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INTRODUCTION

There are many reasons to believe that the method by which a surface is produced can strongly influence the friction and wear properties of such a surface. Conventional machining involving chip removal, either by single point tools or by grinding wheels, can affect the hardness and structure at the surface. Also, nonconventional material removal processes such as electrochemical machining (ECM), electrical discharge machining (EDM), chemical machining (CHM), electron beam machining (EBM), and laser beam machining (LBM) can influence the nature of the surfaces. It would therefore appear that both the physical and the chemical nature of the surfaces are strongly dependent on the conditions under which the machining takes place.

It is the objective of this paper to explore how friction and wear are influenced by the surface state. In order to accomplish this objective, a review is made of the fundamental mechanisms involved in both friction and wear. The adhesion theory of friction is discussed, and the important parameters are pointed out. The strong influence of surface contaminants on the friction process is explored in some detail. The various wear processes (adhesive, abrasive, corrosive, and fatigue) are enumerated and two of them are explored. The influence of crystal structure on friction and wear is discussed, and comparisons are made between the hexagonal and cubic structures as well as differences between single crystals and polycrystalline materials. Finally, the resistance to wear, as influenced by hardness, is discussed. The three principal ways of increasing hardness treated herein are (1) alloying, (2) quenching and tempering, and (3) work hardening.

ADHESION THEORY OF FRICTION

Friction and wear characteristics can best be explained using the adhesion theory of friction. This adhesion theory of friction was advanced by Merchant in the United States (ref. 1) and by Bowden and Tabor in England (ref. 2). This theory of friction is based on strong adhesive forces between contacting asperties. As the load is applied (see fig. 1) the asperities come into metallic contact, with resulting high stresses at the true contact area. The true area of contact is so small that, following elastic deformation, the stress quickly reaches the yield stress of one of the two materials. Hence, plastic flow occurs and a "cleaning" action is obtained at the contact area (some of the surface contaminants are forced out). Because local areas are now somewhat clean and the stress is relatively high, "cold welding" can occur at the junction(s). Moving one surface relative to the other requires shear at these welded junctions.

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The adhesion theory of friction states that the friction force is equal to the sum of two terms: the first a shear term, the second a ploughing or roughness term. The shear term is that force required to shear at the welded junctions (note that shear may take place in the junction itself or adjacent to it in either of the two contacting materials). The ploughing term is that force which results from displacement of the softer of the two metals by an asperity of the hard metal. In many instances, the ploughing or roughness term is negligible in comparison with the shear term.

Friction:

$$F = S + P = \text{Shear} + \text{Ploughing} = As + A'p \qquad (1)$$

where

A real area of contact

A' ploughing area

p flow pressure

S shear strength of junction

Contact area:

$$A = \frac{\text{Load}}{\text{Flow Pressure}} = \frac{W}{p} \qquad (2)$$

Friction coefficient:

$$f = \frac{\text{Friction}}{\text{Load}} = \frac{As}{W} + \frac{A'p}{W} = \frac{s}{p} + \frac{A'p}{W} \qquad (3)$$

When the ploughing term is negligible,

$$f = \frac{s}{p} = \frac{\text{Shear strength}}{\text{Flow pressure}} \qquad (4)$$

Equation (4) shows that reduction of friction coefficient (and usually, reduction in wear) can be obtained if the ratio s/p can be reduced. Reduction of the ratio results from low shear strength, high flow pressure, or both. By the use of low shear strength films (with thicknesses as small as millionths of an inch) on hard base materials, both desirable conditions may be obtained (ref. 2). Thus, low shear strength is achieved without appreciable decrease of the yield strength of the combination. The load will be supported through the film by the hard base material while shear occurs within the soft thin film. These low-shear-strength films can be of the following types: oxides, chemical reaction films (chlorides, sulfides, etc.) metals, fluid lubricants, solid lubricants, etc. In fact, any low shear strength material which is present as a contaminant on the surface can serve to reduce friction and, perhaps, to reduce wear.

INFLUENCE OF SURFACE CONTAMINANTS ON FRICTION

It is possible to show, by use of monolayers and multilayers, that an extremely thin film of contaminant at the surface can be effective in reducing friction. Data from reference 2 show that, with a long chain fatty acid (stearic acid) deposited on stainless steel, the friction coefficient even with a single layer (i.e., a monolayer) is relatively low, about 0.1. The friction however rises rapidly as continued runs are made over the same track. These data are shown in figure 2. With either a monolayer or multilayers of 3, 9, or 53 films the friction coefficient starts at about the same value, 0.1. The greater the number of films, however, the longer it takes to wear off or displace this protective film and consequently the longer the time in which the film is an effective boundary lubricant. The monolayer of stearic acid has an approximate thickness of 20 angstroms (about 10^{-7} inch). It is therefore obvious that any contaminanting film of low shear strength on the surface, even at thicknesses of the order of millionths of an inch, can be effective in reducing friction and also wear.

Additional data are available to show that the effects of contaminants on friction and wear can be quite appreciable, even though the contaminanting film may be as thin and poorly adherent as that obtained from adsorption of a gas. Figure 3, for example, shows the reduction in friction obtained by adsorption of oxygen on outgassed iron surfaces. These surfaces have been outgassed in vacuum (approximately 10^{-6}mm Hg) in order to clean them. These data, from reference 3, show that friction is markedly reduced by admission of oxygen gas even though the oxygen pressure is very low (10^{-4} mm). If the concentration of oxygen atoms is increased by increasing the pressure (in this case, 10^{-3} mm), the friction coefficient is reduced still more. At higher pressures (several millimeters of mercury), the friction coefficient is reduced even further. Finally, figure 3 shows that, if the surfaces are allowed to stand for some period of time, the adsorbed oxygen film becomes more complete and the friction drops still further. An important observation from these results, however, is that seizure of the clean metals is prevented by even a <u>trace</u> of oxygen. It is quite probable that, in these experiments, the oxygen reacted with the clean iron surfaces to form one of the lower oxides of iron (FeO or Fe_3O_4). Both of these oxides are effective in preventing seizure of the contacting iron surfaces.

The effect of chlorine on the friction of iron surfaces is shown in figure 4. As chlorine is admitted at a pressure of about 1 millimeter of mercury, the friction coefficient drops to a relatively low value. This effect is nonreversible, as shown in figure 4 at the point where the chlorine was frozen out and then readmitted. No appreciable change in friction occurs under these conditions. In fact, it was necessary to heat the surfaces to a temperature of about 400° C before the resulting film was broken up and the friction again rose appreciably. Quite probably, with clean iron surfaces, chemical reaction took place at the surface to form an iron chloride film (possibly $FeCl_2$). The indicated necessity to heat the surfaces to 400° C probably corresponds to the temperature required to decompose this film.

The effect of adsorption of hydrogen sulphide gas on the friction of

outgassed iron surfaces is shown in figure 5; the friction is reduced abruptly and appreciably. Seizure of the surfaces was prevented. The reduction of friction with hydrogen sulphide is not as great as it is with chlorine. Note, however, that this film is much more stable and may be heated to over 790° C before decomposition of the film takes place and friction rises. It is possible that, with hydrogen sulphide, reaction occurs to form the iron sulphide, FeS. The iron chloride film, $FeCl_2$, has a lower shear strength then FeS (ref. 2); this difference in shear strength would explain the difference in friction coefficient.

An important point to be noted here is that admission of either oxygen, chlorine, or hydrogen sulphide to the chamber resulted in prevention of seizure of the metal surfaces. Frequently, the prevention of seizure can be more important than reduction of friction, per se. This protection was obtained even though the surface films are extremely thin.

Buckley (ref. 4) has conducted some experiments with oxygen chemisorbed on tungsten. These experiments were done in vacuum (10^{-10} Torr) with tungsten cleaned by electron bombardment. Oxygen was admitted into the vacuum chamber in such a way as to chemisorb atoms to various degrees. The surface exposure, in Torr-seconds, produced oxygen adsorption as shown in figure 6 (from ref. 4). It is noted that the oxygen adsorption was less than a monolayer in thickness until the higher values of surface exposure. Friction data were obtained in these same experiments at various values of surface exposure. With tungsten in the clean state, the friction coefficient was 3.0. With oxygen adsorption equal to approximately a quarter of a monolayer (monolayer thickness $\cong 3$ Å), the friction coefficient drops very markedly from 3.0 to 1.6, figure 6. It is apparent from the data shown in figure 6 that oxygen adsorption equivalent to as little as three quarters of a monolayer is enough to reach a steady state in the friction value, approximately 1.3. These results again prove the point that the surface need be contaminated to only a <u>slight</u> degree in order to have a marked effect on friction.

As is well known, if the surfaces are sufficiently clean, cold welding (friction welding) of the surfaces may take place when sliding is attempted. Figure 7 shows some data obtained at the NASA-Lewis Research Center with the steel 52100 at a vacuum of 10^{-7} Torr obtained by cryopumping. These data, which are from reference 5, as well as much of the data on friction and wear discussed herein are reviewed in some detail in Bisson and Anderson (ref. 6). For the case of the data in figure 7, a liquid-helium condensing coil inside the vacuum chamber condensed the condensible gases such as nitrogen and oxygen. In this manner, the authors of reference 5 felt that availability of oxygen atoms would be markedly reduced from that using ordinary ion pumping. Figure 7 confirms their belief. The friction coefficient showed a slight increase from its initial value of 0.3 to the value of about 1.0 at 30 minutes. At 30 minutes, the friction coefficient rose markedly to a value of about 4.5 after which it continued rising until the specimens welded so firmly that the drive motor of the mechanism was stalled. The initial low friction coefficient is believed to be the result of the presence of the lower oxides of iron (FeO and Fe_3O_4). The time in which the friction coefficient remained relatively low (i.e., less than 1.0) represents the time required to wear these beneficial oxides from the

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surface. After the oxide film has been worn from the surface, it could not reform because of the limited availability of oxygen atoms at 10^{-7} Torr. Hence, cold welding of the surfaces took place.

WEAR

Burwell (ref. 7) has published an excellent survey of possible wear mechanisms; much of the discussion in this section is from this reference. He points out that wear can be classified in at least four principal, distinct, and independent phenomena:

1. Adhesive or galling wear

2. Abrasive and cutting wear

3. Corrosive wear

4. Surface fatigue

Additional experimental and analytical investigations of various wear processes are covered in references 6, 7, and 8 to 14.

Adhesive Wear

In the discussion of adhesive wear, Burwell (ref. 7) notes that one can write an equation for wear as

$$V = kAL \qquad (5)$$

where

    V volume of wear material

    k wear coefficient

    A real area of contact

    L distance of travel

Since the real area of contact is equal to the ratio of load W to hardness H (or flow pressure p), this ratio can be substituted in equation (5). If we make the substitution, we obtain:

$$V = k \frac{WL}{H}$$

or, for a given material,

$$V = k'WL \qquad (6)$$

If we divide both sides of equation (6) by the apparent contact area A_o, the equation becomes

$$h = k \frac{PL}{H} = k'PL \text{ (for a given material)} \qquad (7)$$

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where the average depth of wear $h = V/A_O$ and the average pressure or design stress $P = W/A_O$. In other words, for a given material, the adhesive wear coefficient k' is given by the ratio

$$k' = \frac{h}{PL} \qquad (8)$$

Figure 8 shows the results of some wear experiments conducted by Burwell. We see first that wear is a linear function of distance and, second, that it is an apparent function of load. Figure 9 gives results obtained at the NASA laboratories, which show that wear is a linear function of load and is completely independent of apparent area of contact or apparent contact stress.

The results of figures 8 and 9 show that equation (6) is applicable. In other words, k' is a constant characteristic of the material over a range of speeds and loads.

Burwell continued his experiments at higher loads and found some peculiar results. As the load was increased to a point generally exceeding the range of accepted engineering design, it was found that the adhesive wear coefficient k' was no longer constant but increased rapidly with load (that is, with increase in average compressive stress). These results are shown in figure 10 which is a plot of the adhesive wear coefficient k' against pressure (average stress). These curves show that the value of wear coefficient is constant up to a value of average pressure which is approximately one-third the indentation hardness. Above this pressure, the wear coefficient rises sharply and the curve is finally terminated by the onset of large scale welding and seizure. The curve of figure 10(b) is for the same steel as that for figure 10(a) except that the steel has been hardened to about twice the Brinell hardness. The curve for the hard steel again shows that the wear coefficient is constant up to a value of approximately one-third of the identation hardness. It will be noted in this case, however, that the average pressure is appreciably higher because of the higher hardness.

Archard (refs. 11 to 13) has presented a model of adhesive wear in which he makes two assumptions: (1) each time asperities come into contact to form a junction there is a constant probability that an adhesive fragment will be formed, and (2) each fragment is assumed to be a hemisphere of diameter equal to the junction diameter. Using these assumptions, he develops an equation which is identical to Burwell's with the exception that Burwell's wear coefficient has been replaced with a wear coefficient k/3 which is different by the factor 1/3. This factor is a shape factor applicable in the assumed case of circular junctions and hemispherical fragments.

Rabinowicz (ref. 8) presented a theory of wear involving interfacial energies. In Rabinowicz's theory, which involves stored elastic energy in a wear particle and adhesional energy acting at the interface, a transferred particle comes off the surface only if the elastic energy is greater than the adhesional energy. Using these assumptions, he derived an equation which predicts the diameter of wear particles as a function of two materials properties p (the flow pressure) and W_{ab} the interfacial energy between materials a and b. Rabinowicz's experiments with various materials

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(metals and nonmetals) provide some support for his equations. On the other hand, Tabor has stated that some research results in England and Russia seem to support the view that the formation of wear fragments is primarily by fatigue of the transferred material.

Abrasive Wear

Burwell points out that abrasive and cutting wear are, in general, the same type of damage to the surface. Damage is accomplished by a hard surface plowing or gouging out a softer surface. There are two general situations for this type of wear: (1) the hard surface in question is the harder of the two rubbing surfaces (cutting wear); or (2) the hard surface is a third body, generally a small particle of grit or abrasive caught between the two surfaces and sufficiently harder than these surfaces that it abrades either one or both of them (abrasive wear). Cutting wear is strongly influenced by the choice of the combination of materials while abrasive wear is strongly influenced by the type of foreign object (abrasive) present between the surfaces. Frequently, the foreign objects are formed by chemical reaction with the surrounding atmosphere.

Kruschov and Babichev (ref. 15) correlate resistance to wear with hardness of various "technically pure metals". These results are shown in figure 11. The points labeled "40" and "y12" are for carbon steel specimens of 0.41 and 1.1 percent carbon, respectively. As noted in the figure, these points fall on the same curve as for the technically pure metals.

Spurr and Newcomb (ref. 16) have obtained wear results as a function of hardness which show agreement with Kruschov and Babichev's results.

Kruschov and Babichev also found (ref. 15) that alloys gave results similar to those shown in figure 11 for technically pure metals. Figure 12 (from ref. 15) shows general trends of resistance to wear as a function of three principal ways of increasing hardness: (1) alloying, (2) quenching and tempering, or (3) work hardening. As indicated, alloying to increase hardness should result in an increase in the resistance to wear; similarly quenching and tempering, which results in increased hardness, should also give increased resistance to wear. On the other hand, work hardening (strain hardening), of the surfaces should have no effect since abrasive wear is itself a work-hardening process (according to ref. 15). Hence wear resistance, as measured in abrasive wear tests, is a function of materials in their maximum work-hardened state.

In confirmation of the general trends shown in figure 12, data were obtained with a number of steels of various compositions and various heat treatments. The results of experiments with four steels after quenching and tempering at different temperatures are shown in figure 13 (from ref. 15). The steel labeled X-12, containing a high percentage of carbon and of chromium, shows appreciably higher wear resistance than do the other three steels. The data at maximum hardness for each of the four steels represents the steel "as quenched" or "as-quenched and mildly tempered". For each of these steels, points at lower hardness than the maximum represent different degrees of tempering.

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Figure 14 (from ref. 15) shows relative wear resistance as influenced by work hardening for a number of different materials; these materials include aluminum, copper, nickel, and a steel of 0.45 percent carbon. The nearly horizontal lines at a particular value of resistance to wear represent specimens which have been work hardened to different degrees either by roll hardening or by shot peening. It will be noted that work hardening has no influence on the relative wear resistance.

INFLUENCE OF CRYSTAL STRUCTURE ON FRICTION AND WEAR

Hexagonal Versus Cubic

Recent friction studies in vacuum lubrication at the NASA-Lewis Research Center indicate a marked difference in friction and wear between metals of cubic and hexagonal crystal structures (refs. 17 to 20). Figure 15 shows the atomic arrangement in typical face-centered-cubic (F.C.C.) and close-packed-hexagonal (C.P.H.) crystal lattices. Polycrystalline metals are agglomerates of crystallites that have these basic unit cells; when welding occurs between two metals, the weld is made up of these crystals. When the crystals in the weld shear, they do so along distinct planes and the required shear force depends on the plane being sheared. Shear forces in cubic crystals are normally greater than corresponding shear forces in hexagonal crystals because of work hardening of cubic crystals as well as existence of easy slip planes in hexagonal metals.

The larger the number of slip systems, the greater is the tendency to work harden because of cross slip. Face-centered-cubic metals possess 12 slip systems whereas close-packed-hexagonal metals possess only 3 slip systems at low values of c/a. At higher values of c/a, close-packed-hexagonal metals possess 9 slip systems.

In hexagonal crystals, shear forces are usually the least on the basal plane (i.e., when the shear occurs in the plane parallel to the hexagons, fig. 15). This shearing process is illustrated in figure 16; the top hexagonal plane of the crystal is shown to be displaced from the normal axis by the shear deformation process. As surfaces are moved with respect to one another, deformation, shear separation, and recrystallization occur as a continuing process.

The data of figure 17 show the differences in the force required to shear metals of cubic and hexagonal structures. The crystal form of cobalt at normal temperatures is hexagonal; cobalt, however, transforms from the hexagonal to the cubic structure when heated above 400° C. A marked increase in friction is shown to accompany this crystal transformation (fig. 17). At low temperatures, the sliding is hexagonal cobalt on hexagonal cobalt; at the higher temperatures, it is cubic cobalt on cubic cobalt. The transition from hexagonal to cubic is shown at less than 400° C because friction heating caused the surface temperatures to be somewhat higher than the bulk metal temperatures measured. Adhesive wear rate was about 100 times greater for the cubic cobalt than for the hexagonal cobalt (see the wear rates at temperatures of about 260° and 370° C, fig. 17). Furthermore, at about 480° C, complete welding of these specimens occurred. These data suggest that metals which remain in the hexagonal crystal form over

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the entire operating temperature range are to be preferred for sliding applications.

Additional inquiry showed that the shear force in hexagonal crystals varies with the relative spacing of the atoms within the crystals. In particular, the shear force is controlled by the ratio of the distance c (the spacing between hexagonal planes) to the distance a (the spacing between adjacent atoms in the hexagon). Various metals with hexagonal crystal structures have different values of c/a. Figure 18 shows the variation of friction, in vacuum, for some of these metals. Generally, friction declines with increasing c/a, and those metals that showed low friction give no evidence of gross surface welding. Titanium is well known as a metal subject to severe welding or galling and otherwise having very poor friction properties. The preceding study on crystal structure effects suggested that improved friction properties could be obtained if titanium were alloyed in such a way as to (1) stabilize the hexagonal structure over a greater range of temperature and (2) increase the c/a lattice ratio. This is necessary because its poor friction properties can be related to shear and slip mechanisms which in turn can be related to c/a lattice ratio (ref. 18).

Simple binary alloys of titanium with either tin or aluminum were found to provide the desired structural characteristics. Figure 19 shows friction and lattice ratio for a series of titanium-aluminum and titanium-tin alloys. Increasing the percentage of aluminum or tin produced a number of results: (1) increased the c/a ratio, (2) greatly reduced friction, and (3) minimized surface failure tendencies.

Single Crystal Versus Polycrystal

Buckley points out (ref. 19) that "... polycrystalline materials are aggregates of individual crystallites. The grain boundaries serve as atomic bridges to link the lattice of one crystallite with that of an adjacent crystallite. ... grain boundaries serve as a barrier to the motion of the dislocations of the crystallites. Such a structure could be expected to offer a greater resistance to shear and higher friction than single crystals." Buckley obtained data for single and polycrystal copper in sliding contact with polycrystalline aluminum oxide; these data are presented in figure 20 (from ref. 19). The results of figure 20 show that the friction coefficient at light loads for the single crystal copper is much lower than that for polycrystalline copper at the same load (0.4 versus 1.2). Buckley explains these data as follows: "As load is increased, the interface temperature increases and recrystallization occurs at both contacting surfaces. This condition represents then an increase in friction for the single crystal [and] a decrease in friction for the polycrystalline metal because recrystallization is followed by texturing, which reduces shear stress. At higher loads the friction coefficient should be the same because the interfacial surface films are the same". The results of figure 20 show, in fact, that the curves for single crystal copper and polycrystalline copper do approach one another at the higher load levels.

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X-ray Laue patterns of the crystal surfaces after the maximum load runs of figure 20 indicated recrystallization and texturing of the copper surfaces for both single crystal and polycrystalline copper.

The results with polycrystalline copper and single crystal copper were also checked with two other metals, nickel and iron. Nickel is a face centered cubic metal while iron is a body-centered-cubic metal. The results with these two metals are shown in figures 21 and 22. Both of these figures show that initially, there is a marked difference in friction between the single crystal and the polycrystalline material. Polycrystalline nickel, at light loads, shows a friction coefficient of 1.6 as compared to the single crystal coefficient of 0.4. With iron, the polycrystalline friction coefficient is 1.6 at light loads while the single crystal coefficient is 0.8. Again, as was the case with copper, an increase in load results in a decrease in friction for the polycrystalline material and an increase in friction for the single crystal until, at high loads, the friction for the two different materials is essentially the same. As before, these data are explained on the basis of recrystallization and texturing (orientation) at the interface.

It would appear that any method of production which results in recrystallization and texturing of the surface, may appreciably change the friction and wear properties of a given material. Such changes in friction and wear would be expected whether the surface production methods involve conventional machining methods or nonconventional material removal methods.

Buckley (ref. 20) indicated that there is some correlation between recrystallization temperatures for metals and their sliding friction behavior. In accordance, he attempts to correlate recrystallization temperature for various metals from the literature with their friction behavior. These data are presented in table I. This table presents the loads at which equivalent friction coefficients were obtained for the single crystal and the polycrystalline metals as well as the approximate recrystallization temperature from literature. These data show an apparent correlation.

CONCLUDING REMARKS

It would appear from the discussion in this paper that any process for producing a surface which results in contamination of that surface (with oxides or reaction products) should produce an appreciable difference in friction and wear properties. Also, any method of production of surfaces which changes the crystal size and orientation of crystallites at the surface may appreciably change the friction and wear properties of such materials. The evidence regarding the influence of work hardening of the surfaces is not as clear cut. It would appear, from the Russian researchers, that work hardening does not influence the basic resistance to "abrasive wear". It should be pointed out, however, that the resistance to adhesive wear could be appreciably influenced by the condition of the surfaces with respect to hardness and with respect to any contaminating film. Since friction and adhesive wear are strongly influenced by the yield

strength of the welded asperties, it should be expected that any change in yield strength of the material at the surface should have an appreciable influence on both friction and wear.

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Table I.

RECRYSTALLIZATION TEMPERATURES AND LOADS AT WHICH
EQUIVALENT FRICTION COEFFICIENTS WERE OBTAINED FOR
SINGLE CRYSTAL AND POLYCRYSTALLINE METALS

METAL	LOAD AT WHICH FRICTION IS EQUIVALENT FOR SINGLE AND POLYCRYSTALLINE METAL,* g	RECRYSTALLIZATION TEMP, °C
COPPER	200	100
NICKEL	300	350
IRON	400	450
TITANIUM	1500	700
BERYLLIUM	3500	900
TUNGSTEN	3500	1200

*0.001 CM/SEC, DISK Al_2O_3, 10^{-11} TORR, NO EXTERNAL SPECI-MEN HEATING

CONTACT OF SURFACES THROUGH ASPERITIES

LIGHT LOAD

HEAVY LOAD

OIL

THICK FILM LUBRICATION

THIN FILM LUBRICATION
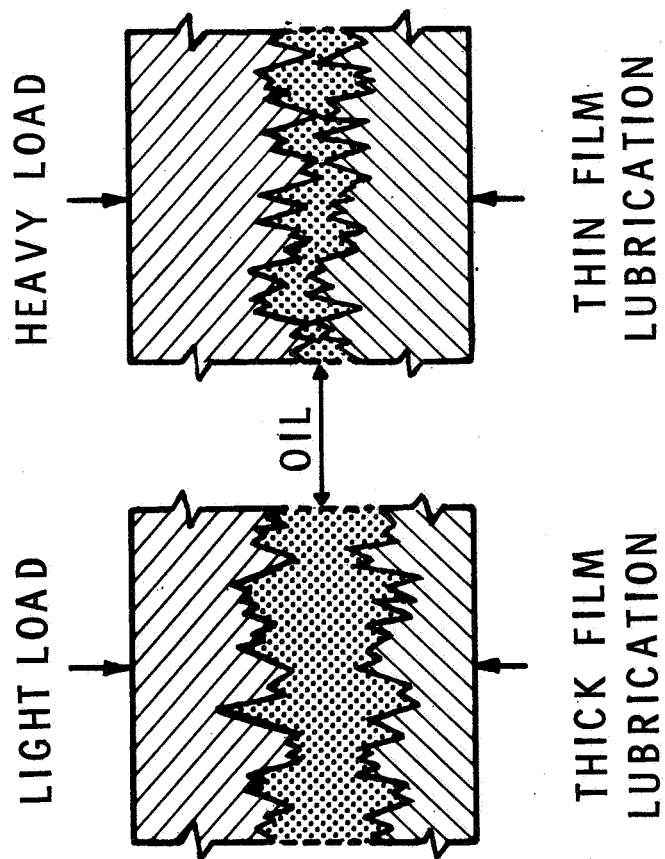
Fig. 1

WEAR OF STEARIC ACID MONOLAYER AND MULTILAYERS
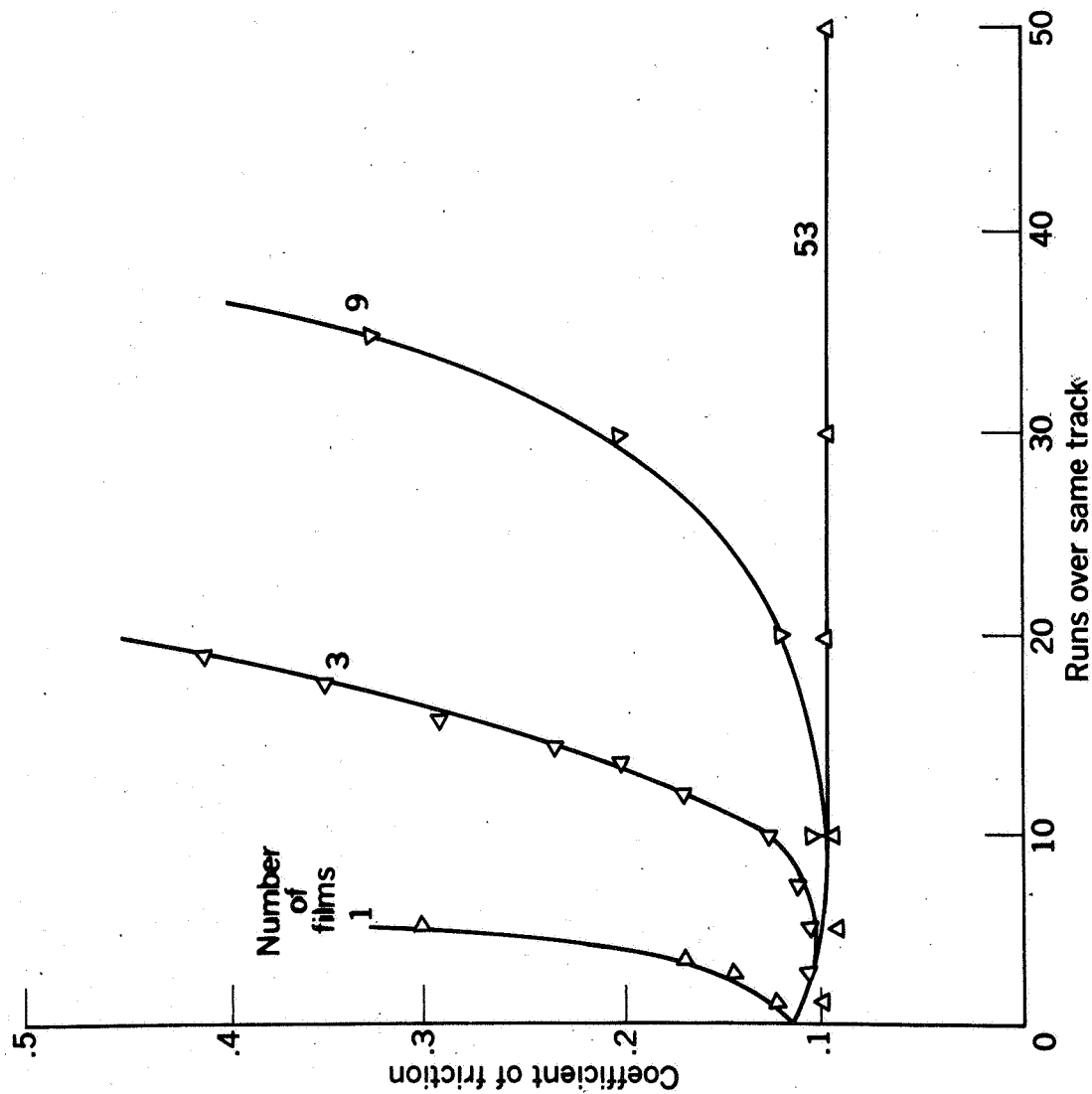
Fig. 2

EFFECT OF OXYGEN ON FRICTION OF OUTGASSED
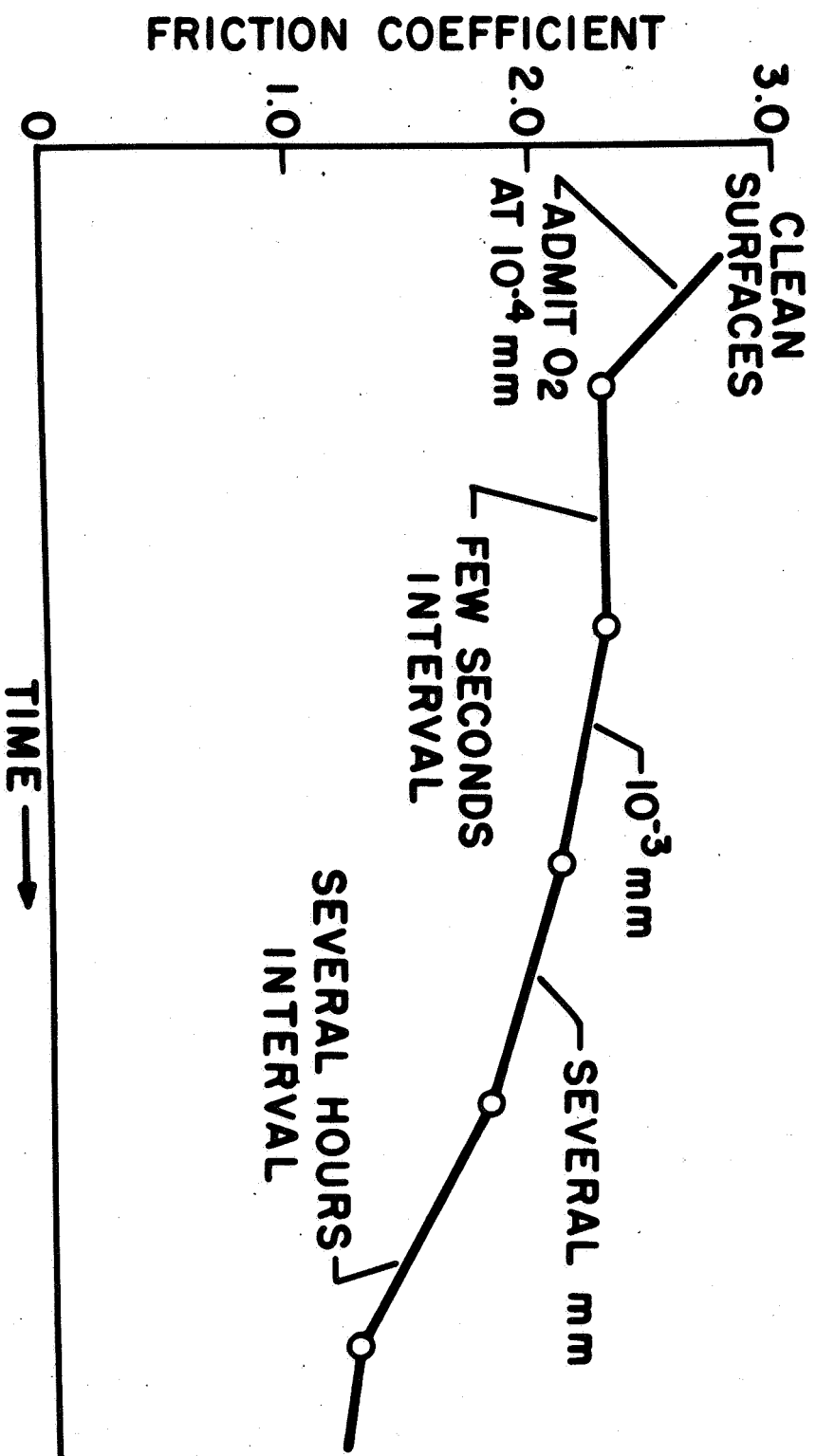IRON SURFACES
(FROM BOWDEN, N.Y. ACAD. SCIENCES, 1951)

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CS-45205

Fig. 3

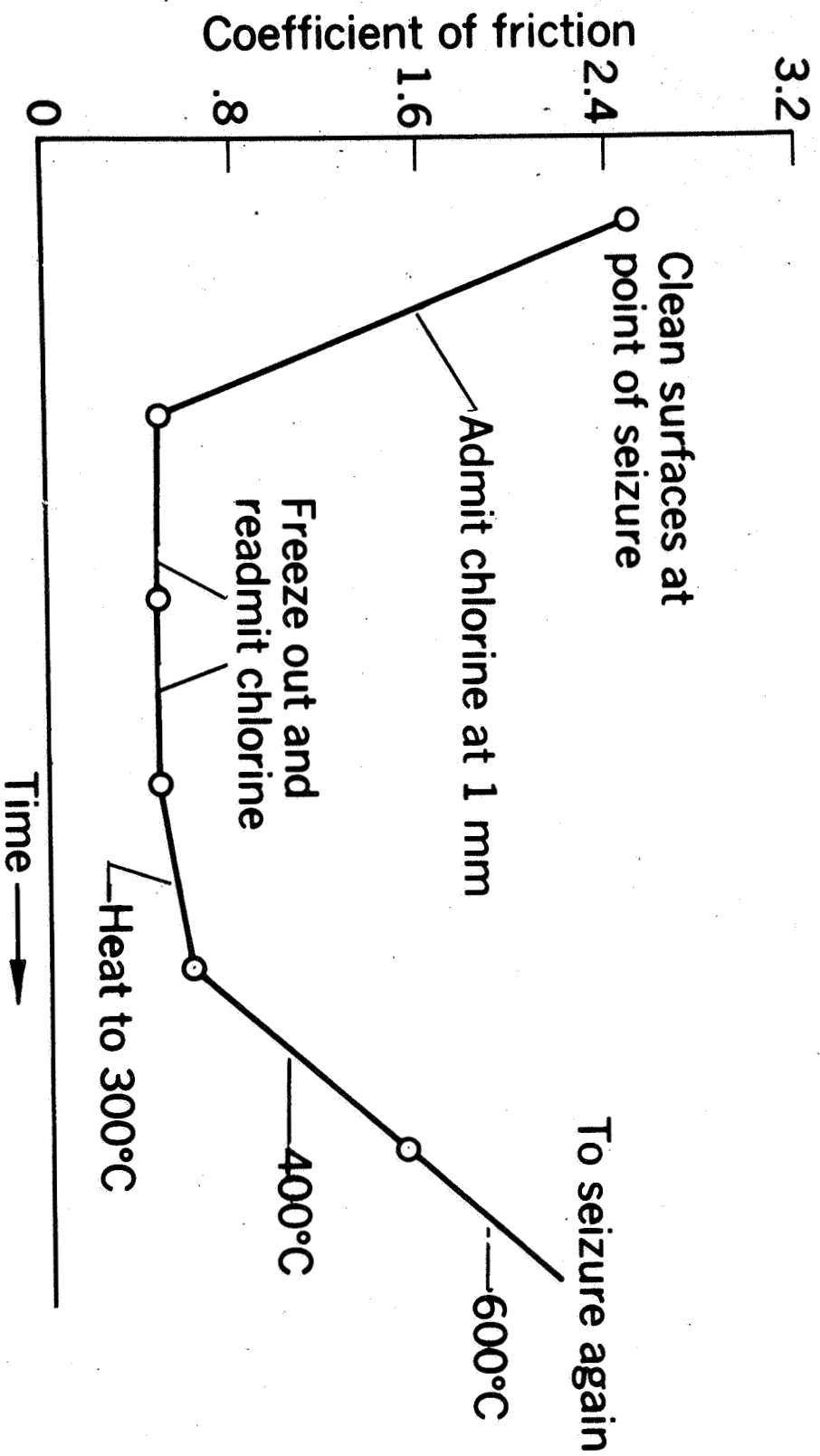# EFFECT OF CHLORINE ON FRICTION OF OUTGASSED IRON SURFACE

Coefficient of friction

3.2

2.4

1.6

.8

0

Clean surfaces at point of seizure

Admit chlorine at 1 mm

Freeze out and readmit chlorine

Heat to 300°C

400°C

To seizure again

600°C

Time ⟶

Fig. 4

EFFECT OF SULFIDE FILM ON FRICTION
OF OUTGASSED IRON SURFACES
(FROM BOWDEN, N.Y. ACAD, SCIENCES, 1951)

FRICTION COEFFICIENT

3.0

2.0

1.0

0

CLEAN
SURFACES

ADMIT H_2S

SEVERAL HR INTERVAL

HEAT
SURFACES
TO 300°C

370°C

500°C

790°C

930°C

TIME
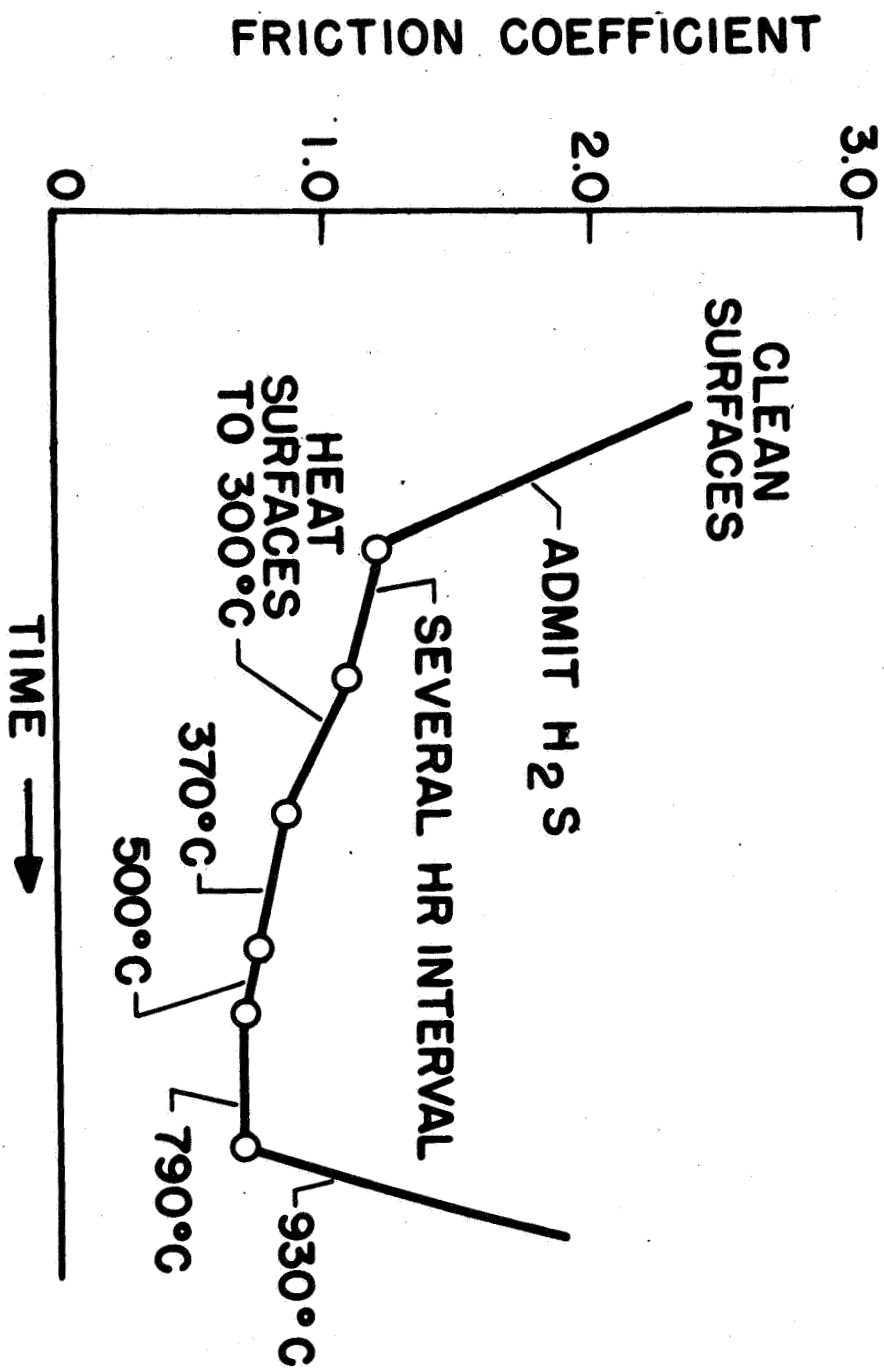
Fig. 5

OXYGEN ADSORPTION AND FRICTION FOR TUNGSTEN SLIDING ON TUNGSTEN

MONOLAYER

1.00×10^{15}

.75

OXYGEN ATOM
ADSORPTION,
(ATOMS/CM^2)

.50

OXYGEN ADSORPTION

.25

0

3.0

FRICTION AS A FUNCTION OF
OXYGEN SURFACE COVERAGE

2.5

RIDER (100) PLANE
DISK (100) PLANE
0.001 CM/SEC, 50 GM
20^0 C, 10^{-10} TORR

COEFFICIENT
OF FRICTION

2.0

6 AND 10×10^{-5}

1.5

1.0

0   .4   .8   1.2   1.6   2.0 2.4×10^{-5}

SURFACE EXPOSURE (TORR-SEC)
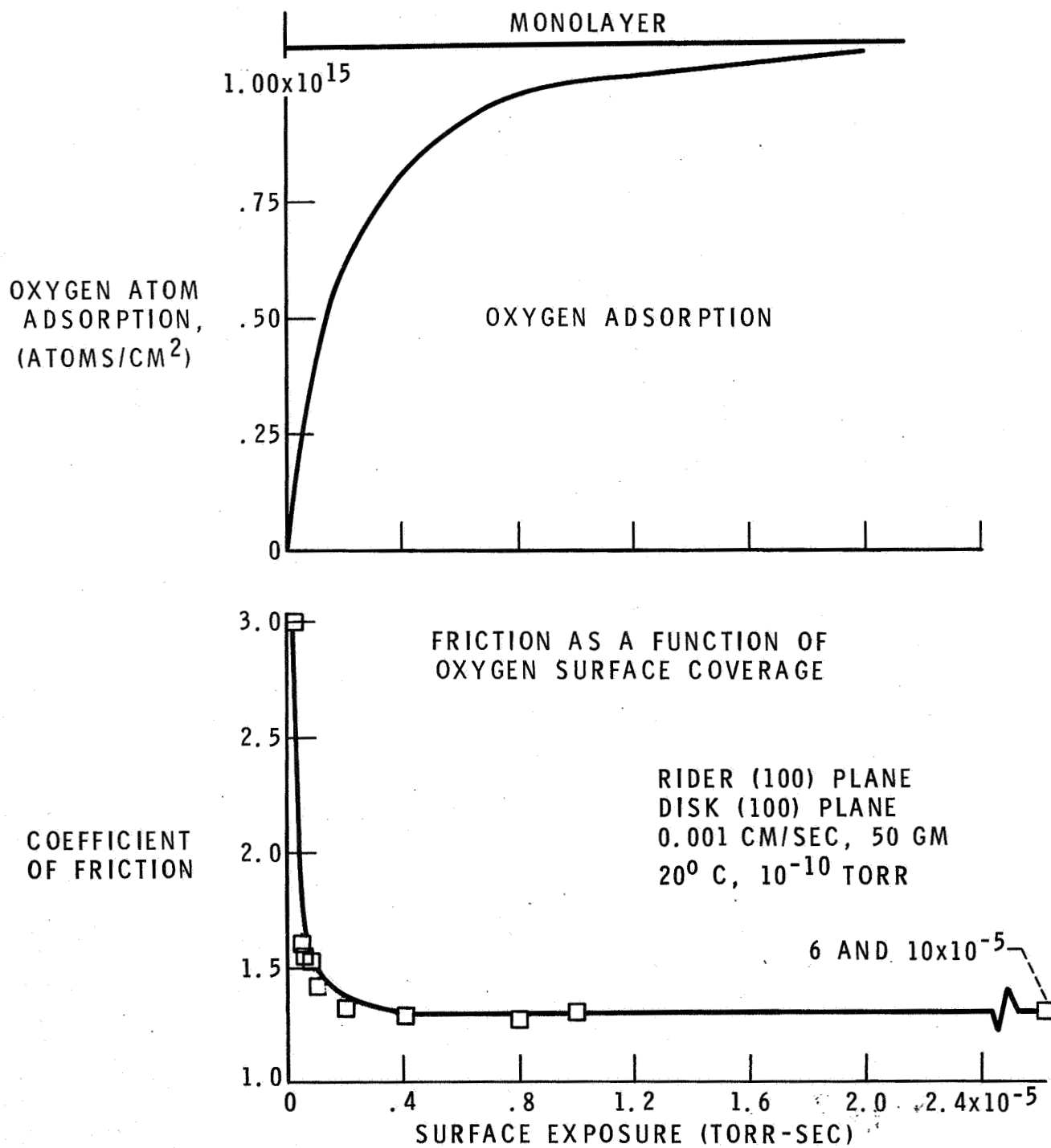
Fig. 6

COEFFICIENT OF FRICTION FOR 52100 SLIDING ON 52100
IN VACUUM WITH LIQUID HELIUM PUMPING
2.0 × 10⁻⁷ mm Hg; 390 FT/MIN, 1000 g
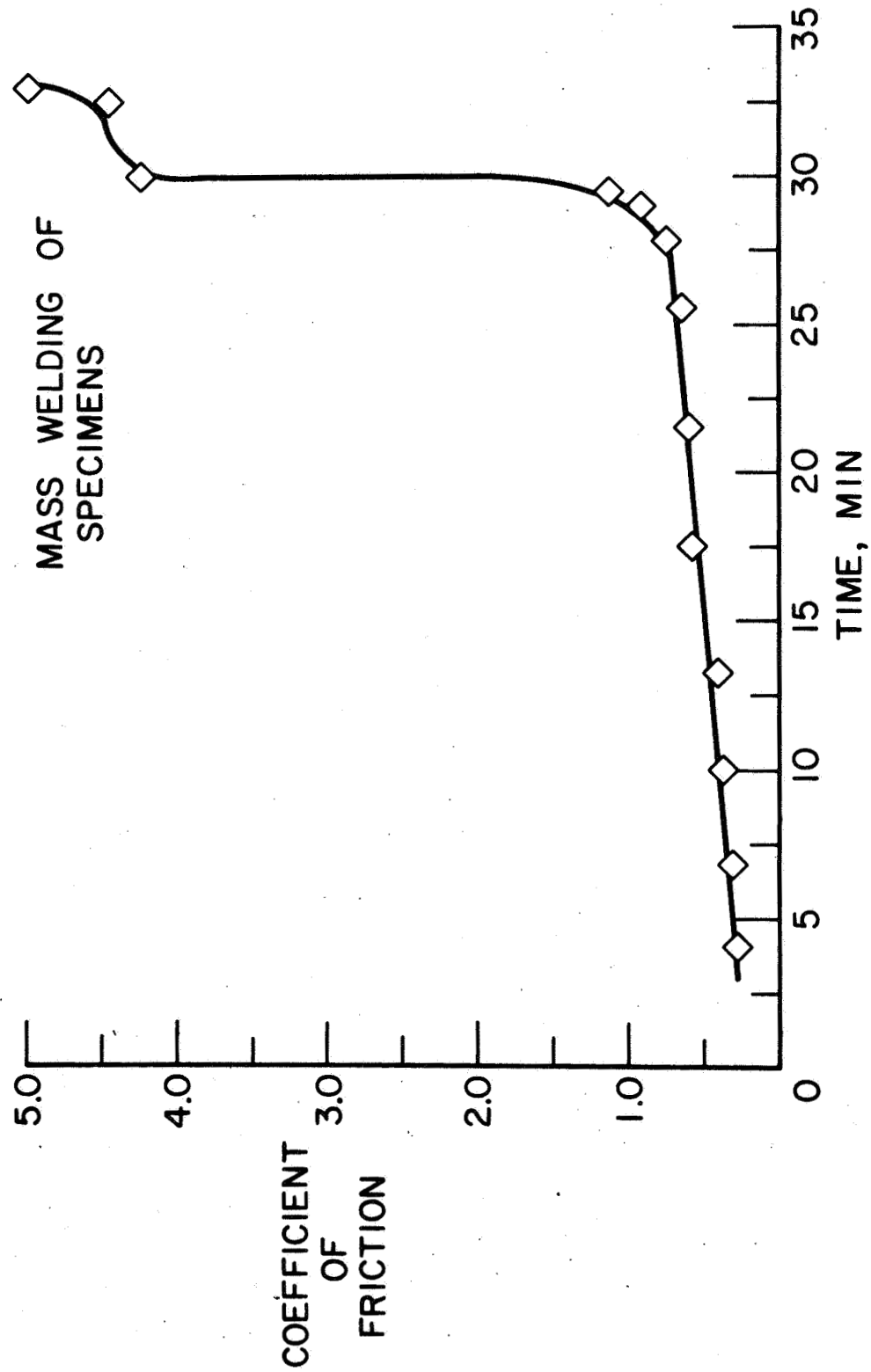
Fig. 7

WEAR VOLUME AS A FUNCTION
OF DISTANCE OF TRAVEL

Load, g

18,000

5280

Wear, cu cm

12×10^{-6}

8

4

0

40

80

120

Distance, cm
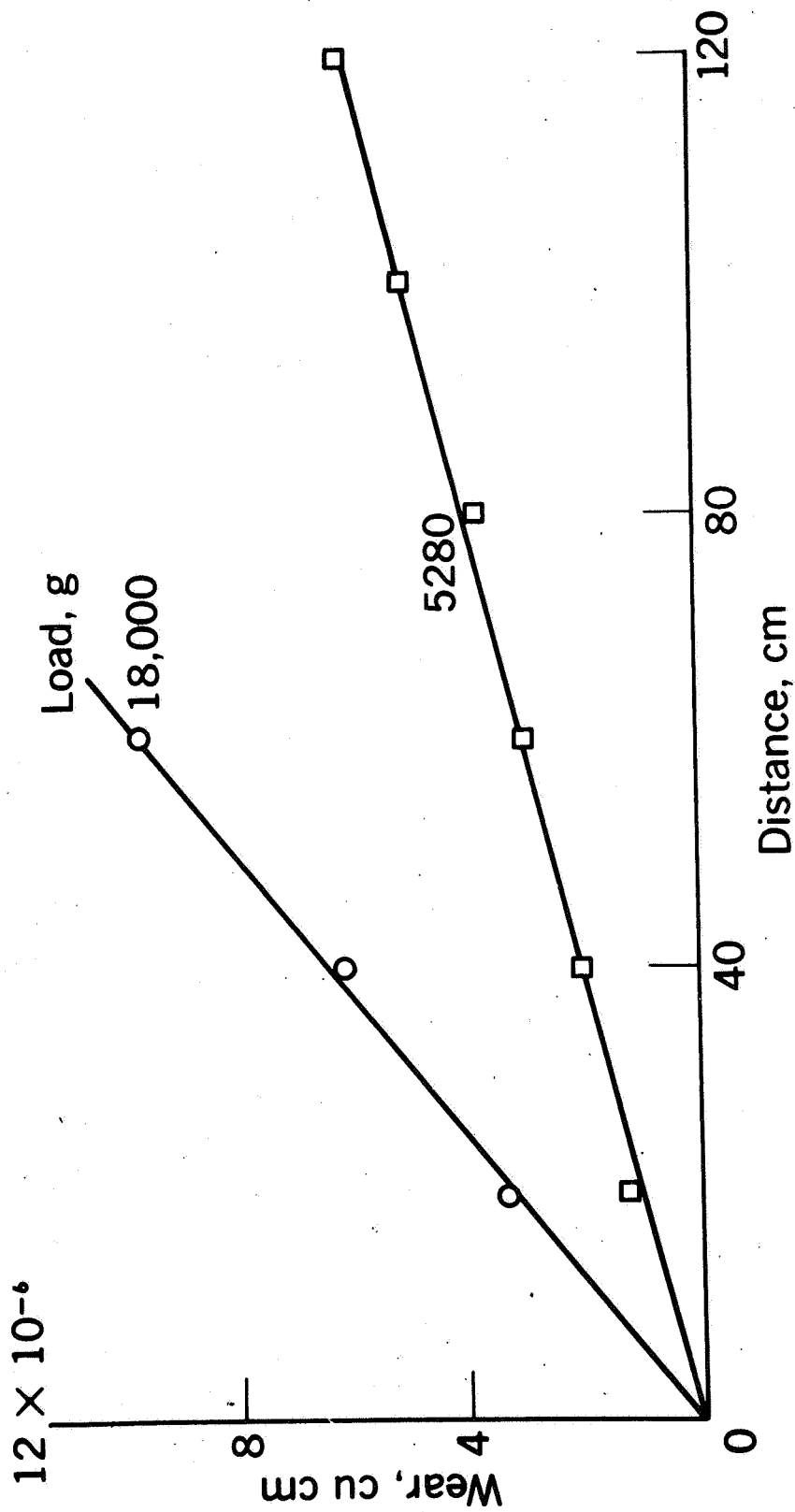
Fig. 8

EFFECT OF TOTAL LOAD AND UNIT LOAD ON
WEAR OF CARBON AGAINST CHROMIUM PLATE

(10,000 FPM, 360° F, DRY AIR)

CYLINDRICAL RIDER SPECIMENS
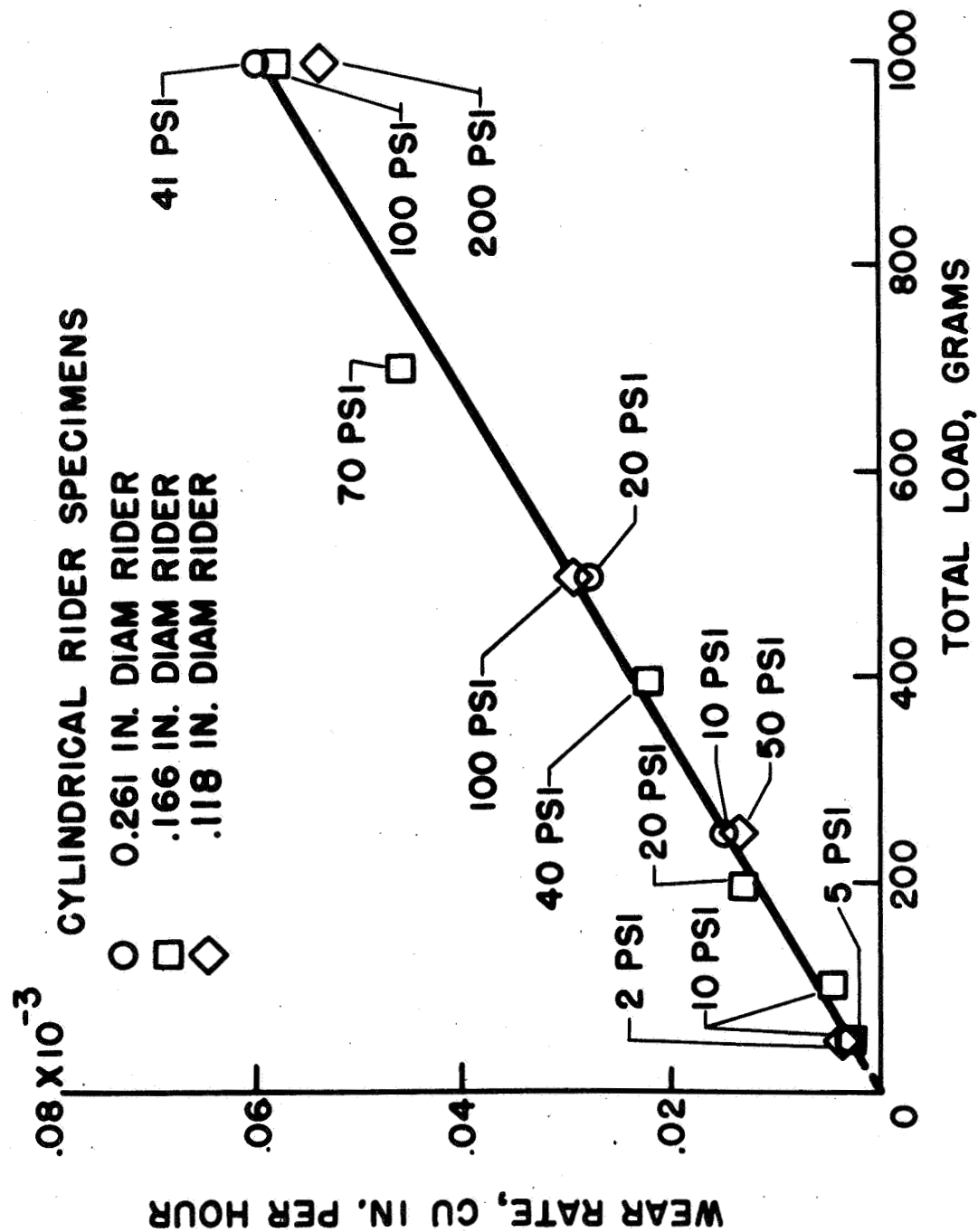
○   0.261 IN. DIAM RIDER
□   .166 IN. DIAM RIDER
◇   .118 IN. DIAM RIDER

WEAR RATE, CU IN. PER HOUR

.08 × 10⁻³

.06

.04

.02

0   200   400   600   800   1000

TOTAL LOAD, GRAMS

41 PSI

100 PSI

200 PSI

70 PSI

20 PSI

100 PSI

40 PSI

20 PSI

10 PSI

50 PSI

2 PSI

10 PSI

5 PSI

Fig. 9

WEAR COEFFICIENT AS A FUNCTION OF
AVERAGE (DESIGN) STRESS
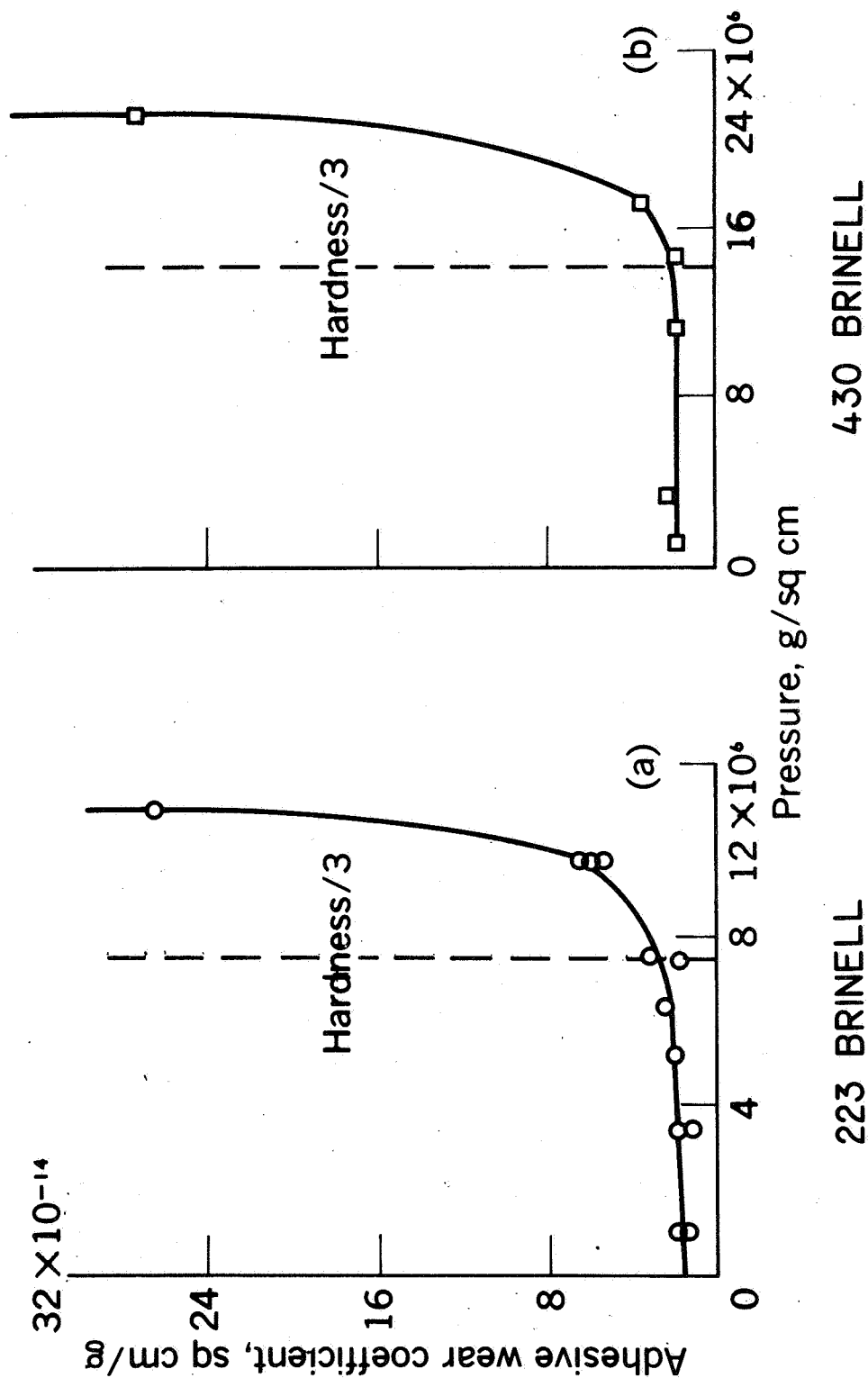
Fig. 10

RESISTANCE TO WEAR AS A
FUNCTION OF HARDNESS
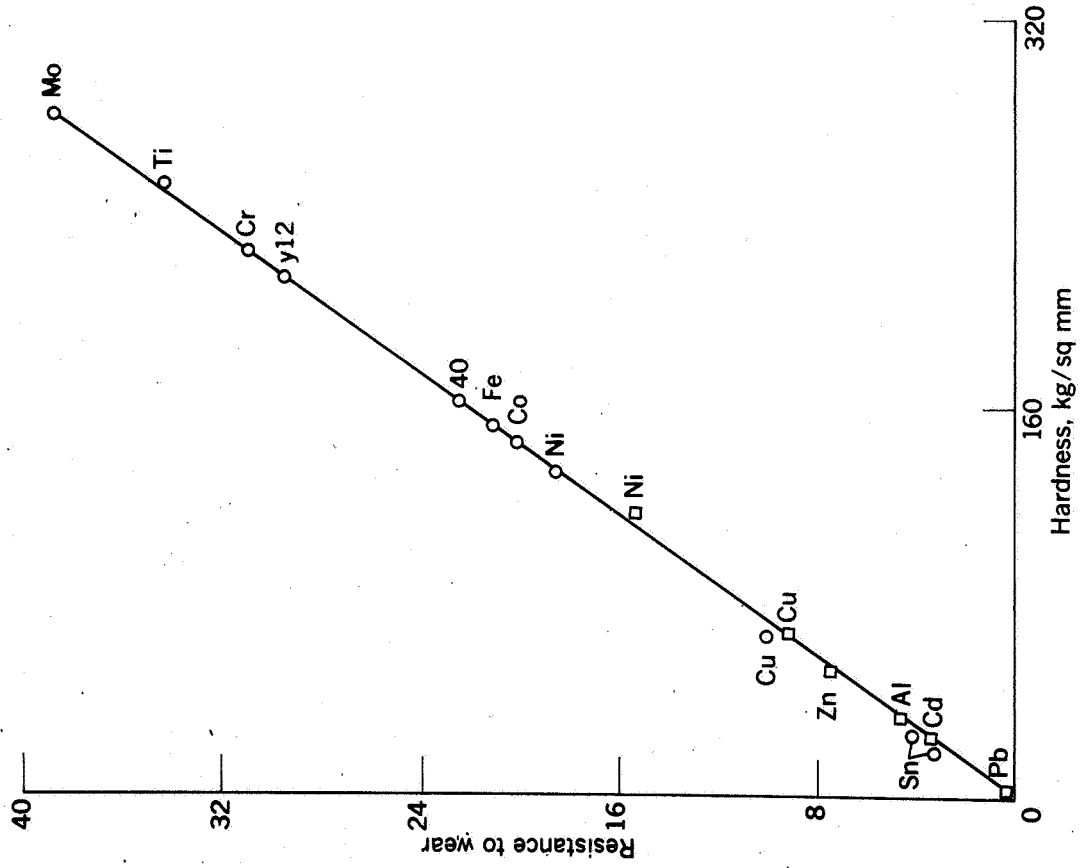
Fig. 11

GENERAL TRENDS OF RESISTANCE TO WEAR AS A FUNCTION

OF ALLOYING HEAT TREATMENT AND WORK HARDENING
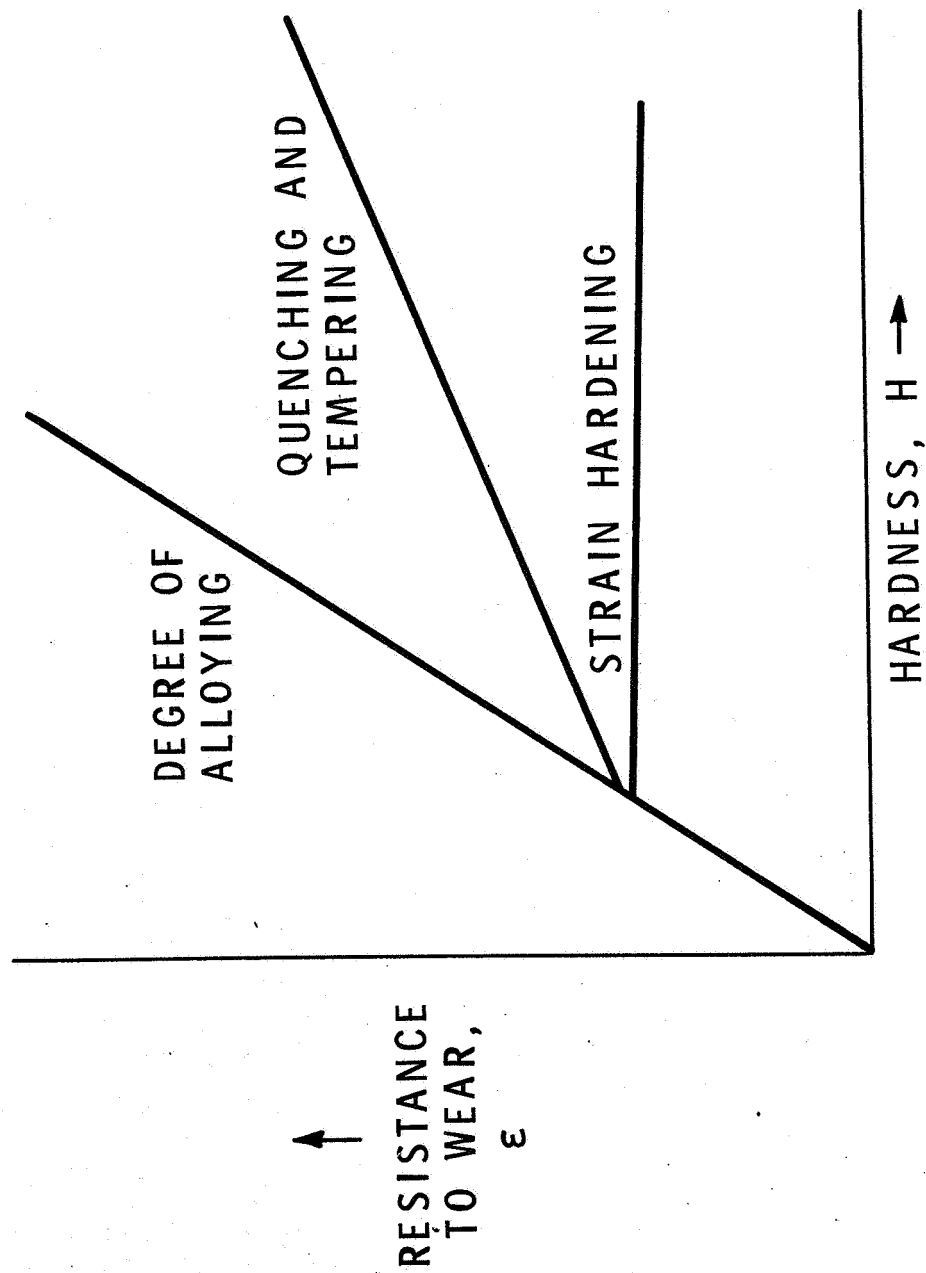
RESISTANCE
TO WEAR,
ε

DEGREE OF
ALLOYING

QUENCHING AND
TEMPERING

STRAIN HARDENING

HARDNESS, H →

Fig. 12

RELATIVE WEAR RESISTANCE AFTER QUENCHING AND TEMPERING AT VARIOUS TEMPERATURES

Fig. 13

RESISTANCE TO WEAR, ε (vertical axis, values 20, 40, 60, 80, 100)

HARDNESS, H, KG/MM^2 (horizontal axis, values 0, 200, 400, 600, 800, 1000)

Curves labeled: X-12, Y-12, Y-8, 40, W, Mo, Fe

X-12	2.35% C
	11.9% Cr
Y-12	1.10% C
Y-8	0.83% C
40	0.41% C
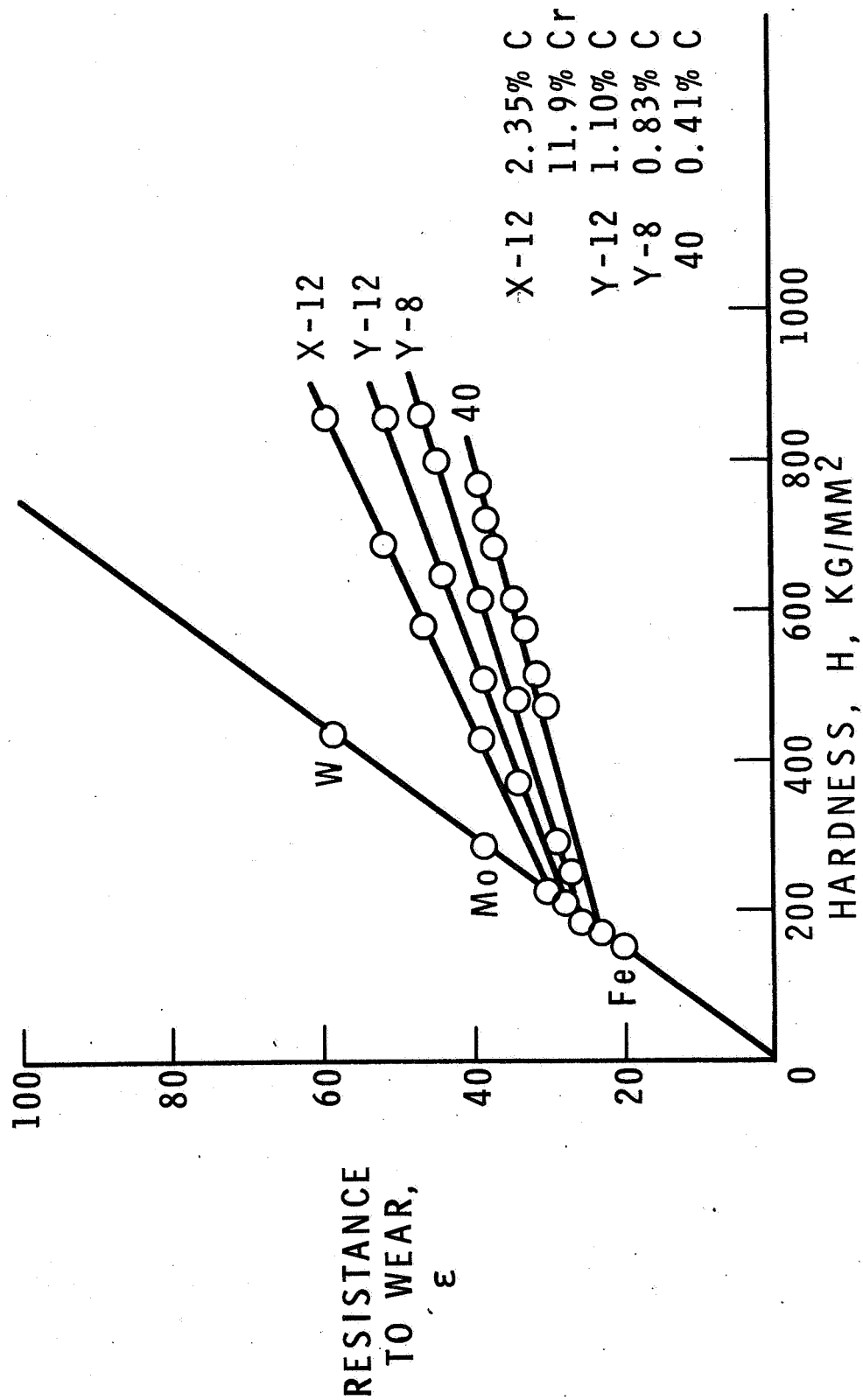
RELATIVE WEAR RESISTANCE AS A FUNCTION
OF WORK HARDENING

Fig. 14

METALLIC CRYSTAL MODELS

HEXAGONAL

CUBIC

Fig. 15

E-4155

NASA
CS-38164
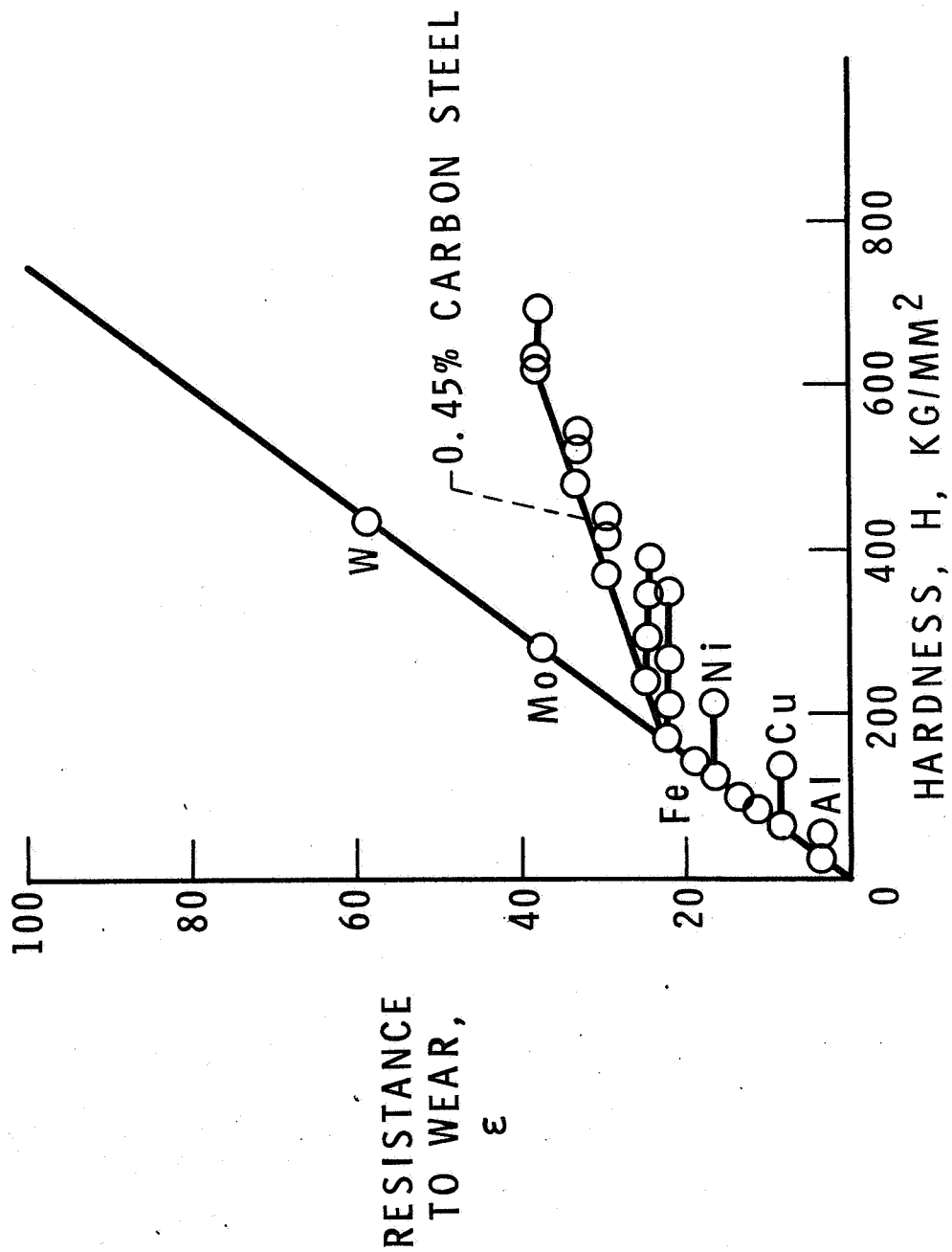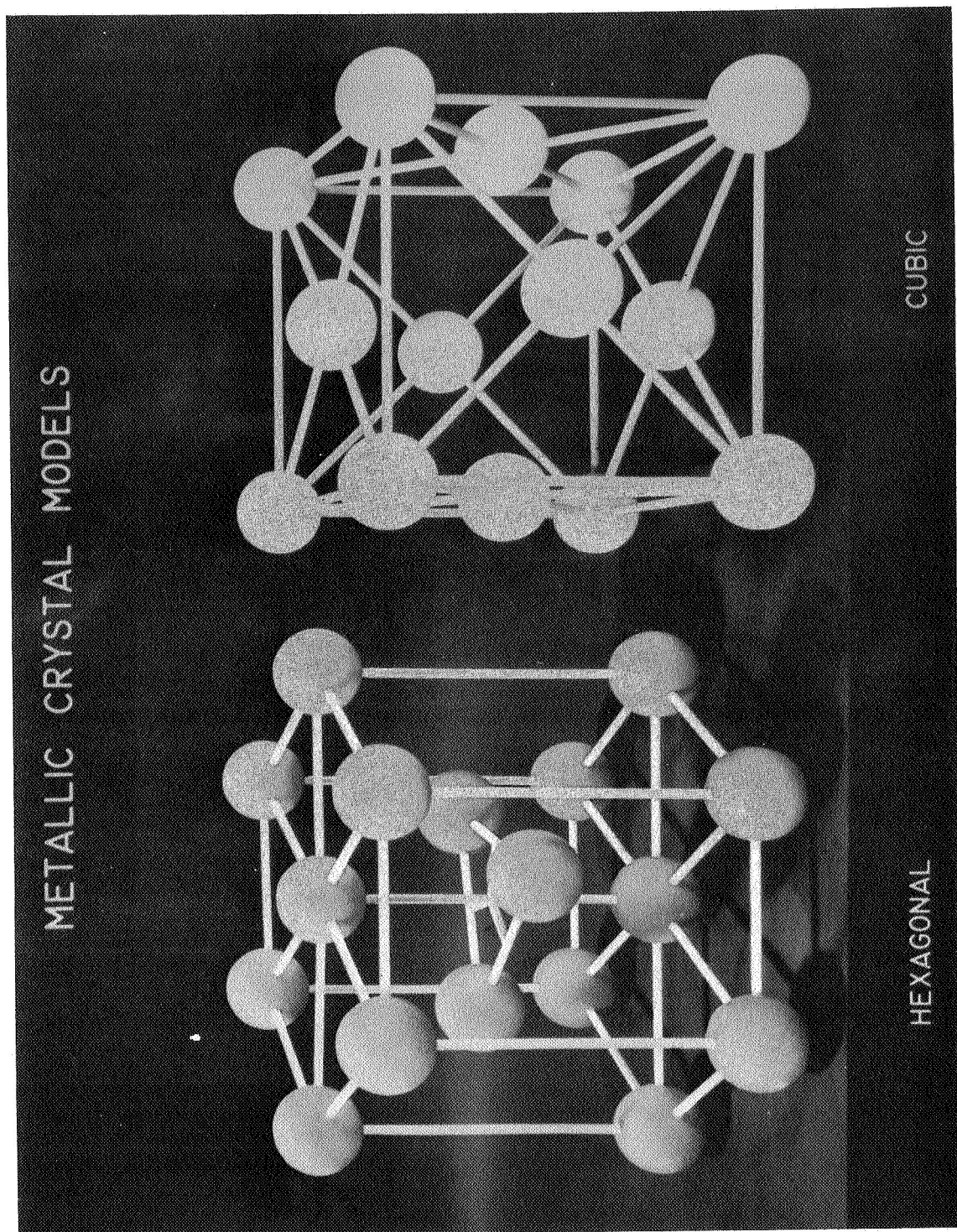
DISPLACEMENT OF PLANES IN HEXAGONAL CRYSTALS WITH SHEAR

I

II

III
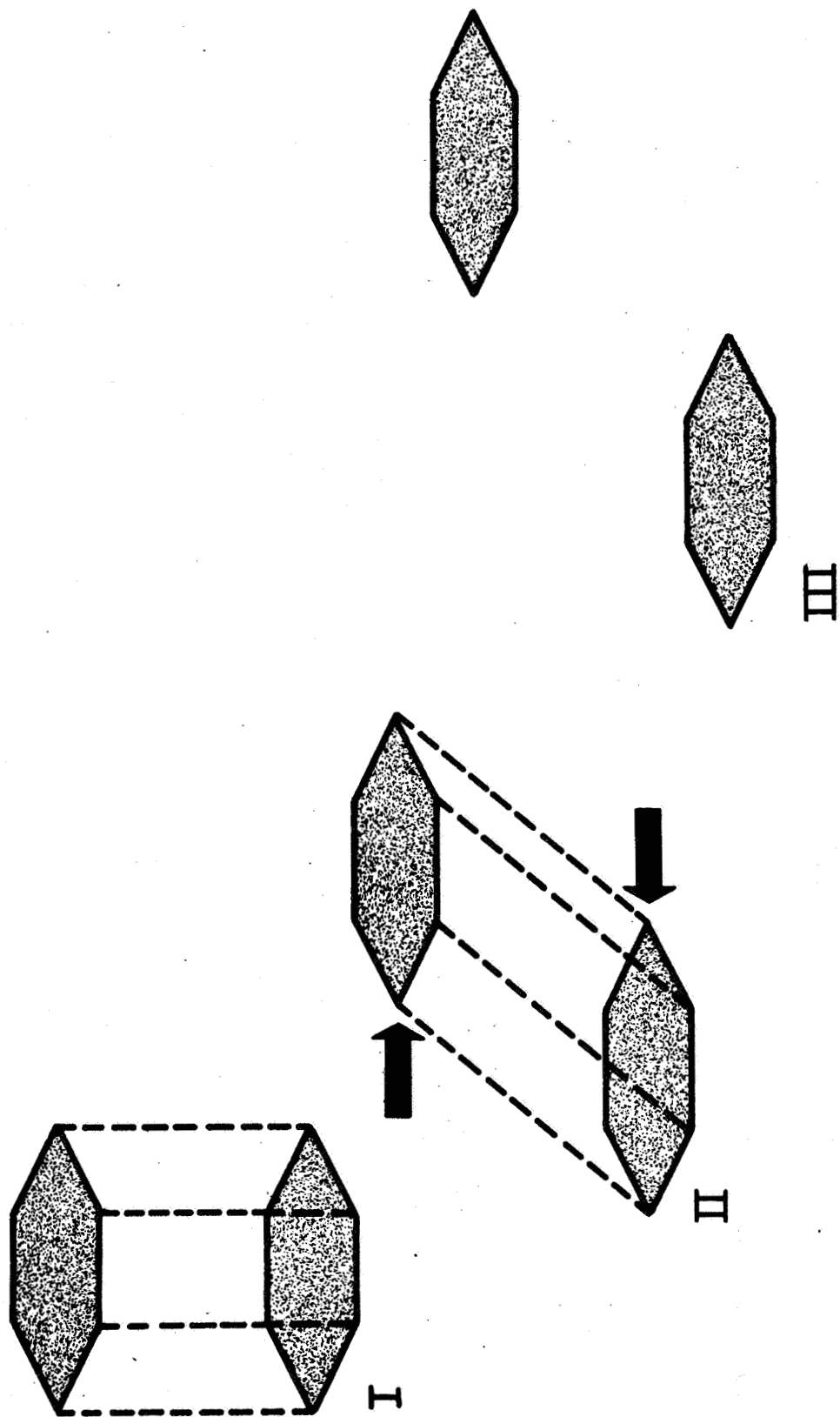
Fig. 16

COEFFICIENT OF FRICTION FOR Co SLIDING ON Co IN VACUUM, AT VARIOUS TEMPERATURES
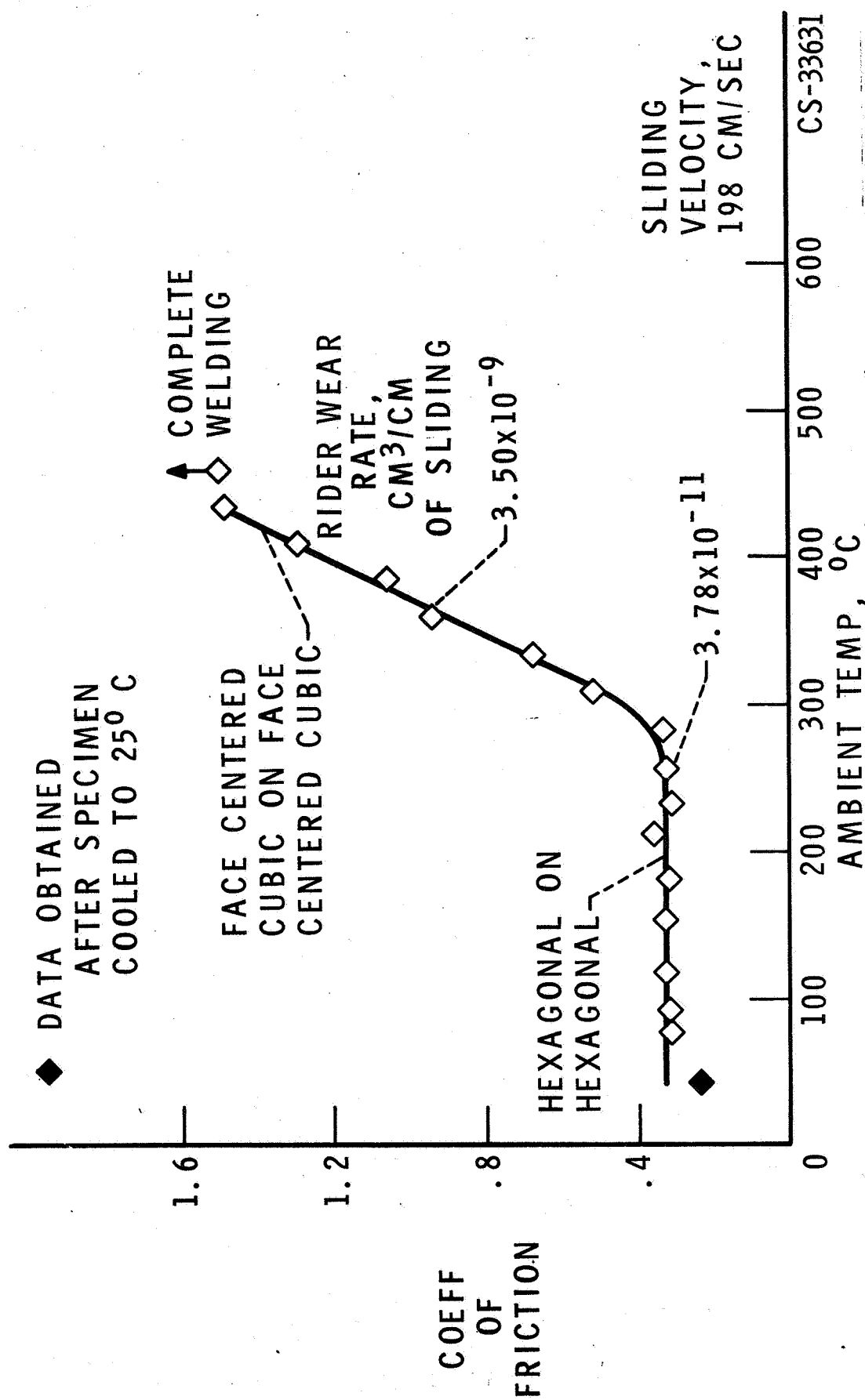
10^{-9} MM Hg, 1000 GM

◆ DATA OBTAINED AFTER SPECIMEN COOLED TO 25^0 C

FACE CENTERED CUBIC ON FACE CENTERED CUBIC

COMPLETE WELDING

RIDER WEAR RATE, CM^3/CM OF SLIDING

3.50×10^{-9}

HEXAGONAL ON HEXAGONAL

3.78×10^{-11}

COEFF OF FRICTION

1.6

1.2

.8

.4

0

100   200   300   400   500   600

AMBIENT TEMP, ^{0}C

Fig. 17

SLIDING VELOCITY, 198 CM/SEC

CS-33631

FRICTION OF VARIOUS HEXAGONAL METALS

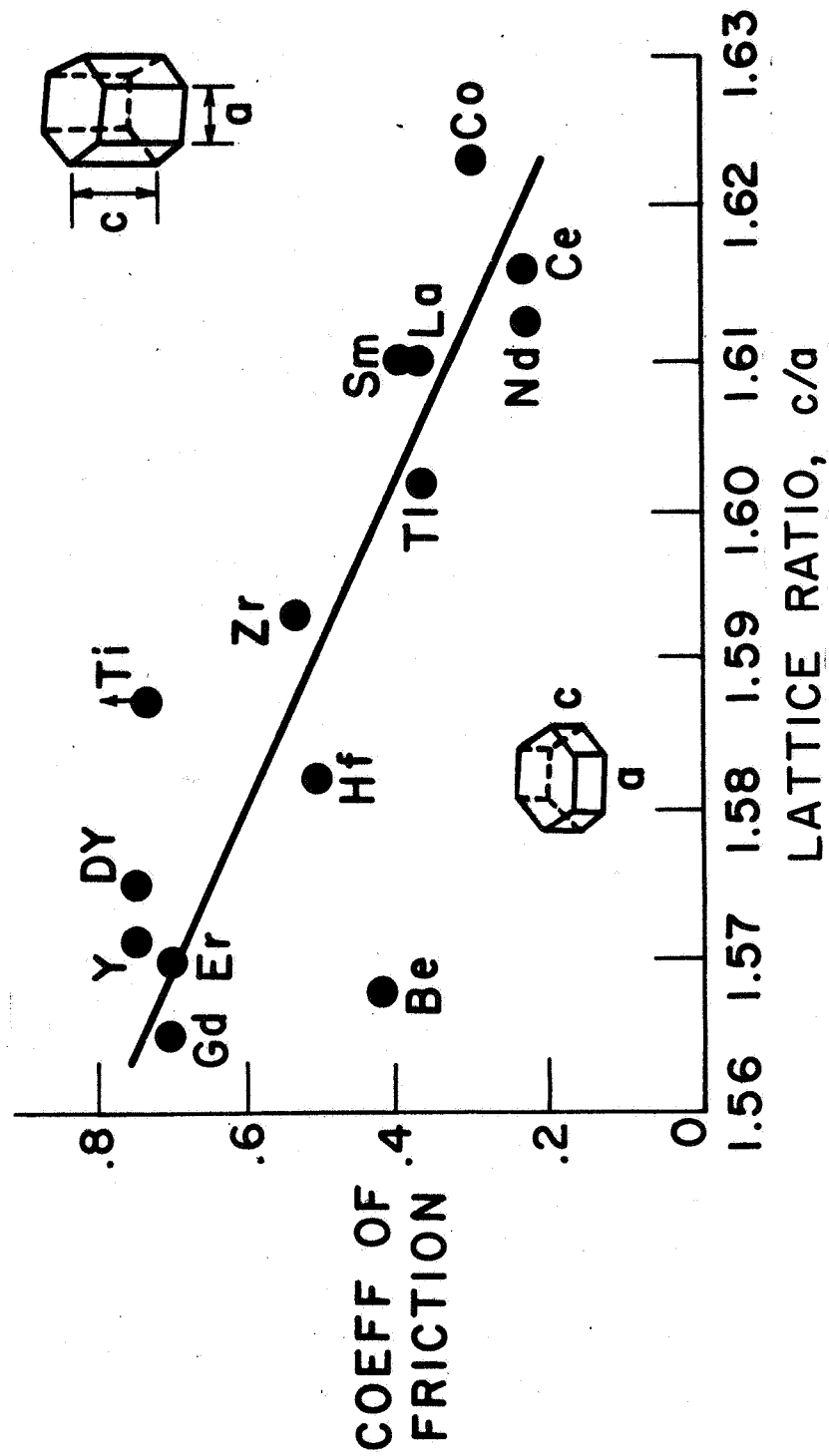10^{-9} mm Hg; LOAD, 1000 GM; 390 FT/MIN, ON 440-C STEEL

Fig. 18
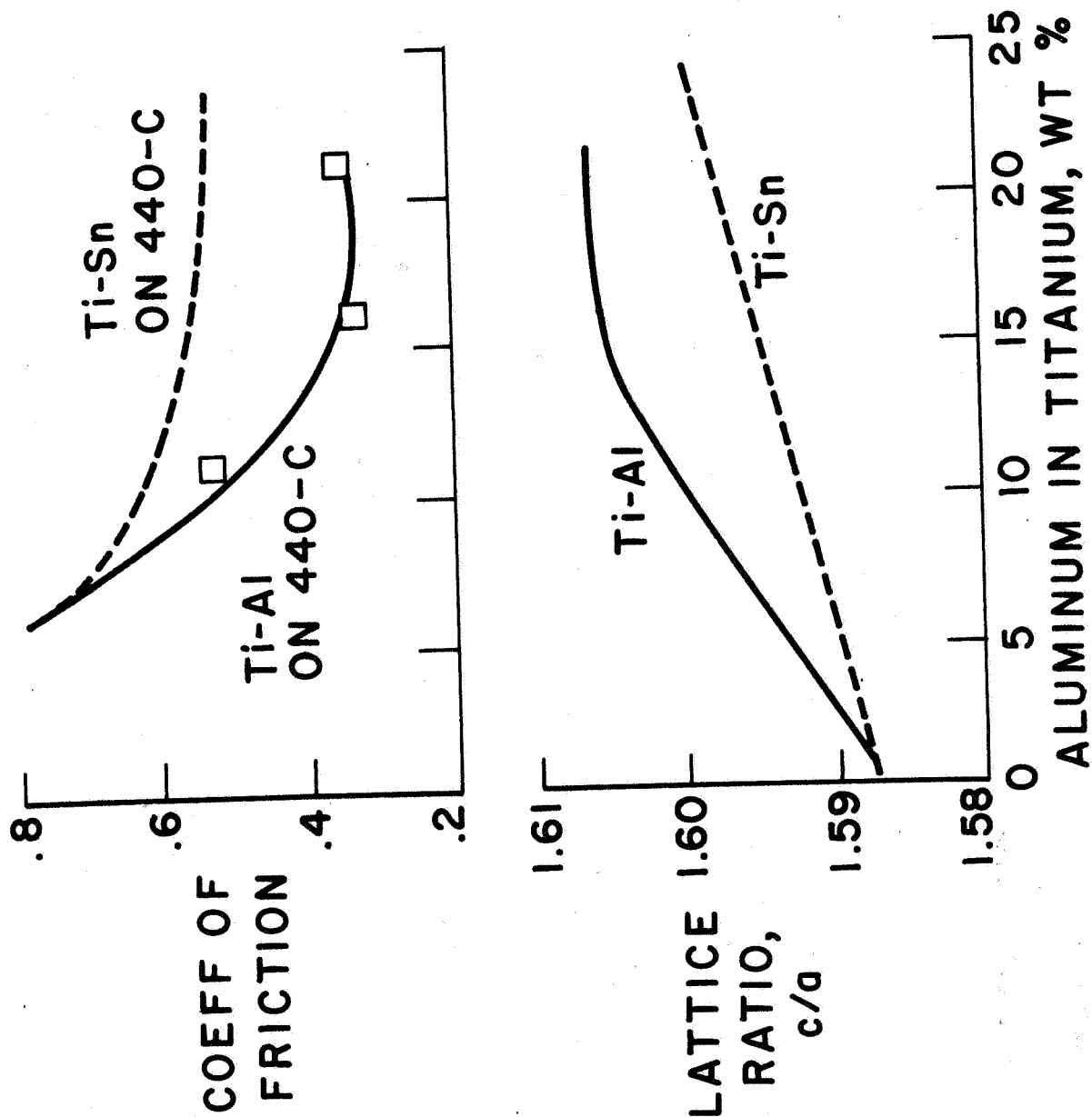
FRICTION AND LATTICE RATIOS FOR TITANIUM ALLOYS

Fig. 19

COEFFICIENT OF FRICTION FOR SINGLE AND
POLYCRYSTALLINE COPPER SLIDING ON POLYCRYSTALLINE
Al_2O_3 IN VACUUM

10^{-11} TORR; Cu ORIENTED WITH (111) PLANE PARALLEL TO SLIDING
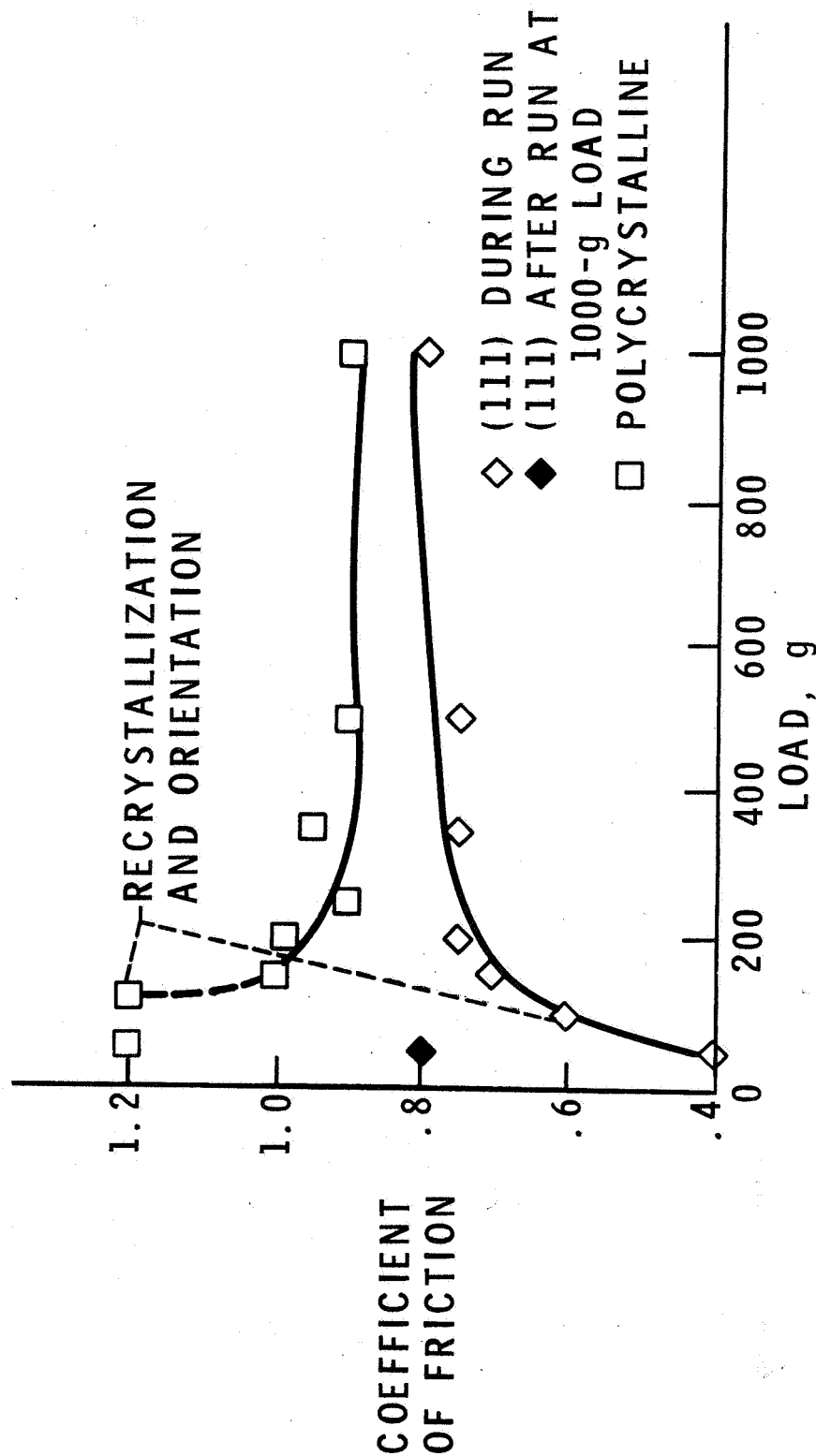INTERFACE; 0.001 CM/SEC; NO EXTERNAL SPECIMEN HEATING

Fig. 20

EFFECT OF LOAD ON FRICTION FOR IRON SLIDING ON
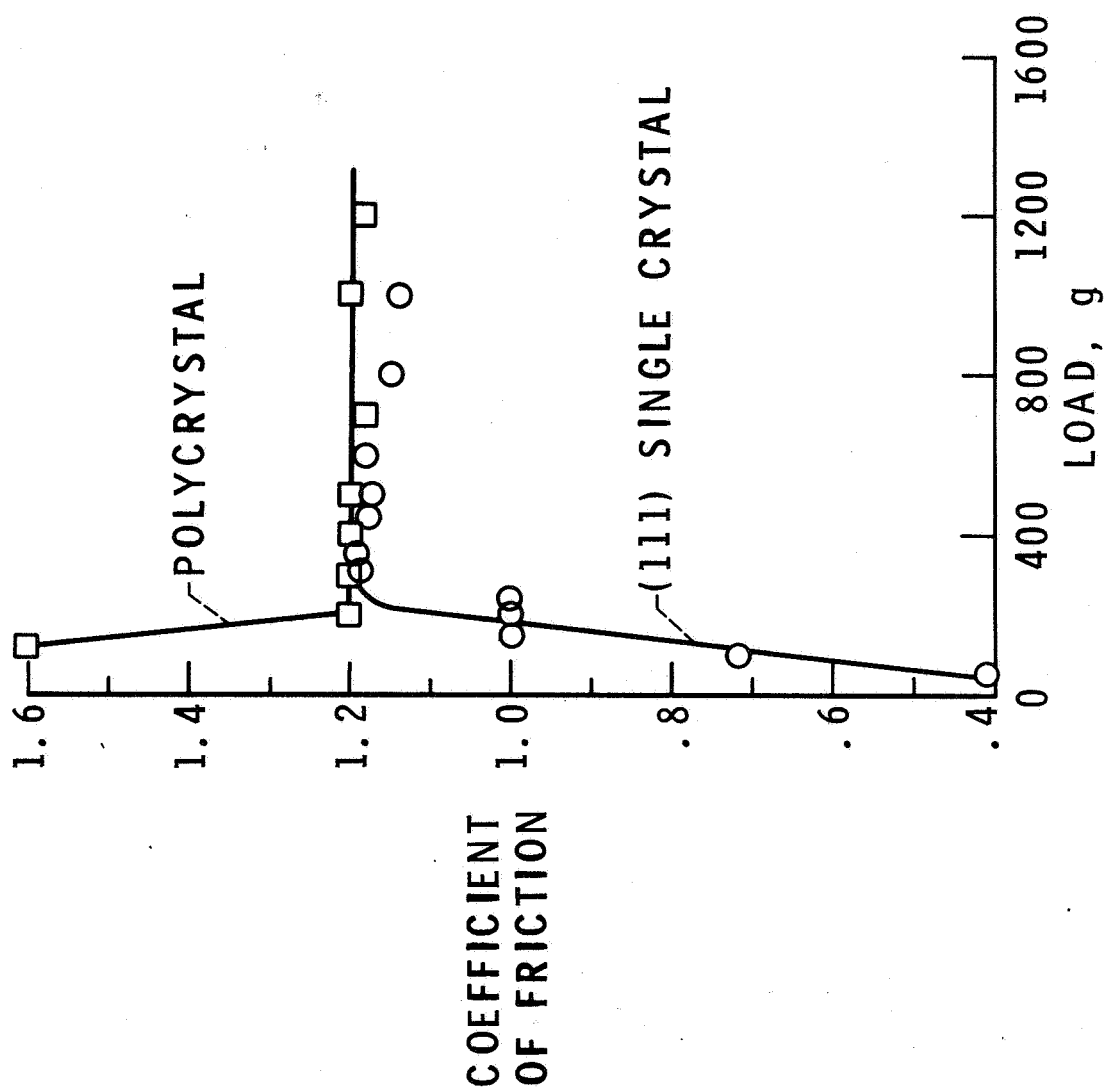POLYCRYSTALLINE ALUMINUM OXIDE IN VACUUM

0.001 CM/SEC, 20° C, 10⁻¹¹ TORR

Fig. 21

EFFECT ON FRICTION OF RECRYSTALLIZATION OF SINGLE CRYSTAL AND POLYCRYSTALLINE IRON SLIDING ON AL_2O_3 IN VACUUM

0.001 CM/SEC, 20^0 C, 10^{-11} TORR

POLYCRYSTAL

(110) SINGLE CRYSTAL

SOLID SYMBOLS INDICATE DATA OBTAINED AFTER MAXIMUM LOAD RUN

COEFFICIENT OF FRICTION

LOAD, g