psi, 20000 psi, and 30000 psi. By looking at Figs. 4-7 it can be seen that the level of porosity was very low and the average porous size small, around 0.6 microns. The porosity was of the order of 5 percent by volume.

Experiments were also carried out to determine the effect of molybdenum on the morphology of the dispersed titanium carbide phase. In particular, an alloy was prepared in which the steel did not contain any molybdenum. The steel consisted of 12 percent chromium, 0.03 percent carbon and 87.97 percent iron. The volume percentage of steel and titanium carbide was 65 percent-35 percent respectively. Samples were sintered at 1455°C for 2 hours. No difference was observed in the microstructure. A dispersed phase consisting of round titanium carbide grains was obtained.

4. Discussion and Conclusions

From the above experiments, it can be said that in order to assess the relative importance of sintering time and temperature, milling conditions, composition, sintering atmosphere and additives in the morphology

of the titanium carbide dispersed in a steel matrix, and in general of any dispersed phase in a suitable binder, the experiments should be carried out in a chronological order. If this is not the case, erroneous conclusions may be formulated. For instance it was first thought of producing a porous steel bonded titanium carbide in which the dispersed phase would exhibit round grains, finely dispersed. Within this context, it was decided to carry out the sintering of the specimens of low temperatures that is as close to the solidus line of the iron chromium - carbon steel as possible. This was done because it was thought that if the amount of liquid present during sintering could be reduced to a minimum, accordingly, the densification process would also be diminished, and hence the possibilities of increasing the porosity enhanced. Likewise, for the same reason it was also decided to sinter for short periods of time. Within this framework, round titanium carbides grains were not present, instead a continuous network of titanium carbide appeared. In an attempt to overcome this difficulty it was decided to look into ball milling procedures, effect of molybdenum and composition (percent by volume of the dispersed phase in the alloy). All this was done without further investigating the time-temperature relationship and its effect on the microstructure. Erroneous conclusions followed, for instance it was first thought that the amount of titanium carbide present would have some effect on obtaining the continuous network. It was later proved that this was not the case within the limits in which the majority of the experiments were carried out - (65% Vol steel 35% to carbide and 50% Vol steel - 50% Vol titanium carbide.).

Additional light was thrown into the problem of obtaining round titanium carbide grains in the steel matrix when it was decided to go in

the opposite direction, that is, to obtain a material of theoretical density. In particular, in this case the sintering temperature and sintering time were increased. This followed from the fact that the higher the temperature, the larger the amount of liquid phase present and hence the larger the densification rates. Likewise, for the same reason it was decided to increase the sintering time. As a result, round titanium carbide grains were obtainable. Hence, it can be said that the reason for obtaining a more or less continuous network of titanium carbide in the first place was due to insufficient sintering temperature and time, and not to amount of titanium carbide present. Once the appropriate sintering time and temperature had been established, the other variables such as ball milling, composition, effect of molybdenum on the titanium carbide grains were re-evaluated and the results mentioned in the previous section followed.

It is interesting to point out that this development follows closely what has been previously observed in the Tungsten Carbide - Cobalt System. In particular....." it has been supposed that during milling of the hard metal powder the cobalt grains partly cohere. These agglomerates remain unchanged at sintering temperatures below the melting point (about 1300°C with dissolve carbon and tungsten carbide.). At temperatures above the melting point of the auxiliary metal, the cobalt becomes liquid and fills the cavities between the tungsten crystals, breaking up the agglomerates."⁶ This same behavior has been observed with the Iron-chromium - Molybdenum carbon and Titanium carbide system. When sintering took place at temperatures of 1370°C, 1396°C and for periods of 30 minutes, the steel appeared as conglomerates in a more or less continuous structure of titanium carbide. However, when the sintering temperature is increased to 1420°C, the steel

conglomerates are no longer present and the titanium carbide begins to form a dispersed phase.

During the experimental development the powders were ball milled for 24 hours. The reasons for preferring 24 hours vs. shorter periods of time are as follows: It has been reported:...." It is clear that the milling time has a profound effect on the properties of the hard metal. Since the shrinkage and behavior during sintering are determined by the cobalt distribution, the porosity, hardness and durability of the cutting edge are functions of the fine degree of comminution of the cobalt, which is reached only after milling for 70 to 100 hours.⁷ Similarly it has also been reported:...." Increased milling time reduces the porosity. There is a greater difference in the results after 25 hours and 96 hours milling time than between milling periods of 96 hours to 200 hours.⁸ Even though these studies have been carried out with the cobalt-tungsten carbide, they may apply to the system under investigation.

In short, it can be said that longer milling periods are preferred because of the reduced porosity and better properties of the material.

Even though no wear tests have been performed with the system under study, it is safe to assume that the higher the content of titanium carbide in the matrix, the more wear resistant the material will be. In particular, the following has been established:...." It may be assumed that with a bond content of over 10%, the tungsten carbide grains can no longer form a completely connected skeleton, because the cobalt completely envelopes a greater part of the carbide structure. These particles were set free by dissolving away of the cobalt, whereby the observed desintegration of the block was caused.... With a cobalt content of over five percent, the wear

rates increases rapidly and over 10% reaches a very high value."⁹ This could also be true of the steel-titanium carbide system under investigation. Even though most of the experimental work has been done with a composition in which the titanium carbide was 35% by volume of the alloy, it can be said that higher volume percentage of titanium carbide would be preferable in order to obtain optimum wear resistant properties. Likewise, it has also been mentioned that the composition has no effect in obtaining a dispersed phase of titanium carbide. This is true within reasonable limits, because if the volume percentage of the titanium carbide is much greater than the steel, chances of obtaining a continuous network of titanium carbide are enhanced.

The fact that no effect was observed in the morphology of titanium carbide when molybdenum was added to the steel can be explained as follows:

The overall amount of molybdenum, 4.5% by weight of steel, is small for any noticeable effect to be present. However, it has been reported¹⁰ that when molybdenum is added to a compound consisting of nickel and titanium carbide in amounts of 18 to 20% by weight, round titanium carbide molybdenum is present.

Even though porosity values up to 35% by volume have been obtained with the use of naphthalene, its use is not recommended because of its non-uniform characteristics. In particular, the specimen shows large holes which may be products of entrapped gases. These large holes will have a detrimental effect on the mechanical properties of the material. Likewise, naphthalene is not nearly as good as a die lubricant as paraffin wax. This causes serious problems during ejection of the specimen from the die. The large amount of friction between the specimen and the die walls, sometimes creates cracks on the surface of the specimen.

5. Summary

The research experience can be summarized as follows.

1. Sintering temperature and time are of utmost importance in obtaining a finely dispersed titanium carbide phase in the steel matrix.
2. Within the volume fraction with which the experiments were carried out, the volume percent of titanium carbide does not have an effect in obtaining round titanium carbide grains.
3. Longer periods of ball milling are preferred, because of reduced porosity and better mechanical properties.
4. Wet milling is preferred because it is more satisfactory with soft malleable substances.
5. Molybdenum does not have any effect in obtaining round titanium carbide grains. It may, however, have some effects on the heat treatability of the material.
6. 410L steel prealloyed powders may be used as a potential binder for the titanium carbide.
7. The use of naphthalene as a means of producing porosity should be considered carefully.

6. Suggested Areas of Further Research

The use of prealloyed powders as a binder for the titanium carbide can be explored a little bit further. In particular, the morphology of the titanium carbide in the 410L prealloyed powders when sintering took place at 1450°C for 2 hrs has been shown in Fig. 9 Appendix III.

However the morphology of titanium carbide in 410L steel prealloyed powder when sintering takes place for shorter periods of time has not been observed. It may be possible that titanium carbide could appear as round dispersed phase in the 410L steel prealloyed powders when sintering takes place for shorter periods of time. If this is the case, then there could be an advantage in using 410L steel prealloyed powder as a binder instead of using elemental powders.

ACKNOWLEDGMENTS

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BIBLIOGRAPHY

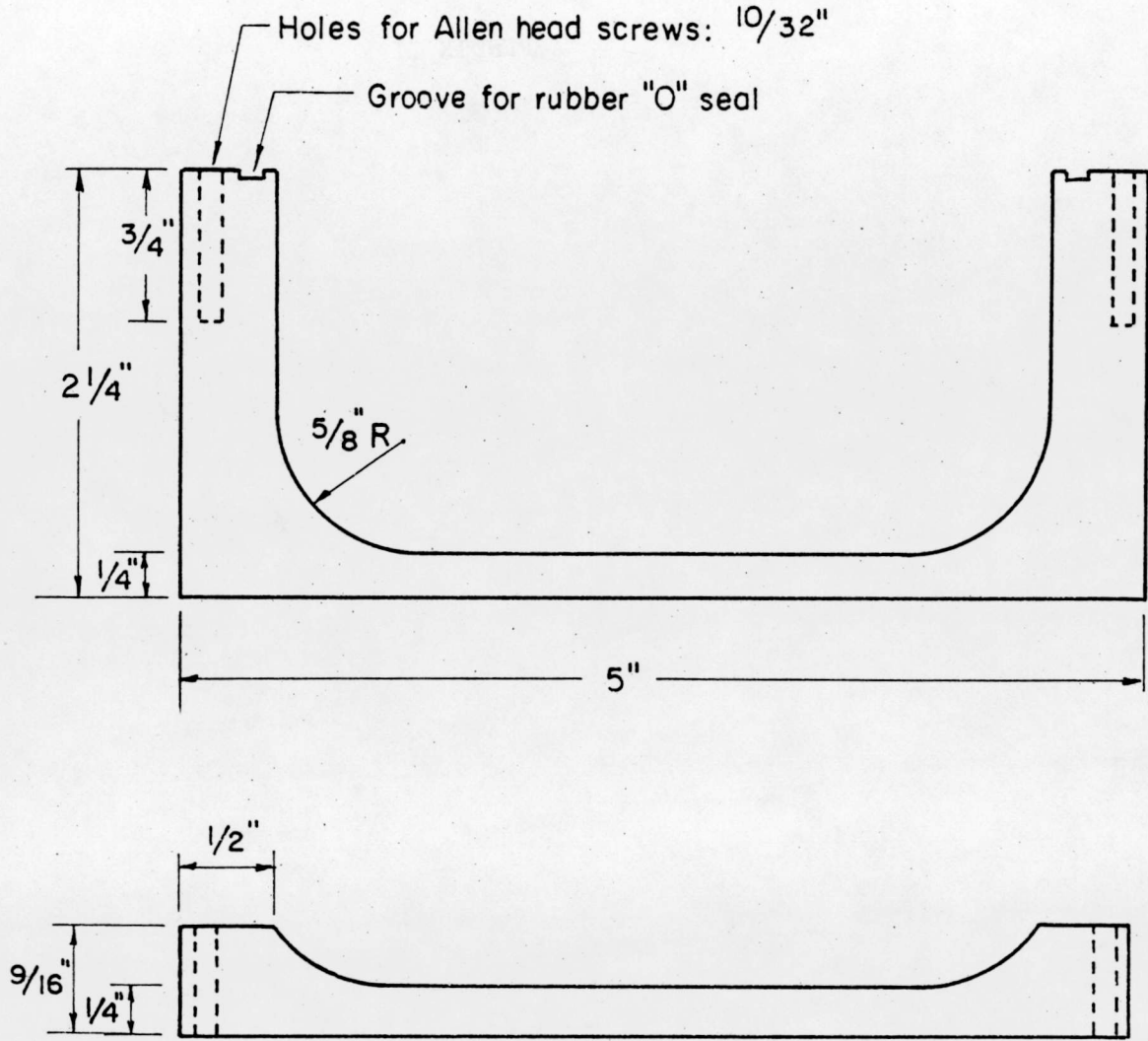
1. Cemented Titanium Carbide Cutting Tools. D. Moskowitz and M. Humenick. Modern Developments in Powder Metallurgy Vol. 3, p 83. Plenum Press, New York - 1966.
2. Development of Dispersion-Hardened Cobalt-base cutting-tool Alloys from atomized powders. O. W. Reen. Ibid, Vol. 2, p. 182.
3. Practical Application of New Carbide Technology. Kalish, H.S. The International Journal of Powder Metallurgy and Powder Technology, 10, No. 4, April, 1974.
4. "Steel bonded Carbides as Engineering Materials". Sintercast Division, Chromalloy American Corporation, West Nyack, NY 10994.
5. Mechanisms and Microstructural Aspects of Liquid Phase Sintering. Whalon, T. J. and Humenick, Jr. Progress in Powder Metallurgy, 1962. Vol. 18, p. 85.
6. Handbook of Hard Metals, by W. Dawihl, p. 31. London, Her Majesty's Stationary Office, 1955.
7. Ibid - p. 62.
8. Ibid - p. 63.
9. Ibid - p. 67.
10. Cerments: I, Fundamental Concepts Related to Microstructure and Physical Properties of Cerment Systems, Journal of the American Ceramic Society Humenick and Parikh, Vol 39, No 2. pp. 60.
11. Warren, R. and Waldron, M. S. Microstructural Development During Liquid Phase Sintering of Cemented Carbides. I. Wettability and grain Contact Powder Metallurgy, 1972, Vol. 15, No. 30.

12. The production of submicron Metal Powders with Grinding Aids. Max Quatinetz, Robert J. Schafer and Charles Smeal. Transactions of the Metallurgical Society of AIME. Vol. 221, Dec. 1961, p. 1105.
13. "The Use of Molybdenum in Titanium Carbide Base Cements". Climax Molybdenum Company, 1270 Avenue of the Americas, New York, NY 10020.
14. Some properties and Metallography of Steel-Bonded Titanium Carbide. Martin Epner, Eric Gregory. Transactions of the Metallurgical Society of AIME.
15. Densification Processes in the Tungsten Carbide - Cobalt System. R. F. Snowball and D. R. Milner. Powder Metallurgy, 1968, Vol. 11 - No. 21.
16. Cermets II, Wettability and Microstructure Studies in Liquid-Phase Sintering. Journal of the American Ceramic Society, Parikh and Humenick, Vol 40 - No. 9 pp. 315.
17. A Study of the Microstructure of Titanium Carbide, Journal of Metals - February 1955; Transactions AIME. Herman Blumenthal and Ronald Silverman.

APPENDIX I

- | | | |
|----------------------|--|----------------------------|
| 1. Iron Powder | Carbonyl, 4+6 μ 's. | 99.97% pure. |
| Supplier: | United Mineral & Chemical Corp.
129 Hudson Street
New York, NY 10013 | |
| 2. Chromium Powder | 325 Mesh. | 99.9% pure. |
| Supplier: | United Mineral & Chemical Corp.
129 Hudson Street
New York, NY 10013 | |
| 3. Molybdenum Powder | 3 μ 's | Stock #1179 |
| Supplier: | Cerac Inc. | |
| 4. Flake Graphite | | |
| Supplier: | The Joseph Dixon Crucible Co.
Jersey City - New Jersey. | |
| 5. Titanium Carbide | 325 Mesh (4+6 μ 's Aug.) | 99% pure.
Stock #T-1151 |
| Supplier: | Cerac Inc. | |

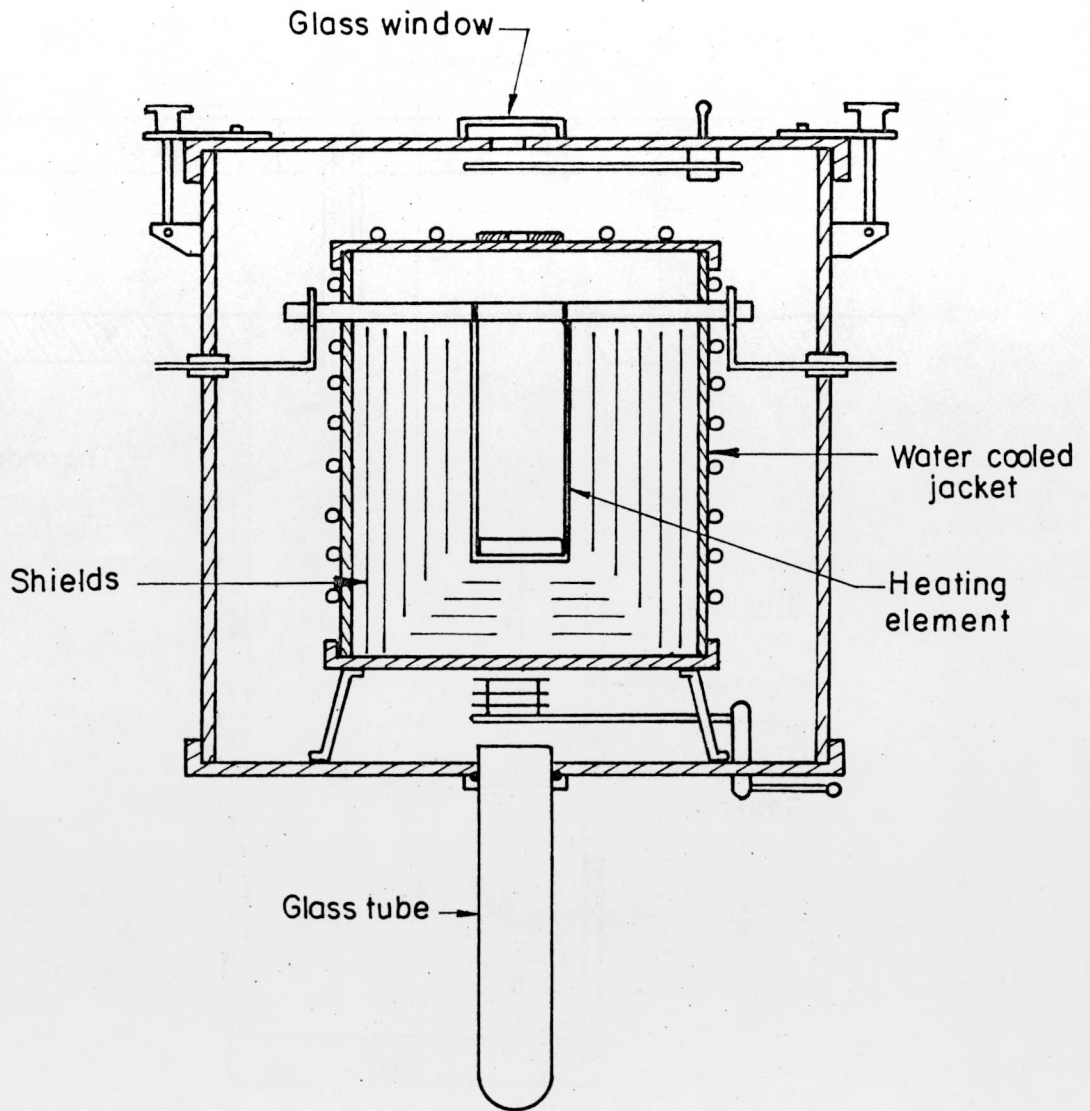
APPENDIX II



Drum Mill

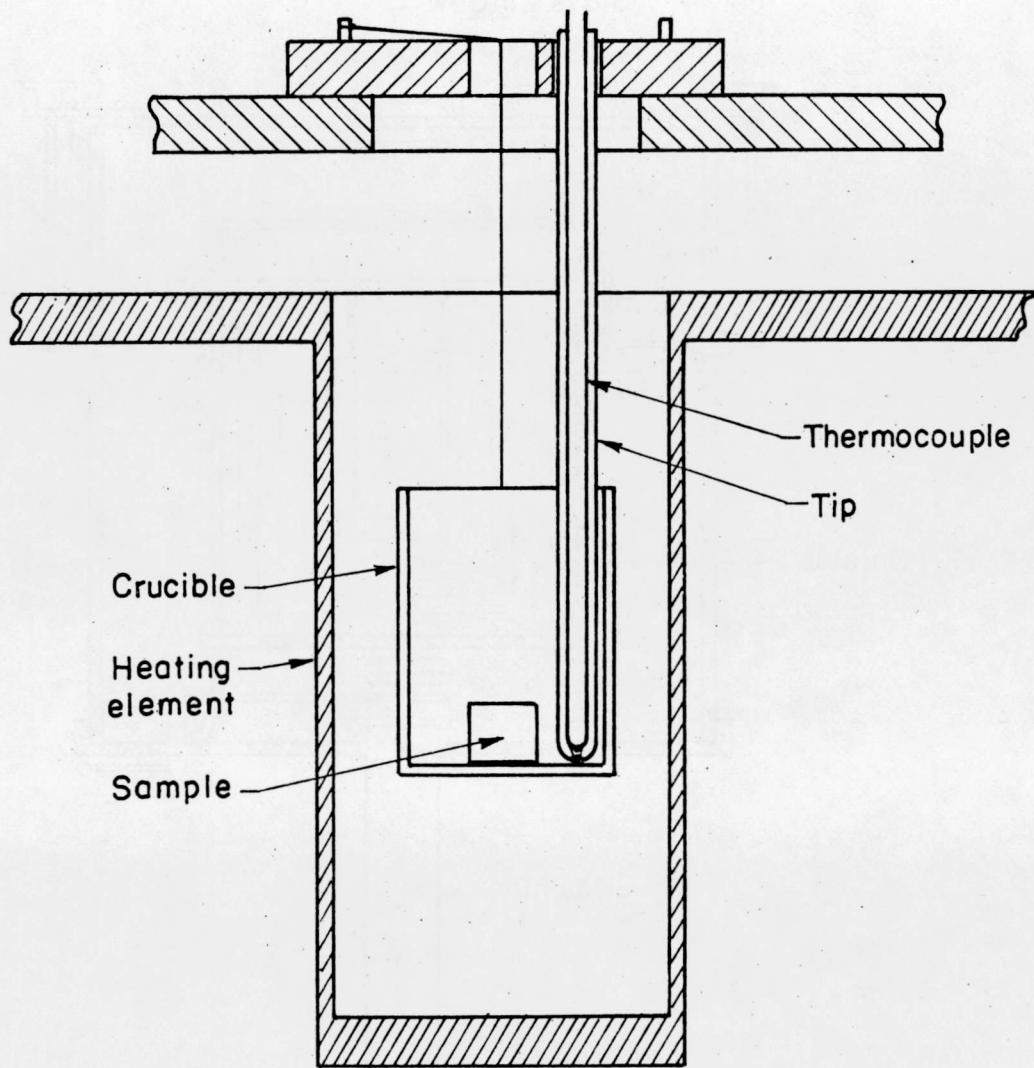
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Fig. 1



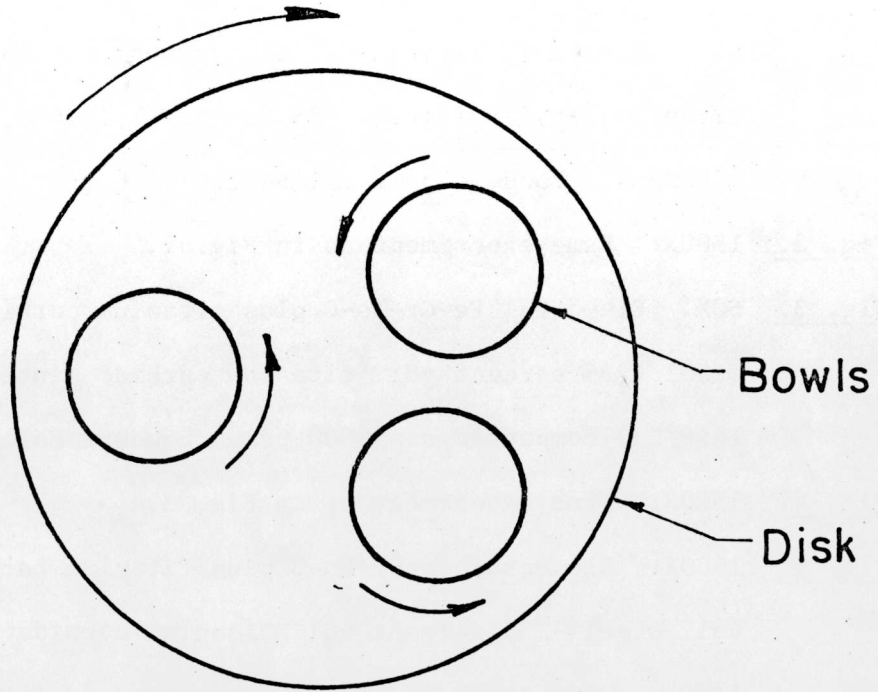
XBL 75 I-5513

Fig. 2



XBL 751-5514

Fig. 3



"Planetary Mill"

XBL 751-5515

Fig. 4

APPENDIX III

- Fig. 1. 60X. Elemental Fe-Cr-Mo-C, plus titanium carbide 65 percent vol. titanium carbide sinter: 35 minutes - 1416°C, compacted at 40000 psi. Powders ball milled for 24 hours.
- Fig. 2. 1500X. Same experiment as in Fig. 1.
- Fig. 3. 60X. Elemental Fe-Cr-Mo-C plus titanium carbide 65 percent vol. steel - 35 percent vol. titanium carbide sinter: 1.5 hours - 1448°C - compacted at 50000 psi. Powders ball milled for 24 hours.
- Fig. 4. 1500X. Same experiment as in Fig. 3.
- Fig. 5. 1500X. Elemental Fe-Cr-Mo-C plus titanium carbide 65 percent vol. steel - 35 percent vol. titanium carbide sinter: 1.5 hours - 1450°C. Powders ball milled for 5 hours.
- Fig. 6. 60X. Elemental Fe-Cr-C-Mo plus titanium carbide 58 percent vol. steel - 42 percent vol titanium carbide sinter: 2 hours - 1468°C - compacted of 70000 psi. Powders ball milled for 24 hours.
- Fig. 7. 1500X. Same experiment as in Fig. 6.
- Fig. 8. 60X. Prealloyed 410L steel powders - plus titanium carbide. 65 percent by vol. steel - 35 percent by vol. titanium carbide. Sinter: 2 hours - 1450°C. Powders ball milled for 24 hours.
- Fig. 9. 1500X. Same experiment as in Fig. 8.

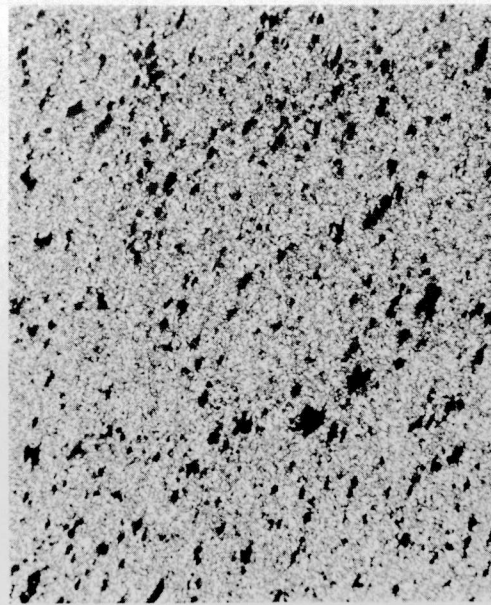


Fig. 1

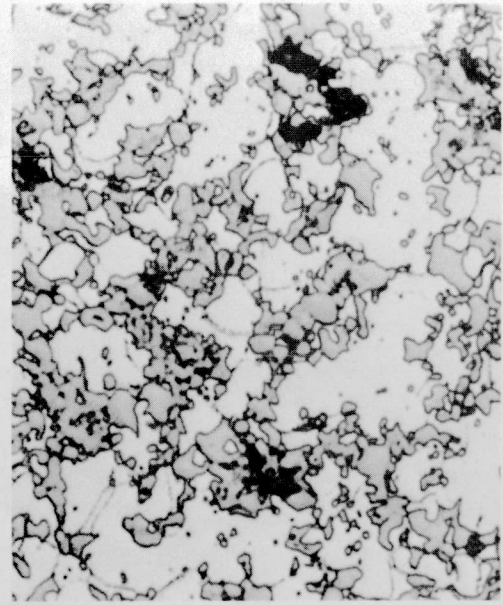


Fig. 2

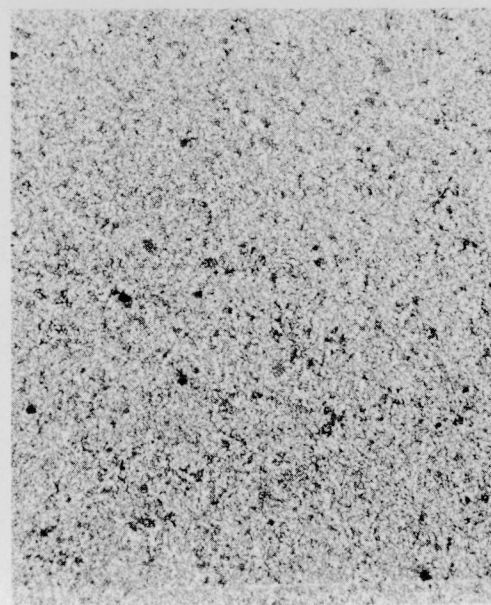
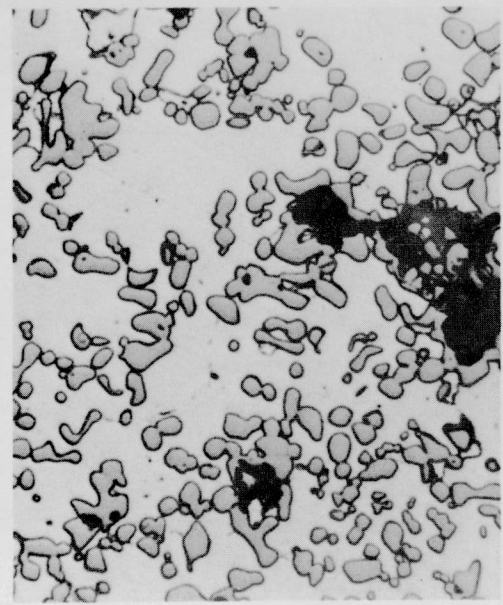
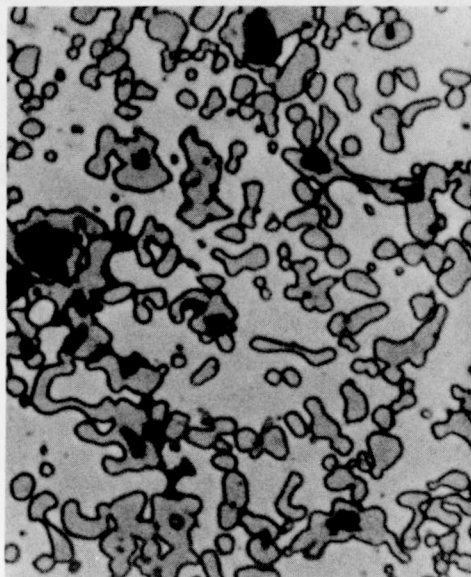


Fig. 3



XBB 751-29

Fig. 4



XBB 751-30

Fig. 5

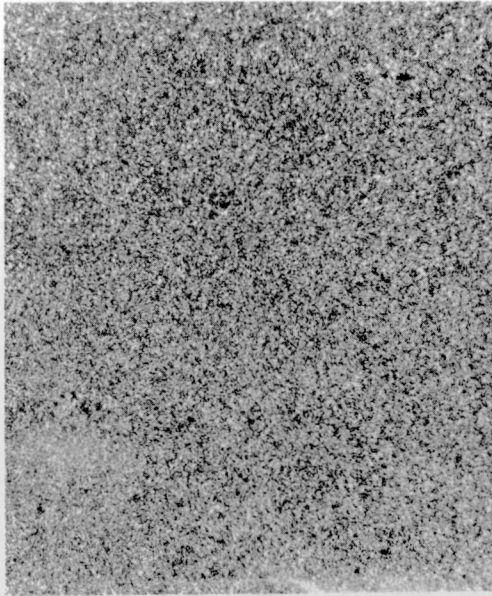


Fig. 6

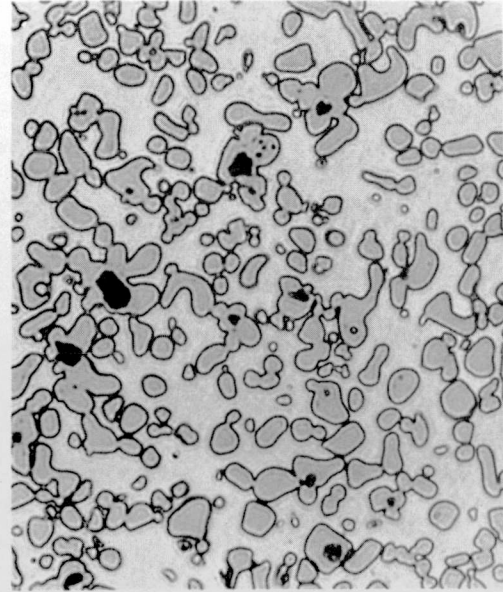


Fig. 7

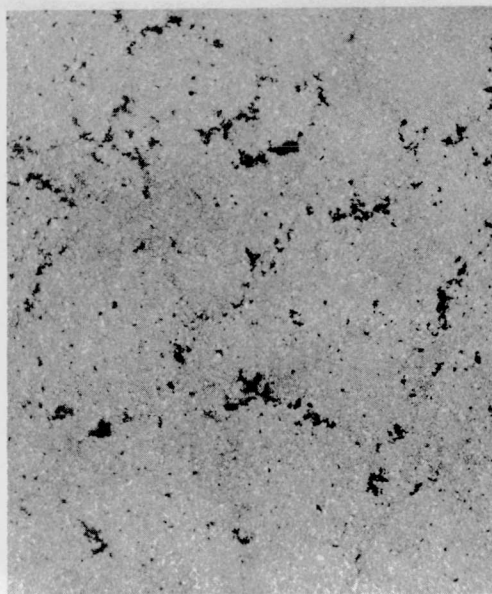
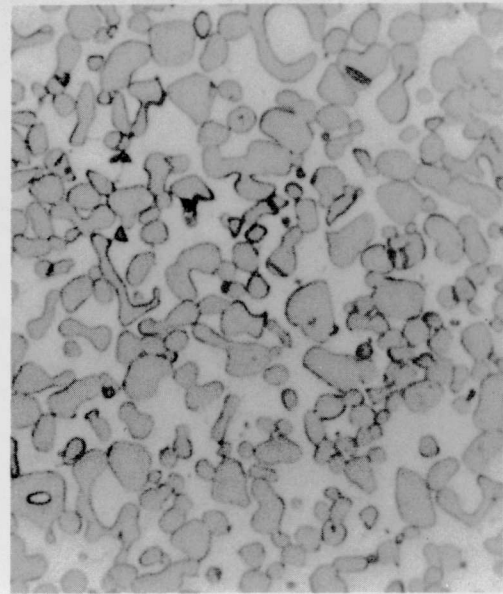


Fig. 8



XBB 751-28

Fig. 9

APPENDIX IV

See Figures 1. through 7.

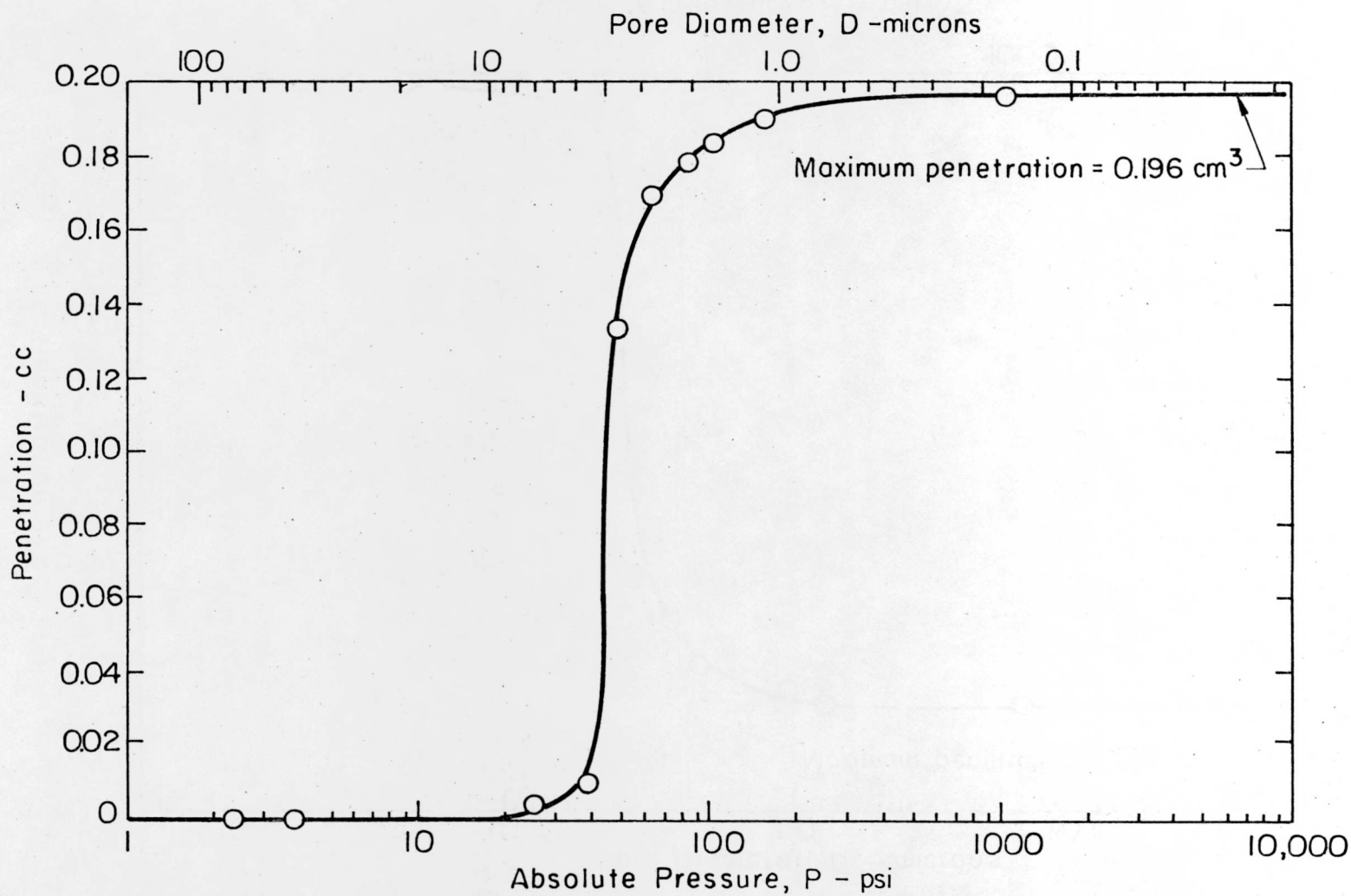


Fig. 1. Sintered 1/2 hr -1402°C. Compacted 10000 psi.

XBL 751-5518

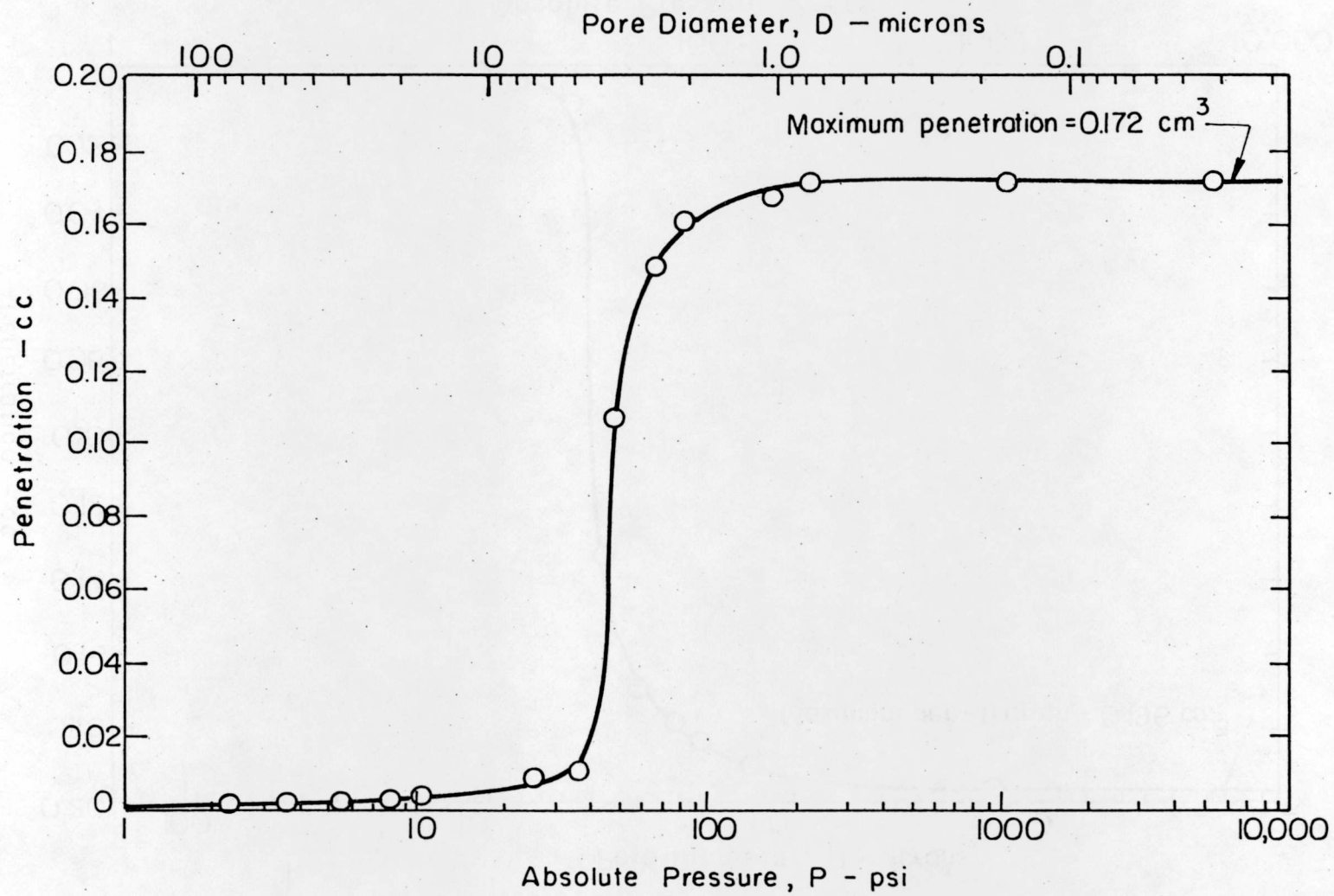
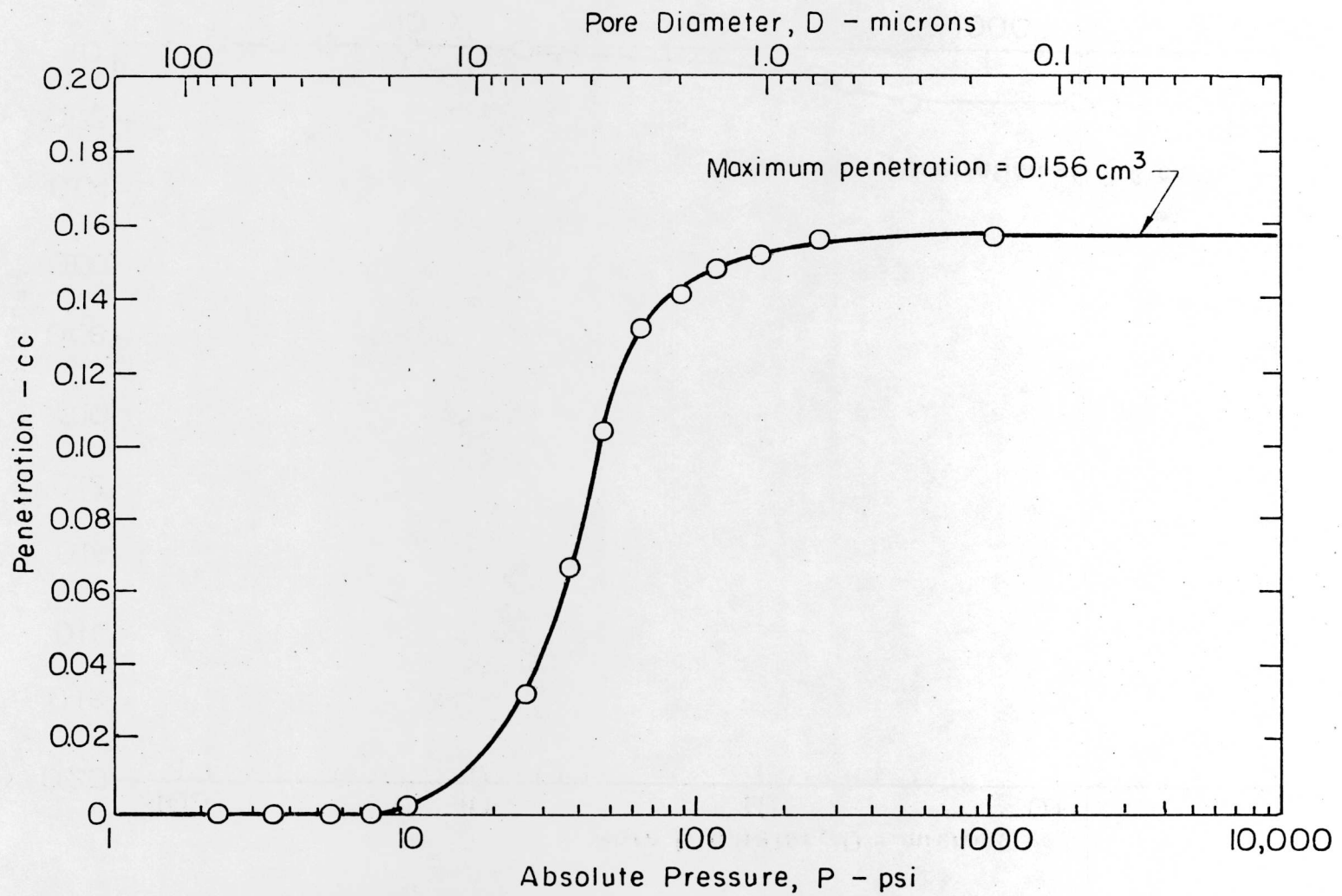


Fig. 2. Sinter 1 hr - 1402°C. Compacted 10000 psi.

XBL 75I-5516



XBL751-5517

Fig. 3. Sinter 1 1/2 hrs - 1402°C. Compacted 10000 psi.

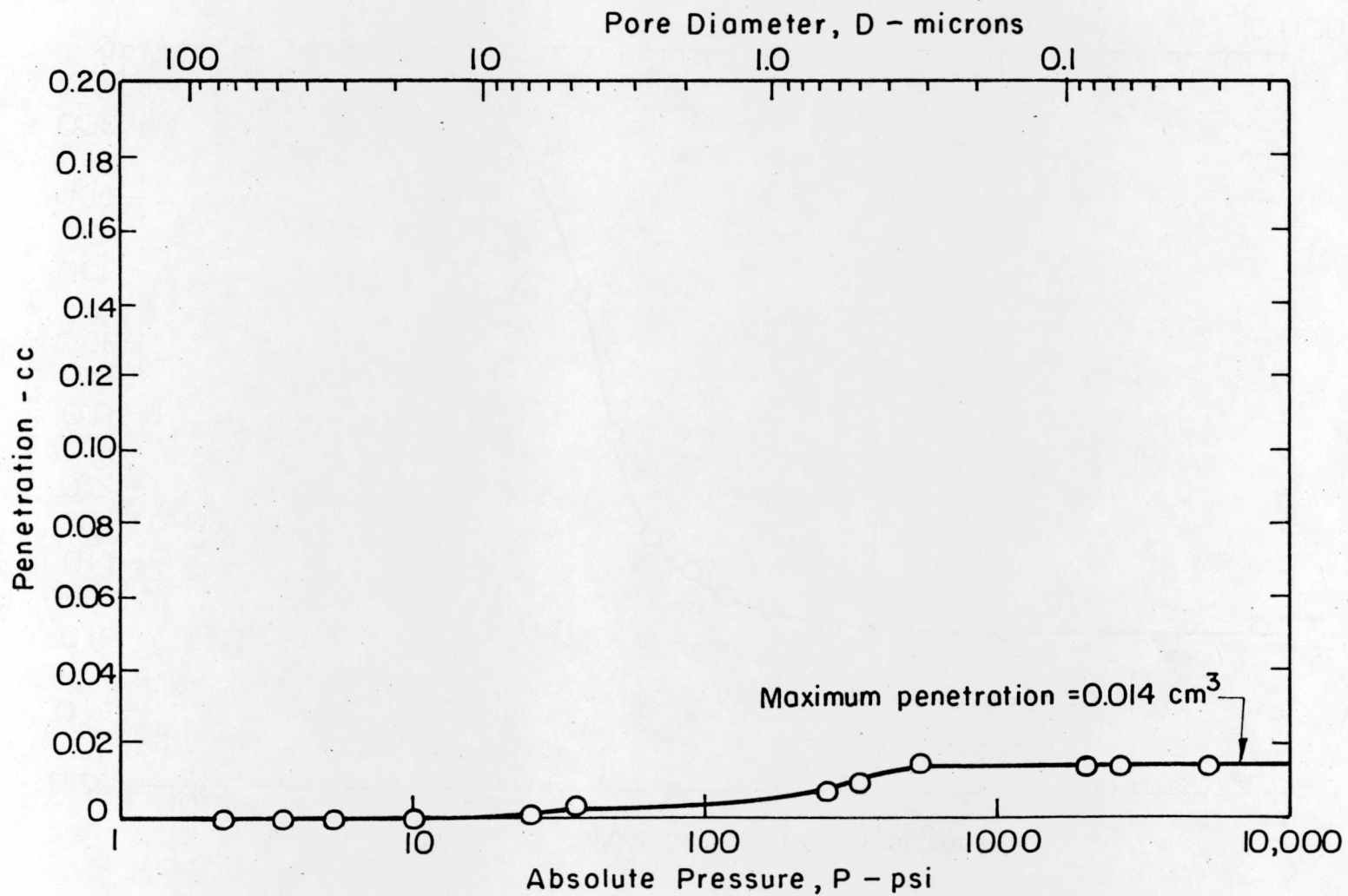


Fig. 4. Sinter 1410°C 35 m/s. Compacted 10000 psi.

XBL 751-5519

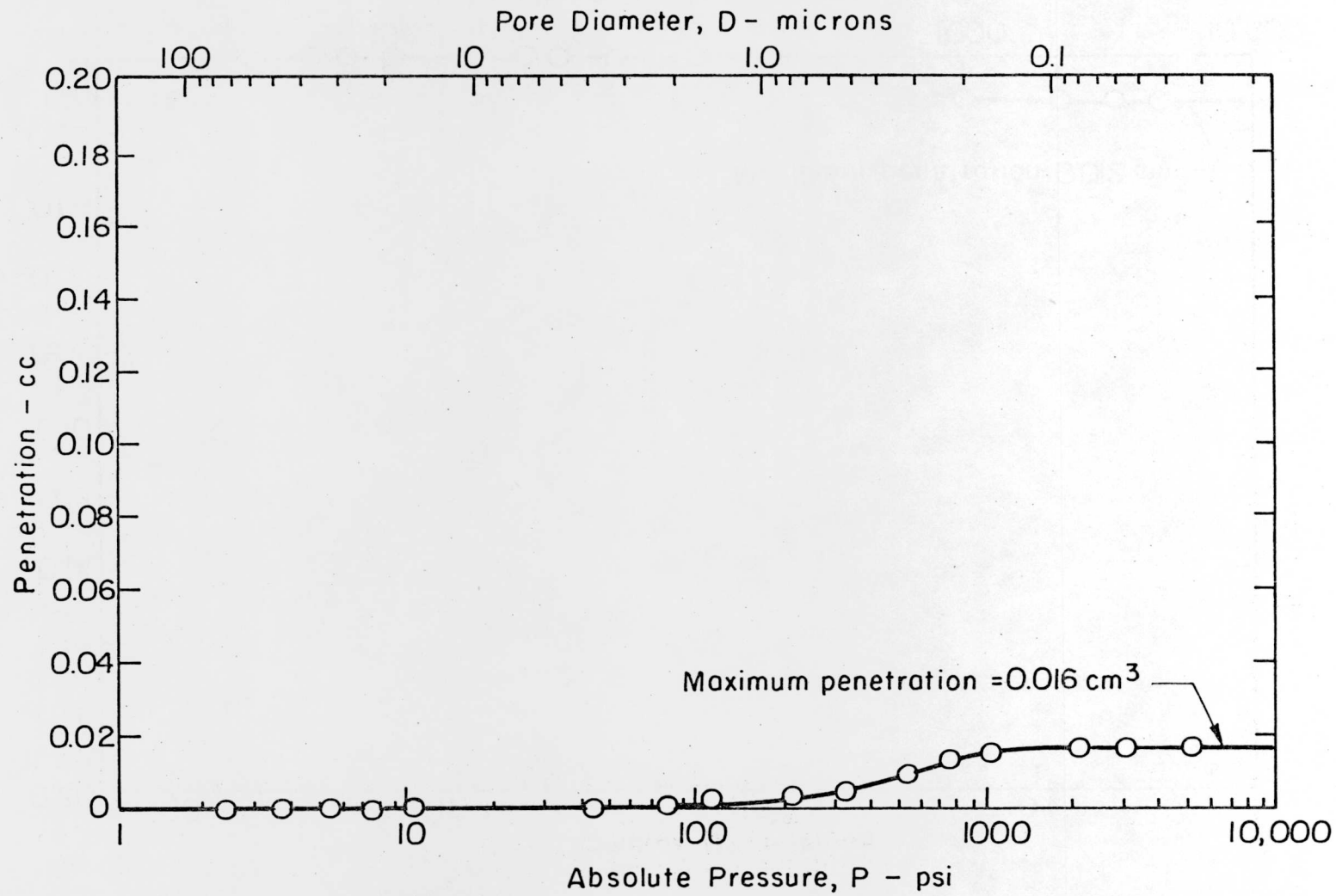


Fig. 5. Sinter 1410°C - 35 m/s. Compacted 15000 psi.

XBL 751-5520

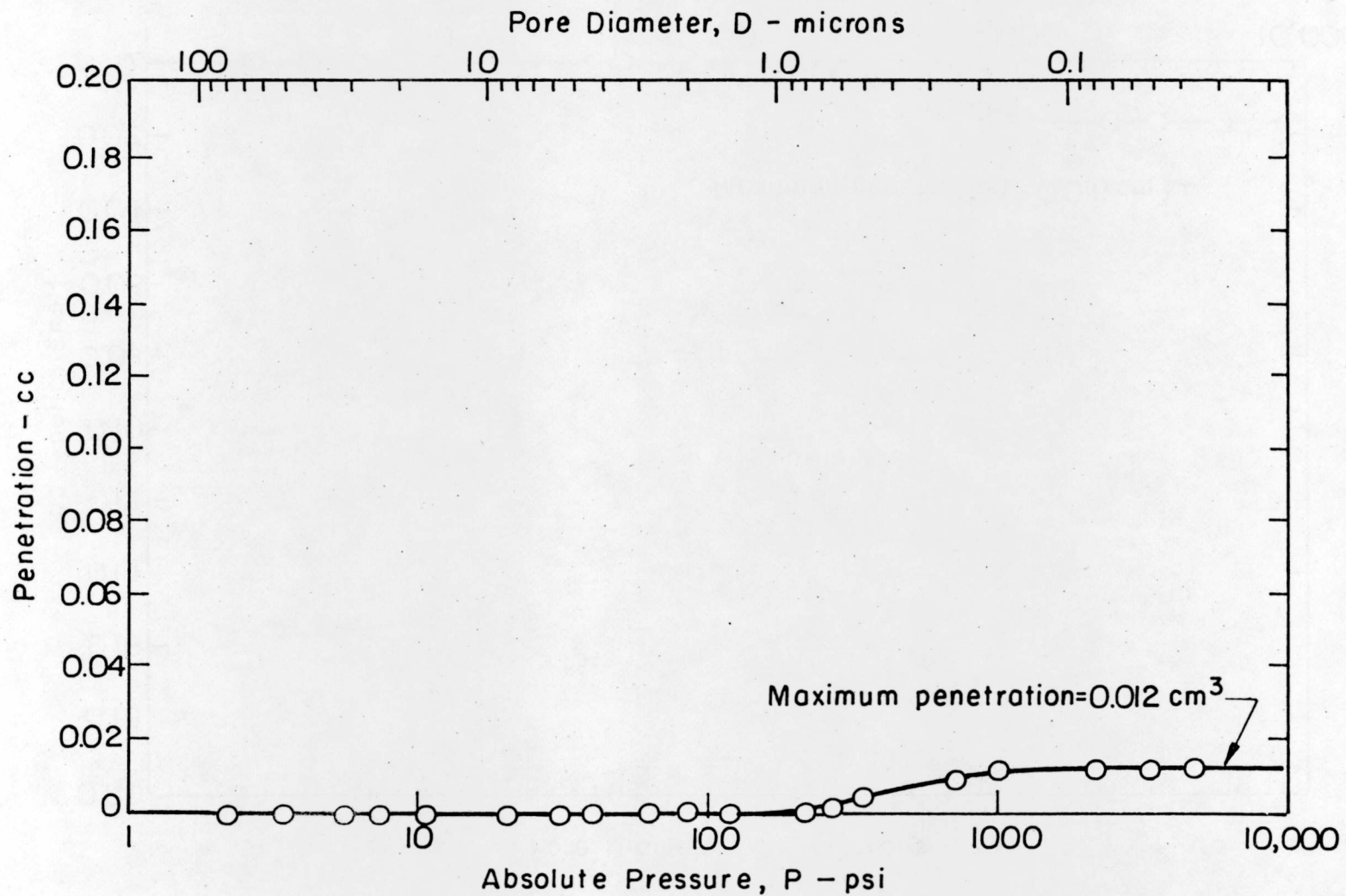
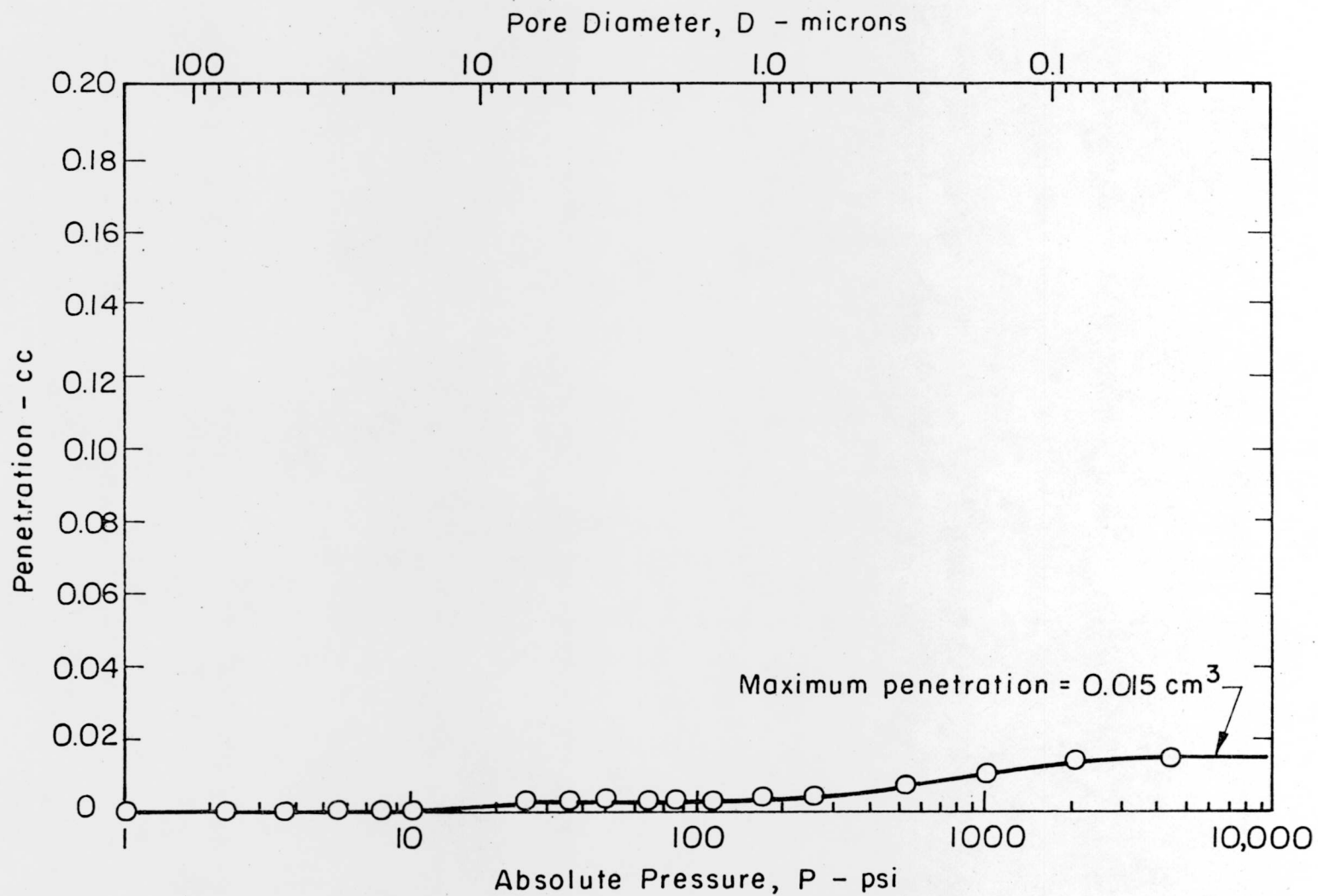


Fig. 6. Sinter 1410°C - 35 m/s. Compacted 20000 psi.

XBL 751-5521



XBL 751-5522

Fig. 7. Sinter 1410°C - 35 m/s. Compacted 30000 psi.

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