Fundamental aspects and technological implications of the solubility for the prediction of running properties

Citation for published version (APA):

Dackus, A. J. G., Klostermann, J. A., & Landheer, D. (1980). Fundamental aspects and technological implications of the solubility for the prediction of running properties. *Wear*, 62, 255-286.
[https://doi.org/10.1016/0043-1648\(80\)90173-8](https://doi.org/10.1016/0043-1648(80)90173-8)

DOI:

[10.1016/0043-1648\(80\)90173-8](https://doi.org/10.1016/0043-1648(80)90173-8)

Document status and date:

Published: 01/01/1980

Document Version:

Publisher's PDF, also known as Version of Record (includes final page, issue and volume numbers)

Please check the document version of this publication:

- A submitted manuscript is the version of the article upon submission and before peer-review. There can be important differences between the submitted version and the official published version of record. People interested in the research are advised to contact the author for the final version of the publication, or visit the DOI to the publisher's website.
- The final author version and the galley proof are versions of the publication after peer review.
- The final published version features the final layout of the paper including the volume, issue and page numbers.

[Link to publication](#)

General rights

Copyright and moral rights for the publications made accessible in the public portal are retained by the authors and/or other copyright owners and it is a condition of accessing publications that users recognise and abide by the legal requirements associated with these rights.

- Users may download and print one copy of any publication from the public portal for the purpose of private study or research.
- You may not further distribute the material or use it for any profit-making activity or commercial gain
- You may freely distribute the URL identifying the publication in the public portal.

If the publication is distributed under the terms of Article 25fa of the Dutch Copyright Act, indicated by the "Taverne" license above, please follow below link for the End User Agreement:

www.tue.nl/taverne

Take down policy

If you believe that this document breaches copyright please contact us at:

openaccess@tue.nl

providing details and we will investigate your claim.

FUNDAMENTAL ASPECTS AND TECHNOLOGICAL IMPLICATIONS OF THE SOLUBILITY CONCEPT FOR THE PREDICTION OF RUNNING PROPERTIES*

D. LANDHEER, A. J. G. DACKUS and J. A. KLOSTERMANN

Technische Hogeschool, Eindhoven (The Netherlands)

(Received July 26, 1979)

Summary

The first Goodzeit rule, which relates resistance to cold welding and seizure with mutual solubility of metal sliding couples, has been investigated. The sliding properties of Au-Ag and Pd-Ag alloys against SAE 1045 carbon steel were observed in ultrahigh vacuum and in argon. The abrupt change in friction and wear coinciding with the composition of the alloys at which the solubility of iron into the alloy vanishes found by De Gee was not confirmed. Instead a continuous decrease in friction coefficient and wear was observed with increasing silver content. Pure silver showed a very limited tendency to metal transfer, although in some cases the transferred layer was removed spontaneously after 200 - 300 cycles. Electrical conduction measurements made during sliding revealed a correlation between the momentary friction force and the electrical conductivity which indicated that variations in the friction force are synchronous with variations in the contact area.

A model is proposed in which the conditions of friction and metal transfer are described in terms of one system and two material properties, *i.e.* the interfacial shear strength τ_i , the shear strength τ^* and the strain hardening exponent n of the junction materials. The friction behaviour is described by a modified adhesion theory: instead of a contaminating layer it is the specific atomic interaction of the contacting metals acting over small regions of atomic contact which controls τ_i . The junction growth mechanism is correlated with strain hardening.

It is suggested that the first Goodzeit rule must be interpreted as a result of atomic interaction in the interface between the contacting metals. In this way prediction of the running behaviour of both metals and alloys would become possible.

*Paper presented at the 4th International Tribology Conference, Paisley, September 10 - 15, 1979.

1. Introduction

It is currently fashionable to correlate subjects of science and technology with energy problems. Except for heating of buildings, industrial processes etc., the most frequent energy path can be represented schematically by

chemical energy → heat → mechanical energy → (electrical energy)
→ useless heat, dissipated by friction

Control of friction is therefore important, but wear control is even more important as wear is a frequent cause of loss of production capacity and also limits the useful life of consumer articles thus causing loss of raw materials. Engineers (supposed "waste makers") have much to gain by control of wear but have only a rudimentary understanding of the physical processes of friction and wear during dry sliding of metals.

Existing theories include the adhesion theory of metals with contaminant films proposed by Bowden and Tabor [1] and the theory of plastic interaction of surface asperities proposed by Green [2] and by Edwards and Halling [3]. In these theories only the plasticity of the metals and the contaminant layer are of importance. In the simple theory of Bowden and Tabor the use of properties of the contacting metals is not discriminative.

Goodzeit *et al.* [4, 5] investigated the scoring resistance of metal pairs and concluded that the metals which have the best score resistance against steel are the B-subgroup metals which are either insoluble in iron or else form intermetallic compounds with iron. As mutual insoluble metal pairs exhibit a comparatively low value of adhesion energy, at least one of Goodzeit's conclusions is in harmony with the approach of Rabinowicz [6]

The work of De Gee [7] strongly suggests that mutual solubility is an important factor. He studied the behaviour of Au–Ag alloys sliding against SAE 1045 steel in argon. Both gold and silver are B-subgroup metals which form a continuous series of solid solutions [8]. Silver and iron are mutually insoluble. Gold and iron have an appreciable mutual solubility in the solid state. De Gee found the first signs of iron solubility in Au–Ag alloys at about 40% gold (Fig. 1). In sliding tests the silver-rich alloys showed low friction and wear. If 40% gold was exceeded a discontinuous rise in friction and wear and the onset of transfer were found.

From the point of view of the adhesion theory of Bowden and Tabor the difference in friction coefficient found by De Gee must be the result of a difference in shear strength of a contaminant layer. However, it is difficult to understand how the formation of a real solution can be of importance for the very slow moving contacts used by Goodzeit *et al.* [4] for metals with a high melting point. In this case frictional heating can only be of the order of a few degrees Celsius as pointed out by Bowden and Tabor [1].

In the theory of Bowden and Tabor it is assumed that the plastic flow of contacts (asperities) causes an increase in the area of contact until the real

Fig. 1. Solid solubility of iron, friction, wear and metal transfer as a function of silver content for Au-Ag alloys sliding on steel in an argon atmosphere at sliding speeds of the order of 1 m s^{-1} (after De Gee [7]).

area of contact A_r is just sufficient to support the load. Under these conditions $A_r p_0 = F_N$ where p_0 is the yield pressure and F_N is the normal load. The regions of real contact will be of the same size as the asperities, of the order of $1 - 1000 \mu\text{m}$. Vandamme [9] measured $1/f$ noise in electrical crossed-rod contacts of brass, gold and tungsten and concluded that the diameter of the conducting spots varied between 0.3 and 26 nm. Therefore, in addition to regions of real contact, it is proposed to introduce regions of atomic contact which are several orders of magnitude smaller. These regions will in most cases be so small that no dislocations that can perform the required plastic deformation will exist in their immediate environment. The maximum dislocation densities in extremely deformed metals are of the order of 10^{16} m^{-2} . Therefore in a region of 10 nm^2 only one dislocation will exist if an even distribution of dislocations is assumed. The dislocations are tangled and dislocation-free regions exist which are large compared with the regions of atomic contact. A relation between the dislocation density, adhesion and friction in very lightly loaded contacts was experimentally demonstrated by Maugis *et al.* [10].

In this paper a model is proposed for the friction of dry sliding metals in which the character of the atoms is involved. Regions of atomic contact are proposed which are so small that in most cases no dislocations can perform plastic deformation in the immediate environment. In the interfaces of atomic contact shear must be performed under stresses which are equal to the theoretical strength of the interface. The model is illustrated in Fig. 2.

In order to investigate (a) how the character of the atoms and the properties of the metals are involved and (b) whether solubility can be important in circumstances under which no solutions can be formed sliding experiments were performed in ultrahigh vacuum (UHV). The results of dry sliding experiments of Ag-Au and Ag-Pd alloys in UHV are reported.

Fig. 2. Comparison of several current models for the description of the friction process with the modified model proposed in this paper (f is the calculated coefficient of friction).

2. Experimental procedures

2.1. Apparatus

Sliding tests were carried out in the UHV pin-on-plate type test rig shown in Fig. 3. An oil-free vacuum better than 4×10^{-10} Torr was maintained as the standard condition after baking out. For some tests very pure argon (not more than 1 ppm oxygen and water vapour) at atmospheric pressure was used.

Fig. 3. Schematic diagram of the UHV friction test apparatus.

A new pin and a fresh track were used for every test. The friction and electrical conductivity (direct current less than 200 mA) could be measured continuously during the reciprocating motion of the pin under constant load. The load was kept constant at $F_N = 3.5$ N and the main sliding velocity v was usually 5.8 mm s^{-1} , interrupted by short periods at $v = 0.29 \text{ mm s}^{-1}$ for detailed analysis.

2.2. Materials and specimen preparation

Materials for the hemispherical tipped pins (4 mm in diameter) are summarized in Table 1. The purity was about 99.99%. X-ray microanalysis showed satisfactory homogeneity of the recrystallized specimen.

The standard plate material was normalized plain carbon steel (0.53% carbon, 0.20% silicon, 0.47% manganese, 0.06% sulphur and 0.01% phosphorus) with hardness $H_V = 2250 \text{ N mm}^{-2}$. In a few instances strips 1 mm thick of cold rolled and annealed gold or silver pin material bolted onto the steel were used as plate specimens. Specimen preparation consisted of machining, mechanical polishing, very slightly etching (2% HNO_3 in alcohol, steel surfaces only), application of wires for measurement of conductivity by spot welding and ultrasonic cleaning in an acetone bath. After mounting the specimen in the friction device the specimen temperature rose to about 300°C during baking out.

TABLE 1
Pin materials

	<i>Au-Ag alloys</i>				<i>Pd-Ag alloys</i>			
Composition (at. %)								
Silver	—	31	60	81	100	—	20	40
Gold	100	69	40	19	—	—	80	60
Palladium						100	80	40
Hardness H_V -100 gf ($N\text{ mm}^{-2}$)	340	580	590	530	440	1190	1100	800
Annealing temperature ($^{\circ}\text{C}$)/time (h)	600/12	600/12	600/12	600/12	600/12	800/14	800/14	700/18
Strain-hardening exponent n	0.49				0.44	0.52		
(<i>cf.</i> Nadai's equation: $\bar{\sigma} = c\epsilon^{-n}$)								

3. Experimental results

3.1. General picture of a test run

The sliding process for each material combination used can be described in terms of a (momentary) local relative friction f given by the ratio of the local frictional force to the normal force F_N and local electrical conductivity C (see Fig. 4). A sliding cycle n is defined as a complete forward and backward pass of the pin over a track of length 17.4 mm on the plate. As well as the coefficient of friction $\bar{f}$ and the average conductivity $\bar{C}$ during a cycle (both calculated as the arithmetic mean of at least 30 local values taken equally spaced along the track), the average coefficient of friction $\bar{f}$ found from the $\bar{f}$ values at a certain number n of all tests with the same sliding pair was determined. As the results in pure argon did not deviate significantly from those in UHV all results are considered together. In nearly all the test runs $\bar{f}$ appeared to increase rather continuously, initially very rapidly and later more slowly. The same behaviour was found for $\bar{C}$ for gold and silver and their alloys, but for pins with a high palladium content $\bar{C}$ levelled off after a few sliding cycles (Fig. 5). The graph of f versus sliding distance appeared smooth during the initial sliding cycles, but marked fluctuations of gradually increasing amplitude and period developed during further sliding. C varied in roughly the same way, apart from the observation of an immediate onset of fluctuations (Fig. 4). The f and C patterns are obviously synchronous for most sliding pairs from about 60 sliding cycles onwards (cf. Section 3.4).

Fig. 4. Portions of the curves of the momentary relative friction f and the momentary conductivity C vs. time for pure gold sliding against SAE 1045 steel under standard conditions taken from the 1st, 60th and 986th sliding cycles. Note that the low frequency fluctuations of f and C become synchronous from about 60 sliding cycles onwards.

Fig. 5. Characteristic examples of the curves of the coefficient of friction $\bar{f}$ and the conductivity $\bar{C}$ vs. the number of sliding cycles n (each of length 2×17.4 mm): load, 3.5 N; sliding velocity, 5.8 mm s^{-1} ; vacuum better than 4×10^{-10} Torr (standard conditions). (a) 40Au-60Ag on steel; (b) 80Pd-20Ag on steel.

Transfer of pin material to the plate was observed in each case. The amount of transferred material usually increased with the number of sliding cycles; the ultimate situation on the plate clearly depended on the pin composition (Fig. 6, cf. Section 3.5). Visual signs of back transfer on the pin are faint for gold alloys and somewhat more pronounced for palladium alloys. In no case was the formation of loose particles observed.

3.2. Anomalous behaviour of silver

Silver may behave exceptionally. In some cases the friction and conductivity gradually decrease and then remain low after a normal process up to 200 - 500 cycles (Fig. 7). The decrease in $\bar{f}$ and $\bar{C}$ is associated with removal of transferred material from the plate and the formation of very elongated grooves on the track in the sliding direction (Fig. 6(b)). This behaviour was found for silver against steel in five out of seven tests and for the 19Au-81Ag alloy in one out of five tests.

3.3. Influence of the pin composition on friction

Despite considerable scatter in the results of several test runs with the same material combination the relation between the average frictional coefficient and sliding distance for one material is clearly different from similar relations for other materials (Figs. 7 and 8). All normal $\bar{f}$ - $\log n$ curves rise continuously and are approximately S-shaped. To facilitate comparison they can be roughly approximated by

$$\bar{f} = a + b \log n \dots \quad (1)$$

where the average slope is given by the factor b (Table 2). Obviously the initial friction $\bar{f}_1$, the rate of increase in friction with sliding distance as

Fig. 6. Scanning electron micrographs of the track on the plate after a standard test run against various Au-Ag alloys: (a) pure gold; (b) 20 at. % silver; (c) 69 at. % silver; (d) pure silver; (e) pure silver. The amount of transferred pin material decreases with increasing silver content. At the highest silver concentrations the transferred layer may eventually be removed from part of the track (middle region of (e)) or from all of the track after more than 200 sliding cycles.

Fig. 7. Variation of the coefficient of friction $\bar{f}$ (average over all tests with a particular material) with the number of sliding cycles n for Au-Ag alloys.

Fig. 8. Variation of the coefficient of friction $\bar{f}$ (average over all tests with a particular material) with the number of sliding cycles n for Pd-Ag alloys.

represented by b and the ultimate friction $\bar{f}_{1500}$ after 1500 cycles generally decrease continuously with increasing silver content of the pin; b and therefore $\bar{f}_{1500}$ tend to reach higher values for palladium alloys than for gold alloys (Fig. 9).

TABLE 2
Some observations concerning the friction of different alloys against steel SAE 1045 in UHV

	Au-Ag alloys					Silver	Pd-Ag alloys				
Silver content of the pin (at.%)	0	31	60	81	100	100	0	20	40	60	80
Number of tests	9	4	5	4 1 ^a	2 5 ^a	2	2	2	1	2	2
Average coefficient of friction $\bar{f}_1$	0.71	0.88	0.64	0.52	0.50	0.50	0.93	0.89	0.56	0.61	0.59
$\bar{f}_n$ during the <i>n</i> th sliding cycle											
$\bar{f}$ 1500	2.40	2.12	1.89	1.50	1.30	1.30	3.8 ^b	2.88	2.10	2.32	2.01
$\bar{f}$ 3000	—	2.25	2.08	1.63 0.50 ^a	1.36 0.54 ^a	1.36	—	—	—	—	—
Results of the regression analysis on $\bar{f} = a + b \log n$											
<i>a</i>	0.62	0.66	0.38	0.39	0.27	0.27	0.67	0.72	0.54	0.43	0.46
<i>b</i>	0.58	0.43	0.43	0.34	0.31	0.31	0.99	0.60	0.51	0.50	0.57
Correlation coefficient <i>r</i>	0.99	0.95	0.94	0.99	0.98	0.98	0.97	0.96	0.99	0.98	0.98
Ratio $\bar{f}_2/\bar{f}_1$ of the average coefficients of friction during the first and second sliding passes in a single test	1.26	1.24	1.08	1.08	1.81	1.81	1.26	1.38	1.43	1.11	1.80
Ratio C_2/C_1 of the average conductivity values during the first and second sliding passes	4.41	3.41	2.11	2.11	2.58	2.58	1.97	1.65	1.78	1.64	2.41

^a Tests with anomalous silver behaviour (*cf.* Section 3.2).
^b At *n* = 1100.

Fig. 9. Average coefficients of friction $\bar{f}_1$ and $\bar{f}_{1500}$ during the first and 1500th sliding cycles respectively and the rate of increase b of friction with sliding distance as a function of the silver concentration in the pin material (cf. eqn. (1)): $\circ$, Au-Ag alloys, $\square$, Pd-Ag alloys.

3.4. Electrical conductivity and friction

In individual test runs the conductivity of a loaded contact always rises at the onset of sliding and increases more rapidly than friction during the first passes (Figs. 4, 10 and 11, Table 2). $\bar{C}$ is independent of the silver concentration for Au-Ag alloy sliding contacts. However, for Pd-Ag alloys the initial conductivity $\bar{C}_1$ clearly depends on the silver content and $\bar{C}$ increases to a saturation level $\bar{C}_s$ during n_s sliding cycles; both $\bar{C}_s/\bar{C}_1$ and n_s seem to increase gradually with increasing silver content. Short-lived fluctuations of relatively small amplitude occur within the fluctuation bands of the f and C signals (examples of relative band width are given in Table 3). These are superimposed on friction fluctuations of larger period and amplitude which exist from 50 - 100 sliding cycles onwards and on conductivity fluctuations which are present from the onset of sliding (cf. Fig. 4). The slow fluctuations are more obvious for gold than for palladium and an increase in silver content tends to suppress them. An interesting similarity is found between the friction and conductivity patterns: slow fluctuation always occurs synchronously and in the same sense. The increase of the signals is often gradual and the decrease is much faster or even abrupt. This kind of similarity may also occur with fast fluctuations.

Fig. 10. The electrical conductivity $\bar{C}$ before sliding, the average conductivity over the first and the second sliding cycles and the maximum average conductivity during a particular test as functions of the pin composition for Au-Ag alloys. The calculated conductivity C_{cl} for a single contact between the pin material and steel is also indicated (cf. point (4) of Section 4.2).

Fig. 11. The electrical conductivity $\bar{C}$ before sliding, the average conductivity over the first and second sliding cycles and the maximum average conductivity during a particular test as functions of the pin composition for Pd-Ag alloys, together with the calculated conductivity C_{cl} for a single contact between the pin and steel (cf. point (2) of Section 4.2). n_s is the number of cycles during which $\bar{C}$ increases.

No fixed quantitative relation between friction and conductivity exists, either between local or average values. This is clear from direct inspection and statistical analysis of the recording; for example Fig. 4 shows that the same friction may occur at several values of conductivity.

3.5. Wear, transfer and surface condition

Because no loose wear particles are formed all the material worn from the pin is found in a transferred layer on the plate. The amount of transferred material generally increases during the test run; however, in the case of anomalous silver behaviour (Section 3.2) most transferred material is ultimately back-transferred.

The net volumetric wear rate $\Delta V/\Delta s$ of gold pins decreases from an initial value of $0.2 \times 10^{-3} \text{ mm}^3 \text{ mm}^{-1}$ (on average over 11 cycles) to a

TABLE 3
Characteristic values of the relative bandwidths of the fluctuations in the friction and conductivity signals during the 1500th sliding cycle

	<i>Au-Ag alloys</i>					<i>Silver</i>	<i>Pd-Ag alloys</i>				
Silver content of the pin (at.%)	0 ^a	31	60	81		100	0	20	40	60	80
Relative bandwidth ^b $\Delta f/\bar{f}$	0.46	0.49	0.44	0.22		0.36 0.15 ^c	0.70	0.50	0.42	0.26	0.34
$\Delta C/\bar{C}$	0.18	—	0.44	0.34		— 0.14 ^c	0.40	0.56	0.50	0.28	0.26
Ratio of variations in friction and conductivity signals ^d $\left(\frac{\Delta f}{\bar{f}}\right)\left(\frac{\Delta C}{\bar{C}}\right)^{-1}$	2.55	—	1	0.65		— 1.07 ^c	1.75	0.89	0.84	0.93	1.31
$\frac{[1 + 0.5(\Delta f/\bar{f})] [1 + 0.5(\Delta C/\bar{C})]^{-2}}{[1 - 0.5(\Delta f/\bar{f})] [1 - 0.5(\Delta C/\bar{C})]}$	1.11	—	0.64	0.63		— 0.88 ^c	0.92	0.53	0.55	0.74	0.84

^a At $n = 1000$.

^b The bandwidth values Δf and ΔC are taken to be twice the standard deviation in the signals.

^c Anomalous silver behaviour (low friction).

^d This should be interpreted physically as the average variation in friction divided by the average variation in real contact area.

constant value of $6.7 \times 10^{-6} \text{ mm}^3 \text{ mm}^{-1}$ for 50 cycles and onwards. In terms of Archard's wear law

$$\Delta V/\Delta s = kF_N/H$$

which implies that k varies from 18.5×10^{-3} to $k \approx 0.65 \times 10^{-3}$. The total amount of material worn during a test decreases continuously with increasing silver content of the pin (*cf.* Fig. 12).

Fig. 12. Rough estimation of the average maximum thickness h of the transferred material on the track on the plate at the end of a test run for Au-Ag alloys.

During the first sliding cycle some shallow ($1 - 2 \mu\text{m}$ wide) and elongated grooves appear on the plate. A few tiny transferred particles (dimensions roughly $1 \mu\text{m} \times 1 \mu\text{m} \times 0.1 \mu\text{m}$) are also visible either in or at the end of the grooves or in isolated areas. The number and average dimensions of the particles then increase (length up to $10 \mu\text{m}$). Very small particles remain observable during the whole run. After some sliding (about 10 cycles for gold) larger secondary particles originate, each covering a few of the primary ones. In turn these secondary particles increase in number and dimensions (especially in length and height); obviously all subsequent contacts are concentrated on them. The shape and pattern of adjacent primary particles are unaffected.

The amount of transferred material decreases gradually with rising silver content. Pure gold and palladium ultimately yield a nearly continuous row of very large secondary particles. However, hardly any secondary particles are observed for silver; only a somewhat smeared primary layer occurs which may be removed on further sliding (Fig. 6). Pin surfaces show shallow furrows, smeared and torn material and, most obvious at a high palladium concentration, back-transferred particles.

3.6. Gold and silver sliding against themselves

Right from the start of sliding both sliding pairs of similar pure metals show a pattern of friction and conductivity similar to that found ultimately for gold against steel. The coefficient of friction is about the same, being about 2.4 for both gold and silver throughout the test (apart from incidental deviations). Under a load of 2.2 N the average conductivity amounts to about $1500 - 2000 \Omega^{-1}$ for the Au-Au system (comparable with gold

against transferred gold on steel at 3.5 N) and about $6000 \Omega^{-1}$ for the Ag–Ag system (about four times the value of $\bar{C}$ for the silver–steel system at 3.5 N).

4. Discussion

4.1. *Running properties of alloys and the solubility concept*

As in the Ag–Au–Fe and Ag–Pd–Fe systems no pertinent intermetallic compounds seem to occur (the Fe–Pd phase diagram shows some wide ranges of order only [8]) the Goodzeit rules simply predict an abrupt change from poor to good running properties at a critical silver concentration c_c where the mutual solubility of the pin material and the iron vanishes. This obviously is not in accordance with the present results for Au–Ag and Pd–Ag alloys sliding on plain carbon steel in UHV. A continuous decrease in the initial friction and transfer is observed with rising silver content. On further sliding these parameters increased. The eventual removal of transferred material at the highest silver concentrations after an appreciable number of sliding cycles does not disrupt this continuity; it can be considered as a secondary phenomenon which occurs after a normal process only.

This contradiction between the general trend in the present experimental results and the predictions of the solubility concept requires reconsideration of the picture of running behaviour.

4.2. *Development of friction and transfer processes*

(1) The initial contact between the pin and the plate material is hindered by adsorbed gases and perhaps by very thin oxide layers. Considering the steep rise in conductivity (Figs. 4, 10 and 11) contaminating films are extensively disrupted during the initial sliding cycles. As the area on the plate covered by transferred material is too small to bear the normal load the contact process is one of direct interaction between the pin material and steel, and interfacial properties may play a dominant part.

(2) On further sliding both the area of the plate track covered and the thickness of the adhering particles, especially the secondary particles, increase (Figs. 6, 12 and 13). Consequently the pin cannot touch the steel. Despite the absence of interfacial interaction between the pin material and steel at this stage the friction and transfer behaviour is different for the various pin materials (Fig. 9). Obviously mass properties have specific effects.

(3) The development of higher friction (Figs. 7 and 8) and of large simultaneous fluctuations in friction and conductivity (*cf.* Section 3.4) roughly correspond to the growth of the transferred layer and the increase in the apparent contact area by wear of the pin tip. After some 1000 sliding cycles the gold–steel system seems to attain a dynamic equilibrium (Fig. 7) in which both the pattern and the level of friction are the same as that of the Au–Au system (Section 3.6). The other Au–Ag alloys possibly tend to a

Fig. 13. Surface profile square to the sliding direction of a pure palladium pin and the corresponding track on the steel plate (three sections) at the end of a test run which shows that direct contact between the pin and the steel surface is impossible for geometrical reasons.

comparable situation which might not be attained because the transferred layer is still too thin at the end of a standard test run. Palladium obviously requires an even greater sliding distance to reach a steady state (Fig. 8). However, the results suggest that there is a positive relation between friction and the thickness of the transferred layer.

(4) It is easily seen that the constriction resistance R_c forms the most important part of the contact resistance after removal of the contaminating films. The conductivity R_c^{-1} of m independent equal clean circular contacts with a total area A_r between pin and plate can be calculated from [1]

$$C_{cm} = 4m^{1/2}(A_r/\pi)^{1/2}(\rho_1 + \rho_2)^{-1} \quad (2)$$

where ρ_1 and ρ_2 are the specific resistivities of the pin and plate material respectively. In the case of contact via a transferred layer of thickness h , ρ_2

should be taken to lie somewhere between the resistivities of the pin material and steel, depending on the relative layer thickness h/r where r is the contact radius.

If single point contact is assumed ($m = 1$ in eqn. (2)) and Bowden and Tabor's [1] simple approximation of real contact area

$$A_r = F_N/H_m \quad (3)$$

is used where F_N is the normal load and H_m is the indentation hardness of the softer partner so that $A_r \approx (0.3 - 1) \times 10^{-2} \text{ mm}^2$ for the materials under test, the calculated conductivity C_{c1} is of the same order of magnitude as that measured after some sliding (Figs. 10 and 11). The resistivity values used are shown in Fig. 14; as ρ was not measured for all materials figures from the literature were used [12, 13, 14]. The high values of ρ of most palladium alloys compared with steel may be responsible for the rapid vanishing of the increase in $\bar{C}$ on the formation of a transferred layer, while the low resistivity of the transferred gold alloys may contribute to the increase in $\bar{C}$.

The distinct synchronous fluctuations in f and c described in Section 3.4 most probably reflect life cycles of single contacts bearing the normal load alone or almost alone. Direct recordings (*cf.* Fig. 4) indicate that such

Fig. 14. Specific electrical resistivities of Au-Ag and Pd-Ag alloys and of steel taken from refs. 12, 13 and 14.

cycles may be sustained for a sliding length s_1 of 1 or a few millimetres of sliding which is equal to some 10 - 30 times the real contact length $l_c = 2r$ for a single circular contact following eqn. (3). Further, the recordings suggest that a life cycle is either terminated by sudden rupture of the junction (fall in C) or by a gradual shift of the load to a new contact.

Relative changes in the contact dimensions of single large junctions can be estimated from recordings of C using eqn. (2) as $A_r \approx C^2$ and $l_c \approx C$. The mean change in C during a contact cycle is about a factor of 2 (*cf.* Table 3) while the maximum changes do not exceed a factor of 4. For initial dimensions following eqn. (3) maximum values of l_c obviously remain well below 1 mm. In view of the length of the path of interaction s_1 it is concluded that junctions undergo growth and may possibly travel along the surfaces.

TABLE 4

Statistical evaluation of the relation between synchronous unidirectional slow changes in friction and conductance assumed to be connected to the life cycle of large contacts

	Average value $\bar{\Phi}$ of Φ^a		Standard deviation σ_Φ in Φ	$\sigma_\Phi/\bar{\Phi}$
Gold on steel ($n = 986$)				
Increasing friction	1.49	(7)	0.52	0.35
Decreasing friction	1.38	(5)	0.21	0.15
Palladium on steel ($n = 950$)				
Increasing friction	0.65	(5)	0.53	0.81
Decreasing friction	0.48	(5)	0.32	0.67
Silver on steel ($n = 100$)				
Increasing friction	1.09	(2)		

^a The number in parentheses is the number of events considered.

$\Phi = (f_2/f_1)(C_2/C_1)^{-2}$ in which the subscripts 1 and 2 refer to the onset and the end respectively of the event considered (see sketch).

No fixed quantitative relation exists between f and C (Section 3.4). Average and local changes in f and C (Tables 3 and 4) indicate that changes in the conductive area, at least in C^2 , tend to exceed those in frictional force.

(5) Removal of the transferred layer from the plate, as is sometimes observed at high silver concentrations after long sliding distances, is assumed to be a fatigue phenomenon in the interface between the transferred material and the steel. Obviously the strength of adhesion is not sufficient to stand

many repeated loading cycles of the layer. This may be associated with a possible continuous decrease of adhesion with increasing silver content, yielding strength values which are too low for permanent adhesion above a critical silver content (depending on the test conditions).

5. Model for the prediction of the running behaviour of metallic sliding couples

5.1. Model for sliding contact between metals

5.1.1. General situation

The real contact between sliding bodies usually occurs simultaneously on a number m of areas ΔA_r within the nominal or apparent contact area A_a [1]. The adjacent material in one or both of the bodies may be deformed appreciably by the action of the forces transmitted through a contact. A zone in which deformation occurs is called a junction, and a situation as shown in Fig. 15(a) ensues: the bulk of body 1 moves with velocity v , that of body 2 is stationary and the material of the junctions has some intermediate motion. The running behaviour of the sliding pair is governed by the processes taking place within the junctions. As the conditions of geometry, material and surface may vary over A_a several types of junction development are possible. In any case, however, a junction has to bridge entirely the difference in the bulk motion of the two bodies (Fig. 15(b)). The way in which a particular junction performs this duty depends at least on the interfacial interaction between the contacting materials and the mechanical response of the junction material (including possibly transferred material) to this interaction (*cf.* Section 4.2). The main features of a model for junction behaviour in which these influences are incorporated are outlined in the next Section. A more detailed discussion is given elsewhere [15].

5.1.2. Junction behaviour and its principal parameters

5.1.2.1. Transmission of tangential load. Let the interfacial shear strength for the junction shown in Fig. 16(a) be τ_i (*cf.* Section 5.1.3) and the initial yield strengths in shear and strain hardening exponents of junction material in bodies 1 and 2 be τ_1 , n_1 , τ_2 and n_2 . The normal load is ΔF_N and the velocity difference across the junction is v .

The forces acting in the contact area will be transmitted through the junction to both bulk materials. It can be argued that after deformation associated with accommodation of the normal load the stress pattern within the junction may roughly develop to a combination of hydrostatic pressure and pure shear stresses parallel to v ; only the latter possibly gives rise to further deformation. On increasing the depth from ΔA_r the zones concerned with transmission of the load are extended so the stress level decreases. Planes of constant maximum shear stress τ just above ΔA_r may be shaped as shown in Fig. 16(b).

5.1.2.2. Strong and weak junctions and the strength ratio. The way in which the junction bridges the imposed velocity difference depends crucially on the ratio T between the strength τ_i of the interface and the strength τ^* of

Fig. 15. Schematic representation of the contact and the velocity distribution for a solid body (1) sliding over another body (2). (a) Real contact occurs at m spots of area ΔA_r within the apparent contact area A_a ; a junction consists of the interface ΔA_r and the adjacent sheared zones in one or both of the bodies. (b) Types of primary velocity distribution over a junction: (1) weak junction, interfacial slip and hardly any subsurface shearing; (2) strong junction, shear concentrated in the upper body and hardly any interfacial slip; (3) strong junction, shear spread over both bodies. u is the local velocity and v the overall velocity. (c) Intermediate junction types (1) with both slip and deformation and (2) with a transition from no slip to slip within ΔA_r .

Fig. 16. Some features of a junction. (a) Normal and tangential loads ΔF_N and ΔF_T transmitted by stresses τ^* in the junction material and in equilibrium with the normal and shear stresses σ and τ acting across the interface of area ΔA_r and length l . The value of τ^* depends on the particular stress condition and the material properties τ_1 and n (initial yield strength and strain-hardening exponent respectively). (b) Rough shape of the isoshear stress planes in a strong junction; the local maximum shear stress τ decreases with increasing distance from the interface.

the adjacent junction material. When both T_1 and T_2 are less than unity (where $T_1 = \tau_i/\tau_1^*$ and $T_2 = \tau_i/\tau_2^*$) the interface is the weakest section of the junction. Such a weak junction will not be sheared; the velocity drop is concentrated in interfacial slip (*cf.* Fig. 15(b1)). If one or both T values exceeds unity, however, the weakest section is found beyond ΔA_r and extensive shear may occur. The deformation process in such a strong junction is spread over both materials if $\tau_1^* \approx \tau_2^*$ and otherwise is concentrated in the weakest member (Figs. 15(b3) and 15(b2) respectively). It should be noted that τ^* itself is not a material property: starting from $\tau^* = \tau_1$ or $\tau^* = \tau_2$ it can be increased by strain hardening and imposed shear rate and possibly be reduced by a rise in temperature and effects such as dynamic recrystallization [16]. Consequently T values may vary during the contact process and over the contact area which possibly results in transitions of the junction behaviour (Fig. 15(c2)) or the occurrence of intermediate behaviour (Fig. 15(c1)). The critical value of T depends slightly on junction geometry.

5.1.2.3. Junction development; strain hardening and adhesion. In order to satisfy the conditions of mechanical equilibrium and thus achieve a stress distribution as indicated in Fig. 16(b) the specific shear and shear rate in a strong junction gradually decrease from maximum values near the interface to zero at the outer boundary of the junction. Thus the depth δ of the deformed zone depends strongly on the strain-hardening exponent: $\delta \approx 0$ for $n = 0$ (ideal plastic behaviour) and an extended deformation zone results for high n values (Fig. 17(a)). Obviously the local velocity u within the junction differs from the bulk velocities (v and 0 in Fig. 15). Consequently a material surplus occurs at the "inlet" side of the junction and a deficit at the "outlet" (Fig. 17(b)). This results in extra compressive stresses and junction growth at the inlet and tensile stresses at the outlet. The latter may either lead to rupture of the contact in the interface at low adhesion or constriction or even cracking (equivalent to onset of transfer) at high adhesion values [15]. These effects are more pronounced for a greater strain-hardening tendency of the material (bulk property) and stronger adhesion (interfacial property). At $\tau_1^* \approx \tau_2^*$ the extra stresses produce tilting of the contact plane and thus enhance the effects. An increase in ΔA_r will inevitably increase the tangential force needed for further sliding.

5.1.2.4. Termination of junction life. The material surplus at the junction inlet resulting from the difference in bulk and junction velocities tends either to increase the normal load on that particular junction (multispot contact) or to increase the normal distance between the sliding bodies (single-spot contact). This situation is stable as long as the material supply to the junction is guaranteed. However, disruption of the contact may occur if the material supply vanishes in both bodies which can be due to several reasons, *e.g.* when the junction meets some topographical irregularity of a depth exceeding that of the deformed zone as it travels along the surface. The governing factor is therefore some ratio of roughness height and a combination of the contact length and strain-hardening exponent.

Fig. 17. Effects of the strain-hardening exponent n on the depth of deformation δ and the growth of a junction. (a) (0), Velocity distribution beyond the deformation zones; (1), (2), deformed zones for low and high n values respectively in a junction with $\tau_1^* < \tau_2^*$; (3) deformed zones for $\tau_1^* \approx \tau_2^*$. (b) Shape of the junctions after a macroscopic sliding distance Δs .

5.2. The interfacial shear strength

5.2.1. Atomic contact area and the interfacial shear strength

The transmission of forces across the interface of two contacting materials A and B occurs by the interaction of forces between the atoms on both sides of that interface. High interatomic forces are only possible between atoms which are in close proximity, say within a distance of about 0.3 nm (cf. Fig. 18). As real surfaces usually have a submicroroughness associated with microsteps, facets etc. superimposed upon the "common roughness" the distance between A and B in several parts of the conventional

Fig. 18. Qualitative picture of the interatomic force-distance relation (curve 1) and the approximation of this relation by a half sine curve (curve 2); the characteristic distance b is of the order of magnitude of the atomic diameter, i.e. about 0.3 nm.

real contact area A_r will be too great to obtain strong atomic interaction. Therefore a subdivision of A_r into the atomic contact area A_{at} in which high interfacial forces are possible and a remaining area A_{sep} in which no or only very small forces can be transmitted is proposed (Fig. 19). The fraction of the real contact area where atomic contact occurs is denoted ϕ where $0 < \phi \leq 1$ so that

$$A_{at} = \phi A_r \quad (4)$$

The value of ϕ may possibly vary during sliding, *e.g.* owing to a process of shearing. For the time being, however, we assume ϕ to be the same for all contacts.

Fig. 19. Subdivision of the conventional contact area A_r in areas of atomic contact A_{at} and areas without atomic force interaction. (a) Contact between the metals A and B on a microscopic scale; the contact is concentrated in real contacts. (b) Schematic representation on the atomic scale of part of the interfacial zone of the real contact depicted in (a); the atomic force interaction across the interface is assumed to occur only in the areas of close approach ΔA_{at} (cf. Fig. 18). (c) Representation of the atomic junctions within the interfacial zone of a conventional junction; as in (b) except that the material is assumed to be a continuum.

The individual atomic contact regions are thought to remain very small, say of the order of $\Delta A_{at} \approx 100 \text{ nm}^2$. This implies that no or hardly any dislocations can be situated within an atomic junction so that its deformation requires stresses equal to the theoretical strength.

The theoretical shear strength τ_{th} of metals is proportional to the shear modulus G [17] so

$$\tau_{th} = \kappa G \quad (5)$$

For face-centred cubic and body-centred cubic metals κ is estimated to be about 1/30 and 1/10 respectively. An estimation of the theoretical shear strength τ_{thi} of the interface is given below. Once τ_{thi} is known the interfacial shear strength which is related to the real area of contact A_r follows from

$$\tau_i = \phi \tau_{thi} \quad (6)$$

The stress concentrations at the edges of the atomic contacts are neglected. For the problem of adhesion the tensile strength of the interface of a junction can likewise be related to the theoretical tensile strength of the atomic contacts.

5.2.1. Estimation of the theoretical shear strength of the atomic contacts

The theoretical strength of an interface between different metals A and B can be calculated if the force of interaction between the metals as depicted qualitatively in Fig. 18 is quantitatively known. Unfortunately this is not the case. It is therefore necessary to obtain a rather rough estimation by assuming a sinusoidal force interaction across the interface (*cf.* Fig. 18):

$$\sigma = \sigma_{thi} \sin\left(\pi \frac{x}{b}\right) \quad (7)$$

where σ is the tensile force per surface unit, σ_{thi} the theoretical interfacial tensile strength, x the distance between the contacting surfaces and b the half wavelength of the sine curve.

The work done by σ on tearing the metals from each other over the interface is assumed to be equal to the increase $\Delta\gamma$ in surface energy associated with the change from one interface to two free surfaces (all the quantities are per surface unit) so that

$$\Delta\gamma = \int_0^b \sigma_{thi} \sin\left(\pi \frac{x}{b}\right) dx$$

which on integration becomes

$$\sigma_{thi} = \frac{\pi}{2} \frac{\Delta\gamma}{b} \quad (8a)$$

In fact both $\Delta\gamma$ and b are required to calculate σ_{thi} with eqn. (8a). b is found by reversing the problem into a case in which σ_{thi} is approximately

known, *i.e.* the cleavage of a metal. Now σ_{thi} is equal to the theoretical strength of the metal and $\Delta\gamma$ is up to twice the surface energy. When these values are substituted in eqn. (8a) for a series of metals a reasonable value of b appears to be $\frac{3}{2}a$ where a is the smallest atomic distance in the lattice. Therefore in the case of different interacting metals $b = \frac{3}{4}(a_1 + a_2)$ where a_1 and a_2 are the smallest interatomic distances in both interacting lattices. Thus eqn. (8a) becomes

$$\sigma_{thi} = \frac{2\pi}{3} \frac{\Delta\gamma}{a_1 + a_2} \quad (8b)$$

The theoretical shear strength τ_{thi} of an interface is related to σ_{thi} as both quantities depend fundamentally on the interatomic forces across the interface. However, both the path of interaction and the direction of the resulting forces with respect to the position of the atoms during shearing deviates appreciably from those observed during tearing. Therefore the derivation of τ_{thi} from σ_{thi} is a complicated affair. We assume provisionally that

$$\tau_{thi} = \psi \sigma_{thi} \quad (9)$$

in which the factor of proportionality at least depends on the types of lattice and their parameters meeting at the interface. For the combinations tested ψ is assumed to be a constant, the value of which might be about 0.15.

5.2.2.1. Estimation of $\Delta\gamma$. Miedema and Den Broeder [18] calculated the interfacial energy of solid-liquid and solid-solid metal combinations. They proposed the following expression for the adhesion energy $\Delta\gamma$ when two different clean surfaces of metals A and B are replaced by an A-B contact:

$$\Delta\gamma = -\gamma_A^{S,0} - \gamma_B^{S,0} + \gamma_{SS}^{AB} \quad (10)$$

in which $\gamma_A^{S,0}$ and $\gamma_B^{S,0}$ are the surface energy of metals A and B respectively at 0 K and γ_{SS}^{AB} is the interfacial enthalpy of the interface between metals A and B.

The interfacial enthalpy can be written

$$\gamma_{SS}^{AB} = 0.15(\gamma_A^{S,0} + \gamma_B^{S,0}) + \gamma_{SS}^{chem} \quad (11)$$

in which γ_{SS}^{chem} is the chemical interaction enthalpy of the atoms of A and B, and $0.15(\gamma_A^{S,0} + \gamma_B^{S,0})$ is the average high angle grain boundary enthalpy (grain boundary enthalpies are estimated to be about 30% of the zero temperature surface energy). Combining eqns. (10) and (11) gives

$$\Delta\gamma = -0.85(\gamma_A^{S,0} + \gamma_B^{S,0}) + \gamma_{SS}^{chem} \quad (12)$$

5.2.2.2. The contact of iron with silver, gold and palladium. An estimation of the theoretical interfacial tensile strength σ_{thi} using eqns. (8b) and (12) and data from the literature is given in Table 5. The values of γ_{SS}^{chem} for the iron/silver and iron/gold interfaces have been calculated by Miedema

TABLE 5

Estimation of the energy of adhesion $\Delta\gamma$ and the theoretical tensile strength σ_{thi} of the contact interface of some pure metals

Contacting metals	Equation (12)			Equation (8b)	
	$\gamma_A^{S,0} + \gamma_B^{S,0}$ (J m ⁻²)	γ_{SS}^{chem} (J m ⁻²)	$\Delta\gamma$ (J m ⁻²)	$a_1 + a_2$ (10 ⁻¹⁰ m)	σ_{thi} (10 ⁶ N mm ⁻²)
Iron-silver	1.6 + 0.9	+ 0.62	-1.51	2.48 + 2.88	5900
Iron-gold	1.6 + 1.1	+ 0.19	-2.11	2.48 + 2.88	8245
Iron-palladium	1.6 + 1.5	-0.24	-2.88	2.48 + 2.74	11555

and Den Broeder [18], but no value for the iron/palladium interface was given in their paper. However, $\gamma_{SS}^{chem}_{Fe-Pd}$ can be estimated from the enthalpy of mixing of palladium and iron calculated by Boom and De Boer [19] to be $\Delta H = -18\,900 \text{ J (g atom)}^{-1}$. Assuming that there are $N_1 = 1.5 \times 10^{19}$ atoms m⁻² on an interface and that these atoms have contact with the other metal over a quarter of their 'surface', it follows that

$$\gamma_{SS}^{chem}_{Fe-Pd} = 2 \times \frac{1}{4} \times \Delta H \frac{N_1}{N_0} = -0.24 \text{ J m}^{-2}$$

where $N_0 = 6 \times 10^{23}$ is Avogadro's number.

As far as single-phase alloys are concerned, which is the case in the present experiments, the effect of alloying can be expected to be a continuous change of σ_{thi} with concentration of the alloying element.

5.2.3. The ratio of the strength of the interface and of the adjacent junction material

The junction behaviour essentially depends on the strength ratio T between the interface and the adjacent material (*cf.* Section 5.1). Comparison of the estimated values of τ_{thi} with the initial strength of the materials under test as indicated by their hardness shows that the atomic contacts in the first instance will act as strong nodes in a junction (Tables 1 and 5). Consequently the junction material, at least that in the immediate vicinity of the interface, will be sheared during sliding so that its strength will increase by strain hardening. The local strength τ^* attained is difficult to estimate but for low temperatures it is assumed that τ^* can approach the ultimate shear strength τ_u . Orowan [20] calculated the shear strength as

$$\tau = \tau_0 + (b/l)G$$

in which τ_0 is the shear strength of the metal without obstacles for dislocations, b the length of the Burgers vector, l the distance between obstructing pinning points to dislocations and G the shear modulus. For an extremely cold deformed metal a value of 10 nm is proposed for l . In this case τ_0 can be neglected and

$$\tau^* \leq \tau_u = \xi G \quad (13)$$

in which the constant of proportionality $\xi = b/l$, e.g. for iron $\tau_u = (0.25/10)G = (1/40)G$.

Two possible formulations of the ratio T which we propose calling the *relative interfacial strength* appear: for the atomic contacts

$$T_{\text{at}} = \frac{\tau_{\text{thi}}}{\tau_{\text{th}}} = \frac{\psi}{\kappa} \frac{\sigma_{\text{thi}}}{G} \quad (14)$$

where τ_{th} is the value of τ for the weakest of the two contacting metals; with respect to the real contacts

$$T_{\text{r}} = \frac{\tau_{\text{i}}}{\tau^*} = \frac{\phi \tau_{\text{thi}}}{\tau_{\text{u}}} = \frac{\phi \psi}{\xi} \frac{\sigma_{\text{thi}}}{G} \quad (15)$$

where τ_{u} is the value of τ for the weakest material (the constants κ , ϕ , ψ and ξ are defined in eqns. (5), (6), (9) and (13) while σ_{thi} follows from eqn. (8)).

T_{at} may possibly be less than unity for material combinations with extremely good tribological compatibility; this implies that even the atomic contacts show weak behaviour so that neither junction growth nor any transfer can take place. The condition $T_{\text{at}} > 1$ and $T_{\text{r}} < 1$ is less discriminating but could still be associated with good running properties; transfer on an atomic scale is possible but the junction as a whole behaves weakly. Strong junction behaviour may be expected for $T_{\text{r}} > 1$.

Owing to the scatter of conditions over the apparent contact area material combinations with a value of T close to unity may show several different types of function simultaneously. At present an exact prediction of the sliding behaviour of a particular metallic couple is not possible because the values of ϕ , ψ , κ , ξ and σ_{thi} are not available with sufficient accuracy. However, comparison of the behaviour of a series of material combinations of the same type is possible as in that case the values of ϕ , ψ , κ and ξ can be assumed to be constant so that the provisional basis for the comparison becomes the ratio σ_{thi}/G (cf. eqns. (14) and (15)). Application of this idea to the contact of silver, gold and palladium with iron yields the results given in Table 6. The T values for the Ag-Fe contact are smaller than those for the Au-Fe and Pd-Fe systems. The conditions for Ag-Fe contact might therefore be in the neighbourhood of the weak-strong transition ($T_{\text{r}} = 1$), while those for Au-Fe or Pd-Fe contact require strong behaviour (the rough estimates of ψ and κ given earlier yield values of T_{at} of 0.90, 1.16 and 1.13 for silver, gold and palladium respectively in contact with iron).

Regarding the behaviour of Au-Ag and Pd-Ag alloys, a continuous decrease of the proportion of strong and weak contacts with increasing silver content can be expected as the quantities used in the above example change continuously with composition.

TABLE 6

Comparison of the characteristic strength properties of some metals and of the interfaces of these metals in contact with iron

Metal (Me)	Material property G_{Me} (10^{10} N m^{-2})	Interfacial property σ_{thi} of the system Me-Fe (10^{10} N m^{-2})	Relative interfacial shear strength	
			Equation (14): $T_{at} = \tau_{thi}/\tau_{th Me}$	Equation (15): $T_r = \tau_i/\tau_u Me$
Silver	2.96	0.59	0.199 ψ/κ	0.199 $\phi\psi/\xi$
Gold	3.2	0.825	0.258 ψ/κ	0.258 $\phi\psi/\xi$
Palladium	4.6	1.156	0.251 ψ/κ	0.251 $\phi\psi/\xi$

5.3. Comparison of the model with the experimental results

5.3.1. Initial friction versus composition

If a roughly comparable fraction ϕ of the real contact area is assumed to be used for strong interatomic interactions in all the materials used the decrease of the relative interfacial shear strength, given by the ratio $T = \tau_i/\tau^*$, with increasing silver content of the pin can predict a gradual change from strong to intermediate or even weak junctions. Consequently the average junction growth is restricted and therefore the friction is reduced with increasing silver concentration (Figs. 7, 8 and 9). As the strain-hardening effects in palladium exceed those in gold the junction growth of palladium is greater, resulting in a higher initial friction. Since T is always comparatively high in the contact of like metals the friction of both the Ag-Ag and Au-Au systems is high.

5.3.2. Development of the transferred layer versus composition

As the strain-hardening behaviour changes continuously with composition the deformation depth δ of the junctions is high for pure gold and palladium (the strain-hardening exponents n are 0.49 and 0.52 respectively) and decreases with increasing silver concentration to a low value for pure silver ($n = 0.44$). Consequently junction development is less intense for the silver-rich alloys, so that in combination with the decrease in T_r the transfer rate decreases with increasing silver concentration [15]. (It should be noted that after some sliding the effective n values are reduced but their order of magnitude is thought to remain unchanged.)

5.3.3. Friction versus sliding distance

When a transferred layer with a hardness which is about the same as that of the actual pin surface and appreciably lower than that of the steel is developed the system gradually changes from deformable *versus* rigid (Figs. 15(b1) and 15(b2)) to deformable *versus* deformable (Fig. 15(b3)). Consequently the mutual interaction of the contact partners is enhanced, which results in a more obvious junction growth and a higher friction for all the

materials. The observation of a greater increase of conductivity than corresponds to the frictional increase during a junction life cycle possibly indicates a reduction of the separation within A_r or an increase in ϕ as a result of better geometrical conformation within the real areas of contact.

6. Conclusions

(1) For Au–Ag and Pd–Ag alloys sliding against steel in UHV both the coefficient of friction and the amount of transferred material gradually decrease with increasing silver concentration.

(2) No abrupt change in sliding behaviour is found at the change of mutually soluble to mutually insoluble partners; definitely insoluble combinations such as silver on steel show metal transfer in UHV.

(3) The above observations reported are not in accordance with the solubility concept for running properties; they also cannot be explained satisfactorily by Bowden and Tabor's theory of the effect of a contaminant layer.

(4) A model for sliding of metals at immediate mechanical contact is proposed which utilizes the atomic interaction in the interface of the junctions and the mechanical response of the junction material to this interaction as central elements.

(5) In this model the intensity of the atomic interaction is related to the enthalpy of binding across the interface.

(6) The actual value of the strain-hardening exponent is thought to be of crucial importance in the behaviour of the mechanical junction; the behaviour also depends on geometry.

(7) For the best sliding properties the enthalpy of binding of the metals and the strain-hardening exponent of the weakest partner should be as low as possible; the ratio of the net interfacial shear strength and the shear strength of both contacting metals after strain hardening should at least be less than unity.

Nomenclature

a	smallest atomic distance in a lattice, m
A_a	apparent area of contact, m ²
A_{at}	atomic contact area (area of strong interfacial forces), m ²
A_r	real area of contact, m ²
ΔA_r	interfacial area of a single real contact, m ²
b	half wavelength of the sine curve describing an assumed force interaction, m
C	(instantaneous) electrical conductivity, Ω^{-1}
$\bar{C}$	average electrical conductivity during sliding cycle, Ω^{-1}
f	ratio of the instantaneous frictional force to the normal force
$\bar{f}$	average coefficient of friction over one sliding cycle
$\bar{f}_n$	coefficient of friction during the n th sliding cycle averaged over all tests with a particular sliding pair

F_N	normal load on the (apparent) contact, N
ΔF_N	normal load on a real contact, N
G	shear modulus, N m^{-2}
H	indentation hardness, N m^{-2}
m	number of simultaneous real contacts
n	strain-hardening exponent in Nadai's law
	number of sliding cycles
s	sliding distance
T_{at}	$\tau_{\text{thi}}/\tau_{\text{th}}$, ratio of the shear strength of the interface to the shear strength of the adjacent material in an atomic junction (<i>cf.</i> eqn. (14))
T_r	τ_i/τ^* , ratio of the shear strength of the interface to the shear strength of the adjacent junction material in a real contact (<i>cf.</i> eqn. (15))
u	local (microscopic) velocity, m s^{-1}
v	overall (macroscopic) sliding velocity, m s^{-1}
γ	surface energy, J m^{-2}
$\Delta\gamma$	increase in the surface energy associated with the change from one surface unit of interface to two free surfaces (adhesion energy), J m^{-2}
δ	deformation depth of a junction (Fig. 17), m
κ	ratio of the theoretical shear strength to the shear modulus of a metal
ξ	ratio of the shear strength to the shear modulus of an extremely cold deformed metal, N m^{-2}
σ	normal stress, tensile strength, N m^{-2}
σ_{th}	theoretical strength of a metal on normal loading, N m^{-2}
τ	shear stress, shear strength, N m^{-2}
τ^*	shear strength of a junction material, N m^{-2}
τ_1, τ_2	initial shear strength of the materials of bodies 1 and 2 respectively, N m^{-2}
τ_i	average interfacial strength of a real contact, N m^{-2}
τ_{th}	theoretical shear strength, N m^{-2}
ψ	ratio of the (theoretical) shear strength to the tensile strength of the interface of an atomic contact

Subscripts

i	interface
1, 2	material of bodies 1 and 2 respectively

Acknowledgments

The authors wish to thank Messrs J. P. M. Faessen and C. J. M. Meesters (cooperators of the Wear Research Section, Technische Hogeschool Eindhoven under the chairmanship of Professor J. H. Zaat) for their contributions to this work.

References

- 1 F. B. Bowden and D. Tabor, *The Friction and Lubrication of Solids*, Parts I and II, Oxford Univ. Press: Clarendon Press, Oxford, 1954, 1958.
- 2 A. P. Green, Friction between unlubricated metals: a theoretical analysis of the junction model, *Proc. R. Soc. London, Ser. A*, 228 (1955) 191 - 204.
- 3 C. M. Edwards and J. Halling, An analysis of the plastic interaction of surface asperities and its relevance to the value of the coefficient of friction, *J. Mech. Eng. Sci.*, 10 (2) (1968) 101 - 110.

- 4 C. L. Goodzeit, R. P. Hunnicutt and A. E. Roach, Frictional characteristics and surface damage of thirty-nine different elemental metals in sliding contact with iron, *Trans. ASME*, 78 (1956) 1669 - 1676.
- 5 A. E. Roach, C. L. Goodzeit and R. P. Hunnicutt, Scoring characteristics of thirty-eight different elemental metals in high-speed sliding contact with steel, *Trans. ASME*, 78 (1956) 1659 - 1667.
- 6 E. Rabinowicz, *Friction and Wear of Materials*, Interscience, New York, 1965.
- 7 A. W. J. De Gee, The friction of gold-silver alloys against steel, *Wear*, 8 (1965) 121 - 132.
- 8 M. Hansen, *Constitution of Binary Alloys*, 2nd edn., McGraw-Hill, New York, 1958.
- 9 L. K. J. Vandamme, On 1/f noise in ohmic contacts, *Thesis*, Technische Hogeschool Eindhoven, 1976.
- 10 D. Maugis, G. Desalos-Andarelli, A. Heurtel and R. Courtel, Adhesion and friction on Al thin foils related to observed dislocation density, *ASME/ASLE Lubrication Conf., Boston, Mass., Oct. 1976, Pap. 76-LC-2B-2, ASLE Prepr.*, 1976.
- 11 R. Holm, *Electrical Contacts*, 4th edn., Springer, Berlin, 1967.
- 12 A. Goldsmith, T. E. Waterman and H. J. Hirschhorn, *Handbook of Thermophysical Properties of Solid Materials*, Pergamon Press, Oxford, 1961.
- 13 F. Gaibullaev and A. R. Regel, Some characteristics of temperature dependence of electrical conductivity of single phase solutions of Ag-Au, In-Pb and Bi-Sb systems in the solid and liquid states, *Sov. Phys.-Tech. Phys.*, 2 (10) (1957) 2082 - 2086.
- 14 W. K. Chen and M. E. Nicholson, Influence of annealing on electrical properties of cold worked Ag-Pd alloys, *Acta Metall.*, 12 (6) (1964) 687 - 696.
- 15 D. Landheer and J. H. Zaat, The mechanism of metal transfer in sliding friction, *Wear*, 27 (1974) 129 - 145.
- 16 J. A. B. van Dijck, The direct observation in the transmission electron microscope of the heavily deformed surface layer of a copper pin after dry sliding against a steel ring, *Wear*, 42 (1977) 109 - 117.
- 17 B. L. Averbach, in H. Liebowitz (ed.), *Fracture*, Vol. 1, Academic Press, New York, 1968, p. 441.
- 18 A. R. Miedema and F. J. A. Den Broeder, On the interfacial energy in solid-liquid and solid-solid metal combinations, *Z. Metallkd.*, 70 (1979) 14 - 20.
- 19 R. Boom and F. R. De Boer, On the heat of mixing of liquid alloys - II, *J. Less-Common Met.*, 46 (1976) 271 - 284.
- 20 E. Orowan, *Symposium of Internal Stresses in Metals and Alloys*, Institute of Metals, London, 1948, p. 451.