

MAR 4 1997

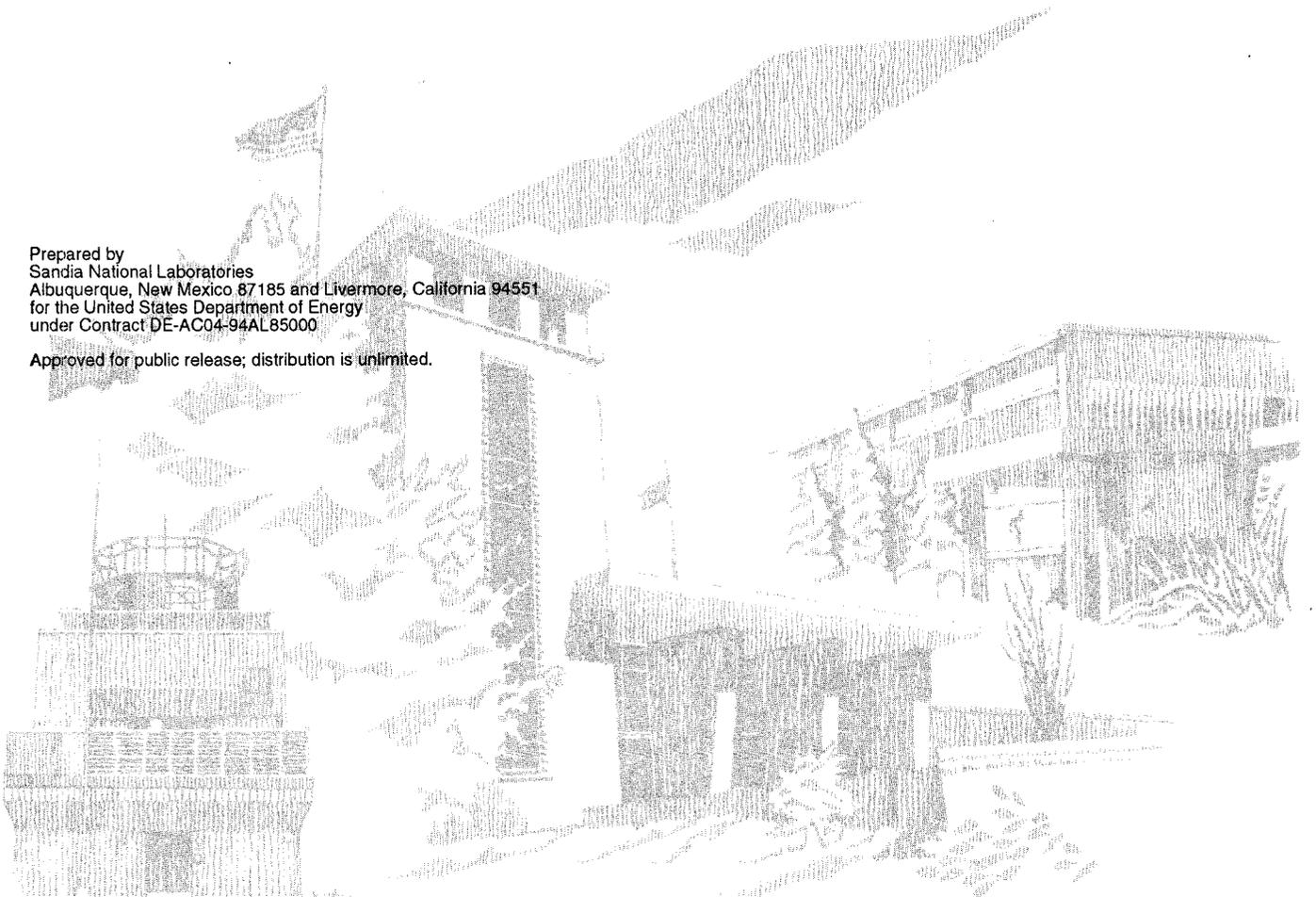
SAND97-8445 • UC-404
Unlimited Release
Printed January 1997

Aluminum-Assisted Joining of Beryllium to Copper for Fusion Applications

C.H. Cadden, and B.C. Odegard, Jr.

Prepared by
Sandia National Laboratories
Albuquerque, New Mexico 87185 and Livermore, California 94551
for the United States Department of Energy
under Contract DE-AC04-94AL85000

Approved for public release; distribution is unlimited.



Issued by Sandia National Laboratories, operated for the United States Department of Energy by Sandia Corporation.

NOTICE: This report was prepared as an account of work sponsored by an agency of the United States Government. Neither the United States Government nor any agency thereof, nor any of their employees, nor any of the contractors, subcontractors, or their employees, makes any warranty, express or implied, or assumes any legal liability or responsibility for the accuracy, completeness, or usefulness of any information, apparatus, product, or process disclosed, or represents that its use would not infringe privately owned rights. Reference herein to any specific commercial product, process, or service by trade name, trademark, manufacturer, or otherwise, does not necessarily constitute or imply its endorsement, recommendation, or favoring by the United States Government, any agency thereof or any of their contractors or subcontractors. The views and opinions expressed herein do not necessarily state or reflect those of the United States Government, any agency thereof, or any of their contractors or subcontractors.

ALUMINUM-ASSISTED JOINING OF BERYLLIUM TO COPPER FOR FUSION APPLICATIONS

C. H. Cadden
Advanced Manufacturing Technologies for Metals Processing Department

and

B.C. Odegard, Jr.
Materials Reliability Department
Sandia National Laboratories/California

ABSTRACT

Five different brazing techniques were evaluated in the process of joining beryllium to copper. Aluminum-based filler metals were used in conjunction with aluminum coatings on both beryllium and copper substrates. This innovative approach was born out of the necessity to inhibit the formation of oxides and intermetallics on the aluminum and beryllium surfaces both before and during the joining process. Several bonding techniques, diffusion barriers, and oxide inhibitors were employed to reduce the bonding problem to that of joining aluminum to aluminum. The volume of aluminum in the joint was found to be an important factor in reducing the segregation of secondary alloying elements at the beryllium interface. Plasma sprayed aluminum coatings were too porous to use in the as-sprayed condition and were further processed using a hot isostatic press (HIP) to accomplish full density. The use of plasma sprayed aluminum coatings, Al-12%Si filler metal (Alloy 718), and the HIP process produced excellent bonds between the aluminum coated beryllium and 1100-Al alloy plate which was explosively bonded to a copper alloy. Bond strengths were measured at 100% of the strength of the 1100-Al plate strength (90 MPa). The ductility of the aluminum bond was sufficient to produce extensive necking prior to fracture.

ALUMINUM-ASSISTED JOINING OF BERYLLIUM TO COPPER FOR FUSION APPLICATIONS

I. Introduction

Beryllium is a leading armor candidate for plasma facing components (PFC's) for the International Thermonuclear Experimental Reactor (ITER). The armor material must be bonded to an actively cooled copper alloy heat sink in order to maintain proper surface temperatures in the 5-10 mm thick beryllium tiles. S-65C is currently the designated beryllium grade for this application. Three copper alloys are potential heat sink candidates: alumina dispersion strengthened copper (Cu-0.2Al₂O₃), CuCrZr (Cu-0.65Cr-0.08Zr) and CuNiBe (Cu-1.8Ni-0.4Be). A successful bonding process must produce a metallurgical joint which permits effective heat transfer while providing adequate mechanical strength.

The single most important characteristic of beryllium from a joining standpoint is its high reactivity. Beryllium forms stable compounds with almost every element in the Periodic Table. These compounds are typically both strong and brittle; their presence in a bond joint can reduce mechanical performance to unacceptably low levels of ductility and toughness. Also, beryllium readily oxidizes in low vacuum pressures of 10⁻⁵ torr. Therefore, in order to form sound metallurgical bonds, a beryllium bonding surface must first be cleaned to remove surface contamination and kept clean throughout all handling and processing steps until the bonding process is complete.

Several elements are compatible with beryllium, that is, do not form beryllides. Among those are germanium, silicon, silver and aluminum. Silver has been successfully used by a number of investigators to join beryllium to copper [1-3] by both brazing and diffusion bonding. However, concerns about activation and transmutation products in the ITER neutron flux environment have led to the decision that silver is unacceptable for plasma facing components (PFC) [4]. This report documents the results of efforts, conducted at Sandia National Laboratories, aimed at finding alternative, silver-free joining processes.

II. Approach

Copper and beryllium form a number of intermetallic compounds which are stable to temperatures in excess of 900°C [4]. Direct bonding of beryllium to copper at temperatures as low as 350-400°C is sufficient to form measurable thicknesses (> 0.1 micron) of the compounds BeCu and Be₂Cu after exposure times of 1 hour [5-6]. Two approaches have been used to circumvent this problem. The incorporation of diffusion barriers [7] can isolate the copper from the beryllium, allowing higher bonding temperatures to be used. Other researchers [3] have used brazing cycles which employ rapid heating and cooling rates to limit the time of exposure at elevated temperatures.

Since aluminum is compatible with beryllium, an approach was taken in the current study to use aluminum as a transitional layer between the beryllium and copper. Aluminum has reasonable thermal conductivity (about 2.4 W/m²·K at room temperature vs. 2.9 W/m²·K for the copper heat sink materials) and good ductility and should act as a compliant layer to absorb most of the thermal stresses generated by the thermal expansion gradient between beryllium and copper. Also, aluminum is routinely explosively bonded to copper, resulting in a high strength joint. For applications requiring high temperature operation, a thin diffusion barrier is

bonded in place between the copper and aluminum to prevent Cu-Al intermetallic phase formation. The barrier material used is typically titanium or tantalum; with the barrier in place the joint is reportedly stable at temperatures near the melting point of aluminum (660°C) [8]. Earlier studies [9] had shown that during the brazing of beryllium to aluminum, solute elements from the filler metal segregate to the beryllium interface during solidification. A preferable microstructure would be one where solute species are more uniformly distributed throughout the braze joint. In the present investigation, beryllium specimens were coated with aluminum, thereby simplifying the bonding task to that of joining two aluminum surfaces.

III. Experimental Procedure

Two explosively bonded aluminum to copper plates were procured from Northwest Technical Industries, Sequim, WA. These plates consisted of a thick copper alloy plate (19 mm) bonded to a thinner 1100 aluminum plate (7 mm) separated by a thin titanium sheet (0.25 mm) which acts as a diffusion barrier; specific details pertaining to the two plates are given in Table 1. Both plates appeared to be well bonded; a partial cross-section of one of the plates is shown in Figure 1. S-65C grade beryllium disks measuring 48 mm in diameter by 5.0 mm in thickness were obtained from Brush Wellman Inc, Cleveland, OH. Both sides of each beryllium disk were subsequently coated with pure aluminum by either of two processes. Three disks were coated at Los Alamos National Laboratory using a vacuum plasma spray process. After first removing surface oxides using a reverse transferred arc cleaning technique [10], a layer of aluminum was deposited on each side. Two other beryllium disks were sputter etched and then coated by Surmet Corporation (Burlington, MA) with 25 microns of aluminum (per side) using a magnetron plasma vapor deposition (PVD) process. Following the aluminum coating, one of the Surmet disks was also coated with a one micron thick layer of silicon.

Table 1 - Explosion bonded plates used for Be-Al joining tests

Cu alloy	Diffusion Barrier	Aluminum alloy
19 mm CuCrZr	1.0 mm Ti	9.5 mm Al 1100
25 mm CuBeNi	0.25 mm Ti	9.5 mm Al 1100

Five different bonding strategies were employed. Bond specimens were fabricated according to the configuration shown in Figure 2 in order to facilitate the machining of specimens for subsequent mechanical property testing. The details of each joining process are described below and summarized in Table 2.

Table 2 - Braze Assemblies and Schedules

Specimen Designation	Al Coating on Be	Bonding Agent	Bonding Temp (°C)	Bonding Time (min)	Bonding Pressure (MPa)
A	0.025 mm PVD	0.25 mm Al-12Si foil	660	3	0.07
B	0.30 mm Plasma Spray	0.25 mm Al-12Si foil	660	3	0.07
C	0.30 mm Plasma Spray	0.25 mm Al-12Si foil	625	15	103
D	0.30 mm Plasma Spray	0.010 mm Electroplated Cu	625	60	103
E	0.025 mm PVD	0.001 mm PVD Si	625	60	103

Vacuum Brazing

Two specimens were fabricated using a vacuum furnace brazing process. An aluminum-silicon eutectic composition (12%Si) filler metal with a melting point of 577°C was used. Cubes measuring 30 mm x 30 mm x 19 mm thickness were removed from the CuCrZr/Ti/Al explosion bonded plate. The aluminum layer was machined flat to a thickness of 1 mm.

Specimen A utilized a beryllium disk with a 25 micron thick coating of sputtered aluminum and Al-12%Si filler metal. Two pieces of braze foil measuring 35 mm x 35 mm x 125 microns thick were used for each of the two Al-Al joints. Prior to assembly, the aluminum bonding surfaces were immersed in an aqueous solution containing 45% HNO₃ and 5% HF to remove surface oxides. The CuCrZr/Ti/Al pieces were etched for 30 seconds; due to concerns about etching completely through the thinner Al coating on the Be disk, it was etched for only 10 seconds. Parts were assembled immediately after etching and placed in a vacuum furnace. A small weight was placed on top of the assembly to provide bonding pressure (0.07 MPa) during brazing. The all-metal vacuum furnace was pumped down to a pressure of less than 10⁻⁵ torr. The specimen was then heated at a rate of 25°C per minute to a peak temperature of 660°C. After holding at temperature for 3 minutes, power to the heating elements was interrupted and the specimen was allowed to cool in vacuum to ambient temperature. A thermocouple placed on one assembly indicated that the cooling rate was approximately 20°C per minute from the bonding temperature down to 400°C. Vacuum leak-up was measured at approximately 7x10⁻⁷ cm³ (STP)/sec.

A beryllium disk with the plasma sprayed aluminum coating was used for Specimen B; the aluminum thickness on this disk was determined to be 300 microns per side. This specimen was prepared identically to Specimen A with the single exception that the etching time for the aluminum coated beryllium disk was increased to 15 seconds. Alloy 718 braze foil was also

used for Specimen B as well as a brazing cycle (660°C/3 min) comparable to that used for Specimen A.

HIP Bonding

Three specimens were bonded using a hot isostatic press (HIP). For these trials, a cylindrical specimen geometry was used. Cylindrical pieces, 48 mm in diameter, were machined from the explosively bonded Cu/Ti/Al plates. Most of the aluminum layer was removed by machining, leaving a flat surface of approximately 0.8-1.2 mm thickness. Stainless steel cans and end caps were machined to dimensions required to permit encapsulation of the individual joint assemblies while maintaining a maximum clearance of 50-100 microns between the can and the bond specimen. After machining, HIP cans were vacuum outgassed at 700°C for one hour to minimize potential outgassing during the subsequent bonding cycle. Assembled HIP cans were sealed by electron beam welding of the end caps in a vacuum of 10⁻⁵ torr or better.

The specimen C configuration consisted of a duplication of Specimen B. Al-12%Si foil was again used as the bonding agent between the pairs of aluminum surfaces to be joined. Etching of bonding surfaces was also identical to the procedure used for Specimen B (15 seconds for coated Be, 30 seconds for Cu/Ti/Al). The HIP schedule consisted of a 15°C/min ramp to 625°C. Gas pressure was initially set at 50 MPa and was allowed to increase with temperature to provide a pressure of 103MPa at the peak temperature. After 15 minutes at 625°C, the specimen was allowed to free cool to ambient temperatures. A cooling rate of approximately 10°C/min was measured from the bonding temperature to 400°C.

Specimen D utilized a thin copper coating (< 10 microns), deposited on the aluminum surfaces, to eliminate the wetting problems associated with aluminum oxide. The copper reacts quickly with the underlying aluminum to form a low melting point eutectic on all bonding surfaces. The copper-aluminum eutectic temperature is 548°C, only slightly below the aluminum-silicon eutectic temperature of 577°C. Intermetallic formation between the copper and aluminum was minimized by keeping the volume of copper low. The solubility limit for copper in aluminum is extremely low. This trial used specimens machined from the CuBeNi explosion bonded plate and a beryllium disk with a plasma sprayed aluminum coating. After machining details as described for Specimen B, aluminum surfaces were given a caustic etch followed immediately by an electroplated copper treatment which was designed to produce a 10 micron thick copper coating. Components were again sealed in HIP tooling and bonded at 625°C and 103 MPa; bonding time was extended to 60 minutes.

A final specimen (Specimen E) was prepared using CuCrZr/Ti/Al details and a PVD aluminum coated beryllium disk. It has been observed that deposited silicon coatings oxidize at a much slower rate than aluminum layers deposited by the same PVD technique [12], possibly eliminating the need for etching of the coated beryllium disk prior to bonding. Therefore, the disk used in this experiment, in addition to the 25 micron coating of aluminum on each face, had been given a final coating of 1 micron of pure silicon on top of the aluminum. The Cu/Ti/Al components were given the "standard" 30 second etch previously described. The aluminum/silicon coated beryllium disk was not etched, but was rinsed in methanol immediately prior to assembly. Specimen E was encapsulated and bonded at 625°C and 103MPa for 60 minutes. Specimens D and E experienced heating and cooling rates similar to those listed for Specimen C.

After bonding, rectangular cross-section bars were removed from each specimen by electro-discharge machining (EDM). The 10 mm x 5 mm x 50 mm bars were oriented transverse to the bond joints; a reduced gauge section was also machined in the bars as shown in Figure 3. Duplicate tensile tests were conducted at room temperature, using a servo-hydraulic machine and a crosshead speed of 0.127 mm/min.

Cross-sections of bonded specimens were mounted and polished. After polishing, specimens were examined using optical microscopy and a JEOL 840-A scanning electron microscope equipped with energy dispersive x-ray spectroscopy (EDS). Etched samples were examined on the metallograph. Fracture surfaces of failed tensile specimens were also analyzed in the scanning electron microscope (SEM).

IV. Results and Discussion

Room temperature tensile test results are presented in Table 3. The vacuum brazed specimens (A and B) were found to have inferior tensile strength; the three HIP bonded specimens (C,D and E) were found to have much higher strength levels consistent with a reported strength of approximately 90 MPa for 1100-Al [11].

Test bars from Specimen A fell apart during machining. Failures occurred at or near the beryllium interface. Metallographic examination of intact areas revealed the presence of a band of aluminum adjacent to the beryllium, approximately 50 microns thick, which was comparatively free of second phase particles (Figure 4). Cracks were seen to extend from the beryllium interface into this band of "clean" material. Beyond this precipitate-free region, discrete second phase particles were found distributed fairly uniformly throughout the aluminum microstructure. These light appearing, iron-rich particles are distributed throughout the 1100-Al alloy. Their composition was not found to be affected by proximity to the braze joint. Higher magnifications (Figure 5) indicated the presence of a nearly continuous phase at the beryllium interface; this structure was found to be enriched in both silicon and iron. Cracking appears to be associated with the boundary between this second phase and the adjacent region of predominantly aluminum composition. No other regions of high silicon content could be found in these specimens.

During the brazing cycle, as the temperature rises above the aluminum-silicon eutectic temperature of 577°C, silicon diffuses rapidly from the molten filler metal through the adjacent aluminum substrates (1100-Al and sputtered aluminum coating). Since silicon does not diffuse past the beryllium interface, it seems probable that most of the sputtered aluminum coating was melted during the braze cycle. Upon cooling of the joint, solidification likely proceeds from the 1100-Al towards the beryllium. Since they both have a low solubility in aluminum, silicon and iron (an impurity in 1100-Al) are rejected into the liquid ahead of the advancing solidification front. The final liquid to solidify, enriched in both silicon and iron, thus becomes trapped between the aluminum alloy and the beryllium, resulting in the nearly continuous layer observed metallographically.

Table 3 - Room Temperature Tensile Results for Bonded Specimens

Specimen Designation	Al Coating on Be	Bonding Agent	Fracture Strength (MPa)	Failure Location
A	0.025 mm PVD	0.25 mm Al-12Si foil	* *	Be interface Be interface
B	0.30 mm Plasma Spray	0.25 mm Al-12Si foil	41.3 30.0	Plasma Sprayed Al Plasma Sprayed Al
C	0.30 mm Plasma Spray	0.25 mm Al-12Si foil	115.4 118.2	Plasma Sprayed Al Plasma Sprayed Al
D	0.30 mm Plasma Spray	0.010 mm Electroplated Cu	114.3 117.3	Plasma Sprayed Al Plasma Sprayed Al
E	0.025 mm PVD	0.001 mm PVD Si	83.3 121.6	Al/Be interface Al/Be interface

* - Test bars from Specimen A broke during machining

The significantly thicker layer of plasma sprayed aluminum used on the beryllium disk for Specimen B resulted in substantially different joint characteristics. Figure 6 indicates that even after a braze cycle near the melting point of aluminum, there remained some porosity in the plasma sprayed coating. The coating thickness was reduced only slightly from an original thickness of about 300 microns to approximately 250 microns thick following the brazing cycle. The bonded specimens possessed strength of about one half the expected level for wrought, annealed aluminum. The fracture surface shown in Figure 7 shows that despite breaking in the aluminum, large areas failed with little or no deformation and appeared intergranular. It seems likely that the fracture proceeded through the plasma sprayed structure, or perhaps at the transition between the plasma sprayed and the fully dense aluminum. Substantial concentrations of silicon were also found at and near the beryllium interface in this specimen. However, rather than a continuous layer, much of the second phase was found in the form of discrete particles at or near the beryllium boundary (Figure 8). Additionally, eutectic constituent silicon was also found distributed throughout the joint (dark phase indicated by arrow in Figure 6), suggesting that some volume of molten braze material filled residual porosity in the plasma sprayed layer. Note that the beryllium interface appears to be less planar than that seen in Specimen A; this effect is likely the result of surface scalloping caused by the transferred arc cleaning process which preceded the plasma spray deposition. These microstructures suggest that while molten material did reach the beryllium interface, possibly through interconnected porosity in the aluminum coating, much of the plasma sprayed layer remained intact, with the voids in the coating acting as sinks for silicon-rich liquid. Further evidence of this was seen in the extremely narrow band of solidified material at the interface between the plasma sprayed and the explosion bonded aluminum material.

Specimen C, the HIP'ed counterpart to Specimen B, possessed excellent room temperature tensile properties. In addition to good strength (115 MPa), tensile bars from this specimen exhibited extensive necking in the aluminum, resulting in a measurable elongation of approximately 15 percent. (Note that only about 6 percent of the gauge section length is aluminum; the actual elongation in the aluminum section is closer to 30 percent). Given this reduction in area, the true stress to fracture was significantly higher than the reported fracture strength. The fracture surface (Figure 9) was completely contained in the plasma sprayed aluminum layer and indicates a ductile failure mode. After the bonding cycle, discrete aluminum grains are clearly visible in the plasma sprayed layer (Figure 10). The remnant coating layer was reduced to barely one half the original thickness during bonding. Higher magnifications showed very little second phase precipitation at the beryllium interface (Figure 11), although there was more of the silicon and iron-rich phase at boundaries between the aluminum "grains". Additionally, these boundaries were decorated with dark phases which were found to contain both silicon and oxygen. Since gas pressure for the HIP cycle was applied at low temperatures, densification of the plasma sprayed coating likely occurred, at least partially, well before the melting point of the filler metal was reached. Consequently, it became difficult for molten material to reach the beryllium interface. At the same time, the high bonding pressures may have ejected molten material from the bond joint, contributing to the reduction in coating thickness observed. Note that the precipitation free band between the outer edge of the plasma sprayed coating and the 1100-Al was less than 20 microns thick, compared to the 45 micron layer seen in Specimen A. The combined effect of these factors was to produce a fully dense joint with silicon and iron impurities distributed fairly uniformly throughout the plasma sprayed coating. This microstructure is considered to be optimal for the selected material system.

Specimen D also produced superior tensile properties at room temperature. A distribution of fine particles was observed near the beryllium interface (Figure 12), although this second phase took the form of discrete particles and, for the most part, was situated a few microns from the boundary. These particles were found to be both iron and copper-rich. Again, no trace of copper was found in precipitates distributed in the 1100-Al structure. The plasma sprayed aluminum coating has few of the distinctive boundaries observed in Specimen C, suggesting that perhaps the entire layer was melted and resolidified. Solidification of this layer, from the 1100-Al towards the beryllium, would also be consistent with the observed concentration of oxygen at the Be interface. The coating layer was found to be only about 125 microns thick after bonding, representing a large reduction in thickness from the original 300 microns. Some fine, dark appearing structures were observed near the transition between 1100-Al and the resolidified coating; composition of these features could not be determined by EDS analysis. Failure of the tensile bars occurred through the plasma sprayed layer, but with less ductility than observed for the Specimen C test bars.

Specimen E, which utilized the thin silicon coating on top of the 25 micron aluminum layer, produced specimens possessing good tensile strength. However, failures were associated with the interface between beryllium and aluminum. Fracture surfaces (Figures 13 and 14) show evidence of limited ductility with areas of failure in the beryllium. Cross-sections of the joints (Figures 15 and 16) revealed a bond region with some second phase particles. Isolated iron- and silicon-rich particles were found at the beryllium interface, but microcracking was often associated with these precipitates. It appears that the thin aluminum PVD coating was penetrated, at least partially, resulting in segregation of solute elements to the beryllium boundary. With less silicon present in this system compared to other specimens (one micron thick coating versus 12 percent of a 250 micron braze foil), much less silicon found its way to the beryllium interface despite the thin aluminum coating. However, microcracking, presumably from the stresses generated during cooling from the bonding temperature, coupled with brittle phases at the beryllium interface resulted in failure initiation at this location.

V. Conclusions

1. The use of aluminum as an intermediate layer facilitates the joining of beryllium to copper. The results of this study indicate that after coating both the copper and beryllium substrates with thin layers of aluminum, the components can be successfully joined by several techniques.
2. The approach of explosively bonding aluminum to copper, coupled with the use of a thin titanium diffusion barrier, results in a metallurgically stable system which can be subsequently heated to temperatures near the melting point of aluminum for up to one hour without experiencing detrimental metallurgical effects.
3. Attempts to vacuum furnace braze using aluminum coatings and Al-12%Si filler metal were unsuccessful for two reasons. The PVD aluminum coating in Specimen 1 was not sufficiently thick to prevent melting of the entire layer, resulting in deleterious precipitation to occur at the beryllium interface during solidification resulting in extremely low strength and ductility. The plasma sprayed aluminum coating employed in Specimen 2 fared better, but performance was still inadequate due to small amounts of porosity which also permitted penetration by liquid filler metal. Additionally, porosity in the aluminum coating was not entirely removed by the small loads applied during the brazing process. Both specimens would probably have been improved with increased aluminum coating thickness and higher bonding pressures. Alternatively, reducing porosity in the plasma sprayed coating, while maintaining the same coating thickness, would lead to increased performance.
4. The HIP process can be used to produce superior Al-12%Si braze joints. By controlling the bonding parameters, consolidation of plasma sprayed deposits can be made to occur early in the cycle, prior to the formation of a liquid phase, resulting in improved microstructures and mechanical properties. However, control of both coating thickness and filler metal quantity is still required to prevent complete melting of the aluminum layer.
5. Copper can serve as a substitute for the Al-12%Si filler metal system. The quantity of copper needs to be controlled to prevent the previously mentioned problems of excessive (or inadequate) filler metal volume at the bonding temperature.
6. With proper control of variables in the bonding process, failure of transverse specimens occurs in the aluminum at strength values consistent with those expected for wrought 1100-Al specimens.

VI. Acknowledgements

The authors would like to thank Mssrs. R. Castro (LANL), S. Sastri (Surmet Corp), W. Bonivert (SNL/CA) and D. Butler (Northwest Technical Industries, Inc) for providing the coated beryllium and copper samples. A special thanks to C. Rood, A. Gardia, and N.Y.C. Yang for their assistance in the evaluation of the surfaces using optical and electron microscopy. Finally, we would like to acknowledge the work done by S. Goods in the set-up, testing, and evaluation of the tensile specimens.

VII. References

1. D.E. Dombrowski and D. Hashiguchi,, "Bonding Beryllium to CuCrZr using Hot Isostatic Pressing", TR-1011, JET Final Report, JDO/14691, 1990.
2. D. Hashiguchi,, "Bonding Beryllium Layer to Copper Substrate", TR-1000, JET Final Report, JD9/13197, 1990.
3. H. Altmann, E.B. Desknis, C. Ibbott,, C. Sborchia,, R. Tiveyt,, and R. Viola,, "An Analysis of Induction Brazed Beryllium on Copper Alloy Substrates", Proc. 15th IEEE Symposium on Fusion Engineering, Hyannis, MA, USA, Oct. 1993, pp 187-191.
4. I. Smid, NET Team, ITER Internal Note, September 21, 1994 (unpublished).
5. T.B. Massalski,, "Binary Alloy Phase Diagrams", ASM International, Materials Park, Ohio, 1987.
6. H. Kawamura and M. Kato, "Reactivity Test Between Beryllium and Copper", Proceedings 2nd IEA International Workshop on Beryllium Technology for Fusion, September 6-8, 1995, Jackson Lake Lodge, WY., pp 204-211.
7. B. C. Odegard Jr. and C. H. Cadden, "Beryllium-Copper Reactivity in an ITER Joining Environment", submitted to the 4th Intl. Sym. on Fusion Nuclear Tech., Japan, April, 1997
8. S. Sastri, "Ductile Joining of Beryllium to Copper", 1994 DOE Phase I, SBIR Project
9. D. Butler, private communication, Northwest Technical Industries, Sequim, WA, USA, 1995.
10. Cadden, C.H., Bonivert, W.D., Odegard, B.C.Jr. and Watson, R.D., "Beryllium-Copper Joining Techniques for Use on Plasma-Facing Components", Proc.16th IEEE Symposium on Fusion Engineering, Champaign, IL, USA, Oct. 1995. pp377-380.
11. R.G. Castro, P. W. Stanek, K.E. Elliott, D.L. Youchison, R.D. Watson, and D.S. Walsh, "The Structure, Properties, and Performance of Plasma Sprayed Beryllium for Fusion Applications", Physica Scripta, Vol T64, 1996, P. 77-83.
12. J.E. Hatch, , "Aluminum Properties and Physical Metallurgy", ASM International, Metals Park, OH, 1984.
13. S. Sastri, Surmet Corporation, Private communication.

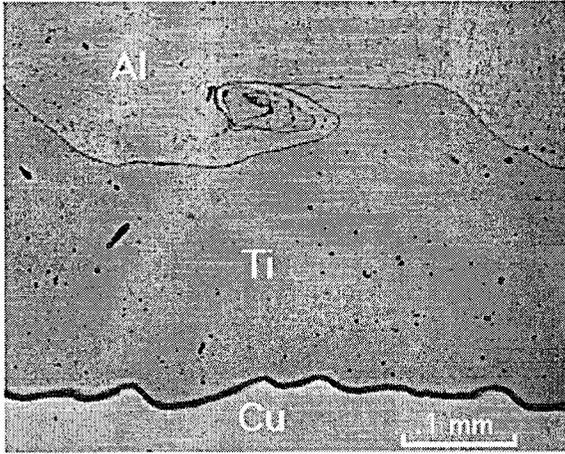


Figure 1. Cross section of the explosive bonded assembly showing 1100-Al aluminum plate bonded to copper alloy CuBeNi with a titanium diffusion barrier.

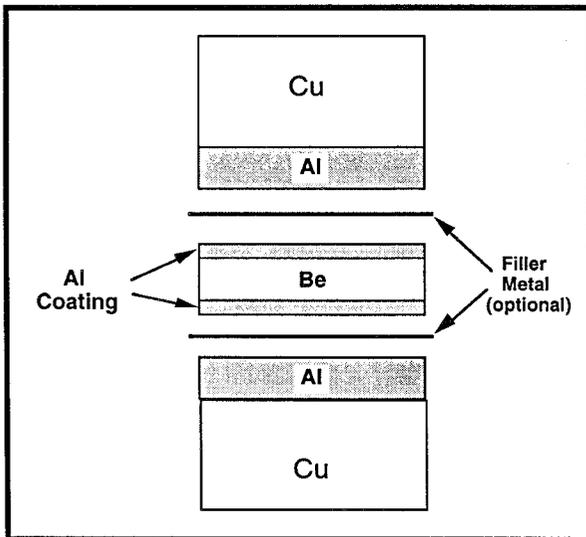


Figure 2. A schematic representation of the beryllium-copper braze assemblies used in this study.

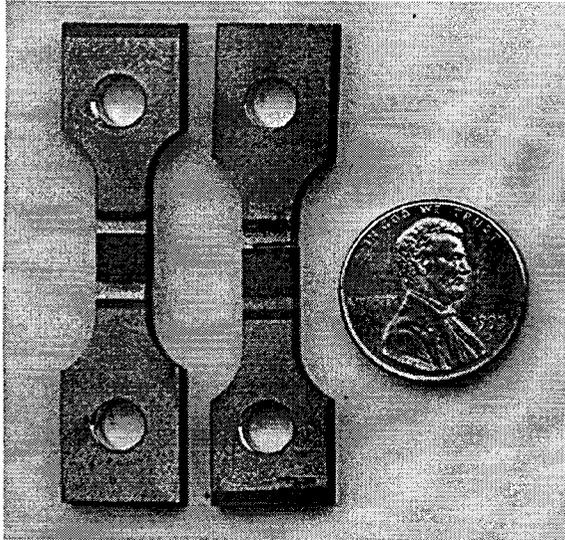


Figure 3. Rectangular cross-sectioned flat tensile bars used for room temperature tensile testing. Electrical discharged machining in a water bath is used to machine the specimens from the HIP assembly.

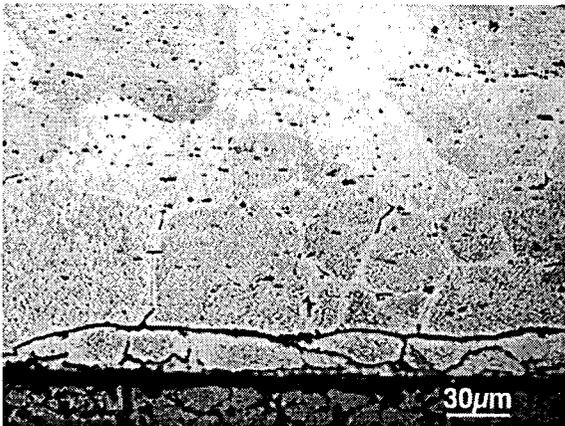


Figure 4. Micrograph showing the cross section of Specimen A; note the intergranular silicon near the beryllium interface (dark lines at the grain boundaries).

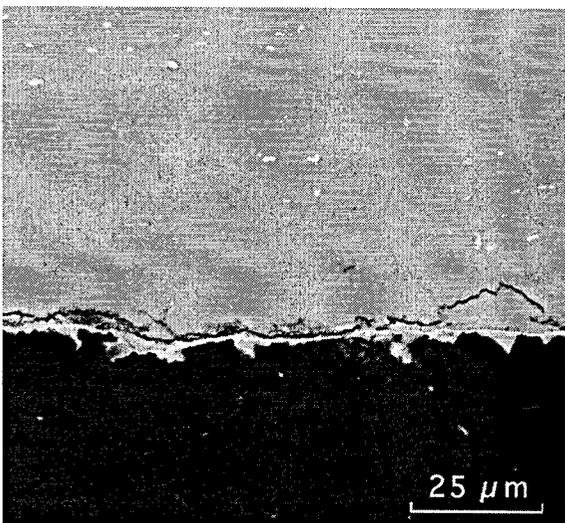


Figure 5. Micrograph showing a higher magnification of Specimen A. The bondline cracking is associated with an iron and silicon-rich layer adjacent to the beryllium interface (light phase at interface).

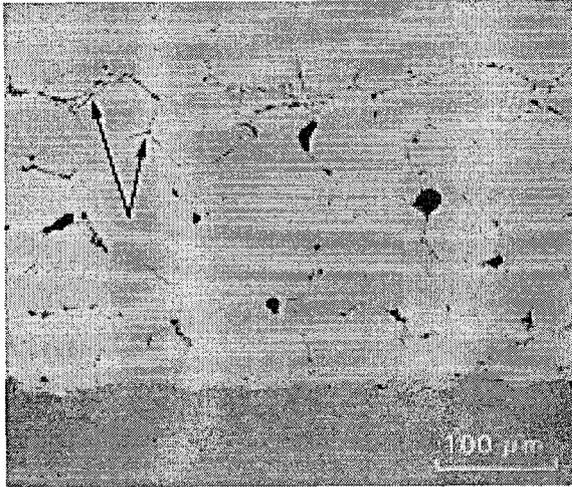


Figure 6. A cross section of the plasma sprayed aluminum (Specimen B) on the beryllium substrate following the furnace braze. Porosity is still evident in the plasma sprayed coating. The Si-Al eutectic structure is seen at prior plasma sprayed aluminum particle boundaries (arrows).

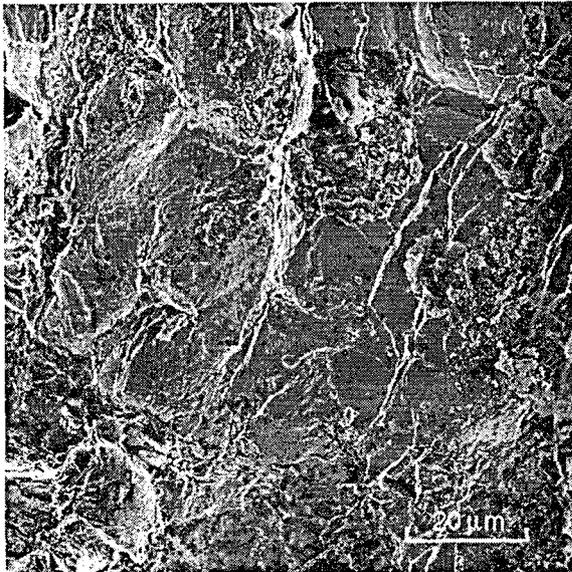


Figure 7. Fracture morphology of Specimen B showing a mixed mode fracture consisting of dimple rupture and intergranular fracture. The effect of the porosity on the fracture surface is evidenced by the intergranular separation.

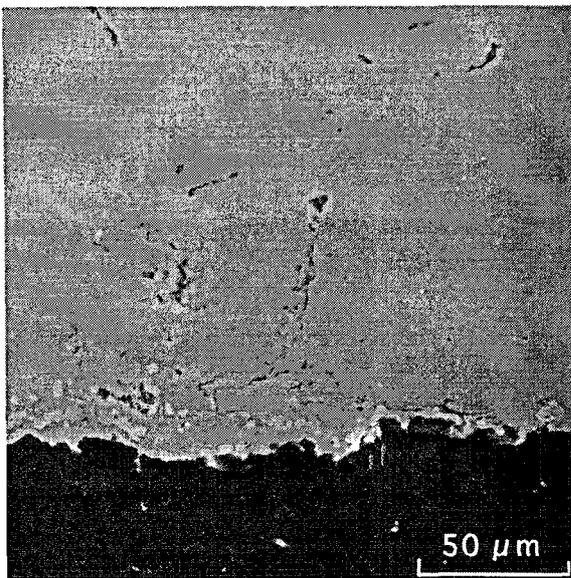


Figure 8. A cross section of the plasma sprayed aluminum (Specimen C) following the HIP cycle. The aluminum coating is fully consolidated. The discrete aluminum and iron-rich particles are seen as a light phase at the interface.

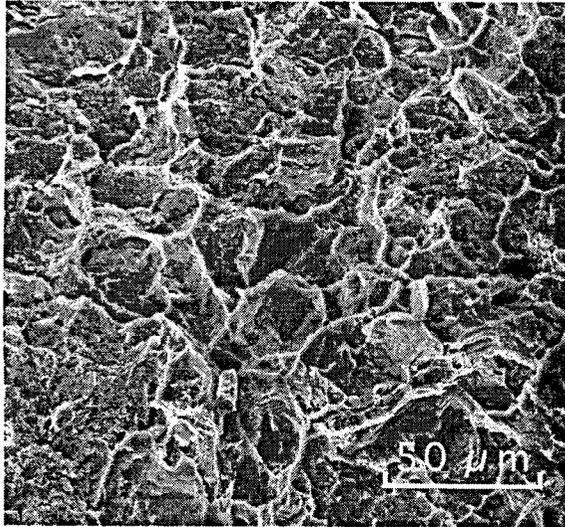


Figure 9. Fracture morphology of a room temperature tensile bar failure from Specimen C showing ductile rupture.

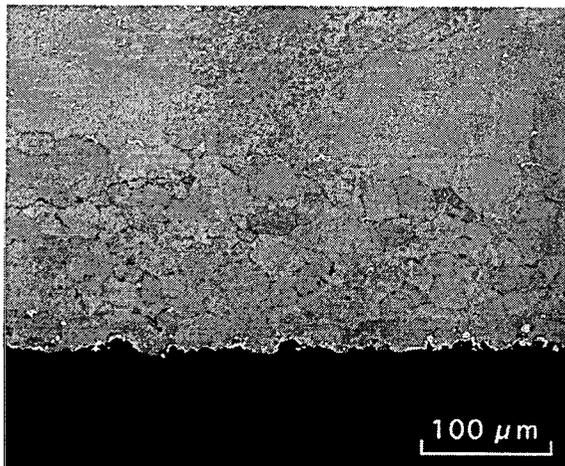


Figure 10. Micrograph showing a cross section of Specimen C. Discrete aluminum grains are visible in the plasma sprayed aluminum coating.

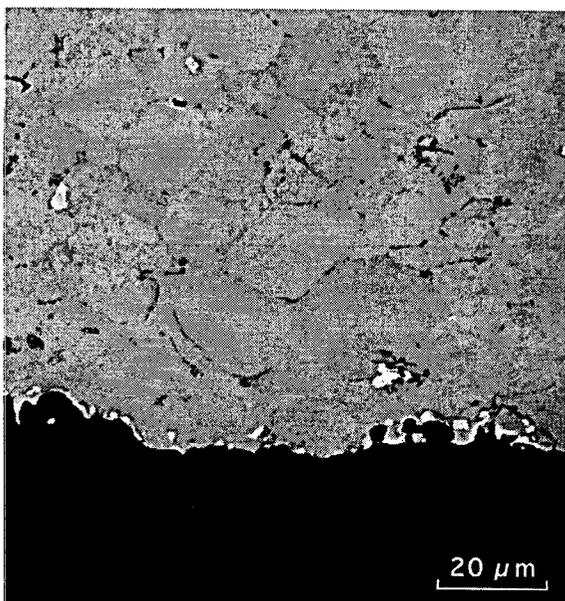


Figure 11. Micrograph showing higher magnification of the reaction zone region between the plasma sprayed aluminum and the beryllium (Specimen C). The lighter phase at the beryllium interface is silicon and iron-rich.

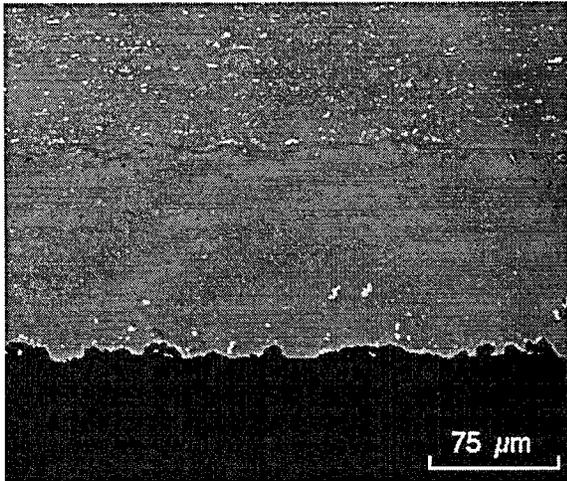


Figure 12. Micrograph showing the cross section of Specimen D. The plasma sprayed aluminum coating is fully dense. The light appearing particles are iron and copper-rich.

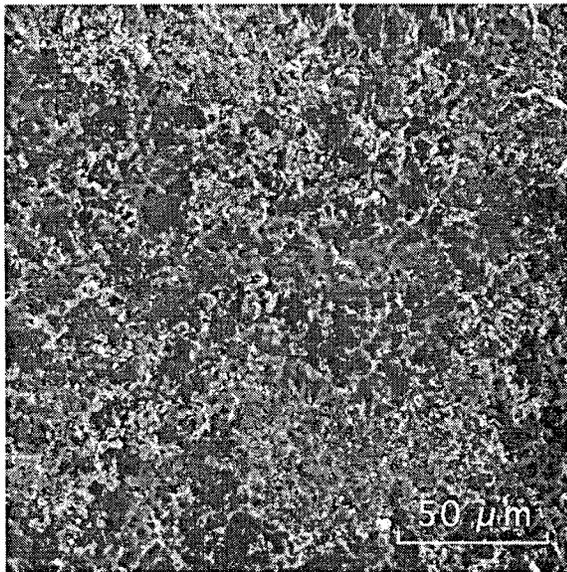


Figure 13. Fracture surface of Specimen E showing a mixed mode fracture. Fracture occurred at the aluminum-beryllium interface.



Figure 14. Fracture surface of Specimen E showing fracture in the beryllium.

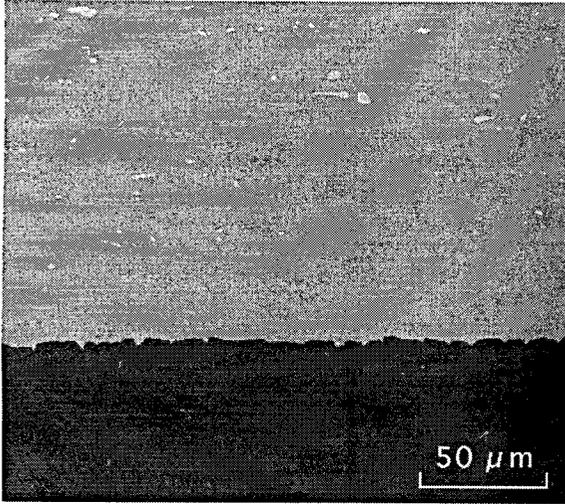


Figure 15. Cross-section of Specimen E showing a nearly precipitate-free aluminum layer adjacent to the beryllium substrate.

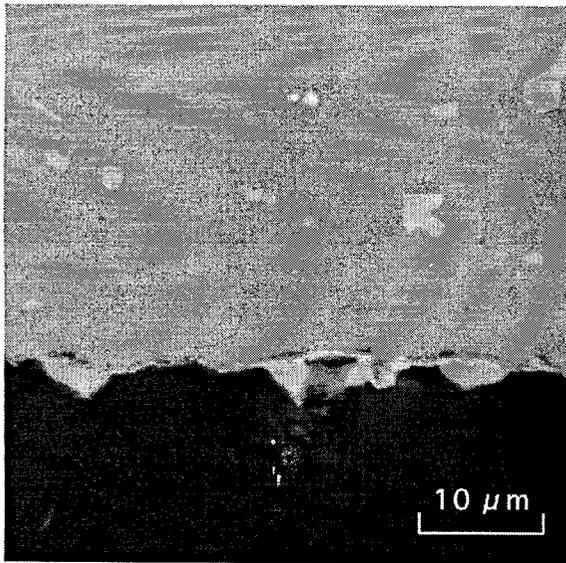


Figure 16. Cross-section of Specimen E showing occasional silicon and iron-rich particles near the beryllium interface.

UNLIMITED RELEASE

INITIAL DISTRIBUTION

BrushWellman Attn: D. Dombrowski 17876 St. Clair Ave. Cleveland, OH 44110	MS 9161 MS 9161 MS 9402 MS 9402	R. A. Causey, 8716 K. L. Wilson, 8716 C. K. Rood, 8715 G. J. Thomas, 8715
McDonnell-Douglas Attn: J. Davis Mail Code:1067220 6143 James S. McDonnell Blvd. St. St. Louis, MO 63134	MS 9402 MS 9403 MS 9403 MS 9403	N. Y. C. Yang, 8715 M. I. Baskes, 8712 S. H. Goods, 8712 B. C. Odegard Jr., 8712 (10)
McDonnell-Douglas Attn: D. Dreimeyer Mail Code:1067220 6143 James S. McDonnell Blvd. St. St. Louis, MO 63134	MS 9405 Attn:	M. T. Dyer, 8700 M. E. Perra, 8711 J. C. F. Wang, 8713 W. G. Wolfer, 8717 M. R. Birnbaum, 8742 P. E. Nielan, 8743 W. A. Kawahara, 8746
McDonnell-Douglas Attn: K. Slattery Mail Code:1067220 6143 James S. McDonnell Blvd. St. St. Louis, MO 63134	MS 9405 MS 9405 MS 9420 MS 9430 MS 9430	W. D. Bonivert, 8230 J. M. Hrubby, 8230 L. A. West, 8200 C. H. Cadden, 8240 (10) A. J. West, 8240
McDonnell-Douglas Attn: G. Wille Mail Code:1067220 6143 James S. McDonnell Blvd. St. St. Louis, MO 63134		
Surmet Corporation Attn: S. Sastri 33 B Street Burlington, MA 01803	MS 9021	Technical Communications Department, 8815/Technical Library, MS0899, 4414
MS 1129 R. D. Watson, 6531 MS 1129 M. A. Ulrickson, 6531	MS 0899 MS 9018	Technical Library, 4414 (4) Central Technical Files, 8940-2 (3)
MS 9001 T. O Hunter, 8000 Attn: J. B. Wright, 2200 M. E. John, 8100 W. J. McLean, 8300 R. C. Wayne, 8400 P. N. Smith, 8500 B. E. Brewer, 8800 D. L. Crawford, 8900		