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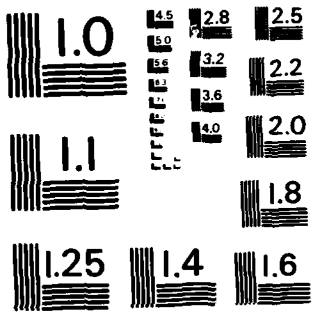

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FINAL REPORT

Contract N00014-82-K-0267

DESIGN AND TESTING OF HIGH PERFORMANCE BRUSHES

Feb. 1, 1982 - May 31, 1983

Submitted to:

Office of Naval Research
800 N. Quincy Street
Arlington, Virginia 2217
Materials Division

Submitted by:

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Charlottesville, Virginia 22901

May 1983

Final Report to Contract N00014-82-K-0267

DESIGN AND TESTING OF HIGH-PREFORMANCE BRUSHES

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SUMMARY

The aim of this contract has been to increase still further the capabilities of high-performance elctrical brushes, as commercially available as well as, indeed with particular emphasis on, the metal fiber and QM brushes invented under the auspices of the preceding contract under the same name, N00014-76-C-1009 from July 1, 1976 to January 31st, 1982.

The result of this research has been to confirm and expand somewhat both the theoretical and experimental advances made in the preceding six years. Additionally contributions were made to ancillary research areas, specifically research on friction and wear, and the dislocation behavior controlling wear.

From a technical standpoint the major success of this present project has been to intensify our understanding of the so-called mode II operation of the traditional silver-graphite brushes (75w/o Ag) which at this point are the best commercially available brushes for high current density applications. It has been clearly recognized that mode II has been previously unknown but can be highly advantageous. Mode II establishes itself automatically if the flash point temperature at the contact spots exceeds some specific level, now believed to be about 300°C. In the past only "mode I" has been employed since, for fear that the brushes would "dust" at catastrophic rates if the temperature were permitted to rise into the region characteristic of mode II, forced cooling was employed and the current density was limited to presumed "safe" levels. However, in mode II, easily obtained in the atmosphere (i.e. without the typically used protective atmosphere) and without cooling, the normally present lubricating surface film disintegrates but a protective film of almost solid silver is established. This second film has a much lower resistance and thus the overall brush resistance drops considerably. Also, as an added benefit, the brush "noise" is strongly reduced. In this condition the current density could be increased far beyond previously accepted limits, while the wear was less than extrapolated for similar current densities in mode I without catastrophic dusting.

Efforts have been made to construct a serviceable QM brush, consisting of copper fibers of about 30 μm thickness, out of each of which would project 19 gold fibers of 2 μm thic-ness. This project has progressed to the point that another two man months or so would be required for completion. It has been laid aside for the time being because it had exceeded the estimated time limit and other work was more pressing.

Our D. Gerdt (very soon to be Dr. Gerdt, i.e. after out spring graduation this month) has screened a variety of metals and atmospheres for possible use as industrially applicable metal foil brushes. Foil brushes were first develop- by P. Haney in his M.S. thesis research. David, before leaving for a position in industry as a research engineer, confirmed the prior results but not suffic- ient new material was gathered to either arrive at industrially worthwhile results of to expand the previously published work.

The experimental research under this contract has been carried out by D. Gerdt, M. Ijaz, W. A. Jesser, L. B. Johnson, R. E. Miller and H. G. F. Wilsdorf

LECTURES/TECHNICAL REPORTS

1) A series of six lectures at the National Bureau of Standards, Center for Materials Science, Washington DC on dislocation behavior and mechanical properties of metals in a wide range of straining conditions.
2) "Two Regimes of Current Conduction in Metal-Graphite Electrical Brushes and Resulting Instabilities", 28th Holm Conference on Electrical Contacts, Chicago, Sept. 23/25 1982, by S. Dillich and D. Kuhlmann-Wilsdorf.
3) "How Readily are Vacancies Annihilated at Dislocations?", Seminar, Oak Ridge National Laboratory (Metals and Ceramics Division), D. Kuhlmann-Wils- dorf, Nov. 11, 1982.
4) "The Physics of Electrical Contacts: Materials for Electrical Brushes", D. Kuhlmann-Wilsdorf, joint meeting of the Oak Ridge Chapter of ASM and the Tennessee Section of the Am. Soc. for Quality Control, co- sponsored by the Society of Women Engineers. Nov. 11, 1982.

PUBLICATIONS

1) " The Role of Vacant Lattice Sites in the Low-Amplitude Failure of Low-Alloy Steel with Inclusions" - D. Kuhlmann-Wilsdorf and P. F. Thomason, Acta Met. 30 (1982) pp.1243-1245.
2) "Theory of Dislocation Cell Sizes in Deformed Metals" - D. Kuhlmann-Wilsdorf and J. H. van der Merwe, Mat. Sci. Eng. 55 (1982) pp.79-83.
3) "Two Regimes of Current Conduction in Metal-Graphite Electical Brushes and Resulting Instabilities" - S. Dillich and D. Kuhlmann-Wilsdorf, Electrical Contacts - 1982 (Ill. Inst. Techn., Chicago, IL, 1982) pp.201-212.
4) "On Orientation Changes, Deformation Banding and Dislocation Structures in Rolled [100](001) Oriented Iron Single Crystals" - D. Kuhlmann-Wilsdorf and E. Aernoudt, J. Appl. Phys. 54 (1983) pp.184-192.
5) "Sub-Surface Hardening in Erosion Damaged Copper as Inferred from the Dislocation Cell Structure and Its Dependence on Particle Velocity and Angle of Impact" - D. Kuhlmann-Wilsdorf and L. K. Ives, Wear 85 (1983) pp.359-371.
6) "Dislocation Cell Formation in Unidirectional Glide of FCC Metals - Part I: Basic Theoretical Analysis of Cell Walls Parallel to the Primary Glide Plane in Early Stage II" - D. Kuhlmann-Wilsdorf and N. R. Comins, Mater. Sci. Engg. 1983, in the press.
7) "Gold Fibre Brushes - Their Promise for Future High-Technology Applications"- D. Kuhlmann-Wilsdorf, Gold Bulletin, 16(1) (1983) pp.12-20.
8) "Extended Performance Limits of Metal-Graphite Brushes at Very High Current Densities" - S. Dillich and D. Kuhlmann-Wilsdorf, Mater. Sci. Engg. 57 (1983) L13-L16.
9) "Performance Characteristics of Silver-Graphite Electrical Brushes (75w/o Ag) Without Cooling up to 1400 A/cm^2" - L. B. Johnson and D. Kuhlmann-Wilsdorf, Mater. Sci. Engg. 58 (1983) L1-L4.

Copies of the title pages of the above papers are appended herewith.

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THE ROLE OF VACANT LATTICE SITES IN THE LOW-AMPLITUDE FATIGUE FAILURE AT INCLUSIONS IN STEEL

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(*Received* 15 *September* 1981)

Abstract—The low-amplitude fatigue limit of low-alloy steels with inclusions of alumina and manganese sulphides lies below the yield stress in unidirectional deformation. Failure occurs without any evidence for irreversible dislocation motion. It is initiated by decohesions at particle/matrix interfaces at the specimen surface, whence fatigue cracks begin to spread in the directions predicted from the applied stress. No decohesion or cracking is observed below the surface. It is suggested that the decohesions are due to the preferential condensation of vacancies at the interfaces at the surface, wherein the driving force is due to the oxidation of the freshly formed surfaces at the decohesions, while the vacancies are formed at jogs of oscillating dislocations.

Résumé—La limite à la fatigue en faible amplitude d'un acier faiblement allié contenant des inclusions d'alumine et de sulfures de manganèse est inférieure à la limite élastique en déformation unidirectionnelle. La rupture se produit sans qu' on observe de mouvement irréversible des dislocations. Elle débute par une décohésion des interfaces entre les particules et la matrice, décohésion quise produit a la surface de l'échantillon; les fissures de fatigue commencent alors à se propager dans les directions prévues d'après la contrainte appliquée. On n'observe ni décohésion, ni fissuration sous la surface. Nous pensions que les décohésions proviennent d'une condensation préférentielle des lacunes sur les interfaces à la surface, la force motrice étant due à l'oxydation des surface de décohésion fraîches, alors que les lacunes se forment sur les crans de dislocations en oscillation.

Zusammenfassung—Die Ermüdungsgrenze bei kleinen Amplituden von niedrig legiertem Stahl mit Einschlüssen von Aluminiumoxid und Mangansulfid liegt unterhalb der Fließspannung bei einsinniger Verformung. Bruch tritt ohne einen Hinweis auf irreversible Versetzungsbewegung auf. Er beginnt durch Dekohäsion an der Grenzfläche zwischen Teilchen und Matrix an der Probenoberfläche, von wo sich Ermüdungsrisse in den von der angelegten Spannung vorgegebenen Richtungen ausbreiten. Unterhalb der Oberfläche wird Dekohäsion ober Bruchbildung nicht beobachtet. Es wird vorgeschlagen, daß die Dekohäsion von der bevorzugten Kondensation von Leerstellen an den Grenzflächen an der Oberfläche herrührt. Die treibende Kraft entsteht durch die Oxidation der frisch gebildeten Oberfläche an der Dekohäsion, während die Leerstellen an den Sprüngen oszillierender Versetzungen erzeugt werden.

The fatigue limit of low carbon steels containing small inclusions of alumina and manganese sulphides lies well below the static yield stress. Even though neither stress-strain curves nor SEM micrographs of specimen surfaces reveal evidence of any irreversible plastic deformation, failure occurs after some 10^7 cycles. By far the greater part of this life time is occupied with crack nucleation. According to previously published results [1, 2] the cracks which lead to failure are formed at the inclusions. Their behavior, once formed, can be readily understood in terms of stress concentrations at those inclusions [1, 3]. However, the process which leads to the nucleation of the cracks is still not entirely understood.

Phenomenologically it has been established [1, 2] that, as a first step prior to fatigue crack nucleation, the inclusions at the surface (alumina as well as manganese sulphide particles) 'decohere' from the surrounding matrix, as seen in Fig. 1. Also the particles frequently reveal cracks through them (Fig. 2). After decohesion has taken place, which in Fig. 1 has occurred over all of the upper interface, cracks as at the upper right in Fig. 1, begin to spread out from the microscopic gaps between matrix and particles where

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Theory of Dislocation Cell Sizes in Deformed Metals

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SUMMARY

Preceding theoretical investigations have demonstrated that dislocation cell structures in deformed polycrystalline materials result from the strain energy reduction of glide dislocations which have become mutually trapped. However, it had been hitherto not possible to give a theoretical reason for the cell sizes actually observed nor for the fact that the average dislocation link length in the cell walls is a constant fraction of the average cell diameter during stage II work hardening. A theory is presented to close this gap. For the example of the simplest known type of dislocation cell structure of very low energy, namely a three-dimensional checkerboard pattern of cubic cells with a common axis of misorientation but alternating sense of rotation, it is shown that the experimentally observed cell sizes and link lengths represent the minimum total strain energy for a fixed dislocation content. This occurs because with shrinking cell size the remnant longer-range cell stresses decrease while the short-range dislocation stresses, i.e. the dislocation line energy, increases whereas the reverse is true when at constant dislocation content the cell size increases.

1. INTRODUCTION

After moderate to large plastic strains, under virtually all conditions of deformation and in an extremely wide range of materials, the microstructure is found to include dislocation cells forming an interconnected network of more or less well-developed low angle dislocation boundaries which enclose volume elements comparatively free of dislocations. The dimensions of the cells are known to be inversely proportional to the applied stress, *i.e.* the cell size shrinks with increasing flow stress. Moreover, as was first pointed out by Holt [1] and Staker and Holt [2], the average cell dimension L relates to the acting resolved shear stress τ, the shear modulus G and the Burgers vector b as

$$L \approx KGb/\tau \tag{1}$$

whereby, in numerous measurements made on a variety of metals, K was estimated to be about 10. From the theory of work hardening this behavior had already been predicted [3], together with a proportionality between the average dislocation link length l in the cell walls and the cell dimension L, *i.e.*

$$g = L/l = \text{constant} \tag{2}$$

Altogether, progress in the understanding of dislocation cells and their role in work hardening has been gratifying. The dislocation cell structure represents the low energy configuration of glide dislocations, which have been generated in response to the applied stress and have become trapped in the course of straining the material [3 - 6]. Further, the fact that the cell size is inversely proportional to the flow stress is easily explained using the argument of "similitude" as follows. The dislocations are mobile under stress; thus the resultant resolved shear stress, consisting of the stresses due to all the

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TWO REGIMES OF CURRENT CONDUCTION IN METAL-GRAPHITE ELECTRICAL BRUSHES AND RESULTING INSTABILITIES

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ABSTRACT

Two distinct regimes of brush behavior in regard to brush resistance and coefficient of friction are observed in silver-graphite brushes. These are explained in terms of two different surface films, one a lubricating layer composed of a mixture of fine silver and graphite particles, the other an adsorption layer of only a few atomic diameters thickness, overlaying a sintered silver layer of negligible resistance. The lubricating film is destroyed through loss of graphite within some critical temperature range above 100°C. It is characterized by relatively high film resistivity (in the order of $3 \times 10^{-11} \Omega m^2$), and a coefficient of friction between about 0.15 and 0.2. The conduction through this film is non-ohmic. The adsorption film, with a film resistivity of about $10^{-12} \Omega m^2$, exhibits ohmic conduction, identifying electron tunneling as the major current transport mechanism. The coefficient of friction associated with it ranges between about 0.3 and 0.4, which is somewhat characteristic for dry sliding of fcc metals. In the high resistance regime, brush noise is high, but it is low in the absence of the lubricating layer. Finally, when the lubrication layer is present, the resistance of the cathodic brush is markedly lower than that of the anodic brush (in the order of 2/3).

INTRODUCTION

In the past several years, metal-graphite brushes, and in particular silver-graphite brushes, have received considerable attention (1-10) regarding their potential use in very high current density/high speed current collection systems for homopolar electrical machines. Use of these machines in high-technology applications, such as high power pulsing generators, energy transfer for possible future fusion reactors, and electromagnetic launchers, places quite severe demands on brush performance. For example, current densities above 18 MA/m^2 at speeds of up to 300 m/s will be needed in pulsed power sources for sub-second operation (11).

Previously, especially through the work of Holm (12) and Shobert (13), the behavior of graphite and metal-graphite electrical brushes has been well investigated at "normal" current densities, say up to 0.5 MA/m^2, but beyond that their properties were essentially unknown. In a preceding investigation (7) of silver-graphite brushes (75w% Ag) a peculiar instability was noted connecting two distinctly different regimes of brush behavior. Namely, when the current was gradually raised beyond some critical value, typically above 100 A, the contact resistance dropped, and over a time interval of several minutes reached another stable but drastically lower value. The current at which this resistance drop occurred decreased with increasing brush velocity, indicating that the critical parameter was the temperature.

By means of an analysis of those measurements, all confined to short-time tests, it was concluded (7) that the brushes could be operated in two different modes, depending on the presence of either of two different surface films. One of these films, associated with the higher contact resistance, was believed to be a fairly thick lubricating layer of silver and graphite debris. The other, associated with the low contact resistance, was tentatively identified as an adsorption film of only a few atomic layers' thickness, perhaps overlaid on a highly conductive silver-rich debris layer. This was presumably the same as the "thin film" described by Schreurs et. al (6,8).

At the time, on the basis of the short-term measurements described in Ref. 7, it appeared as if the coefficient of the friction either remained constant during transition or was mildly decreased when the thicker lubricating film was disrupted. The present investigation was undertaken in order to gain a better understanding of the two different surface films. To this end, firstly, the time per test was lengthened to 90 minutes, as compared to only a few minutes in the previous research in which the transient behavior was the focus of interest. Secondly, the coefficient of friction was carefully monitored throughout. Thirdly, the brush and rotor temperatures were measured concurrently with anodic and cathodic voltage drops. Further, oscillograms were taken of the "noise", i.e. short-term voltage fluctuations.

Throughout, silver-graphite (75w% Ag) brushes were used, not only for the sake of continuity, but also because these brushes have been the subject of careful work by others (1-5) and appear to be the best commercially available brushes for high current density applications.

On orientation changes, deformation banding, and dislocation structures in rolled [100](001) oriented iron crystals

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The texture formation process, dislocation structures, and deformation bands observed in rolled and cross rolled [100] (001) oriented iron crystals have been interpreted from the viewpoint of fundamental dislocation behavior and the properties of low-angle dislocation boundaries. Full agreement between theoretical expectations and published observations was obtained.

PACS numbers: 62.20.Fe, 81.40.Ef

I. INTRODUCTION

Typically, interactions between research on plasticity and texture formation, on the one hand, and on dislocation behavior on the other hand, are not commensurate with the potential benefits of such interactions. Specifically, for example, shear bands and deformation bands are mostly the concern of research in plasticity but they can be readily related to dislocation behavior and dislocation cell structures. One example are deformation bands, formed in uniaxial straining, which are normal to the active glide direction and the lattice within which is rotated about an axis normal to the active glide direction in the active glide plane. In terms of dislocations, such deformation bands result where glide polygonization walls of opposite sign encounter each other in their motion in opposite directions. Note that glide polygonization walls, or in general tilt walls, are the lowest energy form of dislocation surpluses belonging to one slip system.

Another example are microbands, also observed in unidirectional deformation or in rolling, which are parallel to the active glide planes and within which the lattice is misoriented about the same axis direction as above. The dislocation behavior to which these microbands correspond has been investigated in some recent papers.[1-6] Briefly, these bands arise when the tensile or compressive stresses between parallel pile-up like dislocation sequences of opposite sign are relieved by "unpredicted" secondary slip. In such a case the secondary dislocations react with the dislocations in the initial pile-up-like sequences to form tilt boundaries of the described orientation and relative rotation

The investigation of a third example is the subject of the present paper. It is that of rolled iron single crystals originally in [100] (001) orientation. It has been reported[7] that up to 80% reduction of thickness, cell walls with a spacing of about 0.5 μm, parallel to the rolling direction normal to the plane of the sheet are observed. These cell walls mediate a rotation about the normal of the sheet, i.e., are tilt walls with a [001] tilt axis. Beyond 80% rolling, "deformation bands" within which the lattice is in [110] (001) or in [1$\bar{1}$0] (001) orientation are formed. Also these bands are parallel to the rolling direction but according to Grzemba and Hu[7] they are filled with unstructured dislocations. With increasing rolling, such "deformation bands" grow at the expense of the initial cell structure such that at constant spacing the number of parallel (010) [001] tilt walls in the remaining zones between the deformation bands decreases while the angle across the average tilt wall increases. Thus after the formation of alternatively [110] (001) and [1$\bar{1}$0] (001) textured deformation bands, as sketched in Fig. 1, the 90° lattice rotations between them are accommodated by a steadily decreasing number of cell walls in the intervening cell structure.[7] Those regions were called "microbands" by Grzemba and Hu.[7] However, these bands are of a clearly different nature as the microbands discussed before. From recent literature[8] we learn that the name "Transistion Bands" is a more appropriate term for this kind of structural inhomogeneity, and this is used in Fig. 1 as throughout this paper.

II. GENERAL CONSIDERATIONS ON THE EVOLUTION OF THE DESCRIBED ROLLING TEXTURE

Disregarding sideways spreading of the rolled sheet, and assuming glide of type 1/2 ⟨111⟩ {110}, when the rolling direction is [100] and the plane of the sheet is (001), the four glide systems 1/2 [11$\bar{1}$] (101), 1/2 [1$\bar{1}\bar{1}$] (101), 1/2 [$\bar{1}\bar{1}\bar{1}$] ($\bar{1}$01), and 1/2 [$\bar{1}$1$\bar{1}$] ($\bar{1}$01) are initially strongly and equally favored. Spreading of the sheet is accommodated by glide along 1/2 [1$\bar{1}\bar{1}$] and 1/2 [$\bar{1}\bar{1}\bar{1}$] on (0$\bar{1}$1) planes, and along 1/2 [11$\bar{1}$] and 1/2 [$\bar{1}$1$\bar{1}$] on (011) planes. But slip along these systems is limited, the shear stress on them being only 50% of the shear stress on the primary systems. Any glide on the (110) and (1$\bar{1}$0) planes would leave the thickness of the sheet unchanged and thus is not expected to take place.

Rather than continuing to use Miller indices, for the present purposes it is more convenient to employ the labeling clarified in Fig. 1(c). In that nomenclature, in the initial orientation, then, glide will principally and equally take place on 0A(a), 0B(a), 0C(c), and 0D(c), with a minor contribution of 0B(b), 0C(b), 0A(d), and 0D(d) on account of sheet spreading.

As to the principally acting slip planes, glide on (a) and (c) should continue to occur at nearly equal rates throughout. Namely, if slip on one of these, say (a), would significantly exceed that on the other, then (a) would rotate to become more nearly parallel to the plane of the sheet. As a result, the resolved shear stress on (a) would decrease while it would increase on (c), causing glide on (c) to accelerate and rotate the two planes back to the symmetrical orientation. Thus

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SUBSURFACE HARDENING IN EROSION-DAMAGED COPPER AS INFERRED FROM THE DISLOCATION CELL STRUCTURE, AND ITS DEPENDENCE ON PARTICLE VELOCITY AND ANGLE OF IMPACT

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(Received October 9, 1982)

Summary

Previously published measurements of the cell diameters d of dislocation cells underneath copper surfaces exposed to particle erosion were evaluated in terms of the subsurface stresses τ to which they correspond. These were compared with the elastic stresses expected underneath spherical indenters impacting on the surface with different speeds. The inferred stresses differ markedly from theoretical predictions, not only in regard to the dependence on speed and angle of impact but even in regard to their decay along the z axis, the direction normal to the surface. Instead of τ decreasing as z^{-n} with n continuously rising from 0 at a shallow depth to 2 at large depths, as predicted from elastic theory, the stresses follow a z^{-1} dependence throughout the measured range. All data are satisfactorily explained by the relationship $1/d = (7/z)v^{2/5} \sin\alpha \approx \tau/100$ (using MKS units), where v stands for the velocity of the impacting particles, namely rather irregularly shaped alumina particles 50 μm in diameter, and α is the angle of impact. No theoretical explanation has so far been found to account for this result. It is noted, however, that previously a z^{-1} decay of stress had already given excellent results in a theoretical interpretation of the subsurface shear strain underneath a surface subject to sliding wear. Further, $\sin\alpha$ is the direction cosine for the normal force component, whereas a velocity dependence of $v^{2/5}$ is found from elastic theory for the depth scaling of the elastic stress due to a spherical indenter. Suggestions are made for further experiments.

1. Introduction

Recently Ives and Ruff [1] have investigated the dislocation structures below the surface of annealed OFHC copper after bombardment with

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Dislocation Cell Formation and Work Hardening in the Unidirectional Glide of F.C.C. Metals
I: Basic Theoretical Analysis of Cell Walls Parallel to the Primary Glide Plane in Early Stage II

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SUMMARY

Past research leads to the conclusion that dislocation cell formation in work-hardened crystalline materials occurs when dislocations assemble into low energy configurations. On the assumptions that the dislocation cells formed in f.c.c. metals in early stage II also approach the lowest energy for a given dislocation content in the material and that the initial dislocation arrangement before cell formation consists of linear dipolar mats, i.e. sets of similar edge dislocations in coplanar arrays but alternating sign from one mat to the next, the structure of the resulting cells is investigated. It is found that the initial pile-up-like arrays should transform into tilt walls with or without some twist component, such that the axis of relative misorientation is roughly parallel to the original edge dislocations. By a simple consideration of energies it is found that cell formation should begin at or below about $1.2\tau_0$ where τ_0 is the initial critical flow stress. Again if the minimum energy is considered, it is found that for low stacking fault energy materials Lomer–Cottrell locks should form prominently, such that the primary dislocations become rotated roughly normal to the Lomer–Cottrell locks. These results are in good agreement with available experimental evidence.

1. INTRODUCTION

1.1. *Basic properties of dislocation cell structures*

This paper is devoted to an investigation of the dislocation networks which are typically observed in early stage II of f.c.c. metals in single glide and which are (nearly) parallel to the primary glide plane. These are known to accommodate a tilt about an axis nearly parallel to the primary edge dislocations [1-4]. Their structure is obscured by dislocation irregularities when the stacking fault energy is high, but they are known to comprise a significant fraction of Lomer–Cottrell locks [3-5]. An example of a case in which there is low stacking fault energy is shown in Fig. 1. It consists of primary dislocations which on average are normal to a set of "unpredicted" secondary dislocations and Lomer–Cottrell locks. The Lomer–Cottrell locks were evidently formed through a reaction between primary edge dislocations, originally not far from an edge orientation, and the observed unpredicted dislocations.

It is presumed that the comparable networks found nearly parallel to the primary glide plane in early stage II of f.c.c. metals with a high stacking fault energy have much the same basic structure, albeit camouflaged through loops, kinks etc. The word unpre-

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Gold Fibre Brushes

THEIR PROMISE FOR FUTURE HIGH-TECHNOLOGY APPLICATIONS

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A new type of electrical brush has been developed to accommodate the otherwise unmet demands of future high technology. It consists of metal fibres finer than a human hair protruding from a solid matrix very much like the fibres of a camel hair brush. Laboratory tests as well as theory indicate that the new brushes can by far outperform the best currently available brushes especially if the fibres are made of gold or gold alloys. Even better performances are expected to be reached with a second generation brush of similar type embodying yet thinner fibres.

Gold is well established as a contact material. The freedom exhibited by the metal from insulating surface films under a very wide range of conditions and its outstanding corrosion resistance, together with its easy formability coupled with ease of alloying and joining, leave gold without a serious challenger. Thus it is almost taken for granted that wherever contacts must be repeatedly made and broken using light forces, and especially where high reliability is required, the switches will be made of or be coated with gold.

Furthermore, for sliding wire contacts used in the transmission of low currents under small forces such as in guidance systems, gold or gold electrodeposits are unsurpassed for reliable operation. Even so, gold is seldom thought of as a likely material for application in heavy current technology, for example in making electrical brushes conducting hundreds of amperes. The indications are, however, that this situation will change in the future. The reason is that a number of high-technology developments, now on the drawing board or in various stages of experimental design, require electrical brushes with performance characteristics greatly superior to those available so far. Examples are high performance homopolar motors and generators, based either on superconducting or ordinary magnets, energy storage and conversion devices, for example those which would be needed for pulsed fusion energy, and rail launchers or 'mass throwers' (1, 2).

Perhaps the most promising new type of brush proposed to solve this technological bottleneck is the metal fibre brush. This is quite literally a brush, but made of metal fibres rather than bristles, these being often much thinner than a human hair. Figure 1 shows an example. What ordinarily serves as a 'brush', transmitting current across an interface between parts of a circuit in relative motion, is almost always a 'monolithic' (that is one compact piece) metal-graphite brush the components of which are compacted with a binder material.

Much research will still have to be carried out to develop and perfect metal fibre brushes that will reliably satisfy the various requirements of future technology which cannot be met by the present monolithic brushes. Specifically, these requirements include the ability to operate at very high speeds and current densities with low power losses. Thus far metal fibre brushes, and the related metal wire brushes, have been made only in the laboratory or for pilot machinery (3-7). This article presents the argument for the viewpoint that metal fibre brushes are the best possible solid brushes, in principle surpassing in their performance not only monolithic brushes, but also foil (8) and metal wire (6, 7) brushes. Theoretically, brushes of liquid metal might possess superior properties. However, because of design and maintenance complications, such brushes are not likely to be competitive when their metal fibre counterparts with the anticipated properties become commercially available. Finally, it is very probable that a high fraction of future fibre brushes will use gold as the fibre material because of its superb surface characteristics.

Why Metal Fibre Brushes?

It has been stated above that, in principle, metal fibre brushes are likely to be the best possible heavy current conductors. In presenting the case for this point of view it is necessary to begin by examining the origin of the power losses when electrical brushes transmit currents across moving interfaces.

For high-technology applications, 'the best possible brush' means that brush which, with an acceptable lifetime, can conduct the highest current density at the fastest speed and/or with the lowest possible power loss. Since in sustained operation the heat developed in the brush must be transported away by means of some type of forced cooling, there is generally a strong incentive to keep it at a low level. In addition in some brush applications such as homopolar motors/generators and in energy storage devices, very stringent limits are placed on permissible losses lest the devices become uneconomical. However, even if neither of the above conditions are evident and the heat developed in the brushes is of no direct concern, it is indirectly most important when high current densities must be achieved since local melting at contact spots will typically cause brush failure. When at rest, the local temperature at the contact spot is roughly proportional to the applied voltage (9). There-

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Extended performance limits of metal-graphite brushes at very high current densities

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For high performance current collection systems, metal-graphite brushes are almost universally used. Among these, silver-graphite brushes with about 75 wt.% Ag appear to be the preferred choice in applications requiring the lowest possible resistance, especially if current densities are high. Research into the properties of silver-graphite brushes was strongly stimulated by the recent, and still ongoing, development of advanced homopolar electrical machinery which require current collection systems capable of very high current density and high sliding speed operation (see for example refs. 1–5). Partly as a result of this research such brushes are commonly used in protective atmospheres. In particular, CO_2 in the presence of water or hydrocarbon vapors has been found to extend the lifetime and contact performance of 75wt.%Ag-graphite brushes [1, 2, 4]. Further, almost routinely, if needed, the brushes are cooled to maintain the brush holder temperature below about 95 °C. Under the described conditions the brush noise is rather high. However, electrical loads of up to 6 MA m^{-2} (600 A cm^{-2}) could be supported [5], *i.e.* much above those of conventional applications which are typically less than 0.1 MA m^{-2} (10 A cm^{-2}) (see ref. 1).

Recent research indicates that the above are not necessarily the best operating conditions of 75wt.%Ag-graphite brushes if ultrahigh current densities at low losses are desired. Rather, it appears that brush performance at a bulk temperature above 100 °C can be very advantageous. In fact, it was shown that 75wt.%Ag-graphite brushes can be operated in air, with very good performance characteristics, in two distinctly different modes. One of these is the mode at and below about 95 °C which has already been described and will be dubbed "mode I". Mode II occurs in air at a bulk brush temperature of above about 100 °C. The two modes differ on account of two different surface films which the brushes deposit onto the opposite surface [6, 7], *e.g.* a slip ring or commutator, normally of copper.

At low temperatures, in air as well as in protective atmospheres, the surface film is composed of a mixture of fine silver and graphite particles [5–8] with no continuous metallic path through it [7]. Because of the interposed graphite particles, that film has lubricating properties and its friction coefficient μ typically ranges between 0.1 and 0.2. It is overlaid with a very thin adsorption film of only one or a very few molecular layers' thickness, believed to be mainly of water when formed in a moist CO_2 atmosphere [9]. Electrons can tunnel through this adsorption film [5–7, 9] and the film resistivity of it is close to 10^{-12} Ω m^2. However, the underlying lubricating film has a significantly higher film resistivity, of the order of 3×10^{-11} Ω m^2, a fact that has been overlooked in some of the investigations (compare ref. 5). The mode I film is further characterized by much electrical noise, due to structural irregularities, and a polarity effect such that the anodic and cathodic brushes differ markedly in resistance and wear rate [5, 7].

The lubricating film is stable as long as water can be reabsorbed fast enough by the film to replenish the moisture lost through momentary overheating at the contact spots [7]. Once the temperature rises beyond some critical value this balance cannot be maintained, graphite is lost through dusting from the film, and the silver particles sinter

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Performance characteristics of silver–graphite electrical brushes (75 wt.% Ag) without cooling up to 1400 A cm^{-2} in air

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In a preceding communication [1], measurements on silver–graphite electrical brushes (75 wt.% Ag) were reported which highlighted the existence of two different modes of current conduction, dubbed mode I and mode II, the former being the ordinary and well-known one in which graphite forms, or at the least strongly participates in, a lubricating layer, causing a low coefficient of friction. Mode II appears to have escaped previous attention. It is characterized by a surface layer on the opposing contact (the "rotor") which is principally of silver and therefore has a rather higher coefficient of friction. However, that layer is more uniform and therefore causes much less electrical noise, besides having a very low electrical resistance. As a consequence, at very high current densities the total loss, in watts per ampère conducted, is significantly lower in mode II than in mode I. The condition which causes the mode II surface layer is a higher temperature which has the effect of permitting the graphite dust to escape from the surface layer on the rotor, leaving the silver behind. It appears that brushes have been routinely cooled, in the past, to temperatures too low to initiate mode II.

The reason for cooling metal-graphite brushes has been to prevent the loss of moisture for fear that otherwise "dusting" occurs, through which a brush can be worn down within minutes (see for example the discussions of this phenomenon by Holm [2] and Shobert [3]). Somewhat surprisingly, however, dusting never occurred when brushes were used in mode II even though the rotor could be as hot as 200 °C and the brushes themselves considerably hotter. Dusting did occur, however, if the brushes were allowed to arc, generally because of too low a brush force. It is considered that, in mode II, the brush surface is protected not only by being sealed fairly effectively against the opposite side, but also because the silver enriches at the brush surface, the more so the higher the current density [4, 5]. Figure 1 shows this effect by the example of a scanning electron micrograph of a brush with one-half a virgin surface, shaped to the rotor with 600 grit emery paper, and one-half run for a few hours at a current density of 6 MA m^{-2}, at 13 m s^{-1} using an 8.3 N brush force on 0.5 cm^2 area.

Limitations of the apparatus had prevented exploring the limits of current-carrying capabilities of the silver–graphite brushes in mode II. That limit lies at about 6 MA m^{-2} for mode I according to Johnson and Schreurs [4], and this was also the limit to which in ref. 1 the 0.5 cm^2 brushes were tested since, without cooling, the whole apparatus heated up uncomfortably. In fact, it has since become clear that the differential transducer malfunctioned on account of being overheated in tests 7–12 of ref. 1, indicating too fast brush wear. Even so, it seemed to be of the essence to study brush behavior at still higher current densities and, correspondingly, smaller brushes were prepared and the current raised to the highest feasible value. The results are summarized in Table 1 and Fig. 2.

Whereas according to conventional wisdom the brush size should have no effect on the data, provided that the same brush forces are used as was done, in fact it has. This is so because the brush body resistance is included in the measurement R_B and is increased when the cross section of the brush is decreased in accordance with Ohm's law. Thus the brushes

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