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NASA TM X-73431

NASA TM X-73431

(NASA-TM-X-73431) THE METAL TO METAL INTERFACE AND ITS EFFECT ON ADHESION AND FRICTION (NASA) 39 p HC $4.00 CSCL 11F

N76-23415

Unclas
G3/26 28147

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TECHNICAL PAPER to be presented at
International Conference on Colloids and Surfaces
sponsored by the American Chemical Society and
the International Union of Pure and Applied Chemistry
San Juan, Puerto Rico, June 21-25, 1976

E-8584

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ABSTRACT

The interface established by two metal surfaces brought into solid state contact is much more rigidly predetermined than is the interface for the other states of matter contacting themselves or solids. Thus, solid state structural factors at the surface such as orientation, lattice registry, crystal lattice defects and structure are shown to have an effect on the character of the resulting interface established for two metals in contact. The interfacial structural character affects the adhesion on bonding forces of one solid to another. This in turn influences the forces necessary for tangential displacements of one solid surface relative to the other. Because the nature of the metal to metal interface is determined to an extent by the solid surficial layers surface tools such as LEED, Auger emission spectroscopy and field ion microscopy are used to characterize the solid surfaces prior to contacts and after the establishment of an interface. In addition to the foregoing structural considerations many of the properties of matter which influence the nature of the interface of the various states of matter with metals are shown to effect the metal to metal interface. These include metal surface chemistry and the influence of alloying on surface chemistry and bulk chemical behavior. The nature of the interface, adhesion and friction properties of noble metals, platinum metals, Group IV (B) metals and transition metals are considered. Surface chemical activity of the noble and platinum metals are shown to effect metal to metal interfaces as does a valance bonding in the transition metals. With the Group IV (B) metals the degree of metallic nature of the elements are shown to effect interfacial behavior. The effect of surface segregation of alloy constituents such as silicon in iron and its influence on the metal to metal interface is discussed. In addition the effect of alloy constituents on changes in bulk properties such as transformations in tin are shown to effect interfacial adhesion and friction behavior.

INTRODUCTION

Considerable research efforts have been expended in the studies of various types of interfaces. These have included the gas-liquid, liquid-liquid, gas-solid and liquid-solid.[1] It however, has not been until recently that the solid-solid and particularly the metal-metal interface has been examined in any detail. Prior to the advent of field ion microscopy, Auger emission spectroscopy, low energy electron diffraction and scanning electron microscopy much of the information relative to the metal-metal interface was derived from grain boundary studies.[2] Currently a wealth of information relative to such interfaces is emerging.[3-4]

It has become obvious that surfaces can no longer be thought of as extensions of the bulk properties of materials and studies are addressed to the thermodynamics of surfaces,[5] surface structure,[6] and the stresses and strains associated with surfaces.[7] A good deal of metal-metal interface information is being derived from the deposition of films on substrates and examining the resulting interface.[3, 4, 8 and 9] A new terminology is evolving with consideration of the metal-metal interface, lattice disregistry, misfit dislocations, coherency strain and anisotropic interfaces are terms currently in use.

Most of the metal to metal interfacial studies have been conducted with thin metal films deposited by ion plating, sputtering or vapor deposition onto the surface of another metal. The generation of an interface by vapor phase transport of one metal to the surface of the other can, because of the mobility of the incoming species, result in such conditions as interfacial epitaxy.[3-4] These conditions are less likely to occur in those situations where both of the metal surfaces are solids with "frozen" lattices. An important interface is that developed between solid metals contacting in technological mechanisms such as electrical contacts, bearings, gears and seals.

The objective of this paper is to consider the interface that develops between two bulk metals in contacts and the effect of that interface on adhesive bonding, resistance to tangential displacements on friction and the interfacial transport of metal from one surface to another.

METALS IN CONTACT WITH THEMSELVES

If it were possible to bring two metal single crystals of the same surface together with a perfect match of atomic planes and crystallographic directions and the two surfaces were defect free on near touch contact the two single crystals would join to form one continuous interface free metal single crystal. As a practical matter such matching is not experimentally possible and the result is that when two metal crystals of the same orientation are brought into contact the equivalent of defect ladened interfaces develop which are analogous to at best grain boundaries. Such interfaces will contain voids and misfit dislocations.

When two metal single crystals of the same orientation are brought into contact the bonding forces at the interface will depend very heavily on the degree of lattice mismatch across the interface. The greater the degree of lattice mismatch, the greater will be the concentration of misfit dislocations and the greater will be the interfacial energy.[10] Where the misfit is slight the two metal crystal lattices will be pulled into registry at the interface so that very low interfacial energy will exist although because of the mismatch long range elastic distortions into the bulk metals will occur. The minimum or zero energy condition exists in the complete absence of lattice mismatch and an interface.

Adhesion with copper crystals in contact indicate that mismatch of crystallographic directions along a common crystal axis results in a decrease in the force required to pull the interface in tension to fracture. Further, where different crystallographic planes of copper are brought into contact the force required for tensial fracture of the interface is less than where planes of the same orientation are brought in contact.[11]

With respect to the adhesion and bonding of various matched crystallographic planes and directions in general the high atomic density planes exhibit the weakest interfacial bonding and the low atomic density planes exhibit the greatest interfacial adhesion. Thus, for the face centered cubic metals such as copper the (111) orientation in contact with itself yields the minimum in bonding force, while for the body centered cubic metals it is the (110) orientation and for the hexagonal close packed metals it is the (0001) orientation.[12]

Polycrystalline metals when brought into contact with themselves present at the interface in addition to a variety of mismatches in crystal lattice, grain boundaries. These boundaries have their own characteristic energy. It will vary with the mismatch in the orientation of the adjacent grains generating the boundary. In general stronger adhesion bonding forces will be developed between such surfaces than is observed with the high atomic density single crystal surfaces.[11]

DISSIMILAR METAL INTERFACE

The case of a metal contacting itself and the generation of an interface can be compared to the grain boundary in a polycrystalline metal sample. With dissimilar metals in contact such an analogy can not be made. The species generating the interface differ in atomic size, lattice spacing, binding energy and other properties. Surface orientation does exert an influence on the interface formed between the dissimilar metals and the bond strength of that interface.

In Table I various properties for three atomic planes of copper are presented together with adhesion data for a gold (100) surface to those planes. As with copper in contact with itself bonding forces are least to the (111) copper surface. If the copper surface is examined by LEED and AES after the interface is pulled to tensile fracture gold is found to have transferred to the copper surface. AES analysis of the surface indicates the presence of gold in addition to copper peaks in the Auger spectrum. Fig. 1 is a LEED pattern for the copper (111) surface before and after contact by gold.

The LEED pattern of Fig. 1 together with AES analysis indicate epitaxial transfer of gold to the copper. The gold lattice is contracted from that observed for bulk gold. The gold accommodates itself to the copper lattice as indicated in the schematic of Fig. 2.

In Fig. 2(a) the atomic arrangement of copper to gold atoms in a bulk alloy is indicated. For alloy systems in the solid solution region, Vegard's Law can be used to predict lattice spacing.

With two dissimilar metals contacting across an interface as indicated in Fig. 2(b) the atomic arrangement is different than that seen for the alloy in Fig. 2(a). After an interface is established between the copper and gold as indicated in Fig. 2(b) strong bonds are formed between the two elements. Lattice strain occurs in the gold in order that the gold may accommodate itself to the copper lattice as indicated in Fig. 2(c).

When a tensile force is exerted normal to the copper-gold interface fracture will occur in the weakest region. Since on pulling the specimen to fracture gold was found remaining on the copper surface in an epitaxial manner fracture had to occur in the gold. Thus, the interfacial bonds developed between the copper and the gold were stronger than the cohesive bonds in the gold.

The observation that the interfacial bonds formed between two metals in contact are stronger than the cohesive bonds in the weaker of the two metals is a general observation and occurs in other dissimilar metal systems as well.[13] The transfer of metal across the interface does not generally occur in an epitaxial manner but will normally occur in accordance with the general rule. This shall be discussed further when reference is made to the iron contacting noble metal data.

GRAIN BOUNDARY AND ORIENTATION EFFECTS ON INTERFACIAL TRANSFER

Orientation effects not only the nature and strength of an interface between metals in contact but also exerts an influence on tangential displacements such as those associated with sliding. Studies with a polycrystalline slides moving across a copper bicrystal (one grain (111) and the other the (210) orientation) resulted in differences in friction not only on the surface of the grains but in the grain boundary region as well. This effect is manifested in the data of Fig. 3.

In Fig. 3(a) in sliding from the (210) grain to the (111) grain friction is higher on the (210) plane and in the grain boundary region than it is on the (111) plane. Grain boundary effects can be seen much more readily when sliding is initiated on the (111) surface as indicated in Fig. 3(b). There is a pronounced increase in the friction for the slider grain boundary interface. The grain boundary is atomically less dense than the grain surfaces on either side of that boundary.

Examination of the (111) and (210) grain surfaces after sliding with scanning electron microscopy and a single pass of the slider across the surface revealed severe surface disturbance as a result of the contact as indicated in the micrographs of Fig. 4. The micrographs for contacted surface area on both grains are at the same magnification.

On both grain surfaces in Fig. 4 fracture cracks are observed. These cracks are surface initiated. The wear face of the cracks are extremely smooth indicating crack initiation along slip bands. As indicated in the micrographs the size of the cracks are much larger on the (210) surface than on the (111) surface. Sectioning of the wear track and measurement of the crack angle of orientation relative to the surface orientation indicate that the fracture cracks do form along slip bands in the copper grains.

A wake of metal just ahead of the fracture crack stands above the surface of the grain itself. This occurs for both grain surfaces, but again, the amount of metal standing above the surface is greater for the (210) than for the (111) grain.

The metal to metal interfacial mechanism responsible for the manifested friction behavior of Fig. 3 and the surface conditions of Fig. 4 can best be explained with the aid of Fig. 5.

When the copper slider is first brought into touch contact with either grain surface and a load is applied deformation of surface asperities results in penetration of surface contaminating films and metal to metal interface formation with strong adhesive bonding. As tangential motion is commenced fracture must occur in the weakest interfacial region. The weakest region is not at the interface but rather in the cohesive bonds between adjacent slip planes in the individual copper grains. Thus, with tangential force atomic bonds along the copper slip fracture with the formation of a surface initiated crack as indicated in Fig. 5.

With a continued application of a tangential force, at some point the applied force will be sufficient to exceed that of the interfacial slider to grain bond and fracture of that interfacial bonding will occur. The slider will move on until adhesion again occurs.

After the interfacial slider to grain bond has fractured a wake or curl of metal will remain above the plane of the grain surface as indicated in Fig. 5. Subsequent passes result in shear of test surface protuberance of metal with the resulting formation of a wear particle. Thus, for polycrystalline copper in contact with a single crystal (grain) of copper the interface develops bonds which offer greater resistance to fracture than cohesive bonds along slip planes in the copper single crystal (grain).

EFFECT OF CRYSTAL STRUCTURE ON THE METAL TO METAL INTERFACE

A number of metals in the periodic table of elements are polymorphic, that is they exist in more than one crystalline form. This ability to exist in more than one crystalline form raises the question as to the effect of the various crystalline forms of a single metal on the nature of the interface formed with itself and with other metals. Tin is polymorphic, existing as gray tin below 13 °C and white tin above this temperature.[14]

Gray tin has a diamond type of crystal structure, with each tin atom tetrahedrally coordinated by four other tin atoms. White tin has a body-centered tetragonal structure and appears as a distorted diamond structure.

When an interface is formed on contact of the two forms of tin with iron and tangential motion initiated differences in friction behavior are observed. These differences are reflected in the data of Fig. 6.

White tin in Fig. 6(a) has a stick-slip or saw tooth type of friction trace indicating the formation of strong adhesive junctions at the interface between the iron and white tin. With the continued application of a tangential force the interfacial bonds are broken and slip occurs. This is manifested in Fig. 6(a) by the sharp drop in the friction force occurring at regular intervals. After slip adhesion again occurs and the applied tangential force continues to increase until the interfacial bond strength is once again exceeded and slip occurs once again. The process continues to repeat itself. It is a commonly observed behavior pattern for metal in contact with metal where strong interfacial bonds form.

The gray or diamond form of tin exhibited a continuous smooth friction trace (Fig. 6(b)). There is an absence in the trace of the stick-slip or saw tooth behavior seen in Fig. 6(a).

Examination of the iron surface after contact with the two forms of tin revealed that with white tin random islands of tin remained on the iron surface as a result of adhesion and fracture of tin bonds in the bulk tin. With gray tin a dmooth continuous uniform interfacial transport of tin to iron was observed.

The data of Fig. 6 indicate that the nature of the metal to metal interface for tin to iron is different for the two forms of tin. It is not therefore just the atomic character of the two elemental metals which form the interface but also the crystal lattice in which they find themselves.

In gray tin the atoms are in stacked sheets of continuously linked hexagonal rings parallel to the (111) planes with shear taking place along these planes. This then may account for the smooth uniform transfer film of tin observed on iron.

With white tin the tetragonal structure permits slip in two systems, on the (110) planes of atoms in the |001| direction at low temperatures and (110) planes of atoms in the |111| direction at higher temperatures.[15] This multiple slip behavior allows more readily for the type of transfer observed for white tin to iron.

THE NOBLE METAL TO IRON INTERFACE

In attempting to understand the nature of the interface between metals a seemingly logical consideration is the similarity that might exist for those metals having like properties based upon their classification in the periodic table. The noble metals, silver, copper, and gold have many properties in common. When these metals are brought into contact with an iron (001) surface interfacial adhesion occurs for each of these metals to iron.

Separation of the noble metals from the iron and subsequent examination of the iron surface with LEED and Auger emission spectroscopy indicate marked similarity in the adhesive behavior of all three noble metals.

The LEED patterns obtained with all three noble metals are presented in Fig. 7. The basic LEED structures are identical for all three noble metals on iron as indicated in the diagrammatic sketch of Fig. 7. All three noble metals were found transferred to the iron with Auger emission spectroscopy analysis. An Auger spectrum for gold on iron is presented in Fig. 8.

From the LEED and Auger analysis of the iron surface interfacial adhesion of the noble metals to iron occurred. Since the basic LEED structures are the same the iron dictated the structural interfacial arrangement of the noble metals on iron. With the application of tensile forces to fracture the interface the cohesively weaker noble metals were found to have transferred to the cohesively stronger iron. Again as was noted previously the adhesion bonds at the interface are stronger than the cohesive bonds in the weaker noble metals.

It is of interest that all three of the noble metals behaved and transferred to the iron in a similar manner. Such results indicate that basic similarities in the properties of noble metals are reflected in like similarities in their interfacial adhesive behavior.

INTERFACES OF MEMBERS OF THE PLATINUM METALS GROUP

The field ion microscope has been very useful in the study of the metal to metal interface.[16-17] Metals can be brought into the solid state contact and the interfacial results of that contact examined at the atomic level. A number of the members of the platinum metals family have been examined in this manner.

Fig. 9(a) is a photomicrograph of an iridium surface as seen in the field ion microscope, each individual white spot indicating an atom site with some of the atomic planes called out. This is an asperity free surface.

When gold is brought into contact with the iridium adhesion of the gold to the iridium occurs. On tensile fracture of the specimen bonding, gold is found to have adhered to the iridium surface as shown in Fig. 9(b). The white spots are now due to the presence of the adhered gold. There appears to be an ordered distribution of the gold on the iridium surface but without any preference of the gold for a specific atomic plane of iridium. Again, the cohesively weaker metal has transferred to the cohesively stronger.

The adhered gold of Fig. 9(b) can be removed by field evaporation. Where the gold has been removed the original iridium surface is seen. This is shown in the micrograph of Fig. 9(c). The atomic planes of iridium can be reidentified.

The iridium to platinum interface was also studied with field ion microscopy. A field ion micrograph of the iridium surface prior to contact is presented in Fig 10(a) and after contact with platinum in Fig. 10(b).

Platinum transferred to the iridium surface with a fairly high degree of order. There are two nonimaging areas between the (100) and (311) planes of Fig. 10(b). If field evaporation of the platinum was conducted the last region to lose platinum was the (100) region. Examination during field evaporation of the platinum covered iridium surface indicated that the platinum had adhered to the iridium in a near epitaxial manner.

The near epitaxial transfer of platinum to iridium, a sister element in the metals of the platinum family, is analagous to the epitaxial transfer of gold to copper. a sister noble metal. In each case the cohesively weaker of the two metals comes into atomic registry with the cohesively stronger fracture occurring in the cohesively weaker metal. Adhesive interfacial bonding is again stronger than cohesive bonding in the weaker of the two contacting metals.

The interfacial behavior of the platinum metals were examined in contact with a single metal to determine the relative differences in behavior. Loads applied to the surfaces in contact ranged from 1 to 10 grams and then tangential motion initiated with friction forces being measured.

Figure 11 is a summary figure of the friction data obtained for a gold (111) surface in sliding contact with the various metals except osmium. The figure indicates that even though the pin was identical for all platinum metals and the transfer of gold occurred to all of the platinum metal surfaces as indicated by Auger emission spectroscopy differences in friction behavior existed.[18]

The highest friction was obtained with the metals platinum and palladium and the lowest with ruthenium, rhodium, and iridium. All metals were in single crystal form with the highest atomic density, lowest surface energy plane exposed to contact with gold. These atomic planes have the same atomic packing. These orientations were selected to eliminate crystallgraphic orientation as a variable.

Both platinum and palladium are chemically more active than ruthenium, rhodium, and iridium. As the atomic number in period 5 containing the elements ruthenium, rhodium, and palladium is increased, the contribution to bonding of d electrons is increased. Likewise, a similar behavior is observed in period 6 with the elements osmium, iridium, and platinum.[19] Thus, stronger bonding of gold to platinum and palladium would be anticipated from the valence-bond model when that model is applied to metallic systems. As indicated in Ref. 19 there is no reason not to apply it to metal systems since it involves the same basic electronic bonding as is involved in other systems for which the model was originally developed.

In considering the transition elements a knowledge of the contribution of d elections to metallic bonding is necessary. An examination of the heats of atomization of the elements in the periodic table clearly indicates the importance of the d electrons to bonding. The most stable metallic structures are those which use as many d electrons as possible in bonding. The contribution to d electron bonding increases with increasing atomic number; thus, Ru < Rh < Pd and Os < Ir < Pt. With increasing d electron bonding there is a corresponding decrease in the sp electron contribution to bonding.

GROUP IV ELEMENTS

The Group IV elements silicon, germanium, tin, and lead exhibit many common properties. One very interesting property of the elements as a group is the increase in metallic character in moving through the Group from silicon to lead. Germanium for example, is very brittle while white tin has good ductility. Tin itself exhibits differences in metallic character. White tin is very ductile while gray tin is much less so.

Because silicon and germanium are very friable it is difficult to prepare flat surfaces of these elements which readily lend themselves to interface, adhesion and friction studies. Thin films (800 Å) thick of silicon, germanium, tin and lead were therefore deposited by ion plating on a common substrate, namely a (011) nickel crystal surface for interfacial bonding studies. A gold (111) surface was brought into contact with these various films, loads applied, the surfaces separated and the interfaces examined for interfacial transport.

By examining a very thin film, 800 Å, which is just sufficient to form a continuous film over the nickel substrate the effect of the basic chemistry of the elements can be compared without too much concern for differences in the mechanical deformation behavior of the film. Insight into the more fundamental effects of the electronic nature of these elements can thereby be achieved.

Adhesion was greater for the tin and lead in contact with gold than it was for silicon and germanium. Initiation of tangential motion and recording of friction force indicated markedly different interfacial behavior as reflected in friction force data of Fig. 12 for the germanium and tin films.

In Fig. 12(a) high initial friction was noted as indicated by the spike to the left in the friction trace. This reflects very strong interfacial bonding between the gold and the germanium. Once tangential motion has begun the force drops to a very low value. If the movement is stopped and the specimens allowed to stand in contact under load for a period of time and then tangential motion reinitiated high friction is again obtained as indicated in the spikes to the right side of Fig. 12(a).

Identical experiments with tin films yielded the friction results of Fig. 12(b). The stick-slip behavior observed earlier with iron in contact with tin is again seen in Fig. 12(b) with films of tin. As discussed earlier this behavior is characteristic of strong metal to metal interfacial bonding. Observations similar to those presented in Fig. 12 for germanium and tin were observed when silicon and lead were compared.

The experimental results herein indicate that for films of the Group IV elements on a common substrate adhesion and friction are less for the covalently bonded elements than for the metallic bonded metals in contact with a metal. The cohesive binding energies for silicon and germanium are greater than those for tin, lead, and the gold pin. The stronger the interatomic bonding within the element the more closely the valence electrons are held to the nucleus. The covalent bond character of the group IV elements is due to the sp^3 hybrid formation. In this study, the electron pair bonds are strongest in silicon and become weaker with the other elements. The electrons become less and less of a valence type and tend to resemble free electrons more and more when moving from silicon to germanium to tin and finally to lead.

Valence electrons require a greater degree of specificity in interfacial electron compounds than is required with free electrons. Thus, bonding can be expected to occur more readily with free electron elements.

The good adhesion resistance of germanium was recognized in early engineering studies.[20] These early observations were, however, not related to bonding.

THE RELATION BETWEEN METAL BOND CHARACTER AND INTERFACIAL BONDING

Pauling in 1948 formulated a resonating-valence-bond theory of metals and intermetallic compounds in which numerical values could be placed on the bonding character of the various transition elements.[21] While there have been critics of the theory it appears to be the most plausible in explaining the interfacial interactions of transition metals in contact with themselves and other metals.

When two metal surfaces are placed into contact in the atomically clean state the intermetallic bonds that form are going to depend heavily on the character of the bonding in each of the metals. One might predict from Pauling's theory that those metals which have strong d character would be less likely to interact forming strong interfacial bonds with other metals than those metals which do not have this strong character.

Adhesion and friction experiments have been conducted with transition metals both in bulk and thin film form. Results for bulk metal friction measurements are presented in Fig. 13. The surface in contact with each of the transition metals in Fig. 13 was a gold (111) surface. The data of Fig. 13 indicate a decrease in friction with an increase in d character of the metallic bond. Similar results were obtained in adhesion experiments.

When thin films (2000 Å) of some of the transition metals examined in Fig. 13 were placed on a quartz substrate by sputter deposition and examined in adhesion and friction experiments, adhesion and friction decreased with increasing d bond character to iron. With iron and those metals having stronger d character (e.g. platinum) the interface between the transition metal and the quartz substrate was weaker than that between the gold and the transition metal and with tangential motions the metal film separated from the quartz substrate. With iron an abrupt decrease occurred in friction and with all the metals which separated from the quartz the friction was essentially the same as that for gold in contact with quartz.

ALLOYING AND ITS EFFECT ON THE METAL TO METAL INTERFACE

Small amounts of alloying elements can markedly alter the character of metal surfaces via such mechanisms as equilibrium surface segregation.[22-25]

The segregation of alloy constituents to the surface has been found to result in concentrations of alloy constituents on the surface far in excess of the bulk. With copper-aluminum alloys an alloy containing 10% aluminum had in its surface layer pure aluminum atoms with lateral packing equivalent to one-third bulk atomic packing along (111) planes.[24] When two metals are brought into solid state contact the presence of these segregated species can and does alter the nature of the metal to metal interface.

An example of the effect of surface segregation is seen with copper-aluminum alloys contacting a gold surface. The adhesive bonding of gold to a copper - one atomic aluminum alloy resulted in measured adhesive forces five times those for elemental copper in identical experiments. The adhesive bonding forces are identical to those measured for a gold (111) surface contacting an elemental aluminum (111) surface. The sluminum had segregated out of the alloy matrix on the surface of the alloy such that the interface upon contact with gold was one of gold to aluminum rather than gold to an alloy of aluminum in copper. It is for this reason that great care must be taken in using bulk metal properties to predict surface behavior as surfaces may not always be reflections of the bulk.

With the copper-aluminum alloys just described the aluminum, once it has segregated to the alloy surface due to heating or strain remains on the surface. Another type of surface segregation and one which is more illusive to study is that of silicon in iron.

When an iron - 6.55 at. % silicon alloy is heated the silicon segregates from the matrix to the surface. Auger emission spectroscopy analysis of the alloy while being heated indicates growth of the concentration of silicon at the surface. This growth is indicated in the data of Fig. 14.

An examination of Fig. 14 indicates that at temperatures above about 300 °C the amount of silicon on the surface due to segregation from the bulk increases. It continues to increase with increases in specimen temperature to 700 °C. When the specimen is cooled the silicon returns to the alloy matrix as indicated in Fig. 14. Thus, the segregation is reversible.

The adhesion behavior of gold to the iron - 6.55 at. % silicon alloy of Fig. 14 was studied over the same temperature range. The results obtained are presented in Fig. 15 together with data for elemental iron.

The data of Fig. 15 indicate a decrease in the interfacial adhesive bonding of the gold to the iron-silicon alloy as the temperature of 300 °C is approached. It appears from the data that with silicon segregation adhesion forces decrease. There is still strong adhesion but the binding force of gold to silicon is less than it is to iron.

Beyond 300 °C the adhesive binding force of the gold to the alloy remained relatively constant as reflected in the adhesion data of Fig. 15. This is a similar observation to that made with copper-aluminum alloys.[22]

When the specimens of Fig. 15 were cooled to room temperature the adhesion coefficient returned to near the original room temperature value. This is indicated in the single data point of Fig. 15.

The observations of Fig. 15 for iron-silicon are contrary to these normally observed for clean metals in contact. Generally with clean metals in contact the interfacial adhesive binding forces increase with temperature increases. In Ref. 26 strong adhesion for gold was observed to commence at about 247 °C. A similar observation was made for gold to iron. When iton was contacted by gold at temperature above 250 °C binding forces were so strong that separation of the specimens constituted tensile fracture experiments rather than adhesive bond force studies. It must be concluded from Fig. 15 that silicon segregation to the surface of iron-silicon alloys reduced interfacial adhesive bonding.

The interesting aspect of the data of Fig. 15 is that the segregation of silicon is reversible and so is the adhesive behavior. With increase in temperature silicon segregates and adhesion goes down while with a return to room temperature silicon returns to the matrix and adhesion goes back up.

In addition to the segregation of alloy constituents to the surface of alloys influencing interfacial metal to metal behavior there are other effects of alloy constituents which bear upon interfacial behavior. One sure effect is that of alloying elements on the kinetics of crystal transformations. As was discussed earlier in reference to Fig. 6 tin transforms from one crystalline form to another at 13 °C. Some alloying elements have been found to accelerate the kinetics of transformation while others retard it or arrest it completely.

Changes in interfacial friction properties of tin occur with an alteration of the kinetics of crystal transformation. This is demonstrated by the data of Fig. 16 when various elements are added to tin in a concentration of one - at. % alloying element.

The data of Fig. 16 indicates that a decrease in friction coefficient occurs with the transformation of gray tin to white tin. Bismuth arrests the transformation and the data of Fig. 16 indicate that a change in friction is also arrested. Both copper and aluminum accelerate the kinetics of transformation and the data of Fig. 16 indicate marked changes in friction with accelerated transformation, the greatest being noticed with aluminum. Thus, these data indicate that bulk as well as surface effects with alloying will influence metal to metal interfacial behavior.

CONCLUDING REMARKS

There are many properties of metals which influence the nature of the interface developed when two metals are brought into contact. These include surface orientation, lattice spacing, grain boundaries, crystal structure, nature of bond character and alloying elements. The effect of alloying elements can alter interfacial behavior by segregation to the surface of metals or by altering bulk properties such as crystal transformation kinetics.

With dissimilar metals in contact epitaxial transfer from one metal to another has been observed. The bonding at the interface between dissimilar metals is as a general rule, stronger than the bonding in the cohesively weaker of the two metals with the result that on separation of the metals transport of the cohesively weaker to the cohesively stronger is observed.

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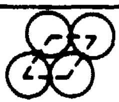
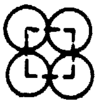
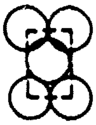
TABLE I. - SOME PROPERTIES OF THREE PLANES OF COPPER TOGETHER WITH MEASURED ADHESIVE FORCES TO THOSE PLANES

Copper surface plane	Coordination number of surface	Atomic arrangement of surface unit mesh	Number of surface, atoms /cm^2	Elastic modulus, dynes/cm^2	Surface energy, ergs/cm^2	Force of adhesion to gold,[a] mg
(111)	9		1.7×10^{15}	19.4×10^{11}	2499	80
(100)	8		1.5×10^{15}	6.67×10^{11}	2892	185
(110)	7		1.1×10^{15}	13.1×10^{11}	----	390

[a] Applied load, 20 mg; Au (100) surface; contact time, 10 seconds.

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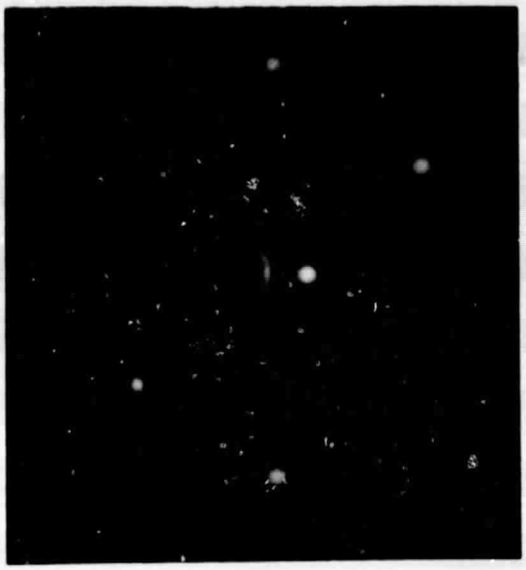
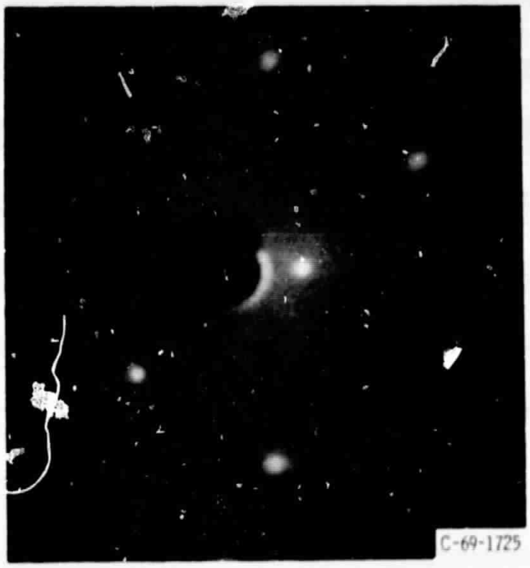
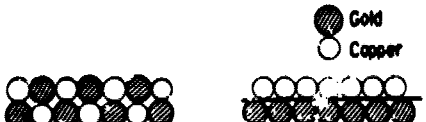


(a) Clean (111).



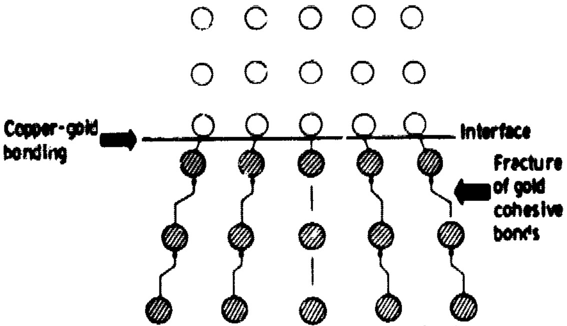
(b) After 20-milligram contact with gold.

Figure 1. - LEED patterns of copper (111) surface before and after adhesive contact with gold.



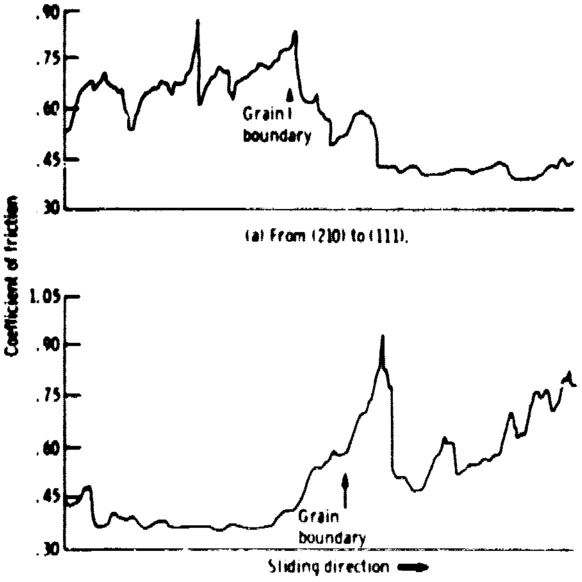
(a) Copper-gold alloy.

(b) Adhesion of gold to cpitaxially.



(c) Lattice strain in gold on adhesion to copper.

Figure 2. - Atomic arrangement and lattice bonding.



(a) From (210) to (111).

(b) From (111) to (210).

Figure 3. - Recorder tracings of friction force for copper slider sliding across grain boundary on copper bicrystal. Load, 100 grams; sliding speed, 1.4 millimeters per minute.

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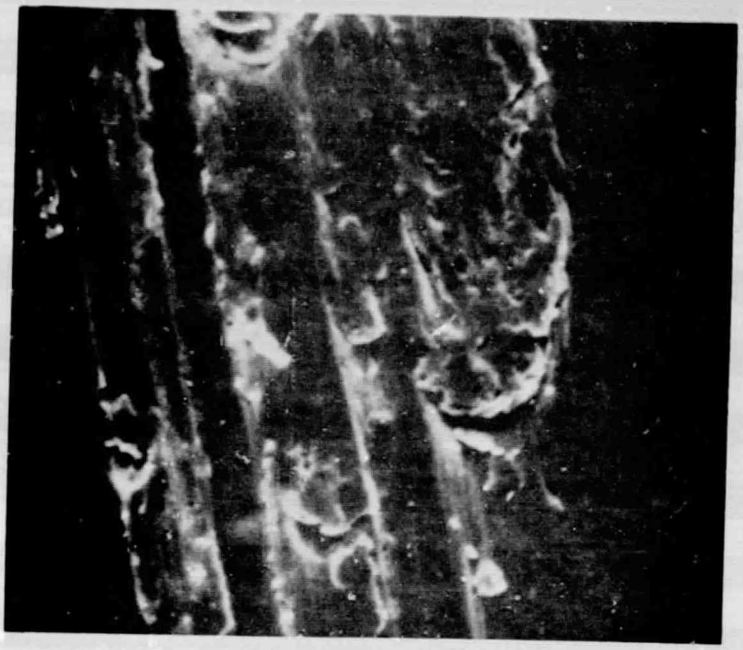
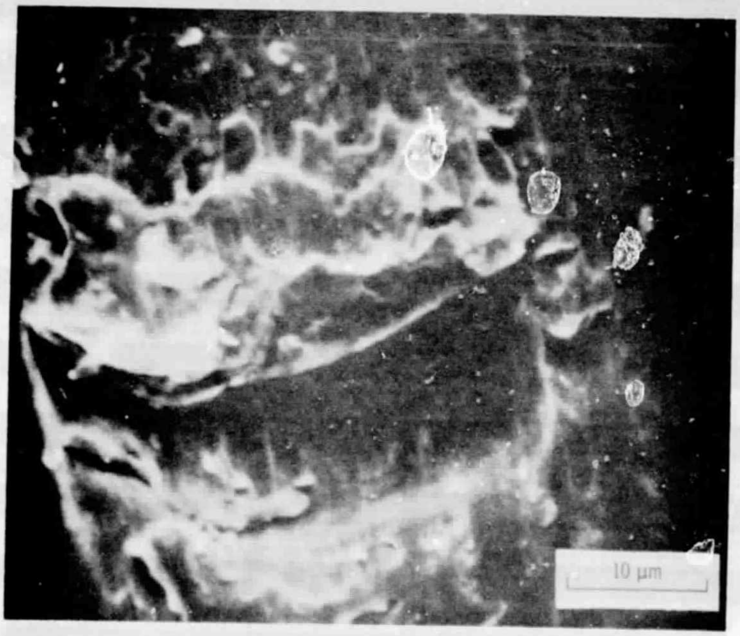


(a) (111) grain.



(b) (210) grain.

Figure 4. - Wear tracks on bicrystal grains. Copper slider; load, 100 grams; sliding speed, 1.4 millimeters per minute.

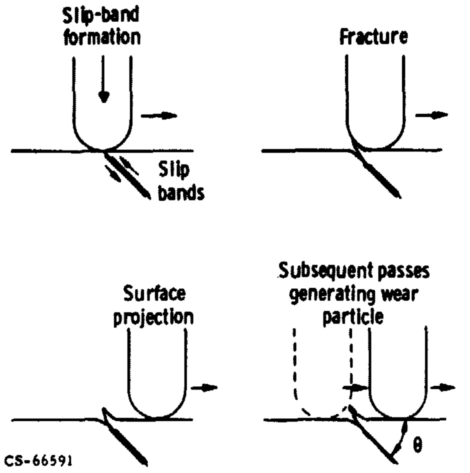
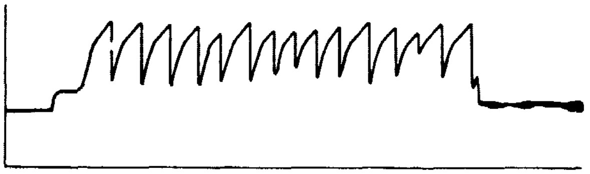


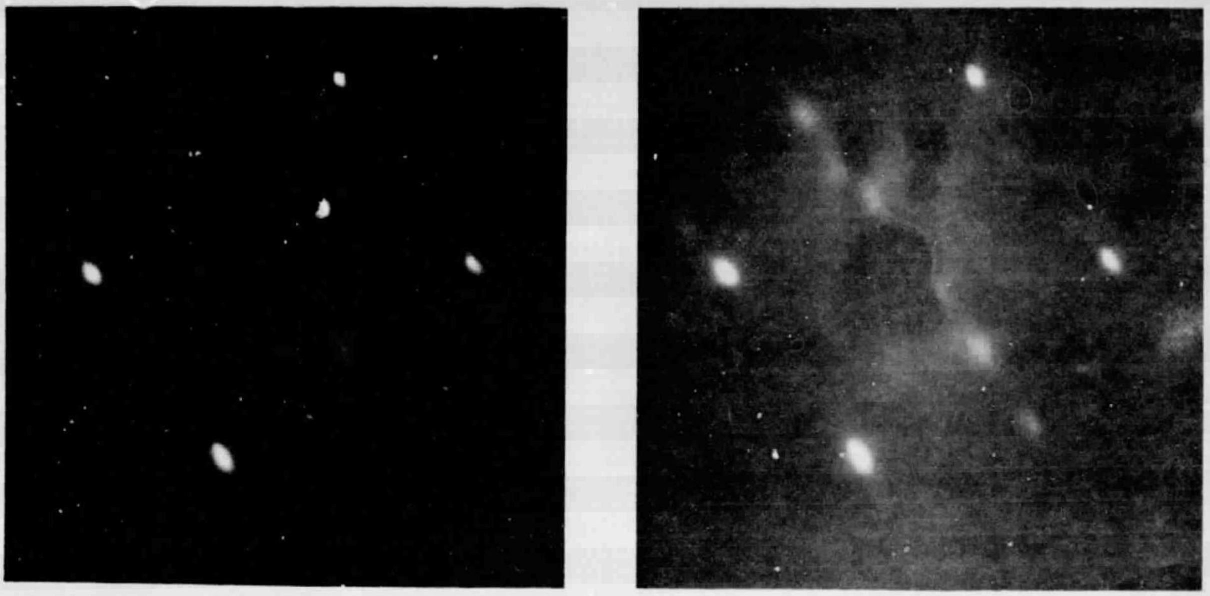
Figure 5. - Origin of surface fracture and formation of wear particle.



(a) Temperature, 24° C; white tin.

(b) Temperature, -46° C; gray tin.

Figure 6. - Friction traces for iron (110) sliding on a tin (110) single-crystal surface at 24° and -46° C. Sliding velocity, 0.7 mm/min; load, 10 g; pressure, 10^{-8} N/m^2 (10^{-10} torr).



(a) Silver.

(b) Copper.

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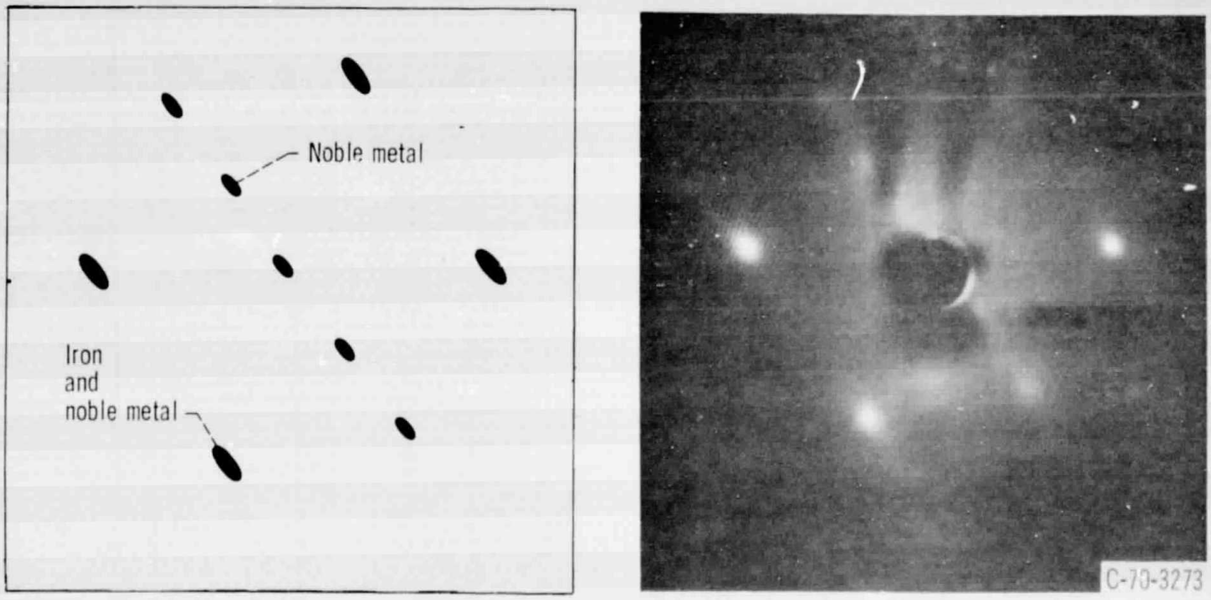

(c) Gold.

Figure 7. - LEED photographs of iron (011) surface after adhesion to noble metals.

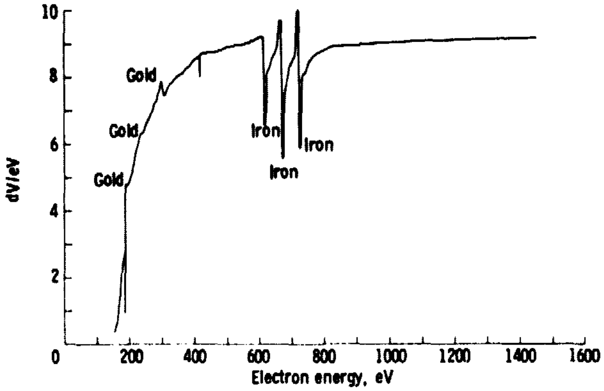

Figure 8. - Auger emmission spectrometer trace of iron (011) surface with adhered gold.

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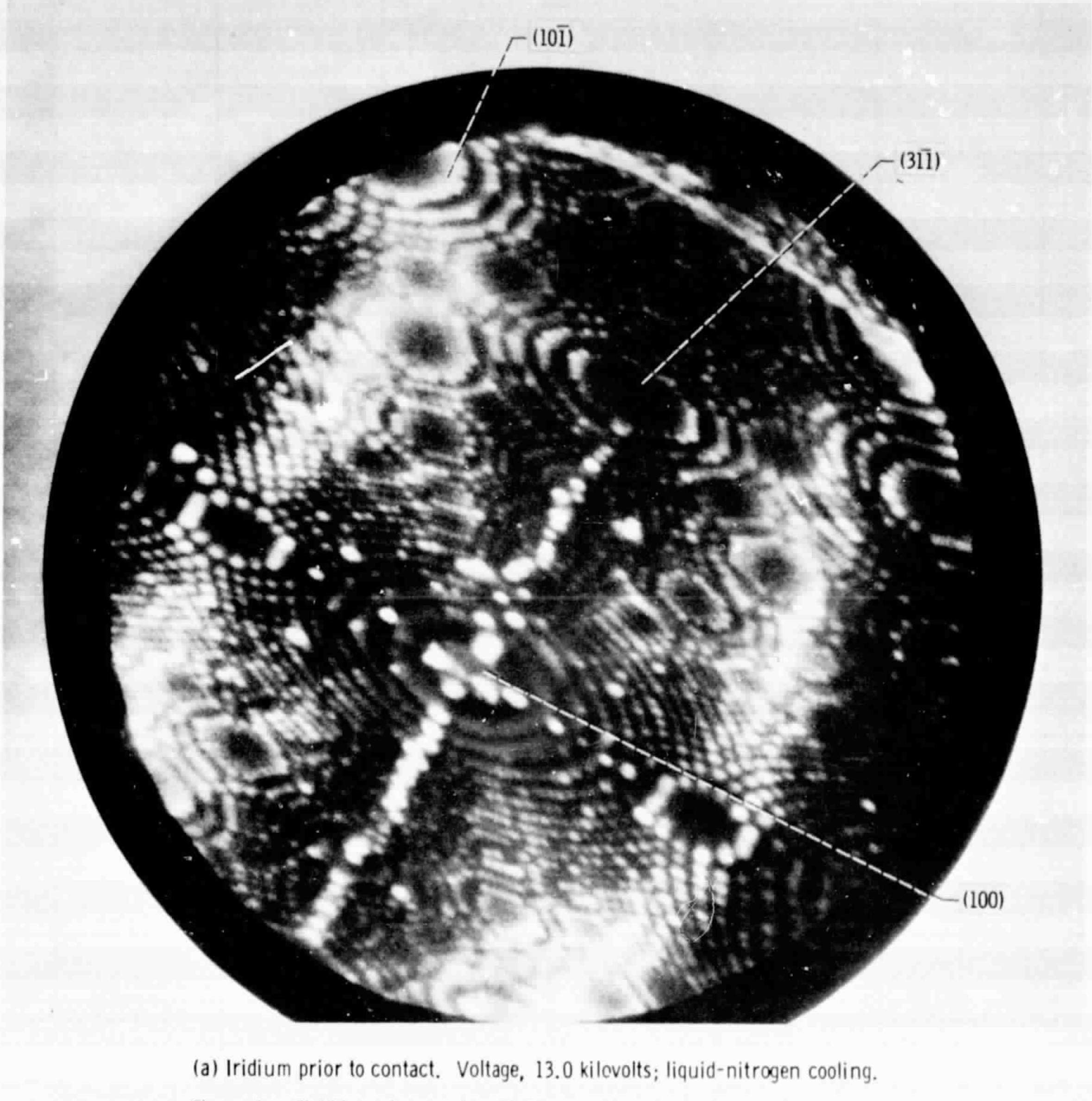

(a) Iridium prior to contact. Voltage, 13.0 kilovolts; liquid-nitrogen cooling.

Figure 9. - Field ion micrographs of iridium-gold contact. Image gas, helium.

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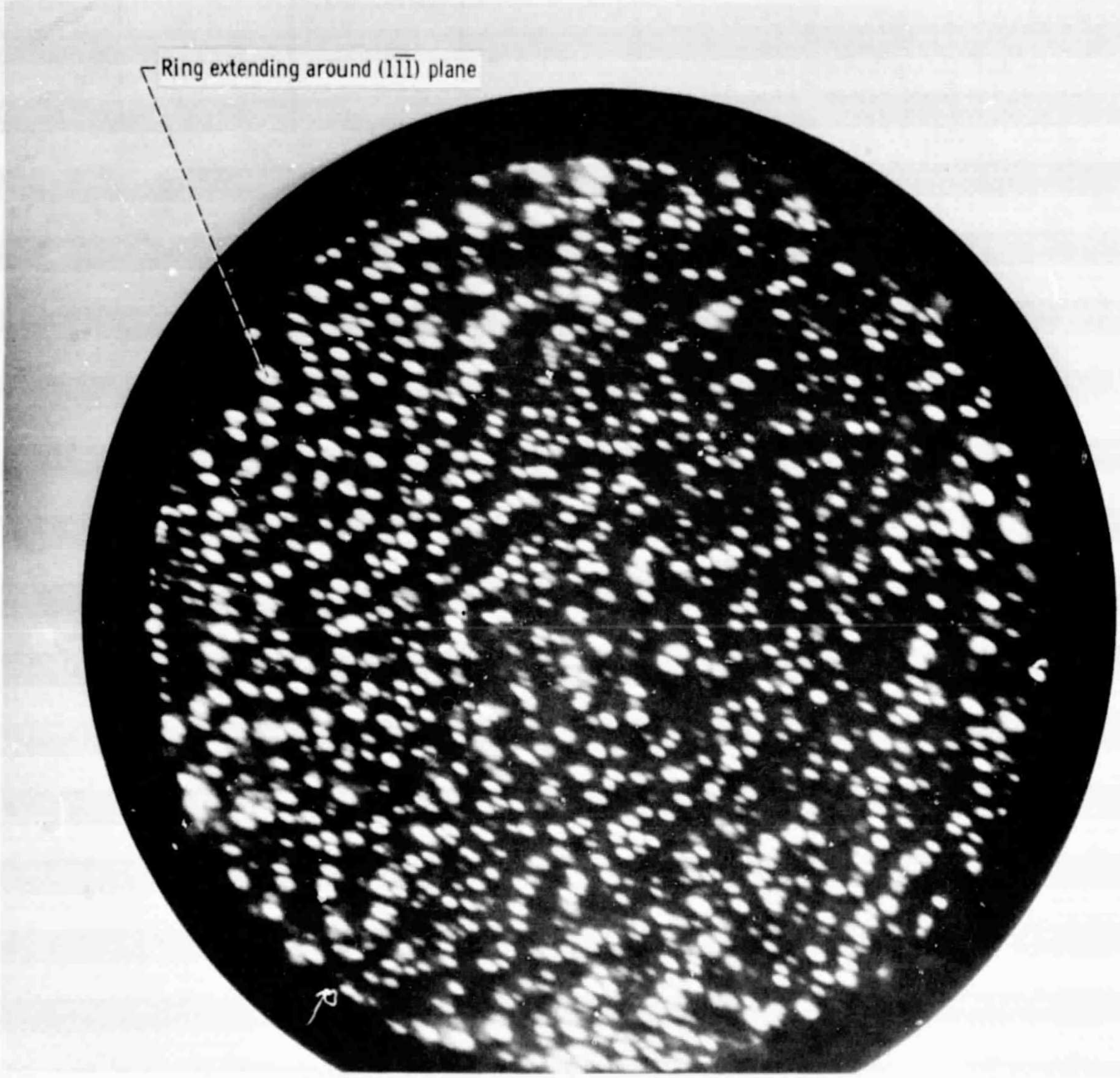

(b) Iridium after contact at 12.0 kilovolts; liquid-helium cooling.

Figure 9. - Continued.

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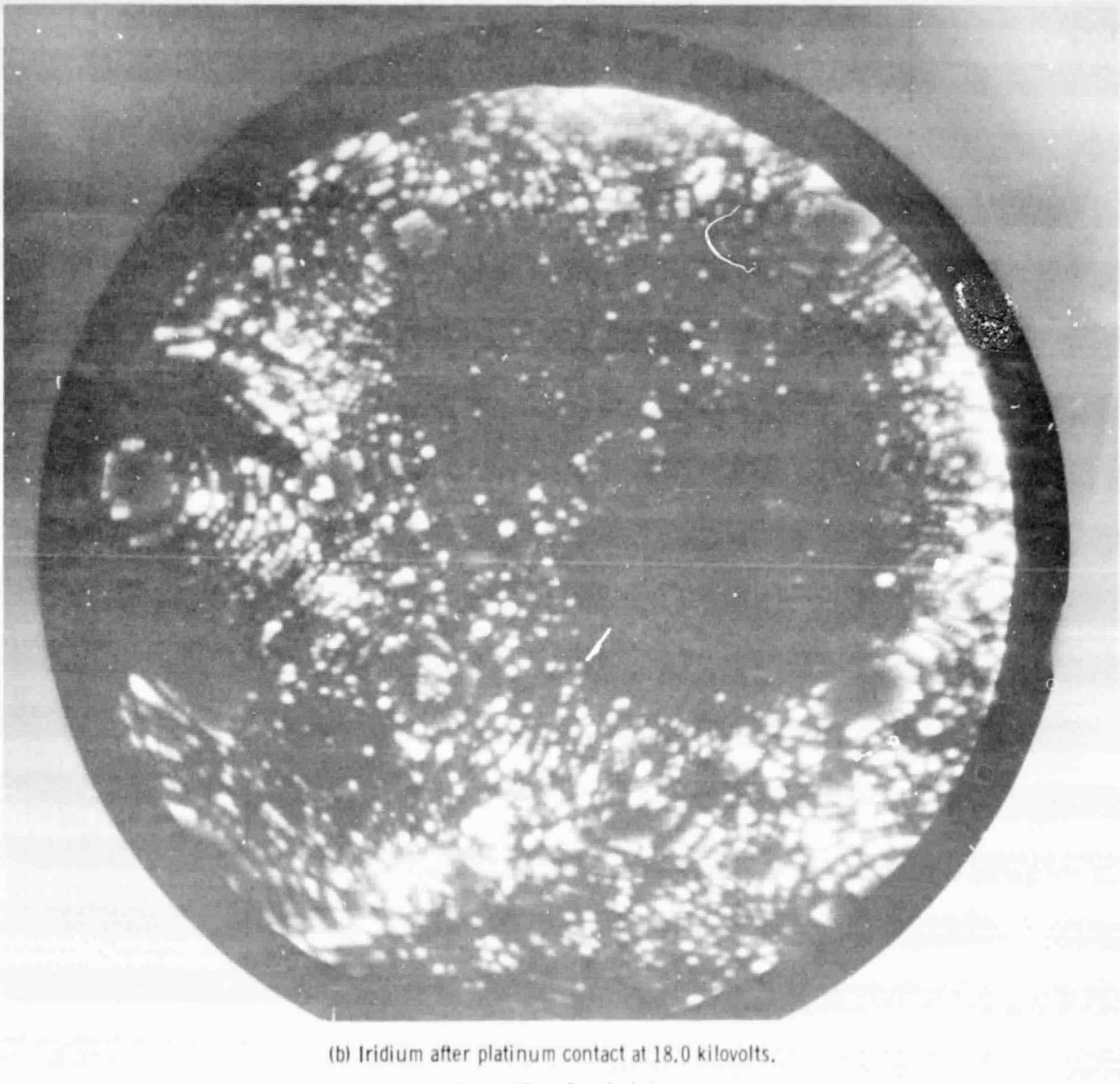
(c) Iridium after gold contact at 13.0 kilovolts with voltage raised to 15.2 kilovolts for 30 seconds; liquid-helium cooling.

Figure 9. - Concluded.



(b) Iridium after platinum contact at 18.0 kilovolts.

Figure 10. - Concluded.

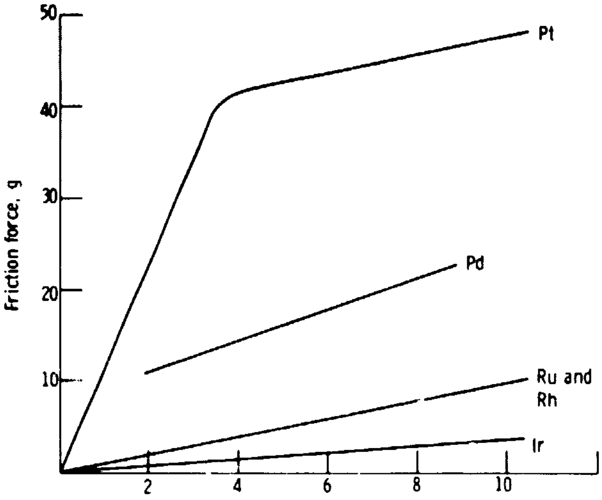
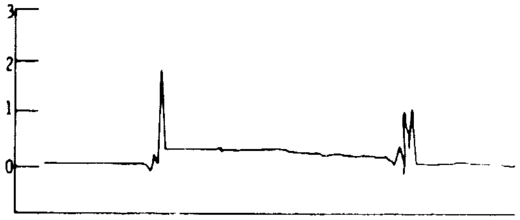
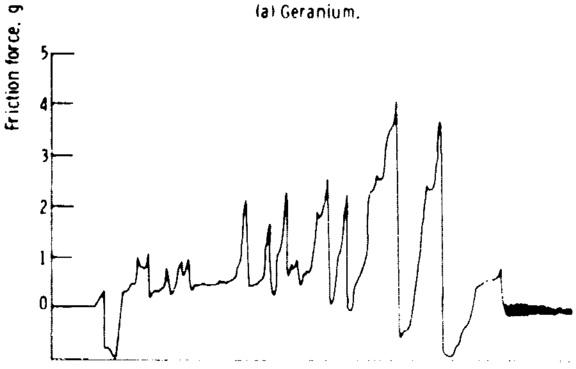


Figure 11. - Friction force as function of applied load for gold (111) single crystal sliding on various members of platinum metals group. Sliding velocity, 0.7 millimeter per minute; ambient pressure, 1.33×10^{-8} newton per square meter (10^{-10} torr); temperature, 23° C.



(a) Geranium.



(b) Tin.

Figure 12. - Friction traces for 8×10^{-8} meter (800 Å) films of geranium and tin films ion plated onto nickel (011) surface. Load, 1 gram; sliding velocity, 7.0 meters per minute; temperature, 23°; pressure, 1.33×10^{-8} newton per square meter (10^{-10} torr).

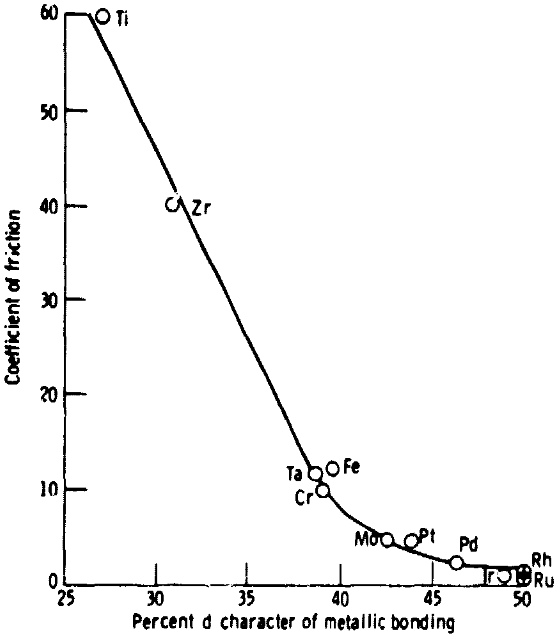


Figure 13. - Coefficient of friction as a function of the percent d bond character for various metals. Sliding velocity, 0.7 mm/min., load, 1 gram, 23^0 C and 10^{-8} N/m^2.

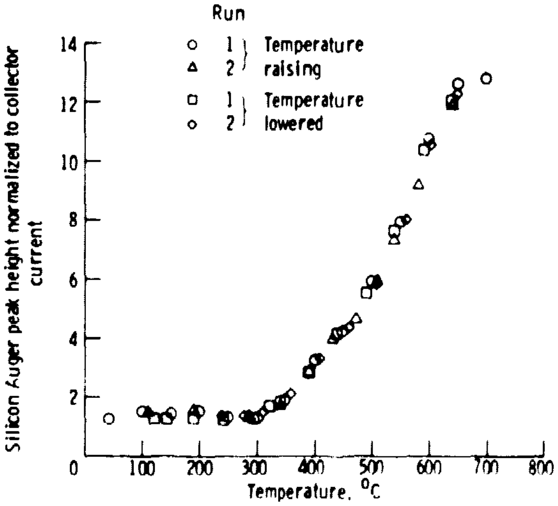


Figure 14. - AES silicon peak-to-peak height normalized to collector current for iron-6.55 at. % silicon (single crystal) against temperature.

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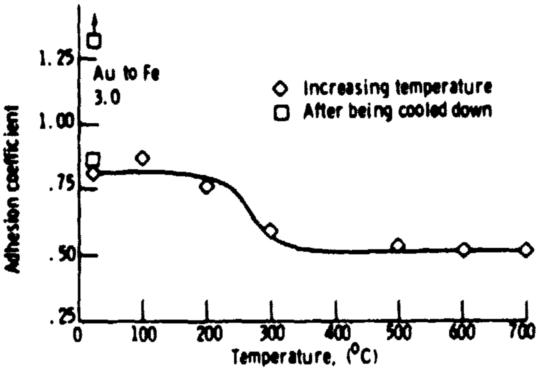

Figure 15. - Adhesion of gold (111) surface to an iron-6.5-percent silicon alloy crystal (110) at various temperatures. Load 1.0 cm and 10^{-8} N/m^2 environmental pressure.

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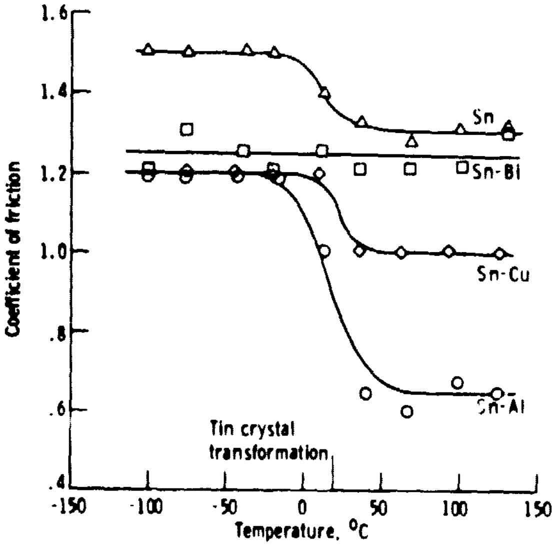

Figure 16. - Coefficient of friction for polycrystalline tin and tin alloys at various temperatures. Sliding velocity, 0.7 mm/min; load, 10 g; pressure, 1.33×10^{-8} N/m^2 (10^{-10} torr).

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