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POSSIBLE RELATION OF FRICTION OF COPPER-ALUMINUM ALLOYS WITH DECREASING STACKING-FAULT ENERGY

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SUMMARY

Friction experiments were conducted in a vacuum (10^{-11} torr) to determine the influence of stacking faults and stacking-fault energy on the shear and consequently the friction behavior of copper-aluminum alloys. The (111) plane of the single crystals slid parallel to the interface. The mating surface was polycrystalline aluminum oxide. Experiments were conducted at loads of 50 to 1000 grams and a sliding speed of 0.001 centimeter per second.

The results of this investigation indicate that the friction coefficient for copper-aluminum alloys is related to the shear stress and stacking-fault energy for these alloys; the lower the stacking-fault energy, the higher the coefficient of friction.

INTRODUCTION

The friction and adhesion characteristics of metals are markedly dependent upon physical and mechanical properties (refs. 1 to 3). In these references, it has been shown that crystal structure exerts considerable influence on adhesion and friction behavior. A marked difference in the work hardening characteristics of metals may be associated with different crystal structures. The hexagonal metals exhibit, with preferred orientation of crystallites or in the single-crystal form, very little tendency to work-harden. Cubic metals work-harden very readily.

Essential difference in the behavior (e.g., deformation, strength, resistivity, etc.) of the hexagonal crystal form of cobalt and the cubic form of the same metal has been attributed to differences in slip systems which, in turn, are related to stacking faults (ref. 4). Stacking faults and stacking-fault energy are described in the section titled STACKING FAULTS. Stacking faults in metals and alloys contribute to work-hardening of metals (refs. 5 to 7). They also influence electrical resistivity (refs. 8 and 9) and

the tendency toward preferred orientations with deformation (ref. 8).

The work-hardening characteristics of metals are extremely important in vacuum, where marked increases in the shear strength of junctions of contacting sliding interfaces can contribute to increases in friction and the occurrence of gross metal transfer. The choice of materials for sliding electric contacts in vacuum is influenced by friction, wear, and electrical resistance. Changes in electrical resistivity of materials (e. g. , a tenfold increase for copper with aluminum additions) are associated with stacking faults and thus are important in determining the behavior of sliding electric contacts.

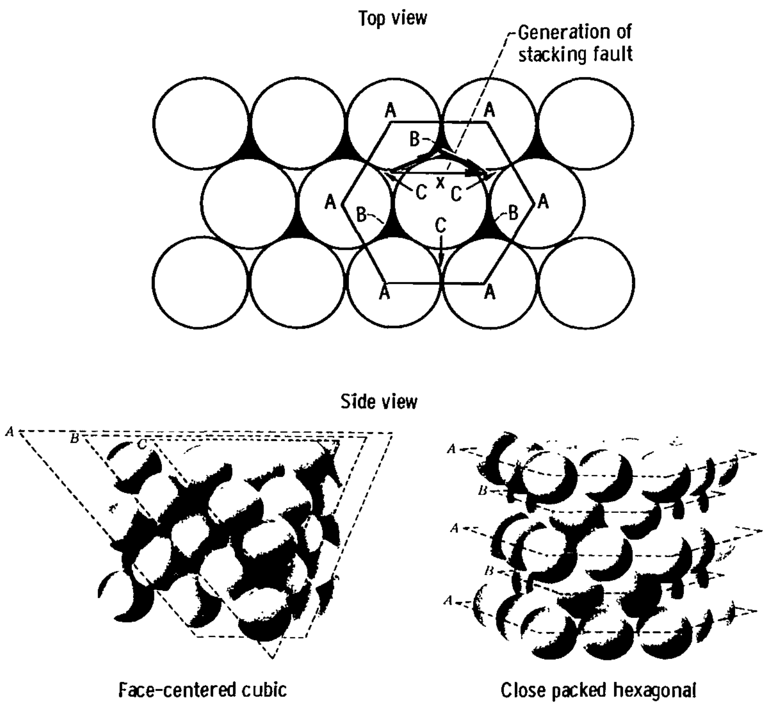
This investigation determined the influence of stacking faults on friction characteristics in vacuum. Single crystals of copper and copper-aluminum alloys were chosen for the study. This alloy system was selected because it shows appreciable differences in stacking-fault energies which have been documented in considerable detail (ref. 9). The single crystals slid, with the (111) plane parallel to the sliding interface, on polycrystalline aluminum oxide. Aluminum oxide was selected as the mating surface because adhesion of metal to aluminum oxide occurs with shear taking place in the metal. This effect is discussed in detail in reference 10. The experiments were conducted in a vacuum of 10^{-11} torr (mm Hg) at a sliding speed of 0. 001 centimeter per second and at loads from 50 to 1000 grams.

STACKING FAULTS

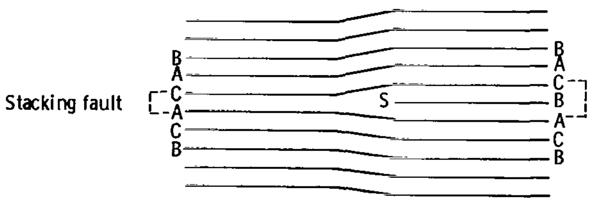
The general topic of stacking faults is effectively treated in chapter 3 of reference 9.

The atoms of closely packed structures, such as the face-centered cubic and closely packed hexagonal, are arranged in specific well-defined layers which repeat throughout the structure. In the face-centered-cubic structure the stacking sequence is ABCABCABC, and in the closely packed hexagonal metals the sequence is ABABAB (fig. 1(a)). Should one-half of one of these layers be removed as shown in figure 1(b), a stacking fault results in the structure. Such faults can develop in the crystal during growth, or they can occur during deformation. Figure 1(a) shows, with heavy arrows, the possible displacement of a layer of C atoms. If in deformation the C atoms move to B sites instead of C, a stacking fault is produced.

Stacking faults are generated in the process of plastic deformation. These faults are generated when dislocations meet at the intersection of planes; for example, intersection of (111) planes generates a Lomer-Cottrell lock. This lock can result in a separation of the intersecting dislocation into partials as shown in figure 2. The spacing between partials is the stacking fault and the area between partial dislocations (ribbons cross hatched in fig. 2) is the faulted area. The potential or stored energy associated with the faults is the stacking-fault energy, which is essentially the energy holding the partials together. The wider the ribbon, the lower the stacking-fault energy. The

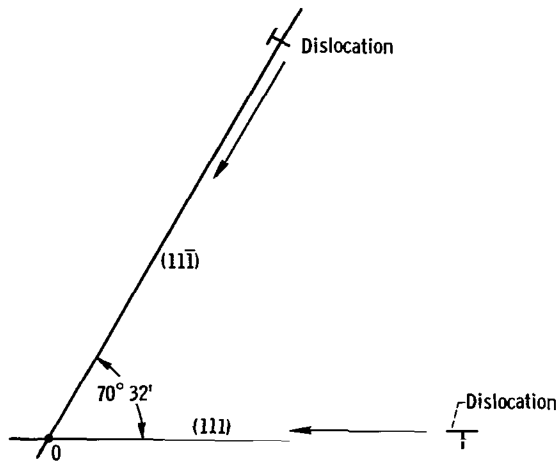


(a) Close packed plane of atoms with two possible stacking sequences, AB AB AB (close packed hexagonal) and ABC ABC ABC (face-centered cubic).

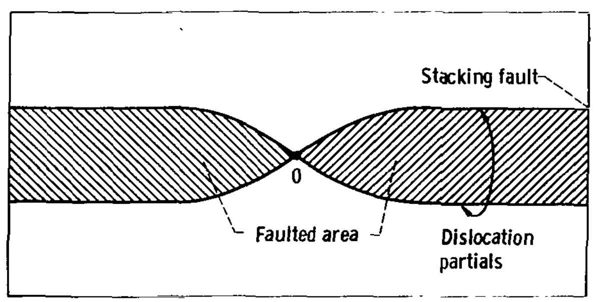


(b) Section view showing stacking fault in face-centered cubic packing of atoms.

Figure 1. - Stacking sequence and faults for atomic planes.



(a) Lomer-Cottrell lock.



(b) Slip plane (111).

Figure 2. - Model of stacking fault.

energy for copper, for example, may be 40 ergs per square centimeter with an approximate ribbon width of 10 atoms while a copper - 7.5-percent-aluminum alloy may have a fault energy of 10 ergs per square centimeter and a ribbon width of 40 atoms.

The greater the width of a ribbon, the greater the possibility of dislocation interaction. A survey of the literature (refs. 5 to 7 and 11 to 14) shows that, when there is a tendency to have a low stacking-fault energy, cold-working tends to work-harden the material at an unusually high rate.

PRACTICAL IMPLICATIONS

A detailed study of stacking faults may appear to be a subject of little direct interest to those concerned with the relation of materials for bearings, gears, or seals. This should not be the case. Since friction, wear, adhesion, fatigue, creep, texturing, and electrical resistivity are of concern to those selecting materials for lubrication systems, stacking faults must also be of concern.

It is well established that the deformation behavior of materials is important to the friction, wear, and adhesion of metals in contact. Deformation is related to crystal structure, and the effect of crystal structure on friction, wear, and adhesion has been discussed (refs. 1 to 3). If the various crystal systems are examined, the differences in their behavior may be related to differences in crystallographic slip with deformation. Further study of the behavior of materials in deformation indicates that it is dislocation on slip planes which influences deformation characteristics. It is these dislocations, or line defects, in the atomic arrangement of materials which give rise to stacking faults. The work reported in this paper is, then, a logical extension of that of reference 3. It is an attempt to go a step farther in the understanding of deformation and work-hardening in metals and their influence on those properties of concern to the lubrication engineer.

MATERIALS

The single-crystal and polycrystalline copper and copper-aluminum alloys used in this investigation were all of 99.999 percent purity. The single crystals all had the (111) plane normal to the rod axis to within 3°. The polycrystalline aluminum oxide was of 99.8 percent purity and high density (99.9 percent), and the average grain diameter was 0.023 millimeter.

After being shaped into specimens, the single-crystal and polycrystalline metals were all electropolished with orthophosphoric acid to remove any worked or deformed surface layer. The orientations were then checked with the Laue back-reflection X-ray technique. The specimens were thoroughly rinsed with acetone and alcohol immediately prior to friction and wear experiments in the vacuum chamber.

The reagent used for dislocation etch pitting in this investigation was the same as that described in reference 15. It consisted of four parts saturated ferric chloride solution, four parts hydrochloric acid, one part acetic acid, and a trace of bromine.

Polycrystalline aluminum oxide was selected as a mating surface because of the results obtained with copper sliding on it in an earlier investigation (ref. 10). Results of this prior investigation showed that, with the metal sliding on polycrystalline aluminum oxide, adhesion of the metal to it occurred with shear taking place in the metal at the sliding interface. The friction-determining factor is then based upon the shear properties of the metal. With the metals sliding on themselves in vacuum, the large increase in true contact area with tangential displacement results in complete welding.

APPARATUS

The apparatus used in this investigation is shown in figure 3. The basic elements of the apparatus were the test specimens $\left(a\ 2\frac{1}{2}\text{-in.-diam flat disk and a } \frac{3}{16}\text{-in.-rad rider}\right)$ mounted in a vacuum chamber. The disk specimen was driven through a magnetic drive coupling. The coupling consisted of two 20-pole magnets spaced axially 0.150 inch apart with a 0.030-inch diaphragm between magnet faces. The drive magnet outside the vacuum system was coupled to a low-speed electric motor. The driven magnet within the chamber was completely enclosed with a nickel-alloy housing (cutaway in fig. 3) and was mounted at the upper end of the shaft. The disk specimen was at the lower end of the shaft.

The rider specimen was supported in the specimen chamber by an arm mounted by gimbals and sealed by a bellows to the chamber. A linkage at the end of the retaining arm from the rider specimen was connected to a strain-gage assembly. The assembly was used to measure frictional force. Load was applied through a dead-weight loading system. A 500-liter-per-second ionization pump and a vac-sorption forepump were attached to the lower end of the specimen chamber. The pressure in the chamber adjacent to the specimen was measured with a cold-cathode ionization gage. A diatron-type mass spectrometer (not shown in fig. 3) was used for determination of gases present in the vacuum system. A 20-foot, 5/16-inch-diameter stainless-steel coil was used for

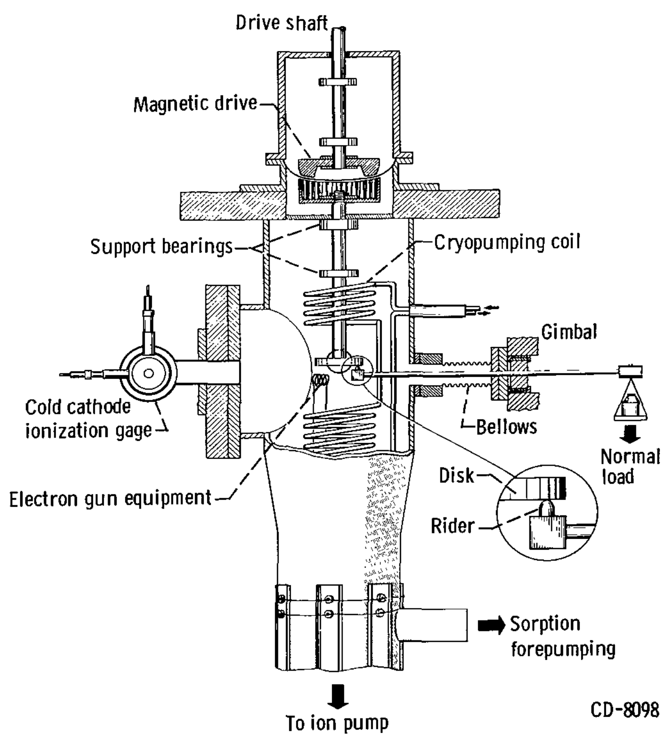


Figure 3. - Vacuum friction apparatus.

liquid-nitrogen and liquid-helium cryopumping of the vacuum system. The vacuum chamber and specimens were baked out at 200° C for 16 hours prior to each experiment.

PROCEDURE

The disk specimens of polycrystalline aluminum oxide were scrubbed with levigated alumina and rinsed with water and then alcohol prior to insertion in the vacuum chamber. The metal specimens after electropolishing were rinsed with acetone and then alcohol prior to being placed in the vacuum chamber. After pumpdown, the entire vacuum system was baked out overnight. The specimens were then electron bombarded for 30 minutes to remove residual surface oxides and contaminants. During electron bombardment, the specimen temperature was 400° C. The specimens were then allowed to be cooled to room temperature and the experiment was started immediately.

RESULTS AND DISCUSSION

The severe plastic deformation and high interface temperature which can be achieved in sliding friction experiments create the probability of interfacial recrystallization. In order to determine the mechanical conditions at which surface recrystallization of copper occurs, friction experiments were conducted in vacuum with single-crystal and polycrystalline copper sliding on aluminum oxide at various loads. The sliding speed of these experiments was very low, 0.001 centimeter per second. The results obtained in the experiments are presented in figure 4 and are taken from reference 16.

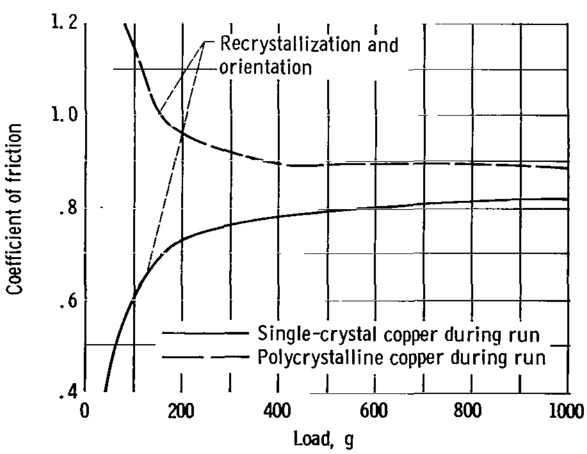


Figure 4. - Coefficient of friction for single-crystal and polycrystalline copper sliding on polycrystalline aluminum oxide in vacuum (10^{-11} torr). Copper oriented with (111) plane parallel to sliding interface; sliding velocity, 0.001 centimeter per second; no external specimen heating.

Figure 4 indicates that, at a load of 50 grams, a threefold difference in friction coefficients exists between copper single crystals with the (111) plane parallel to the sliding interface and randomly oriented polycrystalline copper. Increasing the load results in an increase in the coefficient of friction for the copper single crystal and a decrease in the coefficient of friction for the polycrystalline copper.

If the loading were continued to beyond 1000 grams, at some point the friction coefficients would become essentially the same.

The marked change in friction with increasing load for the two forms of copper may be related to surface recrystallization. X-ray analysis of the sliding surface after sliding at 1000 grams indicated the presence of an oriented polycrystalline film on the wear surface of both the single-crystal and polycrystalline materials. Although the X-ray technique did not detect a recrystallized film at the light loads, electron diffraction revealed a recrystallized surface film on the single crystal at a load of 100 grams. A recrystallized film was not noted at 50 grams. In the subsurface of the wear area a high concentration of Kikuchi lines was observed. Diffraction patterns in the wear area showed severely deformed material.

The difference in friction coefficient for the single-crystal and polycrystalline copper at a load of 50 grams is as anticipated from shear and deformation behavior of single-crystal and polycrystalline materials (ref. 16). It is interesting to note that, although a recrystallized surface film was observed by electron diffraction for the 100-gram load condition, the friction coefficient was still half that obtained for the polycrystalline material. These results indicate that it is not only the thin film at the interface but also the bulk rider structure which is important. This becomes more evident at larger loads, where the friction coefficients continue to approach one another. It would appear that, as the recrystallized interface film thickens, the influence of the bulk material on friction becomes less significant.

The foregoing experiments were preliminary and necessary to determine the operating conditions for single-crystal studies with copper-aluminum alloys. Because of these results, stacking-fault energy studies were conducted with copper-aluminum crystals at a load of 50 grams and a surface speed of 0.001 centimeter per second with aluminum oxide as the mating surface. Since copper did not recrystallize with the 50-gram load, it may be anticipated that the copper-aluminum alloys will not, because alloying increases the recrystallization temperature.

A considerable amount of research has been done on stacking faults and their influence on mechanical properties. Of particular interest in this investigation is their influence on work-hardening and friction. Stacking-fault energy data, critical resolved shear stress, and friction data for copper-aluminum single crystals with the (111) plane parallel to the interface are presented in figure 5.

The stacking-fault energies plotted in figure 5 were obtained from reference 13, and the slope of the curve was confirmed by the data of reference 11. The stacking-fault energy values for the metals and alloys may vary in different references. For example, reference 11 gives a stacking-fault energy for copper of 40 ergs per square centimeter, and reference 8, a value of 70 ergs per square centimeter. All references examined (refs. 4, 12 to 14, and 17) agreed that stacking-fault energy decreased with increasing

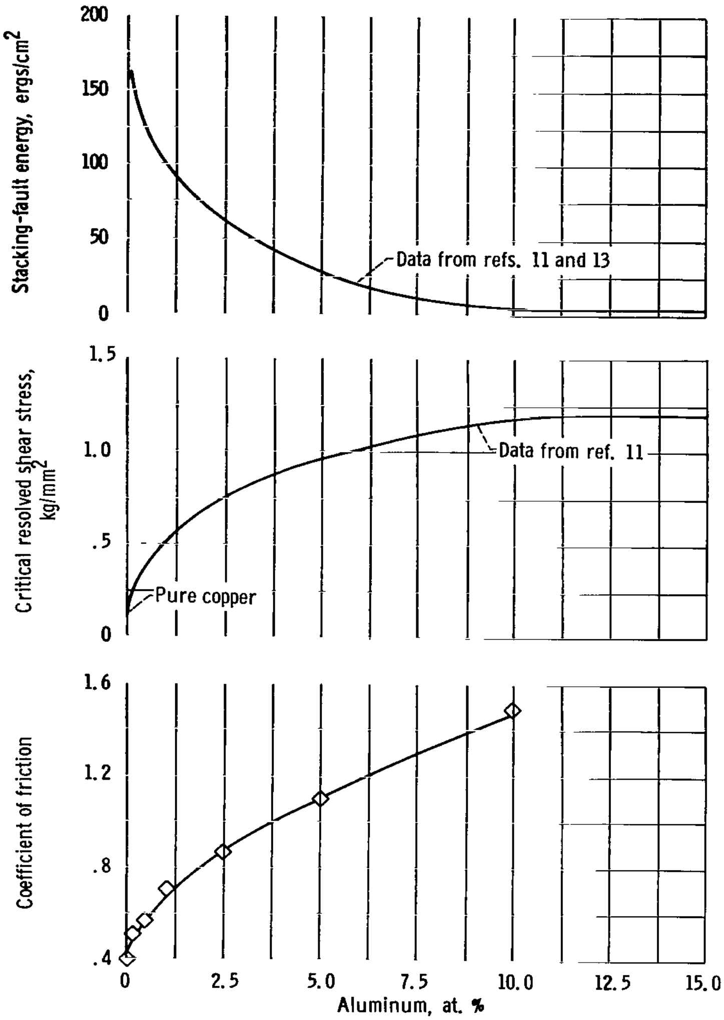


Figure 5. - Coefficient of friction, critical resolved shear stress, and stacking-fault energy for copper-aluminum single crystals sliding on polycrystalline aluminum oxide in vacuum (10^{-11} torr). Rider oriented with (111) plane parallel to sliding interface; load, 50 grams; sliding velocity, 0.001 centimeter per second.

(a) Multiple slip lines in wear area. X250.

(b) Same region as (a) at magnification of 1100.

(c) Slip lines 0.003 millimeters from wear area. X250.

Figure 6. - Unetched wear scars for copper single-crystal rider specimens sliding on polycrystalline aluminum oxide in vacuum (10^{-11} torr). Copper oriented with (111) plane parallel to sliding interface; load, 50 grams; sliding velocity, 0.001 centimeter per second; duration, 1 hour.

aluminum content in copper. The data of reference 13 were used because the crystals were all of the same orientation as those examined in this study.

The critical resolved shear stresses of figure 6 were also obtained from reference 13. They indicate, as might be anticipated, an increase in shear stress with additions of aluminum to copper.

The friction data of figure 5 indicate a marked change in friction properties with the addition of aluminum to copper. The deformation behavior of pure copper and the alloy of 10.0 atomic percent aluminum in copper can be seen in figures 6 and 8. Figure 6 shows the wear area and the pronounced number of slip lines present about it for the 50-gram load.

In figure 6(a) it is interesting to note that there are two sets of slip lines crossing at about 60 degrees. When the rider wear area was completely in view, there could actually be seen three such sets. They formed a triangular pattern about the wear scar. Unfortunately, if the magnification were reduced to show the third set, it would be difficult to show the slip lines photographically. Examination of a sterographic projection for the cubic system onto the (111) plane indicates that the three sets of slip lines intersecting in a triangular form about the wear area are (111) slip planes (70^{o} 32'). In figure 6(b) (higher magnification) the slip lines appear to be very closely spaced adjacent to the wear scar and become separated at greater distances. Since the greatest amount of deformation occurs in the wear area and decreases with increasing distance from the wear area, these results are not surprising.

Attempts were made to measure the maximum distance away from the wear scar that slip lines could be detected (fig. 6(c)) to approximate the depth of deformation in the crystal. An approximation would be 0.003 millimeter. The actual depth is greater than this since optical observation of slip lines indicates stage II of the stress-strain curve for copper single crystals. Electron microscopy is necessary to detect slip lines in the easy glide (stage I) region.

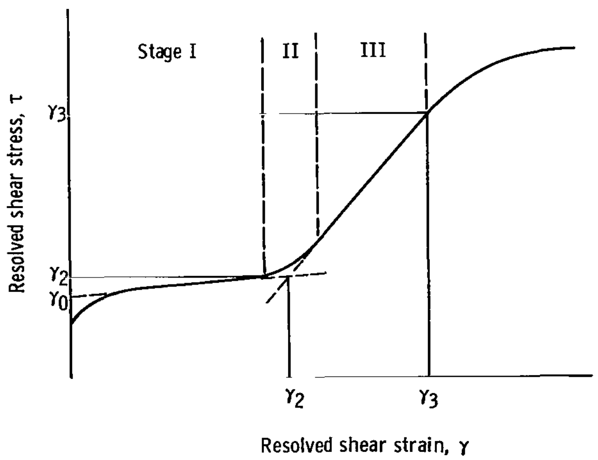
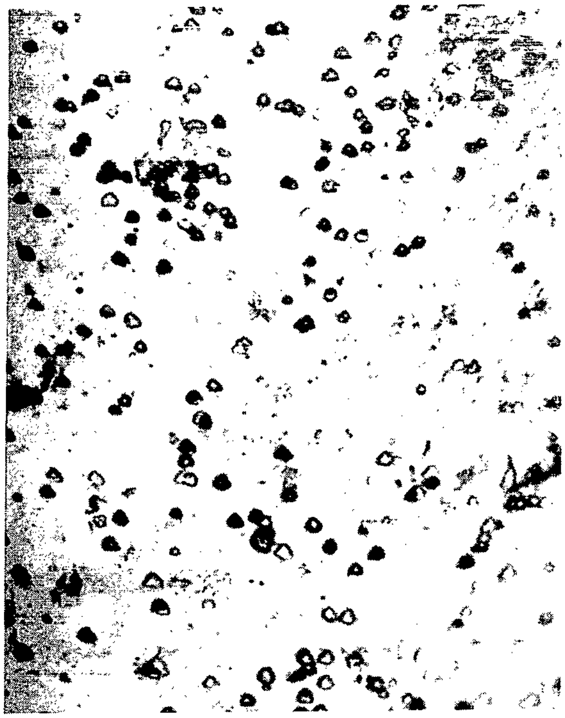


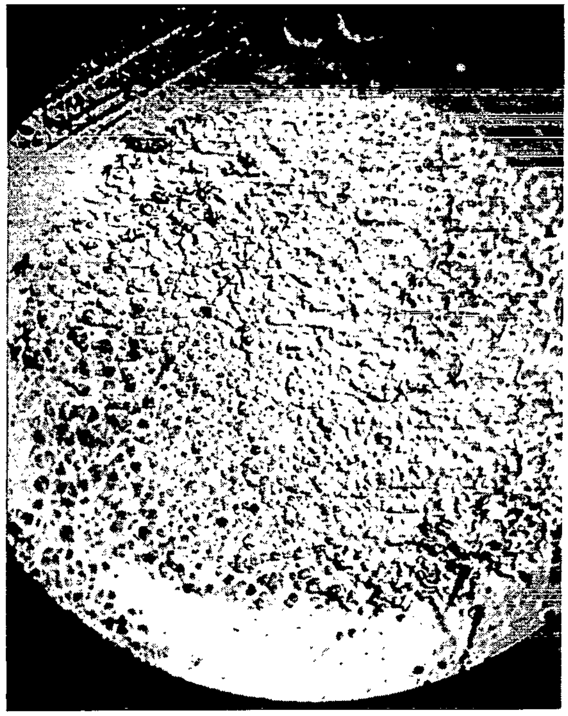
Figure 7. - Flow curve for face-centered single crystals (ref. 18).

The three stages or regions of the copper stress-strain curve are shown in figure 7. They are stage I (the easy glide region), where slip plane dislocations move freely with essentially no hardening; stage II, the region where dislocation interactions occur and cause marked work-hardening; stage III, which is characterized by cross slip.

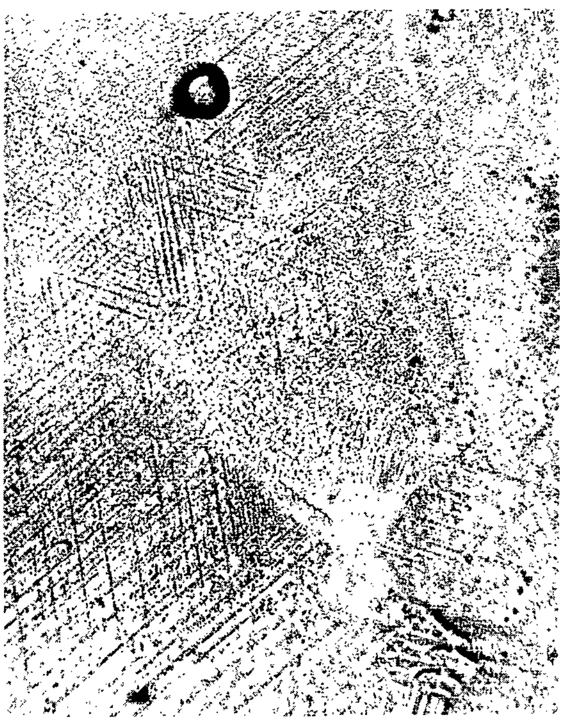
Figure 8 indicates the etch-pitted wear area of the 5.0-atomic-percent-aluminum - copper alloy. The crystal shown in



(a) Annealed (pretest) rider etch-pitted. X1100.



(b) Rider wear area etched and polished. X450.



(c) Rider wear region etch-pitted subsurface. X600.

(d) Rider wear region etch-pitted subsurface. X1100.

Figure 8. - Etch-pitted wear area of 5.0-atomic-percent aluminum - copper single crystals. Mating surface, aluminum oxide; copper oriented with (111) plane parallel to sliding interface; load, 50 grams; sliding velocity, 0.001 centimeter per second; ambient pressure, 10^{-11} torr.

figure 8(a) is etch-pitted prior to the friction experiment. The concentration of etch pits gives some indication of the dislocation concentration in the as-grown crystal. Figure 8(b) shows the wear area etch-pitted and then electropolished for a very short time period. The polish was added because with simple etch-pitting the wear scar was completely black with pits and details of the pits were not readily discernable. Figures 8(c) and (d) whow the wear area subsurface with a high concentration of etch pits along the (111) planes adjacent to the wear area. Close observation will reveal the same triangular pattern of slip lines referred to with the copper single crystals.

Since stacking faults influence the work-hardening characteristics of the alloys examined, the friction coefficients reported in figure 6 are not initial coefficients but values obtained after some sliding. This was done to ensure that the crystals had reached some equilibrium interfacial condition.

The friction coefficients for the pure copper and the alloy of 10 atomic percent aluminum in copper are plotted and presented in figure 9. A marked decrease in friction occurred for copper in the first 30 minutes of the experiment, and the friction then was essentially unchanged with further increases in time. With the alloy, an increase in friction was observed initially, and again the friction became essentially unchanged with further increases in time.

The results of figures 6, 8, and 9 must be examined in terms of dislocation concepts. The essential difference in work-hardening behavior for high- and low-stacking-fault-energy metals is believed to occur at the onset of stage III of the stress-strain curve, and this difference is explained in terms of the cross-slip mechanism in reference 11. For high-stacking-fault-energy materials, stage III begins at a relatively low stress before stage II hardening is extensive. This is in contrast to the low-stacking-fault-

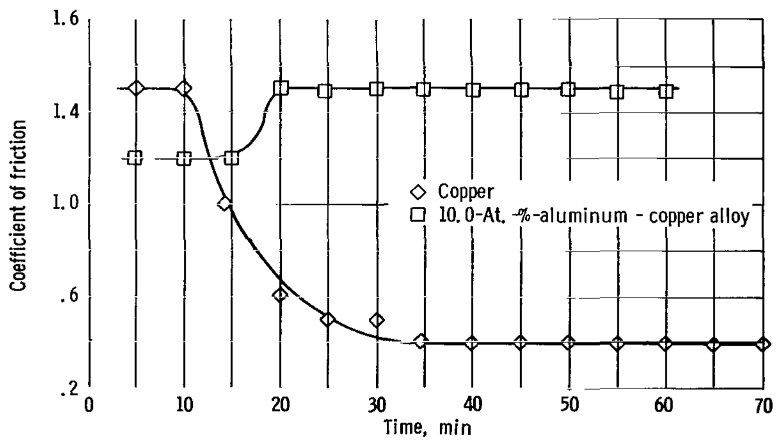


Figure 9. - Coefficient of friction for single-crystal copper and 10-atomic-percent-aluminum - copper alloy sliding on polycrystalline aluminum oxide in vacuum (10^{-11} torr). Copper oriented with (111) plane parallel to sliding interface; load, 50 grams; sliding velocity, 0.001 centimeter per second; no external specimen heating.

energy materials, in which the activation energy for cross slip is markedly higher. Cross slip is a form of softening which may be termed "dynamic recovery" to distinguish it from recovery occuring as a result of thermal activation.

The data of figure 6 are plots of friction coefficients obtained after steady-state conditions had been achieved (as illustrated in fig. 9). Examination of the photomicrographs of figures 7(b) and 8(d) indicates that these data were obtained in stage III. Multiple slip is visually evident in these figures for both the pure copper and the copper-aluminum alloy. The data of figure 9 for the copper-aluminum alloy in the first 20 minutes are believed to reflect the later phase of stage II, namely, continued work-hardening before the onset of cross slip (stage III). For pure copper the decrease in friction represents the effect of cross slip.

The results of the experiments reported herein reflect the influence of the condition of the bulk single crystal on interfacial behavior. Since, as it was pointed out earlier, adhesion and shear of the copper take place at the sliding interface, the metal adhering to the aluminum oxide is of no great concern. The shear properties (influenced by the degree of work-hardening) of the bulk rider about to adhere to a new aluminum oxide surface are of prime importance. Once adhesion again occurs, shear and the friction force are determined in the base metal. Since the entire process of sliding is dynamic, a balance between severe work-hardening and cross slip or dynamic softening must be considered to exist. The easier it is for cross slip to occur and relieve the highly work-hardening condition (high shear stress), the lower will be the observed friction force.

It should be indicated here that the entire process of sliding friction and the type of metal interface obtained are a function of energy input. As observed in figure 4, if the load is sufficiently high, surface recrystallization of the metal will occur. If the load is reduced (less energy), twinning may be observed. If the load is reduced still further, severe deformation and cross slip are observed with single crystals. From the results of this investigation, it can be seen that all these metal states will influence friction behavior.

SUMMARY OF RESULTS

The results of friction experiments conducted in vacuum with single-crystal copper and single-crystal copper-aluminum alloys can be summarized as follows:

Decreases in stacking-fault energy resulted in increases in the critical resolved shear stress and this may be related to friction coefficient. In order to minimize the

effects of work-hardening at interfacial junctions, therefore, it is desirable to have a high stacking-fault energy in metals and alloys used as slider materials.

Lewis Research Center,
National Aeronautics and Space Administration,
Cleveland, Ohio, September 9, 1966,
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