METALLURGICAL PHASE TRANSFORMATIONS

IN THE RUBBING OF STEELS

by

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Metallurgical phase transformations in the rubbing of steels

ABSTRACT

The formation of phase transformed material in the unlubricated rubbing of plain carbon steels using a crossed cylinders machine, in single and multiple rubs, has been investigated.

Previous studies of the dry wear of steels had suggested that if the frictional flash temperatures reached values of approximately 750°C, then a phase transformed material, akin to the martensitic structure of conventional ferrous metallurgy, appeared upon the rubbing surfaces.

Exploratory experiments indicated that phase transformed material appears as a result of a single rub of the surface only under conditions so severe that the calculated flash temperatures are of the order of 1100°C or more.

The fundamental metallurgical considerations applicable to small volumes of material subject to a combination of high temperatures and pressures of short duration which are the special characteristics of flash temperatures, have therefore been re-examined.

The formation of homogeneous austenite, which is the first stage in the production of transformed material, has been considered to occur in two temperature ranges 723 - 910°C and >910°C. Only above 910°C can the transformation be completed in a typical flash temperature duration.

Diffusion of carbon in austenite has been proposed as the rate determining process. Theoretical temperatures of approximately 1250°C are calculated for the formation of homogeneous austenite from lamellar pearlite.

It has been found that the pioneering results of Welsh (1965), in particular his T_2 and T_3 transitions can be given satisfactory interpretations in terms of the new metallurgical theory. An important element in these interpretations is the number of rubs to which any element of the surface is subjected. This factor has been deduced from wear theory.

Further detailed experiments have confirmed the magnitude of flash temperatures required for the production of phase transformed material. The influence of the maximum temperature has been discussed.

Metallurgical and physical analysis indicates phase transformed material to be a fine grained martensitic structure.

PREFACE

This project has been carried out in the Research Laboratories of the Engineering Department, University of Leicester under the supervision of Dr. J. F Archard to whom I give my sincere thanks for encouragement, advice and helpful suggestions.

The work embodied in this thesis is the product of the author's own research, except where otherwise stated.

I am indebted to the technical and workshop staff of the Department, in particular to Mr D. J. Crooks. Thanks are also due to the typing Services, UKAEA and the National Centre of Tribology, Risley for their help in the preparation of the thesis. I would like to thank Mrs K. P. Baglin for many helpful discussions.

I am grateful to the British Railways Board for one terms additional study.

CONTENTS

	<u>Page</u>
CHAPTER 1. Introduction	1
CHAPTER 2. Background review	
2.1 Introduction	4
2.2 The occurrence of white layers	5
2.2.1 Surface layers found in engineering practice	6
2.2.2 Surface layers in simulated or fundamental laboratory experiments	17
2.3 The properties of white layers	24
2.4 Theories of formation	28
2.5 Flash Temperatures	33
2.5.1 Flash temperature theory	35
2.5.2 Flash temperature measurement	47

	<u>Page</u>
2.6 Sub-surface temperatures	54
2.6.1 Stationary and slow moving heat sources	54
2.6.2 Fast moving heat sources	56
2.7 Temperature distribution within the contact	62
2.8 Contact theory for point contacts	67
2.9 Discussion	71
CHAPTER 3. The crossed cylinders machine	
3.1 Introduction	76
3.2 The crossed cylinders machine	76
3.3 Experimental procedure	78
CHAPTER 4. Exploratory experiments with soft steels on soft steels	
4.1 Introduction	80
4.2 Exploratory experiments	80
4.3 Discussion	85
4.4 Conclusions	89
CHAPTER 5. Metallurgical considerations	
5.1 Introduction	90
5.2 The range 723-910°C	91
5.3 The range >910°C	95
5.4 Carbon diffusion theory for lamellar pearlite	98
5.5 Discussion of simple diffusion theory	101
5.6 The diffusion coefficient: its dependence upon concentration and its possible consequences	102

5.7	Simple solution of diffusion equation with limited solubility	103
5.8	Full solution of diffusion equation with limited solubility	106
5.9	The effect of a temperature/time history	110
5.10	Effects of pressure etc., on the metallurgy of steels	114
5.11	Discussion	123
CHAPTER 6.	Effect of metallurgical theory on interpretation of wear patterns	
6.1	Introduction	126
6.2	Review of the wear of steels	126
6.3	Wear patterns of Welsh (1965)	133
6.4	The T_2 transition	137
6.5	The range between T_2 and T_3	140
6.6	The T_3 transition	143
6.7	The influence of carbon content	144
6.8	The influence of externally induced bulk temperatures	145
6.9	Discussion	147
CHAPTER 7.	Experimental	
7.1	Introduction	149
7.2	Materials and specimen preparation	149
7.3	Specimen analysis	151
7.3.1	Metallographic preparation for microscopy	152

7.3.2	Microhardness measurement	155
7.3.3	Scanning electron microscopy	155
7.3.4	Auger electron microscopy	155
7.3.5	Transmission electron microscopy	157
7.3.6	X-ray diffraction analysis	158
CHAPTER 8.	Results	
8.1	Introduction	159
8.2	Crossed cylinders machine experiments	159
8.2.1	Preliminary results, runs 1-19	160
8.2.2	Main results, runs 20-42	163
8.2.3	Friction traces	164
8.2.4	Track appearance	165
8.2.5	Specimen bulk temperature	167
8.2.6	Contact size measurement	168
8.2.7	Flash temperature calculations	169
8.2.8	Appearance of phase transformed material	171
8.3	Identification and nature of phase transformed material	175
8.3.1	Etch response	175
8.3.2	Microhardness measurements	176
8.3.3	Tempering tests	176
8.3.4	Scanning electron microscopy	178
8.3.5	Auger electron spectroscopy	179
8.3.6	Transmission electron microscopy	182
8.3.7	X-ray analysis	184

8.4	Conclusions	185
CHAPTER 9.	Discussion and conclusions	
9.1	Introduction	186
9.2	Discussion of main experiments	191
9.2.1	Preliminary runs 1-19	192
9.2.2	Main runs 20-42	194
9.2.3	Friction records	194
9.2.4	Track appearance	195
9.2.5	Bulk temperature	198
9.2.6	Contact size	199
9.2.7	Flash temperature calculations	205
9.2.8	Appearance of phase transformed material	217
9.3	Identification and nature of phase transformed material	220
9.4	Conclusions	225
9.5	Suggestions for further work	228
APPENDIX 1	Solution to equation 5.9	233
APPENDIX 2	Solution to equation 5.18	236
APPENDIX 3	Flash temperature results	240
APPENDIX 4	Results of etching tests	242
APPENDIX 5	Indexing of Figure 8.42b	245
APPENDIX 6	Maximum flash temperature results	248
REFERENCES		250
<u>NOMENCLATURE</u>		

The meaning of terms used in the text, unless otherwise stated is as follows.

A	Area of contact
B	Constant
C	Carbon concentration
D	Diffusion coefficient
D ₀	Constant
E	Youngs modulus
E'	Reduced elastic modulus
F	Frictional force
H	Hardness
HV	Vickers indentation hardness (kg/mm ²)
K	Coefficient of wear
$L = \frac{Va}{2K}$	Dimensionless parameter (Peclet Number) = τ_1/τ_2
L	Sliding distance
$M = \frac{z^2}{2Kt}$	Dimensionless parameter = τ_3/τ_4
N	Constant
R	Radius of curvature
R'	Relative radius of curvature
S	Gas constant
Q	Rate of heat supply from area A
Q'	Activation energy
T ₁ , T ₂ , T ₃	Wear transitions
V	Velocity
U	Volume of wear
W	Load
a	Radius of circular contact of area A
c	Specific heat
g	Acceleration due to gravity
k	Thermal conductivity

p_m	Flow or yield pressure
$q = Q/A$	Rate of heat supply per unit area
s	Shear strength of junction
t	Time
$x,y,z,$	Distance in x,y,z, direction, where specified
α	Division of heat
β	Correction factor
θ	Temperature
$K = k/pc$	Thermal diffusivity
λ	Wavelength
μ	Coefficient of friction.
ν	Poisson's ratio.
ρ	Density.
σ	Cooling coefficient.
τ_1	Time taken for generated heat to diffuse to a depth of a.
τ_2	Time taken for contact to move a distance a.
τ_3	Time taken for generated heat to diffuse to depth z.
τ_4	Time for which heat has been applied below surface.
ϕ	Angle.

SUBSCRIPTS.

a	As measured.
bk	Bulk.
e	Elastic.
f	Flash.
g	Graphical.

m	Mean
max	Maximum
p	Plastic
r	Multiple
s	Single
sf	Surface
sl	Sliding
B,C,1,2	Properties pertaining to bodies B,C,1 or 2
T	Total

CHAPTER 1

INTRODUCTION

Material and energy conservation are becoming increasingly important in the present and foreseeable future economic situation. In addition to fatigue and corrosion, wear is one of the principal causes of material wastage and component degradation. Any reduction in wear can effect considerable savings, for example by prolonging and increasing machine efficiencies, often leading to reduced power losses and reduced maintenance costs. An indication of the possible savings to be derived from such an approach for industry, in the United Kingdom, was estimated by the Jost Committee report to be around 515 million pounds per annum in 1966. Inflation has dramatically increased this estimate. Alternatively it is assessed that in industrialised societies approximately one half of the gross national product is used to replace the results of wear and similar effects. A recent report on Strategy for Energy Conservation through Tribology (1977) estimates that approximately 11% of the total USA energy consumption can potentially be saved through progress in tribology. However effective progress in wear prevention can be made, only, after a better understanding of the mechanisms by which it occurs and of the controlling factors involved.

One fascinating wear phenomenon, repeatedly reported in the literature, is the formation of "phase transformed material" or "white layers" on rubbing surfaces, usually ferrous. This material can be associated with a marked reduction in wear rate but yet is commonly observed as a concomitant feature on "failed" engineering surfaces. As a phenomenon

the formation of this hardened material is not yet fully understood; on the other hand it has been suggested that its properties are such, that it may have potential use as a commercial finishing or friction-hardening process.

The essence of this thesis has been to study the formation of phase transformed material as a function of the transient surface or flash temperatures generated in a single contact or rub, under well defined experimental conditions. Thence to consider the formation of this material in a number of multiple contacts or rubs which is the condition under which transformed material is usually observed.

Initial exploratory experiments with plain carbon steels suggested that flash temperatures well in excess of those deduced from earlier studies or those expected from the application of conventional ferrous metallurgy are required to produce phase transformed material. From these observations the work developed into formulating a revised metallurgical theory for the production of phase transformations in steels for typical conditions occurring in a tribological contact i.e. short contact times and high contact pressures.

Further crossed cylinders experiments have been performed in which portions of steel surfaces have been exposed to a single rub under well defined conditions; these experiments have been conducted in a manner which is amenable to analysis in terms of the theory of flash temperatures. Analysis of the results of these experiments confirms that the temperatures required for the production of phase transformed material in a single rub are significantly higher than the values assumed

earlier.

The analysis of the experiments also largely confirms the arguments of the new metallurgical theory.

CHAPTER 2

BACKGROUND REVIEW

2.1 Introduction

This chapter primarily concerns a reivev of the occurrence, properties and theories of formation of hardened surface layers often formed when two ferrous surfaces are rubbed together under dry or poorly lubricated conditions. Such surface layers are greatly hardened, generally no structure is revealed by conventional etchants and when examined by optical microscopy appear white or pale yellow. Various terms have been used to describe such layers e.g. "white layers", "white phase", "non-etching layers" etc. In this necessarily selective review of the plethora of literature of such layers, the term used by the respective authors will be followed. The selection has been made both to indicate the range and diversity of the background material but also with special cognisance of the nature of the author's own work which follows. The theoretical and experimental evidence of this thesis concerns the production of hardened surface layers on plain carbon steels. At this stage it is appropriate to note that these layers have been considered to be formed as a consequence of frictional heating, thus having a structure akin to the martensitic structure of conventional ferrous metallurgy and will be referred to as "phase transformed material".

To study the appearance of phase transformed material in terms of frictional heating it has been therefore also appropriate to review the theory and measurement of transient surface or "flash" temperatures

applicable to such experiments. Moreover it is noted that whenever "white layers" etc. have been considered to be formed as a consequence of frictional heating, to date the interpretation has been based upon the average value of flash temperature at the contacting surface, whereas, such layers consist of material of finite depth. To remedy this defect an analysis of sub-surface temperatures, formulated by Archard will be included.

Crucial to such an approach, in terms of the calculation of flash temperatures is the estimate of contact size between the rubbing specimens. Therefore a brief review of contact theory for point contacts applicable to the experimental conditions of this thesis is included.

2.2 The occurrence of white layers

The occurrence of hard surface material, such as white layers can be broadly split into two areas:-

(i) those found in engineering practice, indeed often concomitant with component degradation or failure and as such indicating the importance and diversity of such layers.

(ii) those found in simulated or fundamental laboratory experiments. This classification is not rigid but is useful in the subsequent discussion of hard surface layers.

2.2.1 Surface layers found in engineering practice

Stead (1912) was one of the first to mention the presence of hard layers found on the surfaces of used steel wire ropes. He stated that this was a consequence of momentary frictional heating of the surface layers to redness, followed by quenching, producing martensite. He did not explain, however, the featureless appearance of the hardened layers when etched by mixed nital and picral.

Investigating the causes of failure in cold drawn steel wires Atkins (1927) observed that in some cases wires were covered with a layer, approximately $25\mu\text{m}$ in depth, resistant to nital etch. His conclusions were similar to Stead's, the frictional heating occurring between the die and wire. Based on this evidence it is difficult to justify Atkins conclusions, particularly as nitrides or carbides would have etched in a similar fashion.

Trent (1941) investigated failed wire ropes used in colliery winding gear. Metallographic analysis revealed a large number of them had formed a white etching layer on their surfaces of typical depth $75\mu\text{m}$. This layer usually contained a number of cracks, sometimes extending downwards into the substrate. He deduced that the layer was brittle and cracks formed could be propagated through the wire causing it to fail. He considered these layers to be martensitic, attributing their featureless appearance to an extremely fine grain size produced by the rapidity of heating and cooling. Hitting the wire rope with a sharp blow, cutting or crushing the wire produced similar features. An estimate of the surface temperature based on an assumed temperature gradient, gave values in excess of 1150°C .

In the subsequent discussion it was pointed out that his evidence was inconclusive in terms of the identification of the material and alternative explanations were possible.

More recently in a series of papers, Steininger and Krzeminski (1962), (1965) have studied defects in steel wires and pre-stressed concrete wire, in particular the formation of white non-etching layers, either during service or in drawing. Microhardness and X-ray diffraction analysis indicated the presence of tetragonal martensite in the layers. The low plasticity and brittleness of the structure causing cracks, leading to failure. They referred to such layers as consisting of "amorphous martensite", though it is probably more correct to state that under their metallurgical analysis no grain structure was revealed.

The formation of non-etching layers on rubbing leaf springs and on a broken turbine shaft was examined by Schottky and Hiltenkamp (1936). Chemical analysis showed a small increase in the nitrogen content, whilst scratch hardness tests showed the layers to be harder than the bulk material. They concluded that the hardened surface layers were formed by two concurrent mechanisms, part absorption of nitrogen, part cold working. Also cracks formed in the brittle hardened layer were considered to extend on the leaf springs and cause flaking to occur, whilst cracks on the turbine shaft had caused catastrophic failure.

Fay (1917) studied the formation of a hardened inner surface on gun barrels. Based on its appearance when tempered he concluded the

layer to be martensitic, attributing the formation primarily to a cold working effect in the bore, although austenitisation was also required. Howe (1918) believed that the brief heating period, estimated to be approximately 0.02 seconds per firing cycle (for a 14 inch gun) was sufficient to austenitise the original bainitic matrix at the surface and on cooling by the colder outer material to form martensite. Howe considered the thickness of the hardened layer to increase asymptotically from round to round as a consequence of the repeating temperature cycle, as well as being beneficial in reducing erosion. Howe, like Fay, rejected a theory based on carburisation because of the short times involved, although both authors provided no further evidence to justify the identification of the material.

Wheeler (1922) suggested that due to the high temperatures ($\sim 500^{\circ}\text{C}$) and pressures ($\sim 310\text{ MPa}$) occurring in a gun tube, nascent nitrogen combines with the austenitic case of the bore. Supportive evidence was not provided. De Sveshnikoff (1925) believed the layer to be martensitic in nature and favoured the theory of appreciable carbon absorption from the exploding gases. Some microsections showed a "lath-like" structure in the layer, but others exhibited no observable structure. Chemical analysis of filings removed showed increases in the carbon content, typically from 0.27% to 0.42%. The same filings also showed small increases in nitrogen content, typically from 0.008% to 0.033%. Evidence supporting the nitriding theory was forwarded by Lester (1929) who studied white layers from a 12 inch gun. He found, in addition to a lack of tempering response at 400°C , that X-ray diffraction analysis indicated

similarities between the white layers to those found in nitrided steels.

Snair and Wood (1939) attempted to correlate the various results and theories. They recognised the complex conditions that occur during the explosion and subsequent passage of a shell down the gun tube i.e. very high flame temperatures and pressures together with combustion products. Based on optical microscopy and heat treatment they favoured a nitriding theory occurring at a bore surface temperature of 430°C. The discussion of this paper noted that on the evidence provided, alternative conclusions were possible.

By observing pro-eutectoid ferrite in layers formed on the driving bands of shells, Wrazej (1945) concluded these layers to be martensitic. He proposed that during the firing of the shell the A_{c1} temperature had been exceeded at the surface, but not the A_{c3} temperature. Wrazej demonstrated the ease with which pearlite transformed into austenite by examining a 0.8% C steel cut with a hacksaw and a 0.18% C steel ground and polished. Both revealed martensitic structures. Wrazej did not consider whether gases in the gun barrel can be absorbed or play a part in the surface hardening.

Reviewing the main kinds of grinding damage, Tarasov (1946) identified a continuous white, etch resistant, layer on steels as martensitic, formed as a consequence of the temperatures generated during a grinding burn. A layer 1.3 mm in depth was also reported to be formed on the end of a bar of SAE 1040 due to unsatisfactory

grinding conditions during a cutting-off operation. More complicated grinding burn patterns were identified as occurring when a previously formed martensitic layer was tempered or partly rehardened during a succeeding contact of lesser severity. Further work by Tarasov and Lundberg (1949) reaffirmed these results with hardened tool steels and by taper sectioning identified a transition zone between the white layer and substrate. They were unable, though, to produce white layers by grinding burn on soft steels. Rowland (1949) in the discussion following the paper, reported that a work hardened layer could often be found on hardened AISI 52100 after grinding which etched white and resisted tempering. Latterly Torrance (1978), in studying metallurgical effects associated with grinding, has identified both surface and sub-surface white zones with unburnt plunge ground and cylindrically ground hardened specimens. Under such conditions Torrance considered these zones to be formed by intense plastic shear refining the structure such that it cannot be resolved optically. Vickers hardness values of the white zones were marginally higher than the bulk material (821 HV 100g as compared to 806 HV 100g). The white zones were invariably associated with microcracks. Non-etching layers, surface and sub-surface cracking were observed by El-Helieby & Rowe (1981) in studies on the effects of various grinding parameters e.g. downfeed, on hardened and tempered En 31. The microstructural alterations were considered to be a direct result of heat generated during the grinding process. The observed effects were explained in terms of a heat flow model and the tempering characteristic of En 31.

Etch resistant regions have been commonly found at or near the surfaces of rolling element bearings. There are strong similarities between sub-surface white etching regions and those found at worn surfaces. Styri (1951) found, prior to fatigue failure, sub-surface structural alterations in ball bearing races of SAE 52100 around non-metallic inclusions, which because of their appearance were referred to as "butterflies". He found that inclusions acted as stress concentrations and local shear stresses around the inclusion caused a white etching alteration to develop. Although not attempting to identify the alteration he made the important association of the alteration with the inclusion. Dark etching alterations had been previously reported by Jones (1946) beneath inner ball bearing races operated for extended periods at high loads. Jones considered these features to be stress induced. Gentile et al (1965) identified both dark and light etching constituents in SAE 52100 ball bearing components resulting from overload testing. It was suggested that stress concentrations at the non-metallic inclusions caused the dark etching alterations and thus in dirty steels the structurally changed areas would be more extensive than in clean steels. It was thought that the light etching component resulted from local plastic deformation of the matrix and consisted of cold worked ferritic bands interspersed with precipitated carbides. O'Brien and King (1966), however, found dark etching constituents in steels which were low in non-metallic inclusions.

Scott et al (1967) reviewed the metallurgical aspects of rolling contact fatigue and in particular studied microstructural

alterations in rolling element bearings manufactured from En 31. White etching layers formed at the surface were thought to be a result of high contact stresses, local high temperature flashes causing solution of the carbide followed by rapid quenching under pressure by bulk material and lubricant. The existence of a threshold stress level, above which sub-surface dark etching alterations were formed suggested that the alterations were probably attributable to yielding or plastic flow rather than a tempering effect. Since white etching alterations appeared to be formed by a variety of mechanisms, for example by intensive heating brought about by local dislocations around microcracks or by highly localised stress concentrations, and coupled with their conflicting properties, no firm theory was advanced. In a more recent summary and based on his own experiments, Schlicht (1973) concludes that the appearance of butterflies or white bands is a result of simultaneous high contact pressures and shear stresses developed during rolling. Locally the shear strength of the material is exceeded causing a crack. The flanks of the crack are heated to temperatures in the melting range due to the release of energy associated with the plastic deformation at the propagating crack tip. These calculations were, however, based on crack propagation rates of the order of the speed of sound and a crack tip radius of 10^{-10} m. Self quenching producing a supersaturated untempered martensite of high hardness and distinct boundary with the matrix material.

Kuritsyna (1956) studied white phases formed at the surface of grey cast iron rings after running-in. From X-ray diffraction analysis it was found that the white phase contained a fine distribution of

carbides of the form Fe_3C & Fe_xC dispersed in a ferritic matrix. Annealing at 800°C for 3 hours had no effect on the white phase, whilst annealing at 900°C removed the presence of carbides completely. It was believed that the white phase was formed in the presence of active carbon from decomposed lubricants. It was also proposed that white phase arose with ease due to plastic deformation of martensite either present as the initial structure or as a result of secondary hardening.

Clayton and Jenkins (1947) had earlier reported a white constituent, repeatedly observed on cast iron piston rings prior to or after scuffing, during the running-in of an aero-engine. It was found that scuffing occurred with mineral oils but not with castor oils. They did not fully identify the white constituent but their results indicated it to have a basis of two phases; cementite and a quenched high carbon ferritic phase developed from austenite.

Two type of surface phases were found by Rogers (1969) to be associated with the scuffing of cast iron rings and liners. Examination of the rings revealed bands of holes extending from top to bottom of the bearing surfaces. The holes were attributable to spalling of one of the phases, termed W_1 . Outside these regions a second surface phase, termed W_2 , was observed. The W_2 phase was not only much thinner (typically $1\mu\text{m}$ c.f., $10\mu\text{m}$ for W_1) but also had fewer cracks. Their hardness values were significantly different; 1240 and 655 HV 16g for the W_1 and W_2 phases respectively. W_1 phase began to temper at 720°C , the W_2 phase at 460°C . It was concluded that the W_1 phase was probably a

Fe-C-N compound of high carbon content, its high hardness being a consequence of its atomic structure rather than strain effects. Evidence to support this conclusion was not provided. W_2 was identified as a high carbon phase strained by shear or precipitation. Additional work by Rogers (1970) indicated that when using a pin on disc machine, scuffing of cast iron only occurred when a smear of oil was present. Since the properties (i.e. hardness, etch and temper response) of the W_2 phase found on scuffed diesel engine components were similar to the phase produced during scuffing on a pin on disc machine, they were considered identical. Under dry, fully flooded conditions or just prior to scuffing, other types of wear occurred. No mention was made of the W_1 phase in respect of the other types of wear. Rogers concluded that scuffing is a wear process in itself and involves the continuous production and spalling of the W_2 phase. X-ray analysis indicated that the debris produced during scuffing was Fe_3C formed due to the carbon in the oil reacting with the ferrite of the cast iron. This theory is subject to a number of objections, for instance the W_1 and W_2 layers found by Rogers appears to be similar to those found by Wilson & Eyre (1969) under dry sliding conditions. Also the wear debris produced during scuffing was found to be Fe_3C , yet the W_2 phase itself may have a different structure, which for example is carburised when produced as debris. Finally it was not made clear whether the W_2 phase is a consequence of scuffing or vice versa.

However there is evidence from Shell Thornton Research Centre (private communication) that two separate surface phases are formed

during scoring and scuffing in lubricated sliding contacts. The properties of these phases are similar to those identified by Rogers. Moreover it appears that the phase produced during scoring has been found to precede the onset of scuffing.

Rail corrugations are short wavelength irregularities which form on the head of railway rails in service. The corrugations give rise to the phenomenon of "roaring rails", a characteristic noise created when rails are traversed by traffic. Harisson (1979) has found that the crests of corrugations are covered with a white etching layer of typical depth 130 μm and hardness 900-1300 HV 100g. These surface layers were found to be longitudinally deformed in the direction of traction applied by a locomotive. Corrugation troughs showed no evidence of white phase formation. These results were in accord with earlier work at British Rail. Tempering at 250°C for one hour precipitated the carbides in the white layer. Harisson concluded the white layer to be martensitic caused by frictional heating of the rail surface. Clayton (1976), in investigating vertical rail wear, has found uncorrugated rails exhibit areas of white phase on the running surface. There was evidence of longitudinal cracks below the surface, causing spalling of small areas of white phase.

White etching zones were reported by Zener and Hollman (1944) during the blanking of steel plate. It was thought that the light etching zone was martensite, formed from austenite. The high temperatures were attained by the shear strain rate, characteristic of high speed deformation. Marion and Stock (1970) found that white etching bands, obtained by punching plugs from steel plate, did not temper

in a manner typical of martensite. They remained stable and resisted etching, after tempering at temperatures up to 300°C. Wingrove (1971) has shown that white etching bands, formed during the shearing of steel, neither temper in a manner characteristic of, nor have a structure generally associated with martensite. Nevertheless their lattice spacings (determined by X-ray analysis) correspond to those of martensite. The bands were also found to contain dense tangles of dislocations. The similarity of the temper resistance of the martensitic shear bands to that observed in ausformed steels and physically deformed martensites was pointed out. Similar work by Glenn & Leslie (1971) concerned white streaks generated in steel armour plate by the impact of an armour piercing projectile. Transmission electron microscopy of those white streaks (typical hardness 1025 HV 100g) revealed an extremely fine grain structure ($\sim 0.1 \mu\text{m}$) with a diffraction pattern indicating a BCC structure. It was thought to develop as a result of quenching from deformed austenite, produced at the moment of impact, to form a finely divided, heavily distorted martensite.

Other accounts of interest include the observations of white etching adiabatic shear zones in explosively loaded thick wall cylinders (Thornton & Heiser (1971)) which by X-ray analysis were identified as martensite and the existence of white streaks in a nickel base superalloy subjected to rapid straining (Wallace & Terada (1973)). More recently Turley (1975) has studied white etching layers on reamed holes in ultra high strength steel, produced under both rubbing and cutting conditions. Transmission electron microscopy revealed that both layers were produced by intense plastic

deformation. The layers had a fine subgrain structure which was BCC and there was a general loss of carbides within the layers. These layers were shown to be different from (i) white etching layers produced by abrasion (ii) martensite and (iii) white etching layers produced by abrasive grinding. Although the calculated flash temperatures were insufficient to austenitise the steel, retained austenite was detected in the layers of the reamed holes, which was not detected in the bulk material.

2.2.2 Surface layers in simulated or fundamental laboratory experiments

Mailander and Dies (1943) were probably the first to study white layers in the laboratory in a systematic manner. Whilst rubbing soft steel pins against chromium rings under unlubricated conditions, over a range of loads, they established that at a certain load the wear rate fell dramatically and a white etching hardened "skin" was formed on the pin. They considered the skin to be beneficial and formed by absorption of nitrogen. However they also found that a hard skin was formed in a hydrogen atmosphere (with a corresponding fall in wear rate), while in a reduced pressure atmosphere no skin was formed and a low wear rate was experienced throughout. It is therefore dubious whether, in this case, absorption of nitrogen is the operative mechanism.

Sakmann (1947) studied changes in the lubricated running-in of metal surfaces, using a flat on disc geometry. In many of the run-in chromium steel flats sliding against cast iron, he found ridge-like

features protruding from the surface, by several microns. These ridges were found to be harder than the substrate and tempering caused the ridges to soften, until at 760°C they had reverted to their original structure. Based on these properties it was suggested that the ridges were a carbide rich phase formed by the reaction of nascent carbon from the oil at high local temperatures. As the original structure could be reproduced by tempering and as no other attempt was made to measure the carbon content e.g. by the intercept method, alternative conclusions may be drawn from Sakmann's work.

In the first of two investigations, Welsh (1957) studied the wear of plain carbon steels, using a pin on ring machine. At a load of 10N and sliding speed of 2 m/s, under dry conditions, he found that at low carbon contents (0.12%C) the wear rate was always high. However at higher carbon contents (0.34%C), after an initial period of high wear, the wear rate fell. The time required for this transition to occur decreased with increasing carbon content. Metallographic sections of the pins exhibiting the low wear rate were found to have a continuous, hard structureless layer over their surfaces. Similar sections through pins and rings exhibiting the high wear rate were found to have only occasional hard patches on their surfaces. Welsh demonstrated the hardened patches were formed in a vacuum, while similar steels rubbed on a two disc machine under lubricated conditions would also form hardened layers. These layers were produced under transient scuffing conditions and once formed would protect the surfaces such that 100 times the load could be maintained without further scuffing. He argued that the layers were

formed as a result of high frictional flash temperatures, the layers being austenitised and subsequently quenched to form martensite. This explained why steels of lower carbon content formed hardened layers less easily since increasing the carbon content of steels increases the ease with which austenite forms. Welsh, at first, could not explain the very high hardness of the layers compared to a conventional martensitic structure and performed special tests on steel surfaces in order to reproduce the layers.

Welsh found that hard layers were formed on steel surfaces subjected to sparking under paraffin oil (simulating lubricated conditions) but not when sparked under distilled water. Nitriding a low carbon steel and subsequent sparking under water gave a hardened layer. Based on the similarities between the layers produced by sparking and those found in wear tests, Welsh concluded that atmospheric nitrogen was involved in their formation and this "nitriding" process during wear gave rise to the high hardness of the layers (typically 600-850 HV 100g) though the relative contributions which carbon and nitrogen make to the hardness of the transformed product were not distinguished. In a later investigation, discussed in more detail in Chapter 6, Welsh (1965) identified the general wear patterns of plain carbon steels. Briefly he identified:

- (1) a transition (T_1) from "mild" to "severe" wear occurring at light loads

(ii) a transition (T_2) from severe back to mild wear at heavier loads

(iii) a transition (T_3) of smaller magnitude at loads somewhat higher than the T_2 transition.

Both the T_2 and T_3 transitions (and the region between them) were found to be associated with the appearance of "phase hardening" on both the rubbing surfaces. Welsh avoided contentious issues concerning the nature of the phase hardening, though demonstrated that the calculated flash temperatures at the T_2 transition were sufficient to cause the martensitic transformation at the wearing surface.

Similar transitions in chromium containing steels were identified by Farrell (1969) using a pin on disc machine, under dry conditions. Hard, non etching layers were observed at the corresponding T_2 transition. Electron microscopy showed that the layer contained a fine dispersion of particles, of the order of 50-100 nm in diameter. To account for the formation of the layers, three hypotheses were advanced. First, dispersion hardening of a low carbon martensite by fine carbides. The dispersion occurring due to severe and repeated plastic deformation or due to "autotempering" caused by the raising of the M_s (temperature at which the martensitic transformation starts) as a consequence of plastic deformation at the wearing surfaces. Second, by nitrogen enrichment of surface layers forming a nitro-carbon martensite of high hardness. Third, dispersion hardening of martensite by oxide. Farrell found that with one

specimen tempering for one hour at 800°C under argon caused sub-surface decarburisation of the hardened layers. Since other regions of the same specimen showed no other evidence of decarburisation the evidence of this mode of formation is inconclusive. Farrell also rejected the first two hypotheses as they did not explain the temper resistance of the hardened layer. Farrell gave no firm conclusions except that the layer contained a fine dispersed phase which was probably responsible for its high hardness.

Wilson (1970) observed white layers during unlubricated sliding wear studies of flake and spheroidal graphite cast irons. Mild and severe wear regions were identified in relation to sliding speed and load. The majority of white layers appeared to be formed during running-in. It was observed that white layers found on cast iron surfaces were usually distinct with a sharp boundary. Regions adjacent to the boundary appeared to be work-hardened and distorted. In contrast white layers formed on steel surfaces usually exhibit a heat affected or transition zone between it and the substrate. Transmission electron microscopy revealed that, at a resolution of 2nm, the white layer was structureless. Electron probe microanalysis indicated the layer to be a Fe/Si/C complex with a high carbon and silicon content, compared to the bulk material. Debris collected during running-in often showed that the white layer was attached to the original matrix indicating the layer may not be as brittle as perhaps would be expected. X-ray analysis of the debris identified the presence of austenite, martensite and ferrite. It was concluded that the white layer probably results from the

deformation of high carbon austenite with subsequent back quenching.

Kawamoto and Okabayashi (1971) studied the wear of cast iron in air and vacuum using a rod on ring machine. At sliding speeds > 2 m/s the wear rate decreased, regardless of the atmosphere and examination of the rubbing surfaces revealed hardened patches (~ 1000 HV 50g) of white layers. Although the white layers were generally thicker in vacuum than in air, similar X-ray diffraction patterns were obtained. These also indicated that the white layer was a mixture of ferrite, austenite, martensite and cementite. The hardened patches were therefore attributed to metallographic changes caused by frictional heating and severe deformation.

Surface transformations in the scuffing of En 31 were studied by Torrance and Cameron (1974). Scuffing was induced on a peg on ball machine by incremental loading. Two types of white layer were identified, depending on the conditions. At the higher sliding speed (~ 0.75 m/s) taper sections through scuffed regions of peg and ball revealed white etching surface layers with a dark etching zone or transition region between layer and substrate. The hardness and appearance of the dark etching region suggested that tempering had taken place. Estimates of the surface and sub-surface temperatures were obtained by comparing microhardness/depth profiles across layer and transition zone with the tempering response, over short time intervals (~ 30 seconds) of the bulk material. The authors, while careful to point out the approximate nature of the method, found that the estimated temperature immediately below the white layer was

in excess of 730°C. Estimates of the carbon content of the white layer made by intercept methods, after tempering, revealed negligible difference between white layer and bulk material. Electron diffraction patterns of foils prepared from the white layer identified it to be martensite with a fine grain size (~ 30 nm). The patterns also showed pronounced texture indicating mechanical working of the layer. They concluded therefore that the white layer was developed by the action of frictional shear stresses, causing the surface to be austenitised. A process akin to ausforming was envisaged. Quenching resulting in the martensitic structure.

Slow speed seizure ($\sim 1 \times 10^{-3}$ m/s) produced a second white layer on peg and ball. Although of similar hardness, taper sectioning revealed that no sub-surface tempering of the bulk material had taken place. It was considered that the second white layer was developed by the action of shear stresses breaking up the grain size of the parent (martensitic) material. Further evidence to support the latter conclusion was not provided.

In this section an attempt has been made to give a reasonably balanced "broad brush" review of hardened surface or near-surface layers, often referred to as white layers and typically formed on ferrous surfaces. Other reviews, of varying selectivity and depth, of white layers include those by Eyre and Baxter (1972), Scott et al (1975) and Torrance (1974). At this stage it is important to recognise the diversity of conditions under which white layers are found and ostensibly their differences. Nevertheless this review indicates that white layers do have basic similarities with certain

differences depending on the precise conditions of formation. The following section will concern the properties of white layers and consider some of the parameters which influence its morphology.

2.3 The properties of white layers

By its very nature a white layer is etch resistant, when examined by optical microscopy appears white or pale yellow and is usually featureless. The structure could therefore be simply amorphous or have a grain size too small to be resolved optically. Such a refined grain size, as established in several investigations, would not display the characteristic features of the phase present. Savitsky et al (1964) favour a homogeneous structure with a high dispersion of structural components. The general evidence of the literature, whilst not excluding the first proposal favours the latter two. Wilson (1957) found by performing a hardness test on tempered martensite, re-tempering and etching the specimen, that the region adjacent to the indentation etched light. It was thought that the existing temper carbides could be dissolved by dislocations in the vicinity of the indentation. As the etch response of martensite is thought to originate from the material at the carbide/matrix interface, a reduction in the number of temper carbides will lead to a lighter etching appearance.

Another remarkable aspect of white layers is their extremely high hardness. Various authors report microhardness levels at or near the surface of between 650-1350 HV. These hardness values are virtually always well in excess of the bulk hardness of the parent material; moreover they are harder than could be produced by a conventional

FIG. 2.1 THE VARIATION OF MACROHARDNESS (VICKERS) WITH % wt OF CARBON FOR IRON-CARBON MARTENSITES. (BAIN AND PAXTON (1961)).

hardening process. Figure 2.1 shows the variation of macrohardness with carbon content for iron-carbon martensites. This figure is commonly used to demonstrate that a white layer, by its hardness alone, cannot have a simple martensitic structure. Even if carbon enrichment of the surface layers took place, for example by absorption from oil or gas or by diffusion of carbon to the surface layers (as suggested by Brainin and Seleznoz (1961)) the hardness values obtained would still be insufficient. Some features of figure 2.1 seem worthy of comment. First figure 2.1 is expressed in terms of macrohardness whilst hardness values of white layers are, in practically all cases, quoted as microhardness. This is a consequence of the shallow depth of white layers. Some authors fail to give the indenter load used. These values are not the same and their difference increases with decreasing indenter load. Buckle (1959) surveyed the microhardness values for various metals and found that the most representative values for martensite varied between 950-1100 HV. These values are somewhat higher than implied by figure 2.1. Second, if white layers are assumed to be simply martensitic and therefore of extremely fine grain size, their hardness would be in excess of a conventional martensite value. Hardness is a measure of the yield stress of a material. Petch (1953) has derived a relationship between grain size and yield stress where decreasing the grain size increases the yield stress, though for some reason, as yet unclear, it is inaccurate at grain sizes as small as those implied for white layer. Nevertheless it's qualitative predictions remain the same and small grains result in high hardness. Third, the literature indicates that plastic shear strain may, in many cases, be of importance for white layers. In combination with frictional heating, the process just described is autoforcing, a fact recognised by many authors. Eyre and Baxter (1972), though, have

expressed reservations on the grounds of a lack of complete similarity in the conditions prevailing in the two circumstances. The mechanical properties of ausformed martensite in steels are in general superior to those obtained by normally heat treated martensite. Kenneford et al (1968) have shown that ausformed martensite has, for example, a 30% increase in tensile strength together with increased hardness. A 0.7% carbon steel exhibits a 10% increase in hardness with ausforming, this trend increases with decreasing carbon content. The main reason for these superior properties is thought to be the difficulty of dislocation movement in such fine grained structures. Repeated temperature and deformation cycles, as found in most rubbing systems, will further refine the structure produced, indeed such situations are far more complex than found in present studies of thermomechanical heat treatments.

The appearance of white layers tends to fall into two groups. In the case of steels, white layer usually has a heat-affected region or transition zone between the layer and substrate. White layers on cast iron or formed under conditions of high plastic shear strains are usually more distinct with a sharp boundary. As yet there appears to be no explanation for the distinct boundary with white layers on cast iron.

White layers are resistant to tempering. A conventional martensitic structure begins to temper at between 80-160°C by the martensite gradually losing carbon atoms which combine with the iron atoms to form ϵ -iron carbide. White layers formed under conditions of frictional heating appear to be temper resistant up to approximately 400°C, while those formed under conditions of high shear strains appear to be temper resistant even above 723°C. There is also evidence suggesting that white

layers formed on cast iron are somewhat more resistant to tempering than those found on steel surfaces.

As the tempering response of a white layer is usually assessed in terms of microhardness the earlier discussion of microhardness values of a martensitic structure is, in essence, applicable here. Miller and Breynner (1969) have found that martensite produced during a mechanical working process, akin to ausforming, has a temper resistance up to temperatures of 275°C. Wilson (1957) observed a similar temper resistance of the light etching component surrounding the hardness indentation in tempered martensite. Miller and Breynner suggested that the normal tempering sequence for martensite was inhibited by two principal effects: (i) the attraction of carbon atoms to dislocation cores rather than to developing temper carbides and (ii) the attraction of carbon atoms to vacancies in the lattice produced during deformation. The effect of these two processes was to inhibit the development of temper carbides by effectively removing the carbon atoms from solution and to increase the strength of the martensite by pinning the high dislocation density, through the segregation of carbon atoms to their cores. These mechanisms of temper resistance may well apply to white layers. Kenneford et al (1968) have also observed that ausformed steels are more resistant to tempering compared to similar steels conventionally quenched. They considered the resistance to tempering to be due to a fine carbide precipitation.

2.4 Theories of formation

The selective review of section 2.2 clearly indicates that despite the relatively divergent conditions and materials under which white layers are found, there are probably three main and identifiable modes of formation. Initially these will be considered separately. These are:

- (i) as a consequence of the surface (or sub-surface) shear strains produced during rubbing, the grain size of the structure will be reduced such that it cannot be resolved optically.
- (ii) as a result of frictional heating causing the surface layers to austenitise and on quenching produce a fine grained martensitic structure.
- (iii) the environment surrounding the rubbing surface reacts with it to harden it in some way.

Other authors postulate similar modes of formation some of which contain combinations of these major mechanisms. These modes will now be considered in more detail.

The first mode describes white layers formed under conditions where frictional heating is conventionally thought to be insufficient to cause phase changes and where the environment is shown to be irrelevant. White layers so formed usually have a similar crystallographic structure to the bulk material and generally exhibit no transition zone between layer and substrate. An estimate of the likely strain rate can be made by assuming

the white layer is formed by uniform shear between the rubbing members. The strain rate is then the rubbing velocity divided by the depth of white layer. Turley (1975), for example, observed white layers 1 μm in depth during reaming of high strength at a speed of $1.5 \times 10^{-2} \text{m/s}$. Assuming no shearing at the surface of the reamer, this is equivalent to a strain rate of $1.5 \times 10^{+4} \text{s}^{-1}$, which is high indeed. Some features of this mode of formation seen worthy of comment. First, phase changes in some steels can occur by deformation alone, in particular, the transformation of retained austenite to martensite by the application of high stresses. Hadfield's steel relies on this transformation to form a protective martensitic skin and thus its use in fabrication of such items as dredge buckets, rock crushers etc. Second, such high strain rates can also manifest itself as heat. Shear strain rates of the order of $2 \times 10^3 \text{s}^{-1}$ were estimated by Zener and Hollman (1944) during the blanking of steel plate with concurrent white layer formation. Calculations showed that a shear strain of 5 would raise the temperature, in this region, by 1000°C though assumptions made in this computation were not reported. White bands were observed by Carrington and Gayler (1948) on steel balls after hitting mild steel plate at high impact velocities. They concluded that the regions of maximum shear stress were raised to temperatures in the region of the melting point. Schlicht (1974) considered the concomitant effects of plastic lattice deformation and high hydrostatic pressure on the transformation of steels and concluded that in such cases e.g. butterfly formation in rolling element bearings, plastic lattice deformation will produce high temperatures sufficient to austenitise such zones. In such cases frictional heating produced at the rubbing surface was considered negligible.

There is extensive evidence from the literature to support the second mode of formation, that is frictional heating causing a fine grained martensitic structure. These include the existence of a tempering or transition zone between white layer and substrate, and that as section 2.3 has indicated, the properties of white layers are not dissimilar to that of a fine grained martensitic structure. Retained austenite, untransformed ferrite and undissolved carbide may also be present in such layers. Moreover, various estimates of the transient surface or flash temperatures occurring have been made either directly (e.g. by thermocouple), indirectly (e.g. by hardness changes) or by calculation. Regardless of their respective problems of interpretation the overwhelming evidence points to surface temperatures in excess of 723°C, the minimum temperature of austenitisation for plain and low alloy carbon steels. Flash temperature measurement will be discussed in more detail in the following section.

The third mode principally covers white layer formation by (i) a nitriding mechanism by combination with atmospheric nitrogen (ii) formation of a carbon rich phase by reaction with lubricant or atmosphere and (iii) formation of amorphous material engulfing oxides. These will be considered in turn.

Several authors, in particular Farrell (1969) and Torrance and Cameron (1974) have discussed the relative merits of the nitriding process as proposed in (i); the essential points of which will be reasoned here. First, it should be noted that the nitriding of plain carbon and low alloy steel does not by itself confer great hardness on ferrite. The postulated mechanism of hardening during wear, that is the formation of a

nitrogen or nitrocarbon martensite through rapid cooling from the austenitic range is distinct from that involved in a conventional nitriding treatment. Conventional nitriding involves the diffusion of atomic nitrogen into the ferritic phase in the temperature range 500-590°C over typical durations of 1 - 5 hours. No phase transformations occur on cooling to room temperature. Second, if the compositions are expressed as interstitial atomic percentages it has been found that the hardness of an as-quenched iron-nitrogen martensite is about the same as an equivalent iron-carbon martensite. This is not surprising on account of their atomic size similarity. Thus the addition of nitrogen atoms to an iron-carbon martensite will produce properties similar to those of a high carbon martensite and may therefore contribute to the hardness of the white layer. However, even if this were to occur this would not exert a major effect on the surface hardness. It would also not explain the resistance to tempering of white layers since it would seem reasonable to assume that the tempering behaviour between iron-carbon and iron-nitrogen martensites would be similar. Finally it is evident that the absorption of nitrogen in iron-carbon alloys is not accomplished unless the partial pressure of nitrogen is considerably in excess of that in atmosphere. Although estimates by Farrell (1969) and Torrance and Cameron (1974) indicate that at temperatures of the order of 600°C the pressures occurring at contacting asperities would be sufficient to dissolve nitrogen, as before, it is doubtful whether the addition of nitrogen atoms to an iron-carbon martensite would significantly enhance its properties and account for the properties of white layers.

Carburisation as proposed in (ii) conventionally requires low carbon steels and somewhat higher temperatures ($\sim 825-925^{\circ}\text{C}$). The enhancement

process is caused by a locally increased carbon content, on quenching, producing a high carbon martensite. Whilst several authors favour this explanation in particular under lubricated conditions, apart from rather exceptional conditions e.g. in a gun tube, direct evidence of increased carbon content in white layers is limited. As this will produce, simply, a high carbon martensite, by itself, it is difficult to account for the properties of white layers.

Farrell (1969) observed in tempering a wear transformation product in argon, decarburisation occurred in the layers immediately below. Previous evidence of oxide entrapment in surface layers is provided by Moore and Tegart (1952) who observed oxide inclusions produced in copper after repeated rubbing against a steel slider. A fine grained dark etching product was found. It was thought that this process was largely responsible for the fine grain size of the Beilby layer.

As white layers are, practically, always found associated with an oxide film at the surface, Farrell proposed that due to intense plastic deformation at the wearing surfaces the oxide film will be broken up and ploughed into the sub-layers. Further deformation and possible frictional heating would cause further break-up and eventually a dispersion of oxide would be formed which would harden the sub-layers. The striated appearance of many white layers and the tempering results of Farrell give some credence to this mode of formation. However this evidence is not conclusive; the striated appearance may be due to plastic deformation alone, while the decarburisation evidence was not repeated in other regions of the same specimen. Although oxide particles may well be present in white layers there is ample evidence to show that such layers

are formed in an oxygen-free environment where oxide films will be a minimum or during the early stages of running-in where no appreciable oxide film would be present on the surface.

So far in this review, the principal theories of white layer formation have been considered to occur in isolation of each other. In practice this is unlikely, if not impossible. It is evident that temperature, deformation and environment are important in every case. However it is difficult to define to what extent these factors interrelate in each particular case of white layer formation. Moreover a vital feature of probably all reported examples of white layer is that it has been formed as a result of an unknown number of rubs or contacts of unknown severity. It is therefore proposed to use a crossed cylinders machine in which specimens can be subjected to single and multiple rubs of known severity. Soft plain carbon steels will be used at Hertzian contact pressures of approximately their elastic limit. Unlubricated conditions will be used and thus white layers formed primarily by frictional heating will be considered. Such material will be referred to as "phase transformed material". The following section will consider the theory and measurement of flash temperatures applicable to such experiments.

2.5 Flash temperatures

The generation of high flash temperatures is not new. When man invented the art of making fire by rubbing two sticks together he made far-reaching and intelligent use of flash temperatures. The heat is generated by the formation and separation of junctions at the areas of contact. Nearly all the frictional energy appears as heat and will be

TEMPERATURE

FIG. 2.2 SCHEMATIC TEMPERATURE PROFILE THROUGH RUBBING SURFACE DEFINING SURFACE TEMPERATURES

conducted to the surroundings. If a steady state of heat transfer is achieved a reasonably consistent bulk interfacial temperature will be established. This bulk interfacial temperature of the moving bodies can be measured with relative ease. The highest temperatures which occur close to areas of true contact at which energy is dissipated, are of transient duration and are usually called flash temperatures. These temperatures can, therefore, exercise considerable influence on friction and wear. Despite various attempts which have been made, experimental evaluations of flash temperatures are usually subject to some doubt as to the accuracy of the values obtained. Several heat transfer models have also been proposed and some of them predict values which appear reasonable and are consistent with certain physical changes occurring at the surface e.g. thermal softening. It is prudent to note that any theory used to estimate flash temperatures can only be applicable if the fundamental assumptions of that theory are relevant to the particular case under consideration. Any significant deviation may well give erroneous results.

This review will be appropriate to the requirements of this thesis and will consider the essential points and discuss some of the relative merits of flash temperature theory and measurement. Exhaustive details may be obtained from the extensive reviews in the literature, in particular those by Watson (1964), Tenwick (1971), Richardson (1976) and Winer (1978).

Before discussing flash temperatures the relative terms used to describe such temperatures will be distinguished. Fig 2.2 shows a schematic temperature profile through a rubbing surface. θ_{bk} is the bulk

specimen temperature or substrate temperature and measured well away from the contacting surface. θ_{sf} is the bulk interfacial or average surface temperature. θ_f is the flash temperature; it is the highest temperature occurring close to the areas of contact and is of transient duration.

2.5.1 Flash temperature theory

A simple estimate of the flash temperature developed in sliding was made by Bowden and Tabor (1950) who considered the case of a cylinder moving unidirectionally along a plate under a normal load W at a velocity V . If all the frictional energy is developed as heat then Q the amount of heat developed per second, is given by:

$$Q = \mu W V \quad (2.1)$$

where μ is the coefficient of friction. Assume that the cylinder is receiving a fraction α of the total heat generated, that is αQ . The heat diffusing into the plate will be $(1-\alpha)Q$. The cylinder receives heat by conduction from the interface which will be lost by radiation and convection. If the latter is ignored, in the steady state, there will be a balance between the heat conduction along the cylinder and the heat lost by radiation. Assuming that the cylinder is so long that the top end of it is sensibly at room temperature, Bowden and Tabor arrive at the following expression,

$$\theta - \theta_{amb} = \frac{\alpha Q}{\pi r \sqrt{2k\sigma r}} \quad (2.2)$$

FIG. 2.3 ARCHARD (1959) MODEL (i) SINGLE CONTACT AREA
(ii) SUB - DIVIDED CONTACT AREA

Where θ is the temperature of the flat end of the cylinder in contact with the plate, θ_{amb} is the ambient temperature, r is the radius of the cylinder, k is the thermal conductivity and σ is the cooling coefficient assuming Newton's law of cooling.

In view of the large number of assumptions made in this analysis it can only be regarded as a crude approximation to the calculation of flash temperatures. Nevertheless using these simple considerations alone, it can be found that the surfaces reach quite high temperatures. Using equation 2.2 Bowden and Tabor (1950) estimated that the temperature rise of a 1 mm diameter constantan cylinder sliding on a mild steel plate under a load of 1N at a velocity of 2 m/s as 180° C.

Heat flow phenomena at the areas of true contact have been rigorously analysed by Blok (1937) and Jaeger (1942). Their work was presented by Archard (1959) in a simplified form suitable for application to the experimental work of this thesis. The essential idea is that the contact region is a plane source of heat and all the frictional energy appears as heat, generated at the interface, which is conducted away into the bulk of the rubbing members.

Figure 2.3 shows the model used. A protuberance on the surface of body B forms a circular area of contact $A=\pi a^2$, which moves with velocity V over the flat surface of Body C. Body B therefore receives heat from a stationary heat source while body C receives heat from a moving heat source.

For a stationary contact a good approximation to steady state conditions exists and the heat conduction becomes the thermal analogue of the constriction resistance of electrical contacts (Holm (1948), (1952)). It was shown that the mean rise in temperature over the contact area θ_m could be calculated from the following equation

$$\theta_m = \frac{Q_B}{4ak_B} \quad (2.3)$$

Subscript B refers to body B. For a slow moving heat source there is enough time for the temperature distribution of a stationary contact to be established in body C and therefore

$$\theta_m = \frac{Q_C}{4ak_C} \quad (2.4)$$

At higher speeds this equation no longer holds because there is insufficient time for the temperature distribution of a stationary contact to be established in body C. The speed at which this occurs depends on the magnitude of Jaeger's non-dimensional parameter L (sometimes referred to as the Peclet number).

$$L = \frac{Va}{2\kappa} = \frac{V\rho c}{2k} \quad (2.5)$$

$\kappa = \frac{k}{\rho c}$ is the thermal diffusivity, ρ the density and c the specific heat. To a reasonable approximation equation 2.4 applies for $L < 0.1$.

At higher sliding speeds (say, $L > 5$) the duration of the moving contact at any point on the surface is short compared with the time necessary to establish steady state conditions; therefore (during the duration of the contact) the heat diffuses to a depth which is small compared with the radius of contact a . Under these conditions sideways flow of heat can be neglected and the heat conduction is a close approximation to that given by linear heat diffusion theory. In these conditions if heat is supplied at a constant rate to a plane surface of a semi-infinite solid Carslaw and Jaeger (1947) give the temperature θ of a point on the surface as:

$$\theta = \frac{2qt^{\frac{1}{2}}}{(\pi k \rho c)^{\frac{1}{2}}} \quad (2.6)$$

where q is the rate of heat supply per unit area ($= Q/A$) and t is the time for which it has been supplied. Using equation 2.6 and taking an effective value of $t^{\frac{1}{2}}$ for any point in a circular contact of area A one obtains:

$$\theta_m = \frac{Q_c}{3.25ka} \sqrt{\frac{1}{2L}} \quad (2.7)$$

The results of the above theory can be conveniently summarised as follows:

- (i) For a stationary heat source or for very low loads ($L < 0.1$)

$$\theta_m = 0.5NL \quad (2.8)$$

FIG 2.4 VARIATION OF β WITH JAEGER'S NON DIMENSIONAL PARAMETER L (JAEGER (1942))

FIG. 2.5 AVERAGE TEMPERATURE AS A FUNCTION OF DIMENSIONLESS PARAMETER L . A STATIONARY HEAT SOURCE, B MOVING HEAT SOURCE.

where $N = \frac{\pi q}{\rho c V}$

(ii) For moderately low speeds ($0.1 < L < 5$) the temperature for a moving heat source falls below that given by equation 2.7 and

$$\theta_m = 0.5 \beta N L \quad (2.9)$$

β is a function of L and ranges from about 0.85 at $L=0.1$ to about 0.35 at $L=5$. This value can be obtained from the calculations of Jaeger (1942) or Holm (1948) and is shown in Figure 2.4.

(iii) At still higher speeds ($L > 5$) equation 2.7 applies and

$$\theta_m = 0.435 N L^{\frac{1}{2}} \quad (2.10)$$

From equations 2.8-2.10 a smooth curve of θ_m versus N can be constructed (Figure 2.5).

Archard's method of dealing with the distribution of heat to the two bodies is summarised by the expression

$$\frac{1}{\theta_m} = \frac{1}{\theta_B} + \frac{1}{\theta_C} \quad (2.11)$$

where the temperature rise for each body (θ_B & θ_C) is calculated on the assumption that all the heat (Q) is supplied to it. This equation is obtained from the assumption that the division of heat between the two bodies is such as to make the average temperatures of the two surfaces within the contact of

equal value. Therefore for the model chosen (Figure 2.3) θ_B is calculated from the stationary heat source theory using the appropriate value of L_B . The temperature rise θ_C is calculated from the moving heat source theory using the appropriate value for L_C .

The above method has been used to calculate flash temperatures in the main experiments of this thesis. Archard also presented the above equations in a form where either plastic or elastic deformation was assumed and whether high or low speed conditions were applicable. These derived equations have been used in the exploratory experiments of Chapter 4, being conveniently suited to the simpler analysis of that chapter.

Modifications have been made to Archard's theory to account for the number, size and shape of the individual contact areas over which heat is generated, though the basic postulate of Archard's theory is that there is a single contact area or, at worst, a compact cluster of contact spots which act as one. Using the electrical analogue it has been shown both experimentally and theoretically that the aggregate resistance of a large number of closely grouped constrictions δA (Figure 2.3) is very little greater than the resistance of a large constriction A . Indeed for a single contact region, as between crossed cylinders, the resistance between rough surfaces is very nearly the same as the between smooth surfaces (Archard (1959)). When the contact spots are well separated temperatures are lower; for very low speeds ($L < 0.1$) the temperature with n widely separated contacts will be $n^{-1/2}$ of that of a single one

with the same total area: for high speeds ($L > 5$) the factor will be $n^{-1/4}$. Lancaster (1976) has studied friction and wear transitions in carbons and graphites rubbing against themselves in relation to the flash and total temperatures (i.e. flash + bulk). The equation used was:

$$\theta_m = \frac{0.752 \mu V^{1/2} W^{1/4}}{\epsilon(kpc)^{1/2}} \quad (2.12)$$

where in this case ϵ is a function of contact number, size, shape and mode of deformation.

Other theoretical analyses of varying mathematical complexity include those by Ling (1969) using a quasi-iterative method for computing interface temperature distributions, while Cameron et al (1965) considered contact temperatures in rolling/sliding surfaces. Both authors have found excellent agreement with the flash temperature, for the case of a uniform heat source and have not reported any significant increase in the maximum temperature over Blok's theory. Francis (1970) derived an analytical expression for the steady state interfacial temperature field in a sliding circular Hertzian contact. The particular case considered was for one body stationary, a difference in the bulk temperatures of the bodies and an ellipsoidal distribution of frictional power. The analysis gave a resulting flash temperature approximately 35% higher than that given by Blok's theory. In the discussion of the paper Cheng (1970) noted that for the case considered, an ellipsoidal distribution will give a maximum temperature approximately 20% greater than that

produced by a uniform heat source. It was believed that this difference was mainly due to the non-uniform heat source rather than due to refined matching of the interfacial temperature.

Barber (1967) considered the mechanism of a single interaction between two asperities on opposing surfaces, for the case when $L > 10$. The problem was treated in two stages. First, both initial surface temperatures (before the interaction) were assumed to be zero and the average temperature rise produced on the surface of each of the contacting asperities was calculated. Then an initial temperature difference between the asperities was assumed and an expression produced for the steady state temperature rise in terms of this difference. The combination of the two results gave an expression, equation 2.13, for the maximum temperature obtained in the contact area for similar materials and equal bulk temperatures of

$$\theta_m = \frac{0.84 \mu p_m^{3/4} V^{1/2} W^{1/4}}{(k\rho c)^{1/2}} \quad (2.13)$$

This can be compared with equation 13b from Archard (1959); high speed conditions, plastic deformation:

$$\theta_m = \frac{0.73 \mu p_m^{3/4} V^{1/2} W^{1/4}}{(k\rho c)^{1/2}} \quad (2.14)$$

The equations are of identical form and differ only by a small numerical factor. Powell and Earles (1972), in studies of the surface temperature in high speed unlubricated sliding of SAE 1113,

found that the Barber model predicts total temperatures (i.e. $\theta_T = \theta_m + \theta_{bk}$) which, on the basis of the worst comparison are only marginally different (approximately 60°C difference, in up to 1000°C) than those predicted by Archard's model. Tenwick (1971) reports similar conclusions. A feature of these comparative studies was that the required physical properties of the material were taken at or near the bulk specimen temperature. The choice between these models is therefore of low significance.

Tenwick (1971) derived a flash temperature equation based on a model of square asperities on both surfaces. As the surfaces move the load is passed from one pair of asperities to the next. For identical materials and equal bulk temperatures Tenwick obtained:

$$\theta_m = \frac{0.35 \mu p_m^{3/4} V^{1/2} W^{1/4}}{(kpc)^{1/2}} \quad (2.15)$$

Whilst of identical form, the numerical factor is considerably less than that of Archard's and Barber's equivalent equations. However some assumptions made in this model are erroneous. In particular Tenwick's experiments corresponded to values of $L > 10$ but in the heat flow calculations a hemispherical heat source is assumed. This will only be applicable at values of $L < 0.1$.

Prior to a comparison of flash temperature theory with measurement, some points and problems of the validity of flash temperature calculations will be considered with reference to the proposed experiments on the crossed cylinders machine. An analysis is

appropriate if the subsequent flash temperature calculations are to gain credence.

Two points have already been mentioned. First, the applicability of the model and second, the sub-division of the contact region.

Archard's model is directly applicable to the crossed cylinders machine and in particular to the single rub portion of a typical experiment i.e. a single contact of known severity. Any sub-division of the contact region within such an experiment has been shown to have little, if any, effect on the flash temperature.

Other points to be borne in mind when estimating the flash temperature rise of sliding surfaces are:

(i) Frictional heat may not simply be generated at the interface between two bodies in sliding contact. Some or most of the heat may be generated in the layer immediately below the surface of both bodies. Rigney and Hirth (1979) have proposed a steady state friction model where most of the deformation energy is dissipated in a region of average thickness z (typically 10 μm) below the wearing surface. Such a dispersed heat source will tend to give a lower flash temperature than a localised heat source. Briefly two comments may be made. First, we are generally interested in the maximum flash temperature at the surface and secondly, if experiments are performed where the contact pressures are around the Hertzian elastic limit i.e. orders of magnitude below those required for full plasticity, this effect will be reduced.

(ii) The effects of oxide or surface films has been discussed by Archard (1959). Such surface films will usually be of low thermal conductivity and to have any marked influence on the flash temperature, must be thick compared to molecular dimensions. The general conclusion was that surface films at both low and high sliding speeds would tend to raise the flash temperature and to lower or leave unaffected the bulk temperature.

(iii) Some authors, in particular Dayson (1967) have invoked contact or junction growth, to modify Archard's theory and allow for the fact that a shearing force will tend to increase the junction size. The equation

$$1 + d \mu^2 = \left(\frac{A_1}{A} \right)^2 \quad (2.16)$$

was assumed to be valid. A_1 is the new area incorporating the junction growth effect and d is a constant whose value has experimentally been determined (Bowden & Tabor (1950)) to lie between 3 and 25. Assuming a value of $d=25$ and $\mu=0.5$, this gives a value of (A_1/A) of 2.7 which reduces the estimate of flash temperature by a factor of about 2. However equation 2.16 is strictly true for purely plastic deformation only while such contact area growth would tend to produce separated contact spots which reduce the overall flash temperature in any case.

(iv) In evaluating a flash temperature certain assumptions are made as to the temperature to which the physical (k , ρ , c) and mechanical (p_m) variables must correspond. At low specimen bulk temperatures (say $< 200^\circ\text{C}$) the error in using these variables at room temperature values will be negligible since in the case of the physical properties the heat flow is determined by the substrate at the bulk specimen temperature. Likewise the contacting asperities are supported by the substrate material at the bulk specimen temperature. However at higher bulk specimen temperatures (say $> 200^\circ\text{C}$) the effect of temperature on the variables increases and the appropriate values should be used. Where bulk specimen temperatures are high, it may also be more useful to quote the total temperature.

(v) In repeated rubbing the contact surfaces may undergo changes e.g. work hardening, phase transformations etc. which will tend to invalidate the initial assumptions made. For example an increase in surface hardness will tend to reduce the contact size and increase the flash temperature. However the increased temperatures will tend to lower the hardness. In a single rub experiment these effects will be minimised.

(vi) The temperatures calculated are the maximum average flash temperature θ_m in the contact area. Whilst parts of the contact region may not obtain this value, there will be regions, particularly at high sliding speeds where temperatures in excess of θ_m occur up to a maximum of $\theta_{\max}$. This will be

discussed in more detail in a later section.

2.5.2 Flash temperature measurement

Of the methods employed to measure the surface temperature in the region of the contact, probably the three most often used are (i) the embedded thermocouple (ii) the dynamic thermocouple and (iii) infra-red sensors.

The embedded thermocouple has been extensively used but it does have certain problems associated with it. Usually a hole is drilled in the test specimen in such a way that the thermocouple can be mounted very close to the surface of contact. Since the thermocouple is not at the interface, it is not measuring the flash temperature θ_f but, at best, the interfacial temperature θ_{sf} . The use of embedded thermocouples is also unsatisfactory as the placement of the thermocouple tends to interfere with the heat flow.

Ling and Simkins (1962) used thermocouples to determine the bulk interfacial temperature distribution on a rotating disc with a circumferentially loaded rider. The design was such that the heat fields were two dimensional in the plane perpendicular to the interface. This allowed thermocouples to be mounted on the surface of the specimens and therefore the heat flow was not disrupted by embedded thermocouples. At high values of L , bulk interfacial temperatures of between 125-315°C were measured. Quinn (1968) used embedded thermocouples on a pin on disc machine to study the influence of temperature on the wear of steel in unlubricated

conditions. Bulk interfacial temperatures of up to 248°C were observed (at a load of 36N and speed of 6.25 m/s). In subsequent work Quinn (1978) modified a pin on disc machine and mounted the pin in a calorimeter with precisely located embedded thermocouples. From a heat flow analysis it was possible to deduce the bulk interfacial temperature. Values of up to 250°C were observed (at a load of 26N and speed of 5.1 m/s).

Tenwick (1971) allowed the thermocouple wires to protrude from the pin surface and the junction was formed by the frictional heating when the pin was brought into contact with a rotating disc. Tenwick proposed that the strength, rigidity and location of the junction was improved. He did not expect this junction to respond to individual flash temperatures. Spurr (1979) (1980) has compared theoretical flash temperatures with temperatures measured by thermocouples embedded in wax specimens loaded against various metallic discs. Maximum temperature rises measured were less than 60°C. Perhaps not surprisingly, correlation with temperature calculations from the Archard theory was only fair.

Embedded and trailing thermocouples were used by O'Donoghue and Cameron (1966) to measure bulk interfacial temperatures in rolling sliding contacts between steel discs under lubricated conditions. Values up to 100°C were recorded and both methods gave temperatures which were generally of similar magnitude. Dyson et al (1965) report that the temperatures recorded from a trailing lead thermocouple on steel specimens were as much as 10°C lower than that of an embedded thermocouple 3.2mm below the surface (at temperatures

as low as 50°C). Wannop (1970) has concluded that there is a justifiable doubt as to what temperature a trailing lead thermocouple is recording in addition to effects from the experimental conditions prevailing e.g. the addition of fresh cold oil etc.

The dynamic thermocouple uses the emf generated between two dissimilar materials in relative motion to measure the surface temperature. However this method will generally only provide a measure of the bulk interfacial temperature since the overall emf will be the integrated value of many junctions formed and broken over the apparent area of contact. Discrete temperature peaks tend to be difficult to measure. The dynamic thermocouple cannot be used for fully or partially lubricated systems, while wear and material transfer of one surface to the other will reduce the emf generated since sliding will then be between like metals. The presence and formation of surface oxide films will further alter the emf generated.

One of the first to use this method was Shore (1925) in investigations of surface temperatures at cutting tool/workpiece junctions. Bowden and Ridler (1936) later used this method in a variety of experiments with a constantan alloy pin sliding against a steel disc. They concluded that in many cases the maximum temperatures attained corresponded to the melting point of the more fusible metal. Furey (1964) using a similar system measured both the bulk interfacial temperature and the maximum surface temperature. The latter was considered equivalent to a flash

temperature and compared to theoretical estimates. At a maximum load of 2.5N and speed of 2.25 m/s it was found that the bulk interfacial temperature was 115°C while the maximum surface temperature was 330°C. Theoretical calculation of flash temperature using Archard's theory gave values of 385°C assuming plastic deformation and 536°C assuming elastic deformation at the area of the contact. The general correlation between Archard's theory and measured values was limited.

Dayson (1967) attempted to correlate Furey's divergent results by assuming contact growth and obtained better agreement. In the discussion following Dayson's paper additional features were suggested to account for the observed divergence. In particular in Furey's experiments large scale adhesion, high coefficients of friction and wear took place so that all the heat was not generated at the surfaces nor was it all conducted into the bulk of the sliding members. It is clear that the output from a dynamic thermocouple is difficult to interpret in terms of the flash temperature. More recently Cook and Bhushan (1973) measured the temperatures between a wide range of different metals using the dynamic thermocouple technique. Their general conclusion was that there was fair agreement between these temperatures and their calculation, using a derived Jaeger model, of a bulk interfacial temperature (θ_{sf}).

A third method of measuring surface temperature is by the use of an infra-red sensor which measures the amounts of radiant energy leaving a given area; from this, the temperature can be determined.

One major advantage of this method is that the sensing device can be mounted away from the surface under question, thus causing no interference with local heat conduction. Two disadvantages, however, are that one of the members of the sliding system must be transparent to infra-red radiation while there is some question as to the depth resolution of this method. It is doubtful if the technique could, for example, discriminate between the surface temperature or the temperature, say, 20 μm below the surface. Also if this method is to record flash temperatures, the sensing cell must have a small time constant e.g. $\sim 10^{-4}$ seconds.

Bowden and Thomas (1954) used a lead sulphide cell sensitive to infra-red radiation to measure the flash temperatures between metallic pin specimens rubbing against a rotating glass disc. The cell was located beneath the wear track. Experiments with this apparatus allowed several conclusions to be drawn, all of which support theoretical flash temperature predictions:-

- (i) At a given load and speed the hot spots vary widely in size and temperature
- (ii) The larger the size of a hot spot, the longer it takes to reach the maximum temperature commensurate with the load and speed.
- (iii) The higher the thermal conductivity of the rider the lower is the maximum temperature obtained.

(iv) The duration of a typical hot spot is of the order of 10^{-4} seconds and temperatures up to the melting point of the pin metal were observed.

Winer (1978) used a Barnes infra-red microscope to measure temperatures in sliding ehl point contacts. He was able to determine the temperature distribution in the oil film and on the surface of a rotating steel ball loaded against a sapphire plane. He compared the measured surface temperature with the predicted flash temperature calculated using Archard's theory for lubricated contacts and found good agreement, though the maximum temperature rise was less than 50°C . It is interesting to note that in simulating starved conditions transient temperature spikes as high as 1100°C in less than $8\text{ }\mu\text{s}$ would occur.

Richardson (1976) used a similar apparatus to study temperature rises in the unlubricated rubbing of graphite. Although the surface temperature rises were 100°C there was fair agreement with the predicted flash temperature using Archard's theory. In this case the trend was for the calculated temperatures to underestimate those observed. Parker and Marshall (1948) had earlier used a lead sulphide pyrometer, with a response time of 10^{-3} seconds to measure temperatures generated between railway brake blocks and tyres. The surface temperature of the tyre was determined just as it emerged from the brake block. High flash temperatures 800°C were repeatedly observed. Subsequent metallographic analysis of the tyre identified "heat transformed martensite" in the surface regions.

Other methods of measuring surface temperatures include wax pencils of varying melting points, colour changes of paints, microhardness or micro-structural changes, thermography etc. These methods are conveniently reviewed by Watson (1964). The essential point is that such methods give, at best, the bulk interfacial temperature and cannot be relied on to give the flash temperature.

The last two sections have concerned a discussion of the relative merits of flash temperature theory and measurement. In relation to this thesis the following points seem evident:-

- (i) High flash temperatures of transient duration exist and are in general much higher in magnitude than the bulk interfacial temperature.
- (ii) Of the various techniques to measure surface temperatures, infra-red radiation methods approach the closest to a "true" flash temperature measurement.
- (iii) As yet there appears no method for measuring flash temperatures between rubbing steel surfaces. Inserting a sapphire window at the point of contact and measuring the infra-red radiation emitted has been suggested. However the window will interrupt the heat flow at the contact region and it seems unlikely if this method will yield reliable results.
- (iv) The Archard model for flash temperature calculation is the most widely used and probably the most useful method of

predicting flash temperatures to date, provided that the theory is applied in such a manner that the fundamental assumptions therein remain valid.

2.6 Sub-surface temperatures

Existing flash temperature theory is mainly concerned with surface temperatures. In interpreting the depth of transformed material produced in our experiments we shall require a theory which forecasts the magnitude of transient sub-surface temperatures. We shall use Archard's (1959) theory for the derivation of surface flash temperatures and it is thus appropriate to use an extension of this same theory to the sub-surface region. Because this extension is based upon unpublished notes of Archard (1980) it will be set out here in a little more detail. In this derivation the same assumptions and model (Fig 2.3) of Archard's earlier analysis will be used. It will be assumed that the surface temperatures have already been derived using the earlier theory; therefore an appropriate method of presenting sub-surface temperatures is to derive them as a fraction of the corresponding surface temperature.

2.6.1 Stationary and slow moving heat sources

The surface temperature of a stationary or slow moving heat source ($L < 0.1$) was derived by analogy with the equivalent electrical problem. Sub-surface temperatures will be similarly derived.

In his work on electrical contacts Holm (1958) discusses the potential field surrounding a circular contact of radius a . It is shown that the equipotential surfaces are given by:-

$$\frac{x^2}{a^2 + \mu} + \frac{y^2}{a^2 + \mu} + \frac{z^2}{\mu} = 1 \quad (2.17)$$

where $\sqrt{\mu}$ is the depth of the equipotential surface normal to the plane of the contact area.

The resistance, R , between the contact area and the surface defined by μ is

$$R_{\mu} = \frac{\rho}{2\pi a} \tan^{-1} \left(\frac{\sqrt{\mu}}{a} \right) \quad (2.18)$$

and it is noted, as a check, that the resistance as $\mu \rightarrow \infty$ (the "half constriction" resistance) is

$$R_c = \frac{\rho}{4a} \quad (2.19)$$

If μ defines a surface such that

$$R_{\mu} = R_c (1 - \alpha) \quad (2.20)$$

Noting that this corresponds to a temperature $\alpha\theta_m$ where θ_m is the mean surface flash temperature.

$$\text{Thus } \frac{R_\mu}{R_c} = 1 - \alpha = \frac{2 \tan^{-1} (\sqrt{\mu}/a)}{\pi}$$

$$\tan^{-1} \left(\frac{\sqrt{\mu}}{a} \right) = \frac{\pi}{2} (1 - \alpha)$$

$$\left(\frac{\sqrt{\mu}}{a} \right) = \tan \frac{\pi}{2} (1 - \alpha) = \tan \phi \quad (2.21)$$

The following table can be drawn up for values of α between 0.9-0.5. α is equivalent to the temperature at depth z , θ_z , as a proportion of the surface temperature θ_o .

Table 2.1

$\alpha = \theta_z / \theta_o$	$1 - \alpha$	ϕ°	$\tan \phi = z/a$
0.9	0.1	9	0.140
0.8	0.2	18	0.325
0.7	0.3	27	0.510
0.6	0.4	36	0.726
0.5	0.5	45	1.000

2.6.2 Fast moving heat sources

Sub-surface temperatures at high values of L , are calculated on the assumption that there is a plane source of heat (the area of true contact) at the surface, $z = 0$, generated at time $t = 0$; and that this heat is conducted away into the bulk of the body, originally at zero temperature. It has been shown, earlier, that this is a

problem of linear heat flow, and that heating is limited to depths small compared to the extent of the contact. Under these non-steady state conditions, the equation for heat flow is given by

$$\rho c \frac{\partial \theta}{\partial t} = k \frac{\partial^2 \theta}{\partial x^2} \quad (2.22)$$

a. Physical Derivation

First an expression will be derived for sub-surface temperatures from physical reasoning, using as a starting point Maclaurins expansion

$$\theta(z) = \theta_0 + z \left(\frac{d\theta}{dz} \right)_{(0)} + \frac{z^2}{2} \left(\frac{d^2\theta}{dz^2} \right)_{(0)} + \dots \quad (2.23)$$

As in the low speed case, let q_0 be the quantity of heat generated at the surface, per second; and $q(z)$ the heat flux at any depth z within the body. Then, by definition

$$\frac{d\theta}{dz} = - \frac{q(z)}{k} \quad (2.24)$$

Although, under non-steady state conditions, $q(z)$ is an unknown function of z and t , at the surface the flux $\frac{k d\theta}{dz}_0$ must equal the generated flux q_0 , whence substitution in Maclaurins expansion gives

FIG 2.6 SCHEMATIC SECTION THROUGH SURFACE, AT TEMPERATURE θ_0 DEFINING SUB SURFACE TEMPERATURE θ AT DEPTH z , $\theta - d\theta$ AT DEPTH $z + dz$.

$$\theta(z) = \theta_o - \frac{z q_o}{k} + \frac{z^2}{2} \left(\frac{d^2 \theta}{dz^2} \right)_{(o)} + \dots \quad (2.25)$$

It may be noted that the first two terms of the expansion are those of a body with surface flux q_o , and surface temperature θ_o , under steady state conduction. This implies that the remaining terms are corrections for the heat absorbed by the body under non-steady state conditions.

Let us consider the rate of heat absorption by a layer dz thick inside the body, shown diagrammatically in figure 2.6.

$$\frac{\partial q}{\partial z} = \rho c \frac{\partial \theta}{\partial t} \quad (2.26a)$$

Rearranging, and substitution of equation 2.22 gives

$$\frac{\partial q}{\partial z} = \rho c \frac{\partial \theta}{\partial t} = k \frac{\partial^2 \theta}{\partial z^2} \quad (2.26b)$$

Thus, the third term of Maclaurins expansion $\left(\frac{d^2 \theta}{dz^2} \right)_{(o)}$ represents the rate of heat absorption at the surface. Hence, the first three terms of the series together represent the temperature at any point in a body which is supplied with heat at a rate q_o - (heat absorbed at the surface).

Carslaw and Jaeger (1947) give, for the surface temperature

$$\theta_o = \frac{2q_o t^{\frac{1}{2}}}{(\pi \kappa \rho c)^{\frac{1}{2}}} \quad (2.6)$$

Substitution of equation 2.6 into equations 2.26b, 2.24 & 2.25 and rearranging gives

$$\theta(z) = \theta_o \left\{ 1 - \frac{z}{2\sqrt{\kappa t}} + \frac{z^2}{4\kappa t} + \dots \right\} \quad (2.27)$$

The maximum effect of the heat generated at the surface reaches a depth z after a time

$$T_3 = z^2/2\kappa$$

(Archard (1959)) whereas for any point below the surface the heat has been applied for a time.

$$T_4 = t = x/V$$

Thus a measure of whether a point below the surface is significantly heated is given by

$$M = \frac{T_3}{T_4} = \frac{z^2}{2\kappa t} \quad (2.28)$$

where a small value of M indicates significant heating. Thus, the expression for sub-surface temperature becomes

$$\theta(z) = \theta_o \left\{ 1 - \sqrt{\frac{M\pi}{2}} + \frac{M}{2} + \dots \right\} \quad (2.29)$$

b. Mathematical Justification

For an infinite body, Ingersoll and Zobel (1955) give as a solution to equation 2.22.

$$\theta(z) = \frac{q_o z}{2\kappa\sqrt{\pi}} \int_{\frac{M}{2}}^{\infty} \frac{e^{-\beta^2}}{\beta^2} d\beta \quad (2.30)$$

Where $M = \frac{z^2}{2\kappa t}$; $\beta = \frac{z}{2\sqrt{\kappa(t-T)}}$ and

T is a dummy variable for t, such that the integral is, at a time t, summing all past history from zero time to the time t. Upon integration, equation 2.30 becomes

$$\theta(z) = \frac{q_o z}{2\kappa\sqrt{\pi}} \left\{ \frac{1}{\sqrt{M/2}} e^{-M/2} - \sqrt{\pi} (1 - \text{erf } \sqrt{M/2}) \right\} \quad (2.31)$$

Modifying equation 2.31 to that for a semi-infinite body where heat is constrained to move in only the +ve z direction and substituting for θ_o from equation 2.6 the expression for sub-surface temperature becomes

$$\theta(z) = \theta_o \left\{ e^{-M/2} \sqrt{\frac{M\pi}{2}} (1 - \operatorname{erf} \sqrt{M/2}) \right\} \quad (2.32)$$

For small values of M , $e^{-M/2}$ and $\operatorname{erf} \sqrt{M/2}$ can be expanded as a power series to give

$$\theta(z) = \theta_o \left\{ 1 - \sqrt{M\pi/2} + M/2 - \frac{M^2}{4!} + \dots \right\} \quad (2.33)$$

As shown above, small values of M are valid for a situation in which significant heating takes place i.e., either shallow depths or long values of time. Table 2.2 shows a comparison between equation 2.29 and the full equation 2.33, tabulated as

$\theta_z/\theta_o \sqrt{\frac{M}{2}}$. For small values of $\sqrt{\frac{M}{2}}$, the agreement is excellent, reaching less than 2% divergence for a 64% drop in temperature at $\sqrt{\frac{M}{2}} = 0.5$.

TABLE 2.2

M	$\sqrt{M/2}$	θ_z/θ_o (eqn 2.29)	θ_z/θ_o (eqn 2.33)
0.9	0.671	0.261	0.242
0.7	0.592	0.302	0.290
0.5	0.5	0.364	0.358
0.4	0.447	0.415	0.411
0.3	0.387	0.464	0.462
0.2	0.316	0.589	0.588
0.15	0.274	0.590	0.539
0.07	0.187	0.703	0.703
0.03	0.122	0.798	0.798
0.01	0.071	0.880	0.880

For a representative position in the contact $x=a$ and $t=a/V$. Substituting in equation 2.28

$$M = \frac{z^2 V}{2Ka} = \frac{z^2}{a^2} \left(\frac{aV}{2K} \right) = \frac{z^2}{a^2} L \quad (2.34)$$

Through equations 2.33 and 2.34, the sub-surface temperature θ_z/θ_o may now be determined at high speeds for sliding for various values of z/a .

The combined high and low speed plot of sub-surface temperatures is presented in Figure 2.7. It shows the values of the depth z (expressed as a ratio of the contact radius a) at which various values of the ratio θ_z/θ_o are attained as a function of Jaeger's parameter L . As in the earlier paper (Archard (1959)) the intermediate regions, shown by broken lines, are obtained by curve fitting.

2.7 Temperature distribution within the contact

Section 2.5.1 outlined Archard's (1959) flash temperature theory where equations are derived to calculate the mean rise in temperature over the contact region θ_m . Equations were also given to calculate the maximum obtainable flash temperature i.e. when the total load is borne by plastic deformation at a single area. However the latter calculation still represents a mean rise in temperature over the contact region, but under the worst possible contact conditions. This apparent anomaly arises primarily from equation 2.6 where an appropriate value of the time t for which the heat has been applied to any point within the circular contact is required.

DIRECTION OF VELOCITY OF HEAT SOURCE

FIG. 2.8 SECTION THROUGH CIRCULAR CONTACT AREA
AT HIGH SLIDING SPEEDS DEFINING TERMS.

In the following analysis attention will be confined to a fast moving heat source since this will be applicable to the experiments described later. At high speeds of sliding ($L \gg 10$, say), the vast proportion of heat generated at the interface enters the body which "sees" the moving heat source (Body C in Figure 2.3). It is, therefore, the calculation of this temperature which, under these conditions, effectively determines the flash temperature

To use equation 2.6 in the theory of flash temperatures, the effective value of $t^{1/2}$ for any point in a circular contact is required, since $\theta_c \propto t^{1/2}$. A simple estimate would be to assume a representative time $t = a/V$, therefore

$$t^{1/2} = (a/V)^{1/2} \quad (2.35)$$

Alternatively, the maximum time of heating for a point in a circular contact will be $t = 2a/V$ or

$$t^{1/2} = 1.414 (a/V)^{1/2} \quad (2.36)$$

Substitution of equation 2.36 into equation 2.6 will give the maximum possible temperature rise in the contact.

To determine the effective value of $t^{1/2}$ for all points within the contact, to give the mean temperature rise in the contact, consider Figure 2.8 which represents the area of contact at high sliding speeds at one moment of time. Consider element of width dx , length c (equivalent to the

maximum temperature of that element) and length y (equivalent to the mean temperature of that element).

The average value of $y^{1/2}$ over the contact

$$= \frac{\int_{-a}^{+a} \int_0^c y^{1/2} dy dx}{\int_{-a}^{+a} \int_0^c dy dx} \quad (2.37)$$

Integrating once

$$y^{1/2} = \frac{1}{\pi a^2} \int_{-a}^{+a} \frac{2c}{3}^{3/2} dx$$

From Figure 2.8 $c = 2a \cos \phi$, $x = a \sin \phi$

$$dx = a \cos \phi d\phi$$

and substituting

$$y^{1/2} = \frac{4\sqrt{2}a^{1/2}}{3\pi} \int_{-\pi/2}^{+\pi/2} \cos^{5/2} \phi d\phi \quad (2.38)$$

$$= \frac{8\sqrt{2}a^{1/2}}{3\pi} I$$

$$\text{Where } I = \int_0^{\pi/2} \cos^{5/2} \phi d\phi$$

From standard tables

$$\int_0^{\pi/2} \cos^n x dx = \frac{1}{2} \sqrt{\pi} \frac{\Gamma(\frac{n+1}{2})}{\Gamma(\frac{n}{2}+1)} \quad (2.39)$$

Where Γ represents a gamma function

$$n = 5/2$$

$$I = \frac{1}{2} \sqrt{\pi} \frac{\Gamma(1.75)}{\Gamma(2.25)}$$

Noting that $\Gamma(n+1) = n \Gamma n$

$$I = \frac{\frac{1}{2} \sqrt{\pi}}{1.25} \left(\frac{0.91906}{0.9064} \right)$$

$$\text{Then } y^{1/2} = \frac{16\sqrt{2}a^{1/2}}{15\sqrt{\pi}} \left(\frac{0.91906}{0.9064} \right)$$

$$y^{1/2} = 0.863a^{1/2}$$

$$\text{or } t^{1/2} = 0.863 (a/V)^{1/2} \quad (2.40)$$

This is the effective value of $t^{1/2}$ used by Archard (1959) to give the mean temperature rise (e.g. θ_m of equations 2.7 & 2.10). It may be noted in passing, that for high speed conditions Holm's (1952) value for $t^{1/2} = 1.25 (a/V)^{1/2}$.

FIG 2.9 ISOTHERMS FOR SURFACE TEMPERATURES WITHIN A CIRCULAR HEAT SOURCE UNDER HIGH SPEED CONDITIONS. θ_m IS THE MEAN SURFACE TEMPERATURE OVER THE WHOLE CIRCULAR AREA.

Dividing equation 2.36 by equation 2.40 the result is obtained that for a circular sliding contact

$$\theta_{\max} = 1.64 \theta_m \quad (2.41)$$

Using this result it is possible to determine the temperature distribution within a circular heat source under high speed conditions.

$$\begin{aligned} \text{The mean value of } y^{1/2} \text{ (equivalent to } \theta_m) &= 0.863 a^{1/2} \\ &= (0.745a)^{1/2} \end{aligned}$$

$$\begin{aligned} \text{A temperature } (\alpha\theta_m) \text{ corresponds to } y^{1/2} &= (\alpha^2 0.745a)^{1/2} \\ \text{or } y/a &= 0.745\alpha^2 \quad (2.42) \end{aligned}$$

Table 2.3 shows equation 2.42 evaluated for various values of α .

Table 2.3

$\alpha = \theta/\theta_m$	y/a
1	0.745
1.1	0.901
1.2	1.073
1.3	1.258
1.4	1.460
1.5	1.676
1.6	1.906
1.64 ($\theta_{\max}$)	2

The results of Table 2.3 are plotted in Figure 2.9 as isotherms for surface temperature within a circular contact. The essential and important conclusion from this result is that for a circular contact, at high speed conditions, (as for example between specimens in our crossed

cylinders machine experiments) the vast proportion of the contact will, during the duration of the contact, experience temperatures in excess of the mean calculated temperature, up to a maximum of θ_{max} . Only a small proportion of the contact will experience temperatures less than θ_m (shaded regions in figure 2.9). To what extent temperatures in excess of θ_m can influence changes at the rubbing surface will be examined in subsequent chapters.

2.8 Contact theory for point contacts

It has been mentioned earlier that a further variable in flash temperature theory is the estimate of contact size. This section will briefly review contact theory for point contacts as applicable to the subsequent experiments of this thesis.

Under a light load, W , when two bodies are pressed in stationary contact the deformation will be entirely elastic. The bodies will touch over a defined region. The contact region between crossed cylinders, of equal radii, is circular and Hertz (see Timoshenko and Goodier (1970)) has derived the following equation where the radius, a , of a circular contact is given by:

$$a = 1.11 \left(\frac{WR'}{E} \right)^{1/3} \quad (2.43)$$

In this equation $\frac{1}{E'} = \frac{1}{E_1} + \frac{1}{E_2}$, $E'_1 = \frac{E_1}{1-\nu_1^2}$, $E'_2 = \frac{E_2}{1-\nu_2^2}$

$$\text{and } \frac{1}{R'} = \frac{1}{R_1} + \frac{1}{R_2}$$

Suffices 1 & 2 refer to the two bodies in contact, E is Young's modulus, ν is Poisson's ratio. R_1 & R_2 are the radii of curvature of the two bodies.

Under very heavy loads, plastic deformation will be dominant and the radius of contact will be given, to a good approximation, by

$$a = \left(\frac{W}{\pi P_m} \right)^{\frac{1}{2}} \quad (2.44)$$

P_m is the flow pressure and equal to the hardness, H, of the softer material.

Consider now a simplified account of the intermediate range of loads when the deformation is a mixture of elastic and plastic deformation. This intermediate range will be centred on a critical load W_2 at which both sets of equations would (if they applied) give the same value of contact size. Equating values of a from equations 2.43 & 2.44, gives the value of the critical load W_2 as:

$$W_2 = 69.8 H^3 \left(\frac{R'}{E'} \right)^2 \quad (2.45)$$

The load corresponding to the end of the purely elastic deformation, W_1 , can, in principle, be quite clearly defined. It is the load at which the maximum shear stresses in the stress distribution beneath the contact exceeds the elastic limit. A full analysis combined with the relationship

$$H \approx 2.7Y \quad (2.46)$$

where Y is the yield stress in uniaxial tension, shows that the onset of plastic flow occurs when the contact pressure has a value of $0.4H$. Since over the elastic range $p_m \propto W^{1/3}$, this leads to the convenient approximate relationship.

$$W_2 \approx 15W_1 \quad (2.47)$$

A full analysis of indentation experiments (Tabor (1950)) indicates that a convenient definition of the "full plasticity" range is that it occurs at loads in excess of approximately 200 times at which the onset of plastic flow occurs ($W_3 \gg 200 W_1$). W_3 , therefore, marks the end of the intermediate range and onset of fully plastic conditions and

$$W_3 \approx 15 W_2 \quad (2.48)$$

Using this approach it is possible to provide a simple statement of the theory of point contacts which includes the difficult intermediate

FIG. 2.10 CONDITIONS OF CONTACT AS A FUNCTION OF NORMAL LOAD FOR A HARDENED TOOL STEEL, T1, IN CONTACT WITH A PLAIN CARBON STEEL, C73.

elastic/plastic region. Curve fitting combined with the criteria of equations 2.47 & 2.48 allows a plot of sufficient accuracy for most purposes. Figure 2.10 has been drawn using this method and shows the variation of contact size between crossed cylinders of a hardened tool steel T1 and a plain carbon steel, C73, as a function of load. The significance of this combination as an illustration will be explained in a subsequent chapter. The dimensions and physical properties used may be found in Chapter 7. For completeness the variation of mean contact pressure is also included in figure 2.10.

A feature of figure 2.10 is that at loads between W_1 & W_3 i.e. the intermediate elasto-plastic regime, the estimate of contact size (solid line) using this approach is greater than if either elastic or plastic conditions are assumed (dashed lines).

Figure 2.10 can be expected to give the correct overall dimensions and contact pressures for contact between crossed cylinders i.e. nominal point contact. When two flat surfaces are in contact the situation is more complex. Individual asperity contacts probably behave like the crossed cylinders so that each contact deforms elastically when it carries a small load and plastically when it carries a high one. As the total load increases the surfaces will move together and the number of asperity contacts increases. The new contacts balance the growth of existing contacts such that the average size of contact is unchanged. The number of contacts is roughly proportional to load and the fraction of load carried by elastic contacts will not change, even though the original contacts will have become plastic. Contact between flat surfaces is therefore elastic or plastic depending on the surface

geometry and material properties but does not change from elastic at low loads to plastic at high ones.

The criterion proposed was the plasticity index $\Omega = \frac{E}{H} \left(\frac{\sigma}{\beta} \right)^{1/3}$ where σ is the standard deviation of asperity heights and β the mean asperity radius of curvature. Values < 1 indicate mainly elastic contacts, values > 3 indicate mainly plastic contacts.

Studies of stresses beneath the contact between elastic spheres or cylinders (Hamilton and Goodman (1966)) show that, in the absence of tangential shear stresses due to sliding the maximum shear stress lies below the surface. For a circular contact the depth of this maximum shear stress $z = \frac{1}{2}a$. Sliding shifts the position of the maximum shear stress closer to the surface. Provided that the coefficient of friction $\mu \ll 0.25$ it does not influence its magnitude. When $\mu > 0.25$ the maximum shear stress increases with μ and it now occurs at the surface. Since the onset of plastic flow occurs at the position of maximum shear stress it can be expected that the application of frictional tractions will cause plastic deformation to take place more readily. A more detailed analysis of the effect of surface topography and shear stress distribution on contact mechanics can be found in work by Childs (1981).

2.9 Discussion

The origins of the present work lie in the initial observations of Welsh (1957) and subsequent work by Archard (1959) and Welsh (1965) of the dry wear of steels. Using pin on ring apparatus it was concluded that if the

frictional flash temperatures were sufficiently high (say $> 750^{\circ}\text{C}$) then a phase transformed material, akin to the martensitic structure of conventional ferrous metallurgy appeared on the wearing surfaces. The properties of this material were similar to those of the so-called white layers. Therefore before more detailed studies of phase transformations in the rubbing of steels are performed it has been appropriate, first, to review the appearance, properties and theories of formation of white layers.

The extent of the reported appearances of white layers is in some way reflected by the length of this selective review. From the diversity of instances in which white layers have been found it would be remarkable if this material was of single unique form or structure. This review suggests that temperature, deformation (both magnitude and rate) and environment are major variables in each case of white layer formation. Other factors may include the material and its initial structure, for instance white layers formed on cast irons have a rather different appearance and have been reported to be more resistant to tempering than those found under similar conditions on plain or low alloy carbon steels. A tendency in some earlier reviews to consider white layers as a single unique phase is questionable and has, perhaps, led to some of the controversy. However while certain specific and superficial differences may occur depending on the precise conditions of formation, most white layers do appear to have basic similarities.

Probably the major argument in white layers is whether their formation is principally temperature dependent and therefore martensitic in form or due to deformation and having essentially a refined or distorted

structure. It has been proposed in this review that many of the factors used to discount the martensitic theory are, on closer examination, unsubstantial. The crucial question appears to be whether one would reasonably expect the layers formed under the special conditions of rubbing contact (e.g. transient heating and some concomitant deformation) to be similar in appearance, hardness, tempering response etc. to martensite formed by conventional metallurgical methods. In this author's view it would be surprising if the answer was affirmative.

The second problem has been the measurement or calculation of the surface and sub-surface temperature to account for this transformation. In all the papers reviewed surface temperature estimates have considered the crucial temperature to be 723°C i.e. the AC_1 temperature from the iron-carbon phase equilibrium diagram. This diagram is strictly applicable for time intervals of $> 10^3$ seconds, yet transformations at the wearing surface occur at times of $\approx 10^{-3}$ seconds. Generally measurement of surface temperature gives, at best, the bulk interfacial temperature. These values have been considerably less than 723°C . However infra-red techniques are capable of measuring the transient flash temperatures, even if under rather limited material combinations. These techniques as for example used by Parker and Marshall (1948) have given unequivocal evidence of flash temperatures in excess of 723°C producing phase changes at the surfaces of railway tyres. Indeed these temperatures were measured as the hot spot on the tyre emerged from the brake block and some measure of cooling may have already occurred. Since, as yet, there appears to be no reliable method of measuring transient flash temperatures between rubbing steel surfaces theoretical flash temperature calculations have been examined.

Of the various flash temperature theories used perhaps the most common is Archard's (1959) because of its mathematical simplicity. This theory is typically used to illustrate whether temperatures developed during sliding are sufficient to cause a transformation. The theoretical predictions have been found to be in accord with comparative infra-red measurements and various physical changes at the surface. However this theory has been almost universally applied in a manner which invalidates some of its fundamental assumptions. In particular, since it cannot take into account changes in friction with load and sliding distance nor changes in surface hardness its conclusions for multiple rubbing experiments must be regarded with great reserve.

Whereas white layers formed as result of frictional heating consist of a layer of finite depth, to date the interpretation has been based on the average value of flash temperature at the contact surface. To remedy this defect a sub-surface flash temperature analysis has been included which allows the estimation of the depth of material below the surface which has been subjected to a given temperature rise. It has also been shown that Archard's flash temperature calculations correspond to the mean flash temperature. A temperature profile exists over the contact area and the maximum flash temperature is 1.64 times the mean. The effect of these factors will be considered in subsequent chapters.

If, therefore, theoretical flash temperature calculations are to be used in the study of the formation of phase transformed material, the first requirement is to choose an experimental configuration which will allow phase transformations to occur in a single well defined rub. Multiple rubbing would be a useful link between such experiments and those almost

universally found in the literature. The machine chosen is therefore a crossed cylinders machine. The following chapter concerns a description of that machine and an explanation of how single and multiple rubbing experiments may be performed.

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**Fig 3.1 Detail of Crossed
Cylinders machine.**

CHAPTER 3

THE CROSSED CYLINDERS MACHINE

3.1 Introduction

In an attempt to find phase transformed material produced as a result of one passage of the contact under well defined conditions, a crossed cylinders machine has been used. This chapter concerns a description of the crossed cylinders machine and the procedure followed in a typical experiment.

3.2 The crossed cylinders machine

The essential parts of the apparatus are shown in Figure 3.1.

The two rotating specimens are at A and B in Figure 3.1. In this diagram solid cylinders are illustrated, hollow specimens with internal circulation of oil for temperature control or specimens mounted on mandrels can also be used. As well as rotating, the lower specimen, A, can be traversed, either repeatedly or in a single pass, in a direction at 45° to the specimen axis so that the point of contact traces out a helical path on both rotating specimens. As the two rotations and the traverse are not synchronous, all parts of the specimens are subject to equal wear and thus they retain their cylindrical shape.

To achieve this traverse specimen A is mounted on a carriage C which can be moved forwards or backwards or both, by a hydraulic ram D, whose speed

is variable up to a maximum speed of 0.15 m/s. To study surface transformations occurring in a single rub, no overlapping of the helical single contact track, must be ensured. This is achieved by adjusting the speed of traverse for each rotational speed. The length of the traverse is controlled by limit switches and thus an experiment can be performed on any selected portion of the specimens. The rotation of the lower specimen is effected through a telescopic shaft E and Hardy Spicer flexible couplings. The lower specimen assembly is insulated from the body of the machine so that, if required, the electrical capacity and electrical resistance between the specimens can be measured. In the experiments reported in this thesis, these measurements were not performed. Electrical connections to embedded thermocouples in the specimens are made by mercury contacts J, of a type described by Kenyon (1955).

The upper specimen B, is mounted on an arm G which is pivoted at one end of a gimbal mounting F. The rotating drive is coupled to this specimen by a flexible coupling situated at the intersection of the gimbal axes. A load W is applied to the end of the upper specimen arm and the load between the specimens can vary from a maximum of $10W$ to a minimum of $10W/3$ over the usable portions of the specimens.

In all of the experiments described in this thesis only a short length of specimen was used and therefore the contact load can be considered to be effectively constant.

The friction generated between the specimens tends to rotate the upper specimen about its vertical axis and this movement is restrained by a

rigid vertical rod H. Archard (1958) has shown that the character of surface damage and recorded friction can be greatly affected by the type of restraint used. With a highly elastic restraint the recorded friction varied between wide limits and large relative movement of the specimens occurred during periods of falling friction. More rigid constraints gave more consistent results. Rods of various diameter are available depending on the friction range required. The friction is recorded by measuring the deflection of this rod, usually by means of a Talymin probe connected to a Talysurf 3 recorder. In the course of the experiments of this thesis the rod deflections were also measured using a Talysurf head and a non contacting eddy current device having a very high frequency response (up to 50 kHz). The measured force is a resolved component of the frictional force and it depends upon the ratio of the velocities (V_1 & V_2) of the lower and upper specimens. Calibration of the friction records is by horizontal loads applied at M.

The specimens have a maximum length of 0.13m and are mounted between running centres, which transmit the rotation. Specimen diameters up to 76 mm can be accommodated in the machine. The rotation of the two specimens are independently controlled and can be varied from 0.26 rad/s (2.5 rpm) to more than 314 rad/s (3000 rpm).

3.3 Experimental procedure

The general procedures followed for each experimental test were similar. Prior to the run the specimens were degreased by Soxhlet extraction for approximately 3 hrs, with a 50:50 solution of acetone and iso-propyl alcohol. This was to ensure absolute cleanliness and hence a reasonably

high and consistent value of friction coefficient μ . The specimens were then carefully mounted between centres on the machine. Keeping the specimens separated by means of a scissors jack, under the loading arm, the speeds of each specimen were adjusted to the required value and checked. The ram speed was set accordingly. The limits of ram traverse were adjusted and checked. Bulk specimen temperature and friction recording devices were switched on and checked. The required hanger load was suspended in an oil dashpot from the loading arm. The specimens were allowed to rotate to full speed and then loaded carefully together. After a brief period of repeating rubbing the ram was applied to produce a single well defined track on both rotating specimens followed by a brief period of repeated rubbing at the end of the track as the rotation of the specimens slowed down and stopped.

CHAPTER 4

EXPLORATORY EXPERIMENTS WITH SOFT STEELS ON SOFT STEELS

4.1 Introduction

This chapter concerns some exploratory experiments using the crossed cylinders machine to produce phase transformed material on the surfaces of soft steels, as a result of one passage of the contact under well defined conditions. To further simplify the experimental variables which could affect the formation of phase transformed material, attention has been confined to the unlubricated rubbing of plain carbon eutectoid steels.

The results of this chapter includes experimental work performed by Miss J O'Kelly, Mr N Green and Miss M Feetham in the course of final year undergraduate projects at the University of Leicester. I gratefully acknowledge the use of their results.

4.2 Exploratory experiments

Experiments were performed with specimens manufactured from a BISRA plain carbon steel of the following composition

	C	Si	Mn	S	P
Wt%	0.78	0.18	0.63	0.044	0.035

supplied in normalised bar form with an as received hardness of 275 HV 20kg and grain size of 6-7 ASTM. Some specimens were hardened and

tempered to values of 300 and 400 HV 20kg since earlier experiments with this steel treated in this range of hardness showed a definite severe to mild transition (T_2) in the dry wear of steels (Welsh (1965)). The experimental procedure followed is as described in Chapter 3 and unless otherwise stated tests were performed with both specimens rotating at the same surface speed ($V_1 = V_2 = V_{sf}$). Peak values of friction coefficient μ of 1 or greater were measured during the experiments, both in single and multiple rub and a value of $\mu = 1.0$ was assumed. Welsh used an average value of $\mu = 0.9$ in his flash temperature calculations and ignored higher peak values recorded.

Calculations of the mean surface flash temperatures were made using the methods set out in Archard's (1959) paper, summarised in Chapter 2. In the present work when both specimens were rotating with the same speed, two changes were necessary in the 1959 paper. First the heat source moved with equal velocities (the surface speed V_{sf}) over the surface of both specimens so that the heat was divided equally. Second, because of the configuration of the machine the velocity which determines the rate of heat generation (the sliding velocity V_{sl}) is $\sqrt{2} V_{sf}$. Also a reduction of 10% in the calculated mean surface flash temperature was appropriate as the conditions of the experiments corresponded to the lower limits of applicability when using the high speed equations (i.e. when L equals approximately 5). This reduction is also appropriate in the flash temperature calculations when one body is stationary and L equals approximately 5. For both specimens rotating at the same speed and assuming the deformation is plastic, equation 13b from Archard's (1959) paper therefore becomes

<u>Run</u>	<u>W (N)</u>	<u>$V_{sf}(m/s)$</u>	<u>HV 20kg</u>	<u>θ_m (°C)</u>	<u>Single rub PTM</u>	<u>Multiple rub PTM</u>	<u>Remarks</u>
E1	201	0.8	275	1316	No	No	
E2	299	0.8	275	1453	No	No	
E3	299	0.8	275	1453	-	No	No single rub traverse
E4	299	0.8	275	1495	-	Yes	Upper specimen rotating for 1 minute with lower specimen stationary
E5	299	0.8	300	1552	No	No	
E6	299	0.8	300	1590	-	No	As run 4, duration of rotation 0.5 minutes
E7	299	0.8	300	1590	-	Yes	As run 4, duration of rotation 1 minute.
E8	299	0.8	300	1590	-	Yes	As run 4, duration of rotation 5 minutes
E9	299	0.8	300	1590	-	Yes	As run 4, duration of rotation 15 minutes

Table 4.1 Experimental conditions and results of the first set of exploratory experiments.

FIG 4.1 SPECIMENS USED IN RUNS E5-E9, SHOWING WEAR TRACKS AND CORRESPONDING SCARS

FIG 4.2 MICROSECTION THROUGH SPHERICAL WEAR SCAR IN RUN E9, SHOWING PHASE TRANSFORMED MATERIAL

$$\theta_m = \frac{\mu}{3.25} \frac{(\pi H)^{3/4}}{(k\rho c)^{1/2}} \frac{W^{1/4} V_{sf}^{1/2}}{\sqrt{2}} \quad (4.1)$$

With material properties, at room temperature, $k = 52 \text{ J/msK}$, $\rho = 7850 \text{ kg/m}^3$ and $c = 480 \text{ J/kg K}$ inserted this equation reduces to:

$$\theta_m = 434 W^{1/4} V_{sf}^{1/2} \text{ for } H=275 \text{ HV } 20 \text{ kg} \quad (4.2)$$

$$\theta_m = 463 W^{1/4} V_{sf}^{1/2} \text{ for } H=300 \text{ HV } 20 \text{ kg} \quad (4.3)$$

$$\theta_m = 575 W^{1/4} V_{sf}^{1/2} \text{ for } H=400 \text{ HV } 20 \text{ kg} \quad (4.4)$$

The results of the first set of exploratory experiments are summarised in Table 4.1. This shows the contact load W , the surface speed of the specimens V_{sf} , the Vickers specimen hardness HV , the calculated mean surface flash temperature θ_m (including a 10% reduction) and whether on metallographic sectioning phase transformed material was found in single or in multiple rub or both. As in these experiments the test procedure was altered, a brief description of any variation is also detailed. The upper and lower specimens used in runs E5-E9 are shown in Figure 4.1, the wear tracks and corresponding scars are clearly visible.

A microsection through the spherically worn crater of the stationary specimen from run E9 revealed phase transformed material of approximate depth $15 \mu\text{m}$ (Figure 4.2). The major conclusion from this set of experiments was that despite imposing conditions which should have produced mean surface flash temperatures well in excess of 750°C , no phase transformed material was found in single rub. However some

Run	<u>W (N)</u>	<u>V sf(m/s)</u>	<u>HV 20kg</u>	<u>μ_{av}</u>	<u>θ_m ($^{\circ}$C)</u>	Single rub PTM	Multiple rub PTM	<u>$\mu_{av} \times \theta_m$</u> $^{\circ}$ C
E10	167	0.79	300	0.31	1331	No	No	413
E11	343	0.79	400	0.64	1980	Diffuse	Yes	1267
E12	422	1.1	300	0.63	1982	Yes	Yes	1249
E13	167	0.79	400	0.56	1654	No	Diffuse	926
E14	432	0.92	400	0.56	2264	Yes	Yes	1267
E15	167	0.79	300	0.42	1331	No	No	559
E16	422	1	300	0.65	1943	Diffuse	Yes	1263

Table 4.2 Experimental conditions and results of the second set of exploratory experiments.

(a)

(b)

(c)

Fig 4.3 Structures produced by a single rub.

(a)

INCREASING SEVERITY
↓

(b)

(c)

Fig 4.4 Structures produced by multiple rubbing

evidence of transformed material was found on surfaces which had been subjected to repeated application of rubbing contacts.

After evidence from Shell Thornton Research Centre (private communication) suggesting that higher temperatures might be required, the next set of exploratory experiments used rubbing conditions of greater severity.

Table 4.2 presents the experimental conditions and results of the second set of exploratory experiments. Metallographic sections from the experiments (Figures 4.3 & 4.4) showed the existence of hard phase transformed material in both single and multiple rub regions. Under less severe conditions (lighter loads or lower speeds) transformed material appeared only in the region subjected to repeated rubbing. Under the most severe conditions transformed material also existed on the single rub track. Under an intermediate set of conditions transformed material existed under the track of a single rub, this material revealed a diffuse structure (Figure 4.3(b)). Microhardness measurements on the track which under metallographic examination, showed transformed material, gave hardness values of up to 890 HV 100g, which is consistent with the martensitic transformation.

It is convenient to interpret the appearance or non-appearance of phase transformed material in terms of the calculated mean flash temperature at the interface. On this basis one obtains the following results. When the flash temperatures reach 2000°C a structureless transformed material was found, when the flash temperatures reached 1900°C material having a diffuse structure was observed. At lower values of the calculated flash

<u>Run</u>	<u>W (N)</u>	<u>V_{sf}(m/s)</u>	<u>HV 20kg</u>	<u>μ_{av}</u>	<u>θ_m (°C)</u>	<u>Single rub PTM</u>	<u>Multiple rub PTM</u>	<u>μ_{av} x θ_m</u>	<u>°C</u>
E17	852	1.4	275	0.62	2496	Yes	Yes	1547	
E18	426	0.9	275	0.54	1683	Yes	Yes	909	
E19	175	1	275	0.89	1420	Diffuse	Yes	1263	
E20	166	0.66	275	0.62	1139	No	Diffuse	706	

Table 4.3 Experimental conditions and results of the third set of exploratory experiments.

(say 1600°C or less) no transformed material has, as yet, been detected. In multiple rub the equivalent calculated temperatures are 1900°C for structureless transformed material, approximately 1600°C for a diffuse structure and at temperatures of 1300°C or less, no transformed material has yet been detected in multiple rub.

The final set of exploratory experiments were performed with the steel in its as received (i.e. normalised) state. The experimental conditions and results are summarised in Table 4.3. Despite using the steel in a different condition i.e. lamellar pearlite as opposed to spherical carbide precipitates in ferrite, the magnitude of the calculated temperatures required for the formation of transformed material in both single and multiple rubs was repeated, though the corresponding values are rather lower. In terms of the mean calculated flash temperature, a structureless material was produced in single rub when temperatures of 1700°C were reached, a diffuse structure when temperatures of 1400°C were obtained and at temperatures of say, 1150°C or less no transformed material was detected. In multiple rub the corresponding required temperatures are 1400°C for transformed material and approximately 1100°C for a diffuse structure.

Also of note was that in some runs temper colours were found on the surfaces of the specimens, usually around the multiple rub tracks, inferring the generation of high bulk temperatures.

4.3 Discussion

Although some runs, in the first analysis, gave somewhat contradictory results (i.e. runs E1-E4 compared to runs E17-E20) details of which will be discussed later, the outstanding feature of these experiments is that, for the first time, metallurgically phase transformed material has been produced as a consequence of a single contact between rubbing surfaces. The conditions necessary for the production of transformed material have been shown to be much more severe than was implied by the earlier work of Archard (1959) and Welsh (1965). Some reasons for this difference will now be sought and some experimental problems will be discussed. We may also note that the melting temperature of a 0.8% plain carbon steel is $\sim 1500^{\circ}\text{C}$ and surface temperatures above this value are unlikely.

Firstly, the approximate nature of the flash temperature calculations should be emphasised. A temperature distribution exists over the area of contact and average values both of this temperature and of the time for which it exists have been used. Also contact size determined by plastic deformation has been assumed, although the contact loads used were well below those required for full plasticity. As such methods were also used by Archard and Welsh, such approximations are appropriate for the present exploratory experiments, however a more exact analysis is possible and will form part of the next stage. Secondly, a coefficient of friction μ of unity has been assumed in all calculations, but smaller and larger instantaneous values have been measured during experiments. Indeed a representative average value of friction coefficient μ_{av} for an experiment was generally less than unity (if peak values are ignored)

FIG 4.5 FRICTION TRACE FROM RUN E16

10mm

FIG 4.6 CYLINDRICAL SPECIMENS USED IN EXPLORATORY RUNS E10-E20 SHOWING WELDING SCARS BOTH IN AND EXTENDING BEYOND THE SINGLE AND MULTIPLE WEAR TRACKS

e.g. figure 4.5 (μ_{av} is given in Tables 4.2 & 4.3). If one now uses μ as measured in the exploratory runs of E10-E20, it now appears that calculated temperatures of 1250°C or greater are required for the production of phase transformed material for steel specimens in both as received and hardened and tempered conditions (one exception being run E18). It was also noticed that the coefficient of friction was generally higher in multiple rub than in single rub. Thus the validity of assuming $\mu=1$ in flash temperature calculations must be questioned. Ideally one would like to associate a given site on the specimen track with a given measured value of μ , however this presents considerable experimental problems.

Thirdly it should be emphasised that the calculated flash temperatures are those which should exist at the surface, whereas the transformed material is below the surface. Using the theory developed in Chapter 2 for sub-surface temperatures, it is easy to show that material 15 μ m below the surface which is a typical depth of transformed material in multiple rub, the temperature will be approximately 70% of the surface temperature. While at a depth of 4 μ m, typical of transformed material found in single rub, the temperature will have fallen to approximately 90% of its surface value. Thus the use of surface temperatures can overestimate the chances of producing phase transformed material. If one now applies this reduction to the minimum calculated flash temperature for the production of phase transformed material in single rub (using μ_{av}), temperatures of the order of 1100°C or more are still required, still far in excess of the values which one would expect from the earlier work.

FIG 4.7 CONTACT AREA BETWEEN CROSSED
CYLINDERS WITH EQUAL SURFACE VELOCITIES

Two further limitations of these results must also be pointed out.

Firstly, in general, phase transformed material occurred during weld formation and junction growth, a condition to which flash temperature theory is not strictly applicable. Figure 4.6 shows some cylindrical specimens used in runs E10-E20, where weld marks can be observed in and extending beyond both the single and multiple rub tracks, at an angle of approximately 45° to the tracks. However these soft steel on soft steel experiments are relevant because they apply to conditions below or at the T_2 transition of Welsh.

Thus the contradictory results of runs E1-E4 as compared to runs E17-E20 may be a possible consequence of the effects of peak values of μ or weld formation or both. Indeed the question may be posed:- is welding a prerequisite for the formation of phase transformed material? Secondly experiments performed with both specimens rotating at the same speed cannot be strictly interpreted in terms of simple flash temperature theory even if welding does not occur because the kinematics of the experiments make it difficult to meet the essential requirement that the interface temperatures of both specimens are equal. Consider points on the line OY of Figure 4.7 which represents the area of contact at one moment in time. At O (when $V_1 = V_2 = V_{sf}$) both surfaces have passed through the same time/temperature history (assuming high speed flash temperature theory applies) with $t = a/V_{sf}$. As one moves from O to Y the surface of the upper specimen has passed through a longer period of heat supply than the corresponding point on the lower specimen. To compensate for this as one moves from O to Y a greater proportion of the heat generation at that particular point in space and time will go to the lower specimen, rather than the upper specimen. In the limit at Y the

upper specimen will be, in fact, giving up accumulated heat to the lower specimen. However at this exploratory stage it is important to recognise that with the situation $V_1 = V_2 = V_{sf}$ the system is symmetric and on average, over the whole contact region the division of the total heat generated is equal between the two bodies, this being the essential criterion when considering mean flash temperatures. Also of note is that one could move closer to the simpler situation of Archard's paper by performing experiments with $V_1 \gg V_2$, where the vast proportion of heat goes into the faster moving body.

In earlier work (e.g. Archard (1959), Welsh (1965)) it was deduced that flash temperatures of the order of 750°C were required to produce metallurgically phase transformed material; this broadly corresponds to what might be expected from the application of the concepts of conventional ferrous metallurgy. Despite the problems of the interpretation of the results, which have been discussed above, a major conclusion to be drawn from the exploratory experiments is that flash temperatures significantly greater than these values are necessary for the production of phase transformed material in a single rub. Therefore before proceeding to a description of some more refined experiments it will be appropriate, in Chapter 5, to reconsider the metallurgical arguments. In particular it is necessary to consider the metallurgy of small volumes of material subject to a combination of high temperatures and pressures for short periods of the order of 1 millisecond or less; this is the typical duration of a contact in the crossed cylinders experiments. Related to this (also suggested from the exploratory experiments) is that the formation of phase transformed material can be influenced by the initial state of the steel. We shall see to what

extent the new metallurgical theory which thus emerges provides a justification for the requirement of these higher temperatures which have been deduced from the exploratory experiments.

4.4 Conclusions

1. Hard, phase transformed material appears as a result of a single rub only under conditions so severe that the calculated flash temperatures have values of the order of 1100°C or more i.e. far in excess of values implied by the earlier work of Archard and Welsh.
2. It is suggested that this discrepancy may arise from the special nature of tribological conditions i.e. very short duration of high flash temperatures and pressures.
3. More exact experiments are required to confirm these limited results i.e. the avoidance of welding in single and multiple rub and the alteration of the motions of specimens of the crossed cylinders machine so as to provide conditions closer to those assumed in Archard's flash temperature theory.
4. It would be useful to measure bulk temperatures in subsequent experiments.

FIG 5.1 IRON - CARBON PHASE EQUILIBRIUM DIAGRAM

CHAPTER 5

METALLURGICAL CONSIDERATIONS

5.1 Introduction

In Chapter 4 it was suggested that the magnitude of the required flash temperatures (say, 1100°C or greater) to produce phase transformed material in a single rub on a crossed cylinders machine, was a consequence of the special nature of tribological conditions occurring. This chapter provides a reconsideration of conventional ferrous metallurgy applicable to small volumes of material subject to a combination of high temperatures and pressures for times of the order of 1 millisecond or less. The influence of high flash temperatures will primarily be considered, however the coincidental effect of high contact pressures and/or shear stresses cannot be ignored and will be discussed in the later sections of this chapter. However at this stage it must be emphasised that, as in the role of temperature, the transitory nature of the contact conditions cannot be disregarded. For instance, because the equilibrium phase diagram is influenced significantly by pressures of the order of those occurring in asperity contacts, it does not automatically follow that the same effects occur within the short duration of such contacts. The essential arguments of this and the following chapter are based on a paper by Archard and Rowntree (1979).

Figure 5.1 shows the well know iron-carbon phase equilibrium diagram which will form the basis of the discussion which follows. As is also well known the phase boundaries are influenced by the addition of other

elements, sometimes even in relatively small concentrations. For simplicity, in what follows, the diagram presented here will be used even though the arguments will need modification for any but the most strictly plain carbon steels. In particular the α/γ transformation in pure iron will be taken at 910°C and the eutectoid at 723°C and at a concentration of 0.8% carbon.

5.2 The range 723-910°C

Consider first a eutectoid steel in the normalised condition. In normal metallurgical heat treatment when raised above 723°C it is austenitised according to the change:-

If the steel is raised to a temperature in the range 723-910°C the change represented by equation 5.1 does not occur automatically or instantaneously as one might infer from the phase diagram; there are a number of reasons for this. Firstly, as the phase diagram indicates, the allotropic α/γ transformation occurs in pure α at 910°C . Secondly, dissolution of iron carbide requires solution of excess carbon in iron, which in the α phase has a maximum possible solubility of 0.025% carbon. It seems likely that in this temperature range the process of austenitisation of pearlite must proceed at the boundary between the original α and iron carbide phases either at the interlamellar surfaces or at the pearlitic grain boundaries, corresponding to the lowest free energy of transformation. It is questioned whether it would occur in the short duration of, say one millisecond or less, this being the typical

duration of a flash temperature associated with the single rubbing
exploratory experiments or an asperity contact.

The arguments of the last paragraph are supported by the results of some experiments by Hultgren (1929) who studied the gradual isothermal growth of austenite and the disappearance of ferrite in pearlitic steels. He found that part of the cementite, originally present in the steel, together with the ferrite, did not dissolve at the same time as the ferrite disappeared but could be partly or completely dissolved on longer holding. For instance, with a commercial eutectoid steel, he found that cementite was not completely dissolved until after 1 hour at 770°C. Baeyertz (1942) observed, in quenched, partially transformed plain carbon steels, "pearlitic ghosts" in regions of martensite (previously γ) together with fine pearlite. The pearlite ghosts were continuous with lamellae of the untransformed pearlite. The consensus of opinion in a summary by Bain and Vilella (1948) and in a review by Roberts and Mehl (1943) (on the mechanism of austenite formation) is that any ferrite-carbide interface is a potential site for the nucleation of austenite above the A_{c1} temperature.

More recently, Rose and Strassburg (1956) describe experiments in which specimens of normalised plain carbon steels containing 0.45% and 0.70% carbon were first raised to a temperature below the eutectoid temperature (723°C). The specimens were then rapidly raised to a higher temperature, above 723°C, and then held there for a fixed dwell time before rapid quenching. The progress of the austenitisation of the steels was followed by dilatometry during the high temperature phase and by

The diagram shows the compositions of the steel (originally in the normalised state) after being raised to temperatures in the range 723–910°C for various dwell times.

FIG. 5.2 THE AUSTENITISATION OF A 0.7% C STEEL (ROSE AND STRASSBURG (1956))

metallography of the specimens after quenching. The authors show that austenitisation of these hypo-eutectoid steels proceeds, first, by austenitisation of pearlite followed by austenitisation of the pro-eutectoid ferrite although these two processes overlap. In this work the stages of austenitisation are well defined although the metallurgical details of the processes are lacking.

Typical times required in the austenitisation of the 0.7%C plain carbon steels obtained by Rose and Strassburg are seen by reference to Figure 5.2 which is taken from their work. Thus at 750°C austenitisation can be seen as proceeding in three stages as follows:-

(i) Nothing occurs until a period of 10 seconds has elapsed.

(ii) Between 10 and 130 seconds the structure consists of pearlite plus ferrite plus austenite, the latter being produced mainly by the decomposition of pearlite.

(iii) Between 130 and 600 seconds the structure consists of ferrite plus austenite plus carbide, the latter being the remnants of the pearlite structure.

(iv) After 60 seconds the structure consists of austenite with some carbide remnants remaining; this structure persists for a very long time ($> 10^4$ seconds), the final stage being the dissolution of the remaining carbide and the formation of homogeneous austenite.

Again referring to Figure 5.2 we can see that raising the temperature in the 723-910°C range although shortening the time scale still results in relatively long times for the transformation . Thus at 850°C the stages of the transformation are as follows:-

- (i) Initiation after approximately 0.7 seconds (extrapolation of experimental results).
- (ii) Dissolution of last pearlite after 2.5 seconds.
- (iii) Final dissolution of ferrite after 6 seconds.
- (iv) Formation of homogeneous austenite after a much longer period of between two and three thousand seconds.

The experimental results of Rose and Strassburg therefore confirm the general argument advanced earlier. A typical flash temperature in the range 723-910°C is unlikely to produce significant metallurgical phase changes at asperity contacts.

Further support for a consideration of austenite formation in terms of the temperature ranges used here is provided by Hillert, Nilsson and Torndahl (1971) who have studied the effect of alloying element on the formation of austenite and dissolution of cementite. They suggest five different mechanisms for the dissolution of cementite for a steel in a soft annealed condition, i.e. spherical carbide precipitates. They do, however, consider the formation of austenite from lamellar pearlite and arrive basically at two types of cementite reaction. In their

terminology, type III/IV is, in essence the mode of austenite formation in the temperature range 723-910°C, while type V is the mode of austenite formation in the temperature range >910°C.

5.3 The range > 910°C

Above 910°C the situation is significantly changed. One can now see the austenitisation of a normalised eutectoid steel as consisting of two stages:-

(i) The allotropic transformation $\alpha \rightarrow \gamma$ and

(ii) The dissolution of the carbide with the surplus carbon diffusing into the γ phase (previously α); i.e. equation (5.1) now becomes γ (0.025% C) + Fe_3C (6.67% C) $\rightarrow$ γ (0.8% C) (5.2)

The allotropic transformation of the first stage has a relatively low activation energy. It will be rate dependent (Boedtker and Duwez (1962) but since the second stage involves far larger atomic movements, it seems likely that the overall rate determining process will be the diffusion of carbon in γ iron.

The arguments advanced here have been confirmed by the recent work of Orlich (1974), who studied the processes of austenitisation of a wide range of plain carbon and alloy steels, by rapid electrical impulse heating. This allowed heating rates of up to 2,400°C/s to be obtained and the process of austenitisation was followed by dilatometry during the high temperature phase and metallography of air quenched specimens.

FIG. 5.3 TIME - TEMPERATURE - AUSTENITISATION
CURVES FOR A 0.7% C STEEL (Orlich (1974))

Although these results were expressed in terms of heating rates as compared to isothermal heating times of Rose and Strassburg, an almost identical pattern of results was obtained (e.g. Figure 5.3 taken from Orlich for a 0.7% plain carbon steel). In the range 0.15-0.70% carbon steels, the start of the Ac_3 transformation, beyond a critical rate of heating, did not exceed 910°C . Indeed identical phenomena were produced for a number of alloy steels, but at different temperatures. However one discrepancy between this work and that of Rose and Strassburg was that the Ac_1 line was observed, above a critical heating rate, to be a constant at approximately 800°C for carbon steels. Orlich plotted this line in dotted form as he could not provide an explanation for this level portion. Moreover it may be incorrect to compare "normal" rapid thermal heating results against rapid electrical or impulse heating on account of the different component structures i.e. ferrite, carbide etc. having differing electrical conductivities thus being heated up at different rates. Osborn (1941) established that in induction heating of pearlitic steels an accelerated dissolution of carbides takes place. He suggested that because of the greater electrical resistance of the carbides, they were at a higher temperature and thus the carbides went into solution at a faster rate. Roth (1963) observed in carbon steels that with resistance heating the carbides broke up in one fifth of the time compared to normal heating in a salt bath. His conclusion was identical to that of Osborn. Therefore, for the present initial analysis, this discrepancy has been ignored and the formation of austenite will be considered to occur either in the temperature range $723-910^{\circ}\text{C}$ or $>910^{\circ}\text{C}$. Indeed further evidence is provided by Speich and Szirmai (1969) for the rapidity of the first stage of the transformation at temperatures $>910^{\circ}\text{C}$. They observed, in the formation of austenite from zone refined ferrite,

that complete transformation occurred in laser heated zones (approximate heating rate of 10^6 °C/s) even in those specimens austenitised for only 10 milliseconds at 950°C. In terms of the present work it is unfortunate that evidence does not exist for still shorter dwell times.

The independence of the start of the Ac_3 transformation temperature is inconsistent with the theory of nucleation and growth. It can therefore be assumed that the operative mechanism does not involve long range diffusion. Two mechanisms of this type are the shear or martensitic transformation and the massive transformation. It is beyond the scope of this thesis to attempt to define the exact crystallographic mode of the α/γ allotropic transformation of iron. However suffice it to note that similarities between the allotropic α/γ transformation in iron and shear transformations have been observed by Eichen and Spretnak (1959) and by Zerwekh and Wayman (1965). Though Guy and Hren (1974) consider the α/γ change to be a massive transformation.

Therefore, if the probable rate determining process of the transformation at temperatures $> 910^\circ\text{C}$ is the diffusion of carbon in γ iron, it is appropriate, first, to make an order of magnitude calculation.

Consider pearlite subjected, for a time t , to an elevated temperature (θ) for which the diffusion coefficient, D , will have a significantly higher value than that which exists at the ambient temperature. A representative diffusion distance, $\bar{x}$, through which carbon atoms will diffuse in the time t is given by:-

$$\bar{x} = \sqrt{Dt} \quad (5.3)$$

Taking, from the literature (Van Vlack (1973)) values of D for carbon diffusion in γ iron and assuming that $t \approx 0.5$ milliseconds (c.f. exploratory single rub experiments of Chapter 4 where $t \approx 0.2$ milliseconds) one obtains the results shown in Table 5.1.

Table 5.1 Diffusion of carbon in γ iron for $t = 0.5$ milliseconds

$\theta^\circ\text{C}$	1000	1100	1200	1300	1400
$D \text{ (m}^2/\text{s)}$	3×10^{-11}	8×10^{-11}	1.9×10^{-10}	4×10^{-10}	7.8×10^{-10}
$x \text{ (}\mu\text{m)}$	0.12	0.20	0.31	0.45	0.62

The production of homogeneous austenite from pearlite requires diffusion over distances of the order of $0.5 \mu\text{m}$. Table 5.1 therefore shows that even at temperatures in excess of 910°C complete austenitisation of pearlite requires times longer than the value of 0.5 milliseconds, typical of flash temperatures. Alternatively a flash temperature of 0.5 milliseconds duration would require to have a value of the order of 1400°C to ensure complete austenitisation of the surface layers in the contact region. This is in closer accord with the results of the exploratory single rub experiments of Chapter 4.

5.4 Carbon diffusion theory for lamellar pearlite

Consider for simplicity the transformation of a normalised eutectoid steel (i.e. pearlite). The material exists in a laminar form and we require a solution of the diffusion equation (Ficks 2nd law)

FIG. 5.4 CARBON CONCENTRATION IN LAMELLAR PEARLITE

FIG 5.5 CARBON CONCENTRATION IN AUSTENITE (PREVIOUSLY PEARLITE) PLOTTED AS A FUNCTION OF DISTANCE (x/l) FOR VARIOUS VALUES OF Dt/l^2

$$\frac{\partial C}{\partial t} = D \frac{\partial^2 C}{\partial x^2} \quad (5.4)$$

where C is the concentration of carbon at any value of x and t , the distribution of carbon at $t = 0$ is shown in Figure 5.4. It will be noted that, because of the laminar structure, the problem is one of linear, one dimensional diffusion. Under these conditions the solution of equation 5.4 has been shown to be:

$$C = C^* \left[\frac{l^*}{l} + \frac{2}{\pi} \sum_{n=1}^{\infty} \frac{1}{n} \sin \left(\frac{n\pi l^*}{l} \right) \cos \left(\frac{n\pi x}{l} \right) \exp \left(- \frac{\pi^2 n^2 D t}{l^2} \right) \right] \quad (5.5)$$

The meaning of C^* , l^* , l and x are shown in Figure 5.4. The results of equation 5.5 are plotted in Figure 5.5. This shows C plotted against (x/l) for various values of (Dt/l^2) , a non-dimensional parameter (for simplicity the initial value of C in ferrite has been taken as zero rather than 0.025% as shown in Figure 5.4). It will be seen that diffusion is virtually complete when $(Dt/l^2) \approx 0.2$, since at this stage the values of C differ from the equilibrium diagram by less than 25%. Diffusion is therefore complete at rather lower values of Dt than was implied by equation 5.3.

The exploratory experiments reported in Chapter 4 were planned and performed before the development of the theory presented here. In these experiments the steel was used both in the normalised and in the hardened and tempered condition. In such hardened steels cementite exists as roughly spherical precipitates in a matrix of ferrite so that the simpler

one dimensional diffusion theory of equation 5.5 and Figure 5.4 is not applicable in detail. However at this exploratory stage, it is sufficient to note that, when using well tempered steels, the required distance of carbon diffusion is of the same order as that implied for the calculations of Figure 5.4, even though the geometric configuration is rather more complex.

The diffusion coefficient D is determined by the relationship

$$D = D_0 \exp (- Q'/S\theta) \quad (5.6)$$

where Q' is the activation energy, θ the temperature, S the gas constant and D_0 an experimentally determined value. Using equation 5.6 one can now calculate a theoretical value of temperature necessary to ensure a phase transformation.

In the exploratory single rub experiments, in which phase transformed material was detected, a representative time of contact, t , is obtained by dividing the radius of the circular contact area, a , by the surface velocity of the specimens. This gives $t \approx 0.2 \times 10^{-3}$ seconds. Assuming a value of $Dt/l^2 \approx 0.2$, a representative diffusion distance as $0.5 \mu\text{m}$, gives a required value of D of $2.5 \times 10^{-10} \text{ m}^2/\text{s}$. This corresponds to a temperature of 1250°C which is much larger than that assumed by Archard (1959) and Welsh (1965) and is comparable with the value obtained from the order of magnitude calculation of the previous section.

5.5 Discussion of simple diffusion theory

Although the previous section suggests, by assuming carbon diffusion is the rate controlling process for the formation of homogeneous austenite above 910°C , that reasonable agreement is obtained between the theory and the single run exploratory experiments, the results of section 5.4 require some qualification.

(i) Simple volume diffusion of carbon in austenite has been assumed, the situation would appear to be more complex. For instance under the conditions occurring in a tribological contact i.e. pressures, shear stresses etc. paths of preferential diffusion may be created.

(ii) In equation 5.6, D_0 and Q' were assumed constant i.e. $D=f(\theta)$ only, however D is also a function of the carbon concentration.

(iii) The theory of Figures 5.4 & 5.5 is over simplified because it assumed a starting point of a lamella of γ containing 6.67% carbon (previously cementite). In fact the maximum possible solubility of carbon in γ is 1.8% by weight. Thus the carbon concentration gradients assumed in the present diffusion theory are very much larger than can occur in practice. Since concentration gradient is the driving force in diffusion, processes will be probably somewhat slower than is implied by Figure 5.5.

(iv) The arguments advanced have been based upon average values of the flash temperature at the surface and a representative value of

FIG 5.6 VARIATION OF Q' & D_0 WITH % CARBON FOR
CARBON DIFFUSING IN γ IRON, 760 - 1360 °C
(WELLS et al (1950))

the duration of the contact. This may well be seriously at fault since θ appears in the rate equation 5.6, which through the exponential term gives a strong bias to the influence of high temperatures.

We shall first consider points (ii) - (iv). The influence of pressure and shear stresses on the metallurgy of steels will be discussed later in this chapter.

5.6 The diffusion coefficient: its dependence upon concentration and the possible consequences

It had been assumed that in the relationship

$$D = D_0 \exp (-Q'/S\theta) \quad (5.6)$$

D_0 and Q' were constants (i.e. $D = f(\theta)$ only). However examination of the literature for carbon diffusing into γ iron over a temperature range of 750 - 1360°C showed this assumption to be incorrect (Wells et al (1950)). Both D_0 and Q were found to vary with carbon concentration (Figure 5.6). Thus in the solution of the diffusion equation, as the carbon atoms diffuse from the original lattice sites to those in the γ phase, the diffusion rate should not only vary with the concentration gradient, but also with the absolute values of the carbon concentration.

The variation of D , the diffusion coefficient, with percentage carbon as compared to the variation of D with temperature θ is small. Over the range of temperature 760-1360°C D varies by a factor of 60 or more, while

FIG. 5.7a TRUE SOLUTION REQUIRED

FIG. 5.7b SIMPLE MODEL

over the range of carbon concentration (0-1.8%) D varies by a factor of 3 or less. Since we are primarily concerned with the effect of temperature on diffusion and that allowing for the small effect that carbon concentration has on D would unnecessarily complicate the theory, the effect of carbon concentration has, in the first instance, been ignored.

5.7 Simple solution of diffusion equation with limited solubility

A theory is required which keeps the carbon concentration within austenite to low values (say $< 1.8\%$) as the present theory of section 5.4 tacitly assumes that Fe_3C is transformed into austenite with 6.67%C and such carbon concentrations are very much larger than can occur in practice. Since concentration gradient is the driving force in diffusion the present theory will probably underestimate the time required for the completion of diffusion.

In effect, a theory is required, even if it is a crude approximation, which confines itself to the early stages of diffusion whilst iron carbide still exists as a separate phase. We require this theory primarily to obtain an estimate of how long this initial period might last and to compare it with the state the diffusion would have reached under comparable times using the old theory.

Figure 5.7a sketches the true solution required, whilst Figure 5.7b defines the simple model to be solved in the present analysis. Since $l \gg l^*$ this model should give a first estimate of the time required to complete the dissolution of the carbide. The carbide is being considered

here to dissipate by retaining its dimensions and reducing its carbon concentration, rather than by reducing its dimensions and retaining its carbon concentration (i.e. Figure 5.7a). We seek a solution only for the region $l^* < x < l$ (and similar regions in the periodic structure). The appropriate time to terminate this analysis occurs when the value of C in the region $0 < x < l^*$ has declined to a value of C' i.e. when the carbon lost from this region has appeared in the region $l^* < x < l$. The required condition is therefore:-

$$(C^* - C')l^* = \int_{l^*}^l C dx \quad (5.7)$$

Define $y = x - l^*$, equation 5.7 becomes

$$(C^* - C')l^* = \int_0^{l-l^*} C dy \quad (5.8)$$

Consider the region $0 < y < 2(l-l^*)$, we require a solution of Ficks 2nd law

$$\frac{\partial C}{\partial t} = D \frac{\partial^2 C}{\partial y^2} \quad (5.4)$$

with boundary conditions $\frac{\partial C}{\partial y} = 0$ at $y = l - l^*$, $t \geq 0$

$$C = C' \text{ at } y=0, 2(l-l^*), t \geq 0$$

FIG. 5.8 LIMITED SOLUBILITY SOLUTION FOR CARBON CONCENTRATION IN AUSTENITE IN THE REGION $\bar{x} < \bar{l}$ PLOTTED AS A FUNCTION OF DISTANCE (y/l) FOR VARIOUS VALUES OF Dt/l^2

(i.e. up to when analysis is terminated). However the second boundary condition is non homogeneous. Therefore define $F=C'-C$

Equation 5.4 becomes

$$\frac{\partial F}{\partial t} = D \frac{\partial^2 F}{\partial y^2} \quad (5.9)$$

with boundary conditions $\frac{\partial F}{\partial y} = 0$ at $y = (\lambda - \lambda^*)$, $t \geq 0$

$F=0$ at $y=0, 2(\lambda - \lambda^*)$, $t \geq 0$

(i.e. up to when analysis is terminated). Brief details of the solution are given in Appendix 1. The solution has been shown to be

$$C(y,t) = C' \left[1 - \frac{4}{\pi} \sum_{n=0}^{\infty} \frac{1}{(2n+1)} \sin \frac{(2n+1)\pi}{2} \frac{y}{(\lambda - \lambda^*)} \exp - Dt \left(\frac{(2n+1)\pi}{2(\lambda - \lambda^*)} \right)^2 \right] \quad (5.10)$$

A computer program to solve equation 5.10 on a CBM PET was written. It has been solved for various values of $Dt/\bar{\lambda}^2$ where $\bar{\lambda} = \lambda - \lambda^*$. These have been plotted in Figure 5.8. The area under the curve of $0 < y < \lambda - \lambda^*$ was estimated in the program by a Simpsons approximation and compared against the limiting condition of equation 5.8, where values of $\lambda = 0.5 \mu\text{m}$ and $\lambda^* = 0.06 \mu\text{m}$ were assumed. The critical value of $Dt/\bar{\lambda}^2$ was found to be approximately 0.107 corresponding to a value of Dt/λ^2 in the simple

FIG. 5.9 COMPARISON BETWEEN SIMPLE DIFFUSION THEORY TO THAT OF A LIMITED SOLUBILITY SOLUTION OF SECTION 5.7

diffusion theory of section 5.4 of 0.08. Figure 5.9 shows a comparison between $Dt/\bar{l}^2$ at a value of 0.107 and Dt/l^2 at a value of 0.08.

In this section an attempt has been made to discover the extent to which the limitation of the solubility of carbon in austenite to a maximum value of 1.8% might lengthen the diffusion process beyond the simple analysis given in section 5.4 and Figure 5.5. The model used here (i.e. Figure 5.7b, rather than the true situation of Figure 5.7a) will slightly shorten the time required for the homogenisation of austenite, because, for example, the diffusion distances are very slightly shortened. Nevertheless, as Figure 5.9 amply demonstrates, the limitation of carbon solubility has only a marginal influence upon the times required for carbon diffusion. Therefore the broad conclusions drawn from Section 5.4 remain valid.

5.8 Full solution of diffusion equation with limited solubility

After completion of the work above, a survey of the literature revealed a recent analytical solution by Pabi (1979) for one phase, diffusion controlled, dissolution of uniformly sized and equispaced stacked plates in an infinite or finite matrix in a manner analogous to the momentum integral in fluid mechanics.

A brief formulation of the problem, applicable to the present investigation, will be given here with the essential conclusions. The interested reader is referred to the original paper for the complete mathematical analysis.

FIG. 5.10 TIME / CONCENTRATION PROFILE FOR CARBON DIFFUSING IN AUSTENITE

Uniformly sized and homogeneous plates of a solute rich second phase (i.e. carbide) are assumed to be stacked at equal intervals in an isotropic matrix phase (i.e. austenite) as depicted in Figure 5.4. Positions of symmetry planes across which no transfer takes place are assumed. The half thickness of the precipitate plates is (as before) l^* , at $t=0$ and at $t>0$ is denoted by B_t . The average solute content of the homogeneous solution is $C^\#$ (i.e. 0.8%C). Initial composition of the solvent (austenite) and second phase (carbide) are C^+ & C^* . As in section 5.4, for simplicity, C^+ has been taken as zero rather than 0.025%.

As homogenisation starts C' the solute concentration in the matrix (Figure 5.10) obtains the equilibrium value given by the phase diagram (i.e. 1.8%C). The compositions are normalised such that C^+ is zero, C^* is unity and C' & $C^\#$ have intermediate values i.e. in terms of our investigation $C'=0.27$ & $C^\#=0.12$.

The time/concentration profile adjacent to a dissolving carbide precipitate is schematically displayed in Figure 5.10. Diffusion occurs solely in the austenitic phase and we require a solution of the equation

$$\frac{\partial C}{\partial t} = D \frac{\partial^2 C}{\partial x^2}, \quad B_t \leq x \leq l, \quad t > 0 \quad (5.4)$$

subject to the boundary conditions

$$C(x, t) = C^+ = 0, \quad l^* \leq x \leq l, \quad t = 0 \quad (5.11)$$

$$C(B_t, t) = C', \quad t > 0 \quad (5.12)$$

The mass balance at the interface can be expressed as:-

$$D \frac{\partial C}{\partial x} (B_t, t) = (C^* - C') \frac{\partial B_t}{\partial t} \quad (5.13)$$

To eliminate B_t from equation 5.13 an alternative expression is derived.

Differentiation of equation 5.12 with respect to t gives:-

$$\frac{\partial C}{\partial x} \frac{dB_t}{dt} + \frac{\partial C}{\partial t} = 0, \quad x = B_t \quad (5.14)$$

Eliminating dB_t/dt between equations 5.13 & 5.14 and substituting equation 5.4, it is found that

$$\left(\frac{\partial C}{\partial x} \right)^2 = -(C^* - C') \frac{\partial^2 C}{\partial x^2}, \quad x = B_t \quad (5.15)$$

Pabi's analysis introduces a concept of a phenomenological diffusion layer thickness δt which is defined as the distance from the interface beyond which the initial solute (i.e. carbon) distribution remains practically unperturbed. This penetration distance is analogous to the boundary layer thickness in fluid mechanics and thermal layer thickness in heat transfer. δt increases with time and B_t diminishes. At the crucial moment $t=t_c$ the diffusion layer extends up to the symmetry plane $x=l$ (see Figure 5.10) and stage I of the dissolution process is considered to be complete. If B_t is +ve and equal to B_c at $t=t_c$, then $B_c + \delta t = l$ at $t=t_c$. If dissolution continues beyond t_c , stage II of the process starts. The diffusion flux (i.e carbon atoms) cannot cross the symmetry plane and δt would fail to have physical significance any further. The solution terminates at the equilibrium limit of the second phase (0.8%).

FIG. 5.11 VARIATION OF $Dt_F / 4\ell^2$ WITH NORMALISED C' FOR $C^* < 0.75 C'$ (Pabi (1979))

The essential conclusions for the present study are as follows:-

(i) For dissolution in a finite geometry the dimensionless parameter $Dt_c/4\lambda^{*2}$ depends markedly on $C^\#$ but is almost indifferent to any change in C' (up to a normalised $C'=0.8$).

(ii) On the other hand the dimensionless parameter $Dt_f/4\lambda^{*2}$ (where t_f is the time required for complete homogenisation) is practically independent of $C^\#$ for values of $C^\# < 0.75C'$ but dependent on C' .

Let us consider these conclusions in turn. First, the rationalised value of C' in the present analysis is 0.27, satisfying the criteria for conclusion (i). Figure 5.9 showed the comparison between the simple solution of Section 5.4 of the diffusion equation and the first approximation for the limited solubility case (Section 5.7). At that point, marginally beyond t_c , in the present study, the state of diffusion could be considered to be essentially identical. Assume that these conditions correspond to t_c . The state of diffusion in Figure 5.9 is shown to be practically independent of C' and probably dependent on $C^\#$. This is in exact agreement with the full solution of Pabi (1979).

Second, as the criteria for conclusion (ii) are also satisfied, the value of C' (0.27) corresponds to a value of $Dt_f/4\lambda^{*2}$ of approximately 3 (see Figure 5.11 taken from Pabi). Assuming, as before, values of $\lambda^* = 0.06 \mu\text{m}$ and $\lambda = 0.5 \mu\text{m}$ this corresponds to a value of Dt/λ^2 of approximately 0.17 for complete homogenisation. This can be compared with an approximate value of Dt/λ^2 of 0.2 from the simple diffusion theory for complete homogenisation (Figure 5.5). Since excellent agreement occurs between the simple diffusion theory of Section 5.4 with the full solution

a) Temperature / time history of equation 5.5

b) Required temperature / time history

FIG. 5.12 TEMPERATURE / TIME (θ / t) HISTORIES

of this section, Pabi's solution therefore confirms the results and applicability of the simple theory of Section 5.4

5.9 The effect of a temperature/time history

We have obtained the result

$$C = C^* \left[\frac{l^*}{l} + \frac{2}{\pi} \sum_{n=1}^{\infty} \frac{1}{n} \sin \left(\frac{n\pi l^*}{l} \right) \cos \left(\frac{n\pi x}{l} \right) \exp - \left(\frac{\pi^2 n^2 D t}{l^2} \right) \right] \quad (5.5)$$

from the simple diffusion theory for lamellar pearlite, for the concentration C , after a time t with a given value of the diffusion coefficient D (corresponding to a temperature Θ). This is equivalent to a temperature/time history for the material as shown in Figure 5.12a. In practice the temperature/time history for a point on the rubbing surface will be somewhat as shown in Figure 5.12b. Through equation 5.6 (as the simplest form available) Figure 5.12b will correspond to a diffusion coefficient/time history (D/t).

Equation 5.5 is of the form

$$C = C^* f_1(x) f_2(Dt)$$

In its derivation it is assumed that the temperature, Θ , (or the diffusion coefficient, D) is not a function of x ; in practice the variation of Θ (or D) over distances (i.e. the interlamellar spacing) to which the analysis applies is negligible. Therefore any element of material subjected to a time/temperature (or time/diffusion coefficient)

history can be regarded as being subjected to a sequence of small diffusion increments, each of duration dt , during which the essential variable is Ddt . It follows that, for a total history of duration T , it is appropriate to replace, in equation 5.5, Dt by

$$\int_0^T Ddt \quad (5.16)$$

To simplify the problem attention will be confined to surface temperatures and more specifically surface temperatures which existing during the duration of the contact. In the present analysis the decay of the temperature will be ignored. The simple analysis given in Section 5.4 used estimated values of the mean surface temperature and a representative value of the duration of the contact. This may well be seriously at fault since θ appears in the rate process equation 5.6, which through the exponential term gives a strong bias to the influence of high temperatures.

As a further simplification, Archard's (1959) flash temperature equations for high speed rubbing ($L > 5$: as applicable to the exploratory experiments) will be used. Thus the required temperature/time history for any point of the contact will be

$$\theta = \theta_{bk} + \Delta\theta = \theta_{bk} + K^* t^{1/2} \quad (5.17)$$

where K^* is a constant derived from the linear heat flow equation for the surface temperature of a semi infinite body subject to a heat flux q

$$\Delta\theta = \frac{2qt^{1/2}}{(\pi k \rho c)^{1/2}} \quad (2.6)$$

which is applicable under high speed conditions. θ_{bk} is bulk temperature, its incorporation is important when considering a D/t history. Thus the effective value of Dt in equation 5.5 for any point on the surface subject to the high speed equations will be

$$[Dt]_{\text{eff}} = \int_0^{t^*} D_0 \exp \left(\frac{-Q'}{S(\theta_{bk} + K^* t^{1/2})} \right) dt \quad (5.18)$$

A solution of the integral and insertion of representative values of the constants will give a first indication of the possible importance of a temperature/time history in influencing the diffusion.

Details of the solution are given in Appendix 2. The full solution of equation 5.18 has been shown to be

$$[Dt]_{\text{eff}} = 2 D_0 \left\{ \frac{K e^{-y}}{2y} \left[(2A + K) \left[\left(\frac{1-1}{y} \right) + \frac{2!}{y^2} \left(\frac{1-3}{y} \right) + \dots \right] - (2A + K \left(\frac{1-1}{y} \right)) \right] \right\} \begin{matrix} K/A+t^{1/2} \\ K/A \end{matrix} \quad (5.19)$$

A computer program to solve equation 5.19 on a CBM PET was written. It has been solved for four sets of experimental conditions corresponding to

Run	Designation	W(N)	$V_{sf}(m/s)$	HV 20 kg	Phase transformed Material in		$Dt \text{ (m}^2\text{)}$	$[Dt]_{eff}(m^2)$
					Single	Multiple		
E17	A	852	1.4	275	Yes	Yes	2.09×10^{-13}	2.2×10^{-11}
E14	B	432	0.92	400	Yes	Yes	6.5×10^{-14}	1.4×10^{-11}
E11	C	343	0.79	400	Diffuse	Yes	1.35×10^{-14}	6.5×10^{-12}
E13	D	167	0.79	400	No	Diffuse	7.38×10^{-16}	1.9×10^{-12}

Table 5.1 Conditions assumed in solution of equation 5.19, using results from Chapter 4.

FIG. 5.13 VARIATION OF $[D^\dagger]_{\text{eff}}$ WITH THE FRICTION COEFFICIENT μ . $\theta = 50^\circ\text{C}$

FIG. 5.14 VARIATION OF $[Dt]_{\text{eff}}$ WITH THE BULK TEMPERATURE θ_{bk}

the earlier exploratory experiments of Chapter 4. These details are set out in Table 5.1, together with the conclusions drawn from these experiments, and the values of Dt which would be calculated for these conditions using an average value of the surface flash temperature, an effective dwell time of $t^* (=a/V_{sf})$ and a coefficient of friction of 0.6.

It will be seen that the values of $[Dt]_{eff}$ calculated using the variation of θ across the contact exceed the values obtained using mean values by more than two orders of magnitude. Moreover in analysing the main experimental results of this thesis, to be reported in following chapters, conducted under more well defined conditions, it is obvious that it will be necessary to consider the time/temperature history of elements in the rubbed surface both as a function of position within the track and of depth below the surface.

Therefore, what this section shows, is that the most important feature in the time/temperature history of any element of material is the maximum temperature to which it is subjected.

Figure 5.13 shows on a log/linear plot the results of the solution of equation 5.19 for the conditions of Table 5.1 assuming the bulk temperature θ_{bk} is 50°C for different assumed values of the coefficient of friction μ .

Figure 5.14 shows, also on a log/linear plot, the results of the solution of equation 5.19 for a constant value of the coefficient of friction but

variation in the bulk temperature θ_{bk} .

The conclusions to be drawn from this work are as follows:-

(i) As implied earlier the calculations of $[Dt]_{eff}$ suggest that the use of the calculated mean surface flash temperatures and the representative value of the dwell time grossly underestimate the chances of diffusion.

(ii) The conclusions from the exploratory experiments of Chapter 4 about the experimental conditions necessary for the completion of diffusion must be regarded with great reserve.

(iii) Any theory which is used for comparison with experimental findings must take into account the full temperature/time history which the material has experienced; because of the exponential dependence of diffusion upon temperature the effects are dominated by the highest temperatures which exist during this history.

(iv) Changes in bulk temperature, even though they are small compared with the flash temperatures can have significant influence upon the value of $[Dt]_{eff}$ and thus on the progress of diffusion. Again this is the consequence of the exponential dependence of D upon θ .

5.10 Effects of pressure etc., on the metallurgy of steels

In a tribological contact typical of the exploratory experiments of Chapter 4, as well as the existence of high flash temperatures we must be

aware of the concomitant effects of high contact pressures and related phenomena of elastic and plastic deformations and shear stresses which can occur. This non-mathematical review will primarily consider the effect of high contact pressures, on steel metallurgy and subsequently indicate possible modifications as a result of the related phenomena.

In essence it is possible to divide the effects of pressure on the metallurgy of steels into two parts, the thermodynamic effects and the effects upon transformation kinetics (Hilliard and Cahn (1961)).

Thermodynamic effects include effects upon

- (i) Phase equilibria; where in addition to the possibility of producing new phases, there will probably be some changes in the composition of the phases with increasing pressure.
- (ii) Driving force; pressure induced changes in the driving force are related to changes in the phase equilibria.
- (iii) Equilibrium defect concentration; it is to be expected that creation of both interstitials and vacancies will be associated with an increase in volume. Therefore their concentration will decrease with increasing pressure.
- (iv) Grain boundary and dislocation energy; Hilliard and Cahn (1961) suggest their energies will increase with increasing pressure.

FIG. 5.15 INFLUENCE OF PRESSURE (HYDROSTATIC) ON THE
IRON-CARBON PHASE EQUILIBRIUM DIAGRAM

The effects upon the phase equilibria will now be considered in more detail.

Until the recent development of high pressure techniques, permitting the generation of sustained high pressures and temperatures, phase equilibria and the practical study of transformation kinetics were largely limited to atmospheric conditions. However, using techniques developed by Hall (1960) and co-workers the pressure dependence of the Fe-C equilibrium diagram have been determined up to 6.5 GPa. Hilliard and Cahn (1961) studied the growth rate of pearlite in high purity Fe-C alloys and commercial steels at 3.4 GPa; Radcliffe, Schatz and Kulin (1963) observed the phase equilibria of high purity Fe-C alloys containing between 0.08 - 1.23%C at 4.2 GPa, while Hilliard (1963) determined at 3.5, 5 and 6.5 GPa the phase equilibria of Fe-C alloys. Their results are conveniently shown in Figure 5.15 taken from Nilan (1967). It is easy to show that the plastic deformation of a normalised eutectoid steel, of the type reported in the exploratory experiments, involves asperity contact pressures of the order of 2.7 GPa. Figure 5.15 indicates at this pressure, the α/γ transformation in pure iron is reduced from its value of 910°C at atmospheric pressure to 740°C; similarly the eutectoid point is changed from a temperature of 723°C and a carbon concentration of 0.8% to a temperature of 660°C and a carbon concentration of 0.4%. Therefore, at first sight, it might appear that the required values of flash temperatures deduced in the main body of this chapter should be reduced.

Indeed, in a comprehensive review by Schlicht (1973), the lowering of the critical points with pressure has been postulated as an explanation for

the appearance of white bands or butterflies in rolling elements subject to high simultaneous pressure and shear stress. Thus the effect of pressure on the Fe - C equilibrium diagram suggests a plausible explanation for the formation of white bands or butterflies produced under conditions where, previous conventional metallurgy had indicated that temperatures sufficient for re-austenitisation were most unlikely to be developed; this may be seen in the work of Martin et al (1966) in studies of micro-structural alterations in roller bearing steels under cyclic stressing and in the work of Turley (1975) who studied white layers produced during reaming ultra high strength steel. The major contribution which the arguments of this section make to the problem is to emphasise the important influence of repeated stress and temperature cycles.

The effect of pressure on the transformation kinetics in the literature is limited to the study of austenite decomposition kinetics and the morphologies of structures thereby obtained. No work has been found in the literature on the effect of high pressures on the rapid austenitisation of steel. Nilan (1967) provides a recent summary on this topic, he additionally studied the isothermal transformation of steels containing 0.4 & 0.8% carbon at pressures up to 3.4 GPa. Although the effect of pressure was predominantly the retardation of the transformation, austenite decomposition mechanics were not altered by high pressure, within the limits studied. The decomposition morphologies produced at high pressure were substantially the same as those produced at atmospheric pressure in a steel having comparable phase relations. Similar conclusions were reported by Radcliffe et al (1963) from experiments at a pressure of 4.2 GPa, both in the initiation and rate of

transformation; in addition they found that the reduction in the rate of transformation of austenite with respect to atmospheric conditions was much greater for Fe - C alloys than for plain carbon steels. This behaviour was attributed to the effect of pressure in retarding the diffusion of substitutional impurity elements e.g. manganese in these steels. However, at high pressures, the effect of carbon content on the overall transformation rate was reversed i.e. increasing the carbon content decreased the time required for the isothermal transformation (austenite $\rightarrow$ ferrite + carbide) in Fe - C alloys. Since this effect was accompanied by some morphological differences and that the overall transformation time was still greater than that at atmospheric pressure we shall, in the first analysis, set aside this problem.

Consider therefore, the probable influence of high pressures upon the formation of homogeneous austenite at temperatures 910°C which this work suggests is crucial in the formation of hard phase transformed material in rubbing. Metallurgical considerations suggest that the rate controlling process is carbon diffusion in γ phase. In accordance with Le Chatelier's principle, the application of high pressure can only reduce the available sites for interstitial diffusion and thus retard the formation of homogeneous austenite. Likewise increasing pressure, decreases the rate of substitutional lattice diffusion, principally because of the reduction in the equilibrium vacancy concentration, rather than a decrease in the vacancy mobility (Hilliard and Cahn (1961)). The reduced diffusivity of carbon in austenite with pressure has indeed been observed by Hilliard and Tully (1961) in the decomposition of austenite in Fe - C alloys.

More recently May (1961), Fujita and Suzuki (1974) and Schmidtman et al (1976) have all found the rates of isothermal transformations in both Fe-C alloys and commercial steels are retarded by an increase in pressure.

We now turn to consider the effect on non-equilibrium states on the rate of diffusion. It is well known that plastic deformation and/or shear stresses cause an increase in the equilibrium concentration of vacancies and dislocations, indeed even cause macroscopic tears and holes. However the evidence on the consequential effect on interstitial diffusion is contradictory. Robbins et al (1962) report accelerated spheroidization of eutectoid steels by concurrent hot working. The high rate of spheroidization being attributed to the formation of excess lattice vacancies during plastic deformation. Similar results were obtained by Dolzhenkov et al (1973) in the plastic deformation of a commercial steel in the γ phase. However they conclude that the vacancy mechanism for accelerated diffusion is thermodynamically impossible during deformation and therefore diffusion probably occurs along dislocation channels. Wazzan and Dorn (1965) studied the enhanced diffusion into single nickel crystals during tensile deformation and also ruled out vacancy supersaturations as a possible explanation on thermodynamic grounds. They attributed their results to "pipe diffusion" i.e. the presence of moving dislocations providing high diffusivity paths for tracer atoms. A recent review by Cohen (1970) on strain enhanced diffusivity by plastic deformation favours this interpretation. Ruoff (1967), although not excluding the possibility that excess vacancies or dislocation channels or both may contribute to the acceleration of diffusion during plastic deformation, considers the contribution to be smaller than that observed experimentally. He rejects pipe diffusion and has developed the

concept of "mechanical diffusion". According to this "bundles" or "glide packets" of diffusing atoms are captured during plastic deformation and displaced to a depth approximately equal to the displacement of individual glide packets.

Opposing evidence is, for instance, provided by Matsuyan and Golikov (1971) on the influence of prior cold plastic deformation on the diffusion of carbon in iron and iron alloys. Despite effects of increases in vacancy concentration and in dislocation density, reduced diffusional mobility of carbon atoms was found. Similar results have been reported by Kidin et al (1971) on the diffusion of carbon in austenite and by Dorofeev et al (1972) on carbon diffusion in steel during dynamic hot pressing. Most authors suggest that the reduced mobility is caused by the slowing down or stopping of the interstitial atoms by the lattice defects. Such effects have also been observed by Asano et al (1974) in the diffusion of hydrogen in mild steel. In their work it was suggested that dislocations played a trapping effect on the diffusing hydrogen. Solovyev (1977) has given the mathematical justification for the slowing down of interstitial atomic diffusion by lattice defects. In a review of the effects of plastic deformations on phase transformations in metals Stuwe (1969) suggests that dislocations are effective "sinks" for foreign atoms, with the result that diffusion is slowed down by the lattice until the dislocations are "filled up". It is incorrect to speak of a reduction in diffusion coefficient, what is reduced is the speed with which a concentration rise or change can take place in the material. Schlicht (1974) also notes that despite plastic deformation increasing the lattice defect concentration (i.e. vacancy concentration and dislocation density) the rise in dislocation density

retards the diffusion of carbon in thermo-mechanically heat treated steels.

Two further points deserve comment. First, the accelerated dissolution of precipitates of a second phase e.g. carbides in austenite is usually considered to be assisted by plastic deformation causing the break up of the precipitates (Stuwe (1969)). This results in (a) a possible reduction in the phase boundary energy (usually small) (b) the reaction boundary surface is enlarged and (c) the average diffusion distance would be reduced. However in the exploratory experiments of Chapter 4 (indeed in the subsequent main experiments of this thesis) the loads are in general either just below or just above that required for the onset of plastic flow. Loads for full plasticity and consequent full break up of the second phase are orders of magnitude larger. With corresponding smaller loads it seems probably this effect will be reduced. Second, the formation of phase transformed material in a single rub requires mean surface temperatures of approximately 1100°C or greater i.e. ~ 0.75 of the melting point of the metal. At such temperatures the corresponding thermal defect concentration in the lattice may well be comparable to the increase in defect concentration as a consequence of mechanical deformation, despite the large strain rates occurring.

Based on this evidence it seems likely that the effects of plastic deformations or shear stresses or both, on steel metallurgy are much less influential than might appear.

The possible effects of pressure upon phase transformations in rubbing may be summarised as follows:-

(i) Pressure lowers the critical points of the Fe - C equilibrium phase diagram, suggesting that lower temperatures are required for the formation of austenite.

(ii) However, as in the discussion of flash temperatures, it is doubtful if the duration of tribological contact (say one millisecond) automatically produces the results indicated in the Fe - C equilibrium phase diagram at high pressures.

(iii) Most importantly, pressure retards the diffusivity of carbon in austenite and thus the formation of homogeneous austenite will require a longer time or a higher temperature.

When these conclusions are applied to the production of phase transformed material in rubbing contacts, the effects summarised in (i) are likely to be balanced by those discussed in (ii) and (iii). The influence of pressure seems likely to be much less marked than might appear at first sight. This conclusion is supported by one further consideration. In an asperity contact under the high speed conditions of flash temperature theory, the maximum temperature occurs at the trailing edge of the contact (see Figure 5.12b) where the contact pressures are significantly lower than their maximum value. It has already been noted in Section 5.9 that the production of phase transformed material will be most strongly influenced by the maximum rather than the average value of the flash temperature so that, once again, the influence of contact pressure is likely to be far less marked than expected. Therefore in the first stage of the production of a new metallurgical theory of rubbing contacts (which this chapter represents) it seems appropriate to set

aside the complications introduced by the existence of high pressures and related phenomena.

5.11 Discussion

This chapter has been concerned with a reconsideration of conventional ferrous metallurgy applicable to small volumes of material subject to a combination of high temperatures and pressures for times of the order of one millisecond or less.

It has been shown that because of the short duration of the transient flash temperatures, austenitisation of a plain carbon eutectoid steel requires temperatures in excess of the allotropic α/γ transformation temperature of 910°C . Moreover it has been shown, using an order of magnitude calculation, that even at temperatures above 910°C , the duration of a typical flash temperature is insufficient to complete the diffusion of carbon in γ iron, which is the probably the rate determining process in austenitisation.

A simple one dimensional linear diffusion theory for carbon (at 6.67%) diffusing into austenite (at 0°C) has suggested a temperature of approximately 1250°C . is required for the formation of homogeneous austenite. This temperature is comparable to that required by an order of magnitude calculation and to those required for the formation of phase transformed material in a single rub.

The variation of diffusion coefficient with carbon concentration has been shown to be of low significance compared to the variation of diffusion

coefficient with temperature. More comprehensive solutions of the diffusion of carbon in austenite have merely served to reinforce the conclusions from the simple theory of Section 5.4.

An analysis of the effect of a temperature/time history or more correctly a diffusion/time history of a contact has revealed the important result that the use of average conditions grossly underestimates the chances of diffusion. Thus we must be aware that (a) diffusion is dominated by the highest or maximum temperatures which exist during the contact history and (b) changes in bulk temperature, although small compared to the flash temperature can, by raising the maximum temperature have significant influence on the diffusion history.

The concomitant effects of high contact pressures, and related phenomena of plastic deformations and/or shear stresses on the metallurgy of steels have been considered. Although at first sight the effect of pressure, in lowering the critical points of the Fe - C equilibrium diagram and the effect of plastic deformations and/or shear stresses in raising the equilibrium defect concentration may appear to suggest lower temperatures and that smaller diffusion coefficients are required, it has been argued that these effects may be much less marked. Thus in the present analysis these complications have been put aside.

The arguments forwarded in this chapter are strictly applicable to phase transformations in a single rub. The multiple rub situation is more complex. For instance consider an asperity contact in the multiple rub track, which has been subjected to a flash temperature insufficient to cause a complete homogeneous phase transformation i.e. the carbon atoms

will be moved to a new position related to the temperature and dwell time of that event. The carbon atoms will in effect be "frozen" in their new positions until the next application of heat. Before experiencing the next contact, sufficient in severity (in terms of flash temperature) to now austenitise martensite (previously ferrite) and continue diffusion, the asperity contact may see many contacts which, for instance, may work harden the surface material or cause tempering of the partially transformed material. Indeed such contacts may alter or introduce additional paths of diffusion or may invalidate the simple approach to the effects of contact pressures and related phenomena. Rigney and Glaeser (1978) have shown that heavy plastic deformation of thin surface layers can cause a cellular structure to be formed, such structure providing suitable preferential paths of diffusion, crack generation etc.. Also martensite, by its very nature of formation, has a high defect concentration. On subsequent reheating, sufficient to austenitise the material, these defects may well remain and be inherited by the austenitic phase. The short duration of the austenitic phase, in these experiments being insufficient to anneal these defects. Therefore paths of alternative or preferential diffusion for the carbon atoms may again be created.

However, at this stage, it is appropriate to set aside possible complications and consider, first, the new metallurgical theory advanced. If this theory is to gain credence, it should be possible to re-interpret the previous work, in particular, that of Archard (1959) and Welsh (1965) in terms of this theory. Chapter 6 therefore considers the effect of the new metallurgical theory on the interpretation of wear patterns.

CHAPTER 6

EFFECT OF METALLURGICAL THEORY ON INTERPRETATION OF WEAR PATTERNS

6.1 Introduction

Chapter 6 described a new metallurgical theory for the production of phase transformed material in single and multiple rubs. This has suggested that to austenitise surface layers of steel in times of the order of one millisecond or less the transient temperatures must be raised above the allotropic α/γ transformation of 910°C ; in addition sufficient time must elapse for carbon diffusion to ensure the formation of homogeneous austenite. This chapter considers the implications of this theory on the dry wear of steels and in particular on the classic work of Welsh (1965) who studied the dry sliding wear of plain carbon steels as a function of composition, hardness, load and sliding speed. Prior to this, the evidence on the wear of metals, mainly steels will be briefly reviewed with reference to wear theories and mechanisms.

6.2 Review of the wear of steels

Wear is one of a number of processes which occur when the surfaces of engineering components are loaded together and are subject to motion. Indeed wear can be defined as the removal of material by mechanical and/or chemical action. Taking the approach of Burwell (1957) it is possible to provide the following classification of wear mechanisms:-

(i) Adhesive or galling wear. When two bodies slide over each other, their asperities can adhere or weld to each other. The bond thus formed breaks when the sliding motion is continued. The position of fracture can be at the original surface or in the sublayers of one of the bodies. If welding occurs over large areas of the surfaces, the wear is often called scuffing.

(ii) Corrosive wear. When rubbing takes place in a corrosive environment, either gaseous or liquid, then surface reactions take place and reaction products are found on one or both surfaces. These reaction products are commonly poorly adherent to the surfaces and further rubbing causes their removal. This process is then repeated.

(iii) Abrasive wear. This type of wear covers two types of situation, though in both wear occurs by the ploughing action of a harder surface on a softer material. In the first instance a rough hard surface slides against a softer surface (two body abrasion). In the second case abrasion is caused by loose hard particles between the rubbing surfaces or particles embedded in a soft surface, sliding over another soft surface (three body abrasion).

(iv) Surface fatigue wear. When two bodies in sliding and/or rolling contact are subjected to a loading cycle or when an asperity experiences sufficient contacts without adhesion, cracks may appear in the surface or an asperity will fail owing to metal fatigue, resulting in pitting. Fatigue wear likewise occurs in rolling contact.

(v) Less common effects such as fretting, erosion and cavitation wear. Fretting wear takes place when the contacting surfaces are subjected to an oscillatory tangential relative displacement of small amplitude. Erosive wear is damage caused by sharp particles impinging on the surface and cavitation wear is caused by implosion of vapour bubbles in a fluid or lubricant.

This classification is valuable in concentrating attention upon the different ways in which material can be removed, though as Burwell commented it is only a coarse one. In practice it is likely that one or more wear mechanisms are taking place or are initiating each other. Other wear classifications are possible e.g. Kragelskii (1965), Archard (1981), however the present classification is appropriate for the subsequent analysis of this chapter.

We now turn to a brief review of adhesive wear models, being appropriate in the interpretation of the patterns of wear reported in Welsh's work which are the main concern of this chapter.

(1) Archard model (1953). This model concerns wear under unlubricated conditions. The basic concept is that the worn volume U , produced over a sliding distance L' , can be related to the area of true contact A . If the hardness of the softer material is H and the load W

$$A = \frac{W}{H} \quad (6.1)$$

and the wear rate U/L' can be derived on the assumption that

hemispherical wear particles are produced. Thus:

$$\frac{U}{L'} = KA = \frac{KW}{H} \quad (6.2)$$

The wear coefficient K is usually taken to indicate the proportion of contacts, which results in the production of a worn particle. Alternatively in a continuous wear experiment $1/K$ represents the average number of times each part of the rubbing surface is rubbed before it is worn.

(ii) The Rabinowicz model (1964). Rabinowicz used an energy criterion to estimate when wear particles are formed by an adhesive wear process. Assume two surfaces, under load, sliding over each other; their asperities are subjected to adhesive forces. Consider one asperity on one surface. Postulate that during the sliding process fragmentation occurs in this asperity as a result of fatigue. If the fragment is able to detach and become a wear particle, the criterion is that the elastic energy stored in the fragment as a result of the sliding process must exceed the energy of adhesion between the fragment and substrate. Assuming that the fragment is a hemisphere with diameter d , Rabinowicz showed that there is a minimum wear particle diameter $d_{\min}$, given by

$$d_{\min} = \frac{6 E W_{ab}}{\sqrt{v} \sigma_y^2} \quad (6.3)$$

Where E is the Youngs Modulus of the material, v is Poisson's ratio, σ_y is the yield stress and W_{ab} is the work of adhesion of

the contacting materials. As in most cases the initial fragments formed are too small to satisfy equation 6.3, it is postulated that sub-critical fragments have first to agglomerate before they can form a wear particle.

This model, although providing an elegant criterion for the formation of particles, does not give information on how wear rate depends on load, sliding distance etc. A further disadvantage is that it is difficult to obtain a reliable value of W_{ab} since W_{ab} is strongly dependent on the degree of contamination of the surfaces.

(iii) The Suh model (1973). A recent model has been developed by Suh to describe wear particle formation in adhesive wear at low sliding speeds and low loads. The proposed theory of wear is based on the following steps:

(a) The surface traction between two sliding surfaces deforms the material plastically. Under repeated cyclic loading the plastic shear deformation accumulates. The plastic strain is largest near the surface when the friction coefficient is large and decays to zero at a short distance away from the surface.

(b) As the plastic strain accumulates cracks nucleate around hard rigid inclusions, in materials with inclusions and second phase particles. Cracks also form when the material is a pure single phase material, probably at the walls of dislocation cells, grain boundaries and twin intersections.

(c) Once cracks are present (as a result of either the crack nucleation process or pre-existing cracks and voids) further loading and deformation causes cracks to extend and to propagate, joining the neighbouring ones. The cracks tend to propagate parallel to the surface at a depth governed by material properties and the coefficients of friction.

(d) When these cracks finally shear to the surface (at certain weak positions) long thin sheets "delaminate". The thickness of the wear sheets is a function of the location of crack nucleation points and the surface traction which controls the location of the maximum stress intensity factor.

A wear equation based on this theory, is of the form

$$\frac{U}{L'} = \frac{KW}{L'} \quad (6.4)$$

Where K is a wear factor which is a function of material properties of the surfaces, but is not an explicit function of hardness.

Childs (1981) has suggested in a review of the sliding mechanisms of metals, mainly steels, that Suh's results for steels are only a small part of the broader pattern of the overall wear behaviour established by Welsh.

The correlation between the wear models given here and wear observed in practice is fairly good. All models suggest wear rate is proportional to load and this has been confirmed by Burwell and

Strang (1953) and by Archard and Hirst (1956). The proportionality also roughly holds for lubricated contact e.g. Rowe (1966) and Montgomery (1965). However exceptions occur for instance Hirst and Lancaster (1956), who in studying the wear behaviour of a range of metals rubbing against Stellite in cetane, observed that only for aluminium and 60/40 was the wear rate proportional to load. In the other cases wear rate was a stronger function of load. Archard (1968) has shown for a limited range of materials wear rate is proportional to $1/H$, similar conclusions were reported by Kruschov (1957), though evidence exists where wear rate is not a linear function of $1/H$ (Montgomery (1970)). In terms of Suh's model there is evidence of crack formation and laminar wear particles in the thickness range 1-10 μm (Jahanmir et al (1974)) though further doubts of the applicability of the delamination wear theory have been raised by Torrance (1978) and Beagley (1978).

The brief review given here, of wear mechanisms and models, is commensurate with the requirements of this chapter, further details may be found in the large volume of literature on wear available e.g. Archard (1968), (1981), Eyre (1976) etc.. Before turning to the wear patterns of Welsh, note that several different interpretations of the K factor of wear theory have been forwarded; for example if one assumes wear is primarily a fatigue process then $1/K$ can be interpreted as the required number of contacts or deformations an asperity experiences to produce a fatigue fracture. Alternatively Quinn (1967) has proposed an oxidational wear theory of steel where, on average, $1/K$ asperity contacts are necessary, at a given asperity, for a critical oxide film thickness to build up

FIG. 6.1 EFFECT OF LOAD ON THE EQUILIBRIUM WEAR RATE OF 0.63% PLAIN CARBON STEEL RUBBING AGAINST ITSELF ON A PIN ON RING m / c AT 1 m / s (WELSH (1965))

before it becomes detached to form a wear particle. It is relevant to stress the need for a physical interpretation of K factors.

6.3 Wear patterns of Welsh (1965).

The basic pattern of behaviour which is the concern of this work is shown in Figure 6.1, drawn (in simplified form) from the work of Welsh. The equilibrium wear rate of a pin of 0.63% carbon steel rubbing on a ring of the same material, at a sliding speed of 1 m/s, is shown (on a double logarithmic plot) as a function of load. The main features of the results are as follows:

(i) At light loads, below the transition T_1 the wear rate is low. The surfaces remain smooth and relatively undamaged, and worn debris is finely divided (say 10 nm - 1 μ m) and almost entirely oxidised. This regime of wear has been called "mild wear".

(ii) At a range of heavier loads, between T_1 and a second transition T_2 , the wear rate increases by more than two orders of magnitude. The worn debris consists of relatively large metallic particles (say 10 - 100 μ m, occasionally increasing to 1 mm) the rubbing surfaces are severely torn and damaged. The regime is one of "severe wear". The mild wear/severe wear transition at T_1 is, in essence, similar to that which occurs with a wide range of metals and alloys.

(iii) At loads somewhat higher than the transition T_2 , the wear is initially severe, but as rubbing continues this is followed by a

transition to mild wear. This transition to mild wear is associated with the appearance upon both of the rubbing surfaces of hard metallurgically transformed material. Subsidiary tests, by Welsh, showed that at loads somewhat greater than T_2 , both the hard transformed material and the presence of oxide films are necessary for the maintenance of mild wear; if the surface layers are softened by annealing or if the oxide film is removed, mild wear is only re-established after an initial period of severe wear.

(iv) At a still higher load, corresponding to the transition T_3 (Figure 6.1) mild wear is associated with the presence of hard phase transformed material but oxide films are no longer necessary for the maintenance of mild wear. Above T_3 other changes in wear behaviour also occur. These include a divergence in the wear rates of the pin and ring.

Welsh (1965) followed Archard (1959) in interpreting the T_2 transition as corresponding to the conditions under which the transient flash temperatures reached values of the order of 750°C . Thus hard phase transformed material could be regarded as the martensitic transformation of normal ferrous metallurgy.

The following features refer to effects on the overall pattern of wear shown in Figure 6.1. For clarity Figures 6.2 to 6.4 of the subsequent discourse are plotted in terms of the wear rate of the pin, for which the broad pattern of wear is practically identical to that of the ring (up to the T_3 transition).

FIG.6.2 EFFECT OF CARBON CONTENT (%wt) ON BASIC PATTERN OF WEAR OF PLAIN CARBON STEELS (WELSH (1965))

FIG. 6.3 EFFECT OF INITIAL HARDNESS (VICKERS) ON THE WEAR PATTERN OF A 0.52% PLAIN CARBON STEEL (WELSH (1965))

(i) The influence of lowering the carbon content is both to lower the T_1 transition and raise the T_2 transition, such that at a carbon content of approximately 0.026% (by weight) severe wear occurs over the full range of loads (Figure 6.2). Conversely raising the carbon content has the opposite effect, shrinking the range of severe wear between T_1 and T_2 until at a carbon content of approximately 0.98% mild wear occurs throughout the full range of loads (at a sliding speed of 1 m/s).

(ii) The consequence of changing the initial hardness of the rubbing members by hardening and tempering, for example at a carbon concentration of approximately 0.52% is shown in Figure 6.3. Though the major part of Welsh's work was performed with steels in the annealed condition, Figure 6.3 shows that as the initial hardness of the rubbing members was increased, the severe wear regime, between T_1 and T_2 , was reduced until when the hardness reached a value between 400 and 450 HV 20 kg the severe wear regime disappeared. Mild wear persisting over the whole range of loads, thus illustrating the role of material hardness in wear.

(iii) The influence in lowering the sliding speed was to displace the T_1 and T_2 transitions to a higher load, though within the load range studied (approximately 1 - 150 N) at speeds $\gg$ 0.33 m/s only the T_1 transition was observed. Raising the speed had the opposite effect displacing the T_1 and T_2 transitions to lower loads, such that at a speed of 2.66 m/s only post- T_2 mild wear was observed.

Welsh showed these observations to be in accord with a number of earlier

FIG. 6.4 EFFECT OF EXTERNALLY INDUCED DISC TEMPERATURES ON THE PATTERN OF WEAR OF EN 8 STEEL (DUNKLEY & QUINN (1978))

studies of the wear of steels and more recently Dunkley and Quinn (1978) in studies on the effect of elevated temperatures and speed upon the wear of a En 8 (0.35 - 0.45%C) mild steel have repeated the classical wear patterns of Welsh. Their experiments were performed on a pin on disc apparatus where it was possible to raise the bulk disc temperature to greater than 500°C. Figure 6.4 is taken from their work and shows the equilibrium wear rate of En 8 as a function of load, at a sliding speed of 1 m/s, for various externally induced disc temperatures. It can be seen that the influence of raising the external disc temperature is to displace the T_2 transition to a lower load, until at a temperature of approximately 200°C mild wear (post- T_2) occurs over the range of loads used. It is unfortunate, in terms of this thesis, that no metallographic or physical analysis of the worn pin or discs was performed by Dunkley and Quinn to establish the presence of transformed material, though its presence can probably be inferred from microhardness measurements taken; typically 650 HV at the T_2 transitions. Indeed it is relevant to point out that in their paper (and also in a series of papers of detailed pin and ring studies, carried out by Quinn (1967), (1968), (1978) concerning oxidation wear theory) the conditions may be interpreted as being at or above the T_3 transition, although this point has not been made explicitly by Quinn.

This review is commensurate with the subsequent analysis of this chapter. However the literature contains many detailed papers on various metallurgical aspects of wear (though in respect of hardened surface layers, these have been considered in Chapter 2). Further details may be found in the work of Scott (1966), Eyre (1971) etc.. The influence of temperature on the wear of unlubricated metal surfaces and concomitant

surfaces have been reviewed by several authors in particular by Tenwick (1971) and Hayler (1970). As both authors considered rather different conditions in terms of material, load and speed comparison between their work and that of Dunkley and Quinn is difficult, though Hayler notes that transitions in the sliding process (in respect to En 1A steel) can be shown to be related to the predicted total temperature.

The results and patterns of wear established by Welsh (1965) will now be re-examined. In particular the conditions for the T_2 and T_3 transitions and the region between these transitions will be considered and in what follows the temperatures which are considered are the total transient surface temperatures θ_T i.e. the flash temperature θ_m plus the bulk specimen temperature θ_{bk} .

6.4 The T_2 transition

Figure 5.1 suggests that at total transient surface temperatures below 723°C no phase transformed material is likely to be formed, thus arresting the progress of severe wear. This conclusion is supported by the analysis of the exploratory single rub experiments and the wear experiments of Archard (1959) and Welsh (1965).

The metallurgical considerations of Section 5.2 suggest that, in the temperature range 723-910°C, times very long indeed compared with the duration of a single asperity contact are required to cause the appropriate metallurgical phase changes. Thus at 750°C, a dwell time of the order of 600 seconds is required to cause the complete transformation

of ferrite and a dwell time in excess of 10^4 seconds is required for the complete transformation to homogeneous austenite. At 850°C the corresponding dwell times are 6 seconds (transformation of ferrite) and $2-3 \times 10^3$ seconds (transformation to homogeneous austenite). Therefore phase transformed material cannot be formed as a consequence of a single rub. This is supported by the analysis of the exploratory single rub experiments.

We now turn to the possible influence of flash temperatures in the range $723-910^\circ\text{C}$ upon the wear experiments of Archard (1959) and Welsh (1965). The K factor during severe wear of steels, as derived using equation 6.2, is of the order of 7×10^{-3} giving a value of $1/K$ of the order of 150. Thus the probable maximum high temperature dwell time for a typical element of material on the surface of the wearing surface is of the order of 75×10^{-3} seconds (assuming 150 contacts each with a typical duration of 0.5×10^{-3} seconds). This is shorter than the dwell time required by metallurgical considerations, described in the last paragraph, by two orders of magnitude or more. Therefore, even when the effect of multiple rubs are considered, it is clear that the T_2 transition requires total transient surface temperatures in excess of 910°C .

The flash temperatures calculated in the wear experiments of Archard (1959) and Welsh (1965) are typically in the range $750-800^\circ\text{C}$. If we

- (1) recall that these are average values and remember the important role of the maximum temperature in the temperature/time cycle (as deduced in Chapter 5), and

(ii) include a bulk temperature in the range $100 - 150^{\circ}\text{C}$ which is similar to those measured by Welsh,

it then appears that the effective values of temperatures reached in these experiments could well exceed 910°C . A similar conclusion can be drawn from an analysis of the parts of the crossed cylinders experiments subjected to repeated rubs in the experiments of Chapter 4.

It is interesting to note the results of Kawamoto et al (1973) who studied the wear characteristics of a 0.42% & 0.6% carbon steels rubbing against a 17% Cr stainless steel, on a pin on disc machine, in relation to surface temperature. Flash temperatures were measured by a dynamic thermoelectric means. They observed, both in air and in a vacuum of 10^{-1} torr, a minimum wear rate with loads up to 60 N and sliding speeds up to 7 m/s corresponding to measured flash temperatures in excess of 800°C . Bulk temperatures of the pins, measured at a similar depth to Welsh (3mm), were of the order of 150°C . Again, in terms of this work, it is unfortunate no metallographic analysis of the pins was performed, although the general similarity of their results to those of Welsh imply the formation of hardened surface layers at total temperatures in excess of 910°C .

It remains to be shown that above 910°C the metallurgical transformation could be completed in the 150 rubs typical of severe wear conditions. Assume a temperature of 950°C at which, D , the diffusion coefficient of carbon in austenite, has a value of $1.7 \times 10^{-11} \text{ m}^2/\text{s}$. Assume a typical duration of flash temperature of 0.5×10^{-3} seconds, giving again a total dwell time at the total transient surface temperature of 75×10^{-3}

seconds. Using the arguments of Sections 5.3 and 5.4 we arrive at a representative value of the diffusion distance $\bar{x} = \sqrt{D t}$ of $1.1 \mu\text{m}$ which is significantly larger than the normal spacing of the pearlite laminar structure. Therefore, at total transient surface temperatures immediately above 910°C the conditions are adequate to ensure the production of phase transformed material in repeated rubs and the initiation of the T_2 transition.

6.5 The range between T_2 and T_3

At the T_2 transition the equilibrium wear rate is reduced by a factor of approximately 10^2 so that K is reduced from 7×10^{-3} to 7×10^{-5} and thus the effective number of asperity contacts is increased from, say, 1.5×10^2 to 1.5×10^4 . Immediately above the T_2 transition, mild wear is perpetuated only after an initial period of severe wear during which the metallurgically phase transformed layer is formed. In the mild wear regime there will be a gradual removal of this hard transformed surface layer by wear. Welsh (1965) showed that the perpetuation of mild wear in this regime required the presence of both an oxide film and the hard surface layer. Therefore in the mild wear regime the rubbing conditions are presumably such as to ensure the regeneration of this metallurgically transformed surface layer.

Above T_2 three major changes in the equilibrium rubbing conditions can be identified compared with those below T_2 .

- (i) The reduced value of K means that each point in the surface layers is subjected to a number of rubs which has increased by a

factor of the order of 10^2 . As in the preceding arguments this will correspondingly increase the dwell time to which any element of the surface layer has been subjected to transient high temperatures.

(ii) The reduced value of K (i.e. the increased number of rubs) and the decreased scale of size of the mechanism of wear produces a change in the surface topography which favours elastic rather than plastic deformation of the contacting asperities.

(iii) The increase in the hardness of the surface layers (say, from 200 to 800 HV) may also increase the chances of a regime of elastic deformation.

It is the burden of the argument of this section that the combination of the three factors listed above will barely favour the perpetuation of mild wear. However the assumptions involved in the last of these factors first needs closer examination. Consider a steel of hardness 200 HV rubbing under a load of 20N which is assumed to be close to the T_2 transition. Under severe wear conditions the flash temperature calculations assume that all the load is borne at a single contact under conditions of plastic deformation. With the conditions as defined this contact is a circular area of radius, a , of the order of 50 μm . Consider now an elastic asperity contact area of the same size; the maximum Hertzian shear stress of this contact associated with the production of wear particles is at a depth of the order of $0.5a$, i.e. at a depth of 25 μm . This depth is greater than that of a typical metallurgically phase transformed layer. The conclusion to be drawn from this order of

magnitude calculation is therefore as follows: if, in the mild wear regime immediately above T_2 , the contact conditions are converted to a multiple contact regime with wide dispersal of the load among a large number of contacts then the existence of the hard transformed layer may well have an influence in preventing plastic deformation and a reversion to mild wear. If, however, by mischance, misalignment or any other mechanism of maldistribution of load, it becomes concentrated at a single region or a small number of regions then the hard metallurgically transformed layer will become less effective in preventing the possibility of plastic deformation.

Consider now the influence of the first two factors listed above; i.e. the increased number of rubs and the associated reduction in the likely severity of the contact conditions (wider distribution of the load and increase in elastic rather than plastic deformation). The second of these two factors is likely to reduce the transient surface temperature below the critical value of 910°C but this might be compensated by an increase in the number of rubs to 1.5×10^4 . Consider the transient surface temperature reduced to a value of 850°C as discussed in Section 6.4 above. Using the data of that Section and bearing in mind that a total dwell time is of the order of $1.5 \times 10^4 \times 0.5 \times 10^{-3} = 7.5$ seconds one can see that this is sufficient to cause total transformation of the last ferrite to austenite but unsufficient to produce homogeneous austenite (which requires significantly longer dwell times).

The last two paragraphs have produced a discussion of the mild wear regime immediately above the T_2 transition; this is based purely upon the theoretical aspects of the metallurgical and tribological conditions.

The conclusion drawn from this discussion is that this mild wear regime is essentially precarious. This is also precisely the conclusion drawn from the experimental evidence of Welsh; for example he found that in this regime a reversion to severe wear was produced by either removal of the oxide film or by annealing the hard surface layers below a value of 400 HV. Welsh (1965) thought that this precarious region arose from the difficulty of hardening the transformed layer, whilst Childs (1981) has suggested a cyclic process occurs on a local scale i.e. removal of hardened material by oxidative wear, followed by metallic transfer, hardening and oxidative wear.

6.6 The T_3 transition

Seen in the light of the last section one important feature of the T_3 transition is that it removes the precarious element in the maintenance of a mild wear regime which exists in the T_2/T_3 region. Suggested here are two possible criteria which might satisfy this requirement for the onset of the T_3 regime:

- (i) The production of metallurgically transformed material in a very few (say, one or two) number of rubs.
- (ii) The certainty of the production of phase transformed material as a result of multiple rubs at elastic contacts during the mild wear regime.

The rapid formation of phase transformed material as proposed in (i) will require temperatures well in excess of 910°C (Section 5.3), whilst the

formation of phase transformed material in a very few number of rubs implies that the presence of an oxide layer is now unnecessary for the perpetuation of mild wear.

Support for (i) is given by Welsh (1965), who, if one includes bulk temperatures of between 100 - 150°C, calculates total transient surface temperatures for the T_3 transition in excess of 1000°C.

Criterion (ii) proposes that due to the severity of the conditions the increased depth of the transformed layer is effective in preventing the possibility of plastic deformation and that despite the gradual wear, the K factor is such that renewal of the transformed layer will take place in a small number of rubs at elastic contacts.

It is now possible to provide an explanation for the divergence in the pin and ring wear rates above T_3 . As a result of criterion (i) being reached earlier in the pin (because of its higher bulk temperature) permanent phase transformed material will be formed in the pin first. Thus one would expect (as a result of this) the ring wear rate to increase until permanent phase transformed material is formed on the ring itself. The probably dominant factor will then become the thermal softening of the pin due to its bulk temperature distribution. This will result in a decrease in the ring wear rate and conversely an increase in the pin wear rate.

6.7 The influence of carbon content

Welsh showed that a reduction in the carbon content below the eutectoid

composition caused, amongst other features, a raising of the T_2 transition (Figure 6.2). Without indulging in a detailed discussion, it is very easy to see how this trend can be explained in terms of the approach used here. The production of homogeneous austenite in a eutectoid steel is seen here as occurring above 910°C by the diffusion of carbon within the pearlite structure. Now consider a hypoeutectoid steel (in its normalised state), the austenitisation of the pearlite grains must be followed by production of homogeneous austenite which includes those elements of the material which at room temperature were pro-eutectoid ferrite grains. The diffusion distances involved are therefore significantly larger and higher transient surface temperatures will be required. In this way the T_2 transition will be raised.

Indeed Welsh also showed that the T_2 transition for a steel of a given composition was lowered by the use of the material in a less severely tempered condition i.e. higher hardness values (Figure 6.3). In Welsh's work the implied influence of tempering upon the pattern of behaviour is its direct influence upon hardness. One consequence of the work described in this thesis is that the influence of tempering upon the required diffusion distance also requires consideration. In Figure 6.3 the T_2 transition is displaced to a lower load with increased hardness - up to approximately 400 HV where mild wear exists throughout the load range studied.

6.8 The influence of externally induced bulk temperatures

Dunkley and Quinn (1978) have shown (Figure 6.4) that raising the bulk temperature caused, amongst other features, a lowering of the T_2

FIG. 6.5 EFFECT OF CARBON CONTENT ON THE M_s & M_f
TEMPERATURES OF PLAIN CARBON STEELS
(HUME & ROTHERAY (1966))

transition. In simple terms, again without indulging in detailed discussion, it is easy to see how this trend can be explained as a result of the approach used here. If the production of homogeneous austenite requires temperatures in excess of 910°C (and thus on quenching phase transformed material) an increase in the externally induced bulk temperature will lower the corresponding flash temperatures required for a phase transformation. In terms of Figure 6.4 this results in a lowering of the T_2 transition with respect to load. The situation is more complex, however, by the metallurgical requirement of a minimum temperature, M_s , before the start of the martensite transformation (Figure 6.5). An En 8 steel has a M_s of the order of 400°C , whilst a 0.78% plain carbon steel of the experiments of Chapter 4 has a M_s of approximately 200°C . At first sight it appears therefore that the influence of high externally induced temperatures is limited to low carbon steels. However the effect of applied stress, especially plastic stresses, usually increases the amount and start temperature of the martensitic transformation (Christian (1975)). Conversely if the original phase is cold worked in a temperature range where it is stable i.e. at temperatures sufficiently above M_s but below approximately 500°C , the resultant deformation can inhibit the transformation.

As any point on the rubbing surface (or sub-surface, at depths comparable to that of the phase transformed material) experiences possibly both temperature and deformation cycles, the effect is complicated. However, at this stage, it is sufficient to note that the lowering of the T_2 transition with externally induced bulk temperatures is in accord with the new metallurgical theory presented here.

6.9 Discussion

This chapter has considered the effect of the metallurgical theory of Chapter 5 on the interpretation of Welsh's classical wear patterns and in particular on the T_2 , T_3 transitions and region between these transitions.

Attention has been confined to temperature induced transformations and it has been shown that temperatures in the range 723-910°C are insufficient to cause the T_2 transition since the maximum high temperature dwell time for a typical contact is shorter by orders of magnitude than that required for the production of austenite. Thus at the T_2 transition temperatures in excess of 910°C are implied.

In the range between the T_2 and T_3 transitions the reduced value of K , i.e. the increased number of rubs an asperity can experience, in conjunction with surface topographical and hardness changes suggest that temperatures less than 910°C are sufficient to cause partial austenitisation; however, in terms of this approach, homogeneous austenite requires higher temperatures. The T_3 transition, where oxide films are now unnecessary for the perpetuation of mild wear, have been compared to the single rub experiments of Chapter 4, where, again, temperatures in excess of 910°C were implied.

In the interpretation of wear patterns described in this Chapter the crucial role of the K factor of wear theory will be noted. The interpretation of the K factor as representing the number of temperature cycles to which an element upon the surface has been subjected provides

the essential link between the single rub experiments and the multiple rub conditions of typical wear experiments. It is thought that this is the first time that this feature has been introduced in the interpretation of the production of phase transformed material in multiple rub experiments.

The influence of carbon content, hardness and externally induced bulk temperatures on the wear transitions have all shown to be in broad accord with the metallurgical theory of Chapter 5.

These conclusions give further credence to the new metallurgical theory of the previous chapter. However Chapters 5 & 6 have been based upon some simple exploratory experiments which were performed before the development of the present theories. Detailed interpretation of these exploratory experiments was found to be difficult; therefore, further more detailed experiments have been performed using the crossed cylinders machine. These experiments and the methods used in their interpretation will be described in the following chapter.

CHAPTER 7

EXPERIMENTAL

7.1 Introduction

This chapter concerns the details and techniques used in the author's experiments which form the main experimental evidence of this thesis. These experiments were designed to avoid the limitations in the finer interpretation of the exploratory experiments of Chapter 4.

Techniques employed in the study and identification of phase transformed material are also given. The experimental procedure followed in a crossed cylinders experiment is as described in Chapter 3.

7.2 Materials and specimen preparation

In the present tests plain carbon steels were rubbed against a hardened tool steel, in dry conditions. This combination of materials was chosen for two reasons; firstly, by using differing materials with large differences in hardness, weld formation in both single and multiple rub would be reduced and secondly, by the nature of their metallurgy, the physical properties of high speed tool steels remain practically constant over a large temperature range thereby increasing the validity of using the material properties at room temperature in flash temperature calculations.

<u>Designation</u>	<u>C</u>	<u>Si</u>	<u>Mn</u>	<u>S</u>	<u>P</u>	<u>Ni</u>	<u>Cr</u>	<u>W</u>	<u>V</u>	<u>HV 20kg</u>	<u>Grain Size ASTM</u>
C52	0.52	0.19	0.58	0.043	0.036	0.07	0.07	-	-	206	5-6
C73	0.73	0.27	0.36	0.009	0.010	0.08	0.07	-	-	235	5-6
T1	0.75	-	-	-	-	-	4.0	18.0	1.1	887	9-10

C52 and C73 machined, as received.

T1 hardened after roughing out. Preheated at 850°C, austenitised at 1300°C quenched into oil, double tempered at 560°C for one hour.

Table 7.1. Chemical compositions (% wt), hardness and mean grain size of steels used.

<u>Designation</u>	<u>C52</u>	<u>C73</u>	<u>T1</u>
Modulus of Elasticity E(GN/m)	210	210	203
Density ρ (kg/m ³)	7850	7850	8750
Specific heat c(J/kg K)	480	480	402
Thermal conductivity k(J/msK)	52	52	24
Poisson's ratio ν	0.293	0.293	0.295

Table 7.2 Physical properties of steels used at room temperature.

0.52% and 0.73% plain carbon steels were used, supplied by Firth Brown as 50 mm round bars, in normalised condition and designated C52 and C73 respectively. These are research steels and have no equivalent specification. The high speed tool steel used was BS4659 BT1, designated T1 and supplied by Edgar Allen Balfour in 50 mm round bar. The material compositions and tool steel heat treatment details are shown in Table 7.1. The physical properties of these steels, at room temperature, are given in Table 7.2.

For ease of analysis, hollow cylindrical specimens were manufactured of length 19.05 mm bore 15.875 mm and external diameter 34.92 mm and clamped on mandrels. The specimen manufacture involved roughing out, heat treatment of the tool steel specimens followed by final machining. The specimens were ground to length, the bores were lapped to size to ensure a good fit on the mandrels and the external diameter was circumferentially ground to size, with a surface roughness of approximately $0.3\mu\text{m}$ (i.e. best finish possible with a standard wheel). These specimens are larger than those employed in Chapter 4. In these experiments high bulk temperatures had been inferred by the generation of temper colours on the track surfaces. The effect of bulk temperatures has also been considered in Chapter 5 and its effect on diffusion. It was therefore considered desirable to manufacture specimens with a larger thermal capacity and measure bulk temperature rises by embedded thermocouples. Chromel/Alumel thermocouple wire was used and the thermocouples were located at a depth of 4.76 mm beneath the surface, in the centre of the specimens. Thermocouples were connected to a Comark electronic thermometer linked to a recorder.

FIG 7.1 VIEW OF SPECIMENS, LOCATED ON MANDRELS, MOUNTED ON
CROSSED CYLINDERS MACHINE

In the author's experiments the specimens under test were clamped in the centre of the mandrels and supported by smaller specimens, fibre washers were inserted between each specimen for ease of specimen removal. In this position a loading ratio of 5:1 occurs with an error of $\pm 5\%$ over the length of the specimen. Therefore the loading ratio has been considered to be constant.

Figure 7.1 shows the specimens clamped on the mandrels and in position on machine. The mandrel containing the plain carbon steel specimen was located in the loading arm (upper specimen) and the hardened tool steel specimen was located on the ram mandrel (the lower specimen). A ratio of 20:1 was maintained between the surface speeds of plain carbon steel and hardened tool steel specimens respectively; this resulted in the generated heat distribution within the contact being almost identical to that assumed in Archard's (1959) flash temperature theory. The resultant sliding velocity of the two specimens is therefore equal to the surface velocity of the upper specimen (error of 0.13%). The friction was recorded directly from the strain rod (error of 0.13%).

All experiments reported were carried out in the normal laboratory environment of 45-60% relative humidity and at a temperature of 18-25°C.

7.3 Specimen analysis

The following details of specimen analysis primarily refers to specimens of C52 or C73 in which the formation of phase transformed material was being studied. The opposing T1 specimens were not, in general, analysed.

After an experiment the specimens were removed and visually examined. Single and multiple rub track widths were measured from profiles using a Talysurf 4. As the area of contact between crossed cylinders of equal diameter is a circular contact, a measure of the width of the single contact track will give an estimate of the diameter of the area of contact, corresponding to the test conditions. This assumes elastic deformation. However, if heavy plastic flow occurred (full plasticity) the material at the rear of the contact would not recover and, in the limit, the load would be supported on the front half of the contact which would be semicircular in shape. However, as will be seen later in most of the experiments reported in Chapter 8, the loads only exceeded those required for the onset of plastic flow by factors of 1.5 to 9. (Full plasticity occurs at loads of the order of 200 times the onset of plastic flow (Tabor (1950))). Therefore conditions were only just outside the elastic range. An optical microscope was also used to check track widths. These measurements are of importance for flash temperature calculations.

7.3.1 Metallographic preparation for microscopy

The success of preparation of edges to examine changes which may have occurred locally in very thin layers depends upon the following criteria:-

- (i) The mounting medium must hold all the surface layers intact and prevent spalling.
- (ii) The mounting medium must support the edge to prevent rounding and possible difficulty in optical focussing.

FIG. 7.2 VIEW OF CYLINDRICAL SPECIMEN
DEFINING SECTIONING PLANES

Problems in near surface metallography have recently been discussed by Torrance (1978) and Ahn et al (1979). They conclude that little attention has been devoted in the wear literature to damage and artefacts which can result from inadequate metallographic practice. In particular, in the formation of surface and sub-surface cracks and effects of over-etching.

Three sectioning planes have been used in the present work and are as defined in Figure 7.2. A chordal section through a cylindrical specimen has the same advantages and disadvantages as taper sectioning and gives a mechanical magnification perpendicular to the edge of D/d where D is the diameter of the specimen and d the length of the chord. However, the taper ratio varies as the field viewed is moved away from the edge and thus this method is less easy than normal taper sectioning to use quantitatively (Gifkins (1970)). Indeed the interpretation of taper sections can be difficult unless used in conjunction with normal sections. Normal sections give a better indication of the amount of surface flow taken place, whilst axial and chordal sections give a more representative picture of the depth of transformation and amount of deformation taken place.

Various mounting materials and edge retention techniques were attempted. It was found that for axial sections mounting in "Plastimet" gave reasonable edge retention and support. However for chordal and normal sections the procedure adopted was to nickel plate the surface. Although electrolytic plating processes have some advantages e.g. uniform coating over complicated shapes, an electro-deposition process adopted here gave satisfactory results.

The plating solution consisted of Nickel Sulphate 124 g/l, Ammonium Chloride 37 g/l and Boric acid 25 g/l, in equal volumes. Plating took 10 hrs at 50°C at a current density of 200 A/m² with an applied voltage of 3 volts. Constant agitation with a polythene stirrer improved the plating quality. Prior to immersion in the plating bath, it was found that good adhesion was obtained by cleaning by Soxhlet extraction for approximately 3 hrs. More vigorous forms of cleaning e.g. pickling or ultrasonic degreasing were found unnecessary, indeed these methods can to some extent damage the surface itself. A thick layer of nickel was deposited (about 150 µm) which held the surface layers intact and allowed subsequent rotary disc grinding, during specimen preparation without damage.

The specimens were then mounted in "Epofix", a cold mounting compound and allowed to cure for 24 hrs at room temperature. A cold mounting compound was used for two reasons; firstly normal mounting compounds e.g. bakelite etc., require the application of pressure, typically 10 MPa, at a temperature of, say, 150°C which could modify the structure of the transformed material and secondly, during such mounting the nickel plate occasionally broke loose from the surface of the specimen, probably due to effects of differential expansion.

The specimens were then carefully turned on a lathe, with ample coolant, to provide a damage free surface, followed by rotary grinding with 240, 400, 600 and 1200 silicon carbide papers flooded with water. Three polishing stages were used 6µm, 1µm and a final polish of ¼ µm diamond paste. Once prepared they were etched with 4% Picral and examined on an Olympus BH optical microscope.

7.3.2 Microhardness measurement

Hardness profiles were measured by using a Leitz Miniload Vickers microhardness tester. A load of 25g was used.

7.3.3 Scanning electron microscopy

Scanning electron microscopy can be used at magnifications above those obtainable from an optical microscope. Some specimens were, therefore, further examined by an ISI 60 scanning electron microscope, linked to an energy dispersive X-ray analysis system (EDAX). The mounted specimens were made conductive by evaporating a very fine layer of gold, under vacuum onto their surface, thus preventing electrostatic charges building up and reducing resolution. A comparison between a specimen mounted in conductive bakelite revealed no observable loss in definition between the two methods. The EDAX system allows qualitative identification of elements up to and including sodium in the periodic table. The total emitted X-ray spectrum was recorded between the range 0 to 10 keV. A target spot of 1 μ m in diameter may be analysed with confidence.

7.3.4 Auger electron spectroscopy

In order to identify any changes in the surface elements of phase transformed material, a Varian 10 keV scanning Auger electron spectrometer was made available at Loughborough University. Auger spectroscopy can detect and measure, with accuracy, changes in low

atomic number elements e.g. oxygen, nitrogen and carbon and thus has considerable advantages over electron probe micro-analysis. The spectrometer was used to give a composition depth profile through phase transformed material formed in a single rub and compared against an unworn surface.

The method involved rotating a 10 mm diameter hardened steel ball with suitable diamond paste to fashion a well defined spherical crater in the sample surface. The depth and width of which could be accurately controlled and measured. The specimens were cleaned as thoroughly as possible prior to insertion into the vacuum chamber and then cleaned in situ by argon ion bombardment using a 3 keV Varian scanning ion gun with an ion current density of $75 \mu\text{Acm}^{-2}$, for a short period of approximately 1 minute. Primary beam energies of 3 keV and beam currents of $2 \mu\text{A}$ were used and the beam diameter was approximately $10 \mu\text{m}$. The background pressure in the system during analysis was always less than 5×10^{-9} torr. The energy was scanned over a range of 20 to 1020 eV.

Composition depth profiles for iron, carbon, nitrogen and oxygen were then obtained by both elemental line scans and point to point analyses down the crater wall. Walls, Hall and Sykes (1979) give further details of the ball cratering technique for composition depth profiles using a scanning auger microprobe.

7.3.5 Transmission electron microscopy

A conventional 'Denison' pin on disc machine, was used to produce phase transformed material in multiple rubbing for analysis by transmission electron microscopy. No analysis of phase transformed material produced on the crossed cylinders specimens was attempted because of difficulties in the production of foils from the surfaces of curved specimens.

Pins of length 32 mm, having a final diameter of 3 mm, were machined, as received, from C73 and loaded against an 82 mm diameter hardened high speed tool steel disc manufactured from T1. The disc was prepared by unidirectional grinding using a standard wheel and both pin and disc surfaces were abraded by a 400 grit silicon carbide paper to provide a standard pre run finish, having a Ra of approximately 0.2 μm . Prior to run both pin and disc were degreased in a 50:50 solution of acetone and iso-propyl alcohol. Tests were performed at a speed of 2m/s and at a load of 29.5N. Tests were run for 15 minutes before removal, microhardness measurements indicating if transformed material was present on the surface of the pin. One pin was nickel plated, as before, and mounted in "Epofix" at an angle of 6° to the vertical to produce a taper magnification of 10:1.

For examination by electron microscopy a microslice was used to cut sections, parallel to the surface of worn pins, of approximate width 0.2mm. By suitable masking of the worn region, using PTFE tape

thinning was allowed on a "Metalthin" automatic polisher from one direction only. Comparison foils were also prepared from an unworn pin and a conventional martensitic pin. The electrolyte used was 5% Perchloric acid in methanol. The operating voltage was around 45V and was determined by locating a current plateau at 70-90 mA. Foils were examined in a Joel 100 CX transmission electron microscope operating at 100 kV.

7.3.6 X-ray diffraction analysis

As little, if any, debris was collected during a typical run, no X-ray analysis of the wear debris was attempted. However with the techniques available at Leicester University, phase transformed material formed in multiple rub was gently abraded from the surface of C73 specimens and compared against unworn material removed from the surface of an identical specimen in a similar manner. A PW1010 Philips Diffractometer was used, the abraded material being irradiated with monochromatic Cu K α radiation. The K β component was suppressed with a Ni filter. The ASTM powder file and direct comparison with standards were used in the identification of chart peaks. The Bragg relation $n\lambda = 2d \sin \theta$, where n is an integer, λ the radiation wavelength (Cu K α = 0.154178 nm) and θ the angle, enabled determination of interplanar spacings (d values) (Cullity (1959)).

The full results of the crossed cylinders machine experiments and subsequent metallographic analysis performed, is presented in the following Chapter 8.

CHAPTER 8

RESULTS

8.1 Introduction

This chapter is broadly split into two parts. The first part describes the main results of the experiments in which soft 0.52% and 0.73% plain carbon steels rubbed against a hardened tool steel, in dry conditions, on a crossed cylinders machine in single and multiple contacts. The major objective was to determine the conditions (i.e. loads and speeds) under which phase transformed material was produced in a single rub, under conditions where flash temperature theory could be used with confidence; i.e. compared to the exploratory experiments described in Chapter 4, one was concerned with the avoidance of large macrowelding in the single rub track and a uniform generated temperature distribution within the sliding contact.

The second part of this chapter gives the results of supplementary experiments concerned with the identification and nature of phase transformed material using various techniques of physical analysis.

8.2 Crossed cylinders machine experiments

The full experimental conditions and results obtained in rubbing specimens of C52 and C73 plain carbon steels against a hardened high speed tool steel, T1, under dry conditions are summarised in Table 8.1. This shows for each run performed the contact load W (assumed constant),

Run	W(N)	V _{SL} (m/s)	μ_s	μ_r	θ_{bk} °C	2a _a (mm)	Single Rub P.T.M.	Multiple rub P.T.M.
25	250	3.8	0.34-0.54	0.24-0.83	31-35	0.54	No/Yes	Yes
26	250	2.9	0.11-0.2	0.6-0.7	36-40	0.62	No	Yes
27	250	3.5	0.2-0.29	0.5-1.3	26-32	0.61	No	Yes
28	152	5.5	0.34-0.42	0.35-0.56	27-32	0.46	No/Yes	Yes
29	152	4.6	0.45	0.35-0.56	23-28	0.44	Yes	Yes
30	152	3.8	0.64	0.56-0.8	16-21	0.43	Yes	Yes
31	152	3.8	0.34-0.35	0.37-0.7	21-27	0.46	No	Yes
32	643	3	0.26-0.32	0.26-1.1	21-31	0.94	No/Yes	Yes
33	643	2.5	0.31-0.46	0.31-1	26-38	1.04	No/Yes	Yes
34	643	2	0.38-0.54	0.38-0.95	28-35	1.00	No/Yes	Yes
35	790	0.8	0.22-0.37	0.56-1	25-31	1.08	No	Yes
36	790	2	0.56-0.59	0.37-0.9	25-33	1.02	Yes	Yes
37	790	1	0.31-0.37	0.3-0.87	24-27	1.05	No	Yes
38	790	1.5	0.56-0.59	0.2-0.7	25-27	1.01	Yes	Yes
39	790	1.1	0.57-0.67	0.5-1	23-26	1.01	Yes	Yes
40*	348	3.8	0.28-0.46	0.42-2.2	26-30	0.76	No/Yes	Yes
41*	348	3	0.28-0.35	0.3-1.3	15-22	0.72	No	Yes
42*	348	2	0.35-0.4	0.42-0.5	27-31	0.71	No	Yes

Table 8.1 Cont. Experimental conditions and results of main crossed cylinders experiments. * denotes runs using specimens manufactured from C52.

Run	W(N)	V_{SL} (m/s)	μ_s	μ_r	θ_{bk} °C	$2a_a$ (mm)	Single Rub P.T.M.	Multiple rub P.T.M.
1	250	1	0.1-0.4	0.2-2	30-31	0.60	No	Yes
2	348	1	0.3-0.53	0.3-1.9	23-25	0.68	No	Yes
3	446	1	0.27-0.6	0.27-1.8	26.5-2.8	0.84	No	Yes
4	544	1	0.23-0.75	0.4-2	23.5-25	0.92	No/Yes	Yes
5	692	1	0.28-0.6	0.3-1.5	26-27	0.85	No/Yes	Yes
6	839	1	0.32-0.64	0.3-1.5	24.5-28	1.20	No/Yes	Yes
11	348	1	0.17-0.3	0.2-0.66	21-25	0.70	No	Yes
12	495	1	0.16-0.27	0.16-1	22-28	0.76	No	Yes
14	741	1	0.08-0.24	0.25-1.1	21-33	1.00	No	Yes
15	348	1.5	0.1	0.2-1.2	22-30	0.70	No	Yes
16	495	1.5	0.13	0.35-0.75	20-28	0.80	No	Yes
17	741	1.5	0.48	0.4-1.1	26-44	0.92	Yes	Yes
18	741	1	0.3-1	0.4-0.8	22-30	1.0	No/Yes	Yes
19	250	0.5	0.7-1	0.6-1.1	24-28	0.54	No	No
20	348	3.8	0.28-0.3	0.4-0.85	28-37	0.68	No	Yes
21	348	4.6	0.25-0.34	0.3-0.92	24-32	0.66	No/Yes	Yes
22	348	5.5	0.25-0.34	0.4-0.7	28-30	0.64	No/Yes	Yes
23	250	5.5	0.34	0.5-0.6	29-39	0.50	Yes	Yes
24	250	4.6	0.34-0.39	0.24-0.83	25-31	0.56	No/Yes	Yes

Table 8.1 Experimental conditions and results of main crossed cylinders experiments.

(a)

(b)

FIG 8.1 FRICTION TRACES FROM RUNS 2 and 3

the resultant sliding speed of the specimens V_{s1} (equal to the surface velocity of the plain carbon specimens) and the variations in the coefficient of friction in single rub μ_s and in the multiple rub μ_r . Also given is the change in the bulk temperature of the plain carbon steel specimens θ_{bk} , the average value of the measured single contact track width, $2a_a$, and an indication of whether, on metallographic sectioning, phase transformed material was found in single or multiple rub tracks. In some specimens the single rub tracks analysed gave both results; i.e. parts of the track were found with and parts of the track were found without phase transformed material (for example runs 3-6). Runs 1-39 refer to experiments where specimens of C73 were rubbed against specimens of T1 whilst runs 40-42 concerns experiments where specimens of C52 were rubbed against specimens of T1.

Before a detailed description of the results, certain features of runs 1-19 will first be discussed; this discussion should give further credence to the main results of the later runs 20-42. In runs 1-19 various experimental problems were encountered, the effect of differing cleaning procedures and alternative methods of measuring the displacement of the frictional strain rod were attempted.

8.2.1 Preliminary results, runs 1-19

In runs 1-6 the contact load was varied between 250 to 839 N at a sliding speed of 1 m/s. Frictional displacements of the strain rod were recorded by a Talysurf head connected to a Talysurf 3 recorder. Figure 8.1 shows the friction traces from runs 2 and 3 and are representative of the friction variations found. Several features

may be identified. First the coefficient of friction in multiple rub μ_r varies between wide limits e.g. in run 2 between values of 0.3 to 1.8 and μ_r is generally greater than the friction coefficient in single rub μ_s . A factor of 2 is typical. Although the single rub portion of the friction trace is clearly defined, both in this portion and in multiple rub a cyclic variation on the friction trace appears to be superimposed. This was found to be due to the effect of the specimens being slightly eccentric (± 0.025 mm) on the mandrels. This was caused by poor relocation of the specimens onto the mandrels, after the specimens, once ground, had been removed for storage. The procedure adopted for subsequent tests was not to remove the specimens after grinding and therefore each set of specimens was ground true (i.e. concentric to the running centres) and to size prior to each run.

To increase the sensitivity of the frictional system, in order to measure the frictional variation in single rub, in runs 7-10 a "Distac" non contacting eddy current probe was used to measure the displacements of the strain rod. The probe has a frequency response up to 50 kHz and the output was displayed on a Brush UV recorder. The sliding speed was 1 m/s and two runs were each made at loads of 250 and 348 N.

No analysis of the results of runs 7-10 was performed as it was found that measuring frictional displacements with a high frequency response probe gave problems both in the repeatability and interpretation of the recorded friction. In both sets of runs with identical conditions, the recorded friction traces were markedly

FIG 8.2 FRICTION TRACE FROM RUN 10, USING 'DISTAC' PROBE

FIG 8.3 FRICTION TRACE FROM RUN 15, USING TALYMIN PROBE

different. Indeed the character of the recorded frictional variations in multiple and single rub appeared to differ. Figure 8.2 shows a section of the friction trace from run 10. In multiple rub μ_r varies between values of 0.28 to 0.78, whilst in single rub μ_s varies between values of 0.02 to 0.86. This can be compared with the friction trace, using identical conditions, from run 2 (Figure 8.1) using a Talysurf head. As this method also involved re-calibration of the probe prior to each run, it was no longer used.

Runs 11-16 used similar conditions to runs 1-6, although in runs 15-16 the sliding speed was increased to 1.5 m/s. Run 13 is not listed in Table 8.1, as the applied ram speed was insufficient, resulting in overlapping tracks in single rub. As originally utilised by Archard (1961) a Talymin probe was used to measure the displacements of the strain rod (frequency response approximately 200 Hz).

A typical trace is shown in Figure 8.3, taken from run 15. The frictional variations are more clearly defined and in this run μ_s is practically a constant at a value of 0.1. The cyclic variations in single rub, observed in runs 1-6, are no longer present and in general $\mu_s < \mu_r$. However μ_r still varies between wide limits (between 0.2-1.3). From Table 8.1 two general features may be identified from the results of runs 11-16. First, compared to runs 1-6 much lower values of μ_s (and to some extent μ_r) were measured, indicating the effect of specimen eccentricity on the recorded value of μ . Second, as a probable consequence of this, phase transformed material was not found in single rub, though as in runs 1-6 it was found in multiple rub. Runs 11-16 had been cleaned in

the usual manner i.e. Soxhlet extraction for approximately 3 hrs. Therefore in an attempt to increase μ_s and consequently produce phase transformed material in single rub, the specimens in runs 17-19 were first cleaned ultrasonically in a solution of 5% Teepol in distilled water at 40°C, followed by rinses with distilled water and with acetone, prior to Soxhlet extraction.

Table 8.1 shows that for runs 17-19, with contact loads of 741 or 250 N and sliding speeds of 1.5, 1 and 0.5 m/s respectively, that μ_s was increased, up to a value of 1 in runs 18 and 19. Subsequent metallographic analysis found phase transformed material in the single rub tracks of runs 17 and 18. However in all runs, in single rub, gross or macrotransfer of material from the C73 plain carbon steel specimens to the hardened tool steel specimen occurred. (This feature will be further described later in this chapter).

Archard's (1959) flash temperature theory explicitly assumes no damage or transfer at the interface and as such the application of flash temperature theory to the results of runs 17-19 is subject to error.

8.2.2 Main results, run 20-42

We shall now consider, in detail, the results of runs 20-42, which form the major experimental evidence of this work. Where possible features of the preliminary results will be included. Runs 20-42 were performed using a Taymin probe to measure the frictional displacements and a cleaning procedure of Soxhlet extraction for

(a)

(b)

(c)

FIG 8.4 FRICTION TRACES FROM (a) RUN 27 (b) RUN 32 and (c) RUN 37

approximately 3 hrs was used. For specimens of C73 contact loads of 152, 250, 348, 643 and 790 N were used with sliding speeds up to 5.5 m/s, comparative tests were made with specimens of C52 at a load of 348 N with sliding speeds up to 3.8 m/s.

8.2.3 Friction traces

Representative friction traces from the main results are shown in Figure 8.4. These are traces from runs 27, 32 and 37 with contact loads of 250, 643 and 790N respectively. The coefficient of friction in single rub μ_s for each run is clearly shown. The pattern of frictional variations is similar to that observed in the preliminary results. Over a load range of 152 to 790N the coefficient of friction in multiple rub μ_r is generally greater than in single rub μ_s . Also as before the variations in μ_r are much greater than the variation in μ_s . In runs 20-42 values of μ_s typically lie between 0.25 and 0.45 whilst μ_r varies typically between values of 0.26 to 1.3. However in runs 36, 38 and 39 with the heaviest contact load used (790N) values of μ_s up to 0.67 were recorded. Subsequent examination of the specimens showed, as in runs 17-19, gross or macrotransfer from the plain carbon to the hardened tool steel specimens. This also occurred in run 30 where the cleaning procedure was accidentally allowed to continue for 5 hrs, resulting in a value of μ_s of 0.64.

→
Tracks →
→

1mm

FIG 8.5 RUN 27, POLISHED SINGLE RUB TRACK, NOTE MICROWELD IN TRACK

→
Tracks
→

0.5mm

FIG 8.6 RUN 23, NON-POLISHED SINGLE RUB TRACK

(a)

(b)

FIG 8.7 TALYSURF TRACES THROUGH (a) A POLISHED SINGLE TRACK AND (b) A NON-POLISHED SINGLE RUB TRACK ON PLAIN CARBON STEEL SPECIMENS

1mm

FIG 8.8 RUN 27, SINGLE RUB TRACK ON HARDENED T1 SPECIMEN, CORRESPONDING TO A POLISHED TRACK ON OPPOSING SPECIMEN. NOTE SMALL TRANSFERRED PARTICLE DUE TO MICROWELD

1mm

FIG 8.9 RUN 23, SINGLE RUB TRACK ON HARDENED T1 SPECIMEN, CORRESPONDING TO A NON-POLISHED TRACK ON OPPOSING SPECIMEN

(a)

(b)

FIG 8.10 TALYSURF TRACES ACROSS SINGLE RUB TRACKS OF T1 SPECIMENS WHEN OPPOSING SINGLE RUB TRACK EXHIBITS (a) A POLISHED APPEARANCE (b) A NON-POLISHED APPEARANCE

8.2.4 Track appearance

A complicated pattern in the appearance of the wear tracks was exhibited. In single rub the tracks of the C52 and C73 specimens were broadly classified as either "polished" in appearance for example as in Figure 8.5 (from run 27) or "non-polished" in appearance as for example shown in Figure 8.6 (from run 23). Talysurf traces across the tracks are shown in Figure 8.7a and b respectively. Although the contact load in both runs was identical (250N) there is a marked difference in the profiles. In Figure 8.7a the polished track is comparatively smooth with respect to the original surface roughness of the specimen and some indentation by the hardened T1 specimen has occurred. However in Figure 8.7b the non-polished track is rough as compared to the specimen surface, again some indentation has occurred but the edges of the track are above the general level of the surface; rather more severe plastic flow is indicated.

A similar difference in appearance on the single rub tracks of the T1 specimens occurred. Figure 8.8 shows a track, for example from run 27, corresponding to a polished single track on the opposing surface. Little transfer or surface modification is indicated. Figure 8.9 shows a track, for example from run 23, corresponding to a non-polished single track on the opposing surface where some surface modification is apparent. Talysurf traces across these tracks are shown in Figure 8.10 a and b. Figure 8.10a confirms that, when the opposing single track is polished in appearance, practically no transfer and little modification occurs in the single

1mm

FIG 8.11 RUN 34, ALTERNATING POLISHED AND NON-POLISHED
SINGLE RUB TRACK

track of the T1 specimen. Conversely when the opposing single track is non-polished in appearance, Figure 8.10b shows that material is worn or removed in single rub from the T1 specimens.

Figures 8.5 and 8.6 when compared to Figures 4.1 and 4.6 also show that, using the present combination of materials, no large or macrowelds occur in single rub. Thus macrowelding, as observed in the exploratory experiments, is not a prerequisite for the formation of phase transformed material. However some microwelds were observed to occur in the single rub track e.g. as in Figure 8.5, resulting in a small volume of material transferred to the T1 specimens. Figure 8.8 shows the opposing track of Figure 8.5 with a small transfer of material corresponding to the microweld. The propensity of microwelding in single rub tended to increase with increasing load. Friction peaks measured in single rub could often be correlated to microwelds formed in the track.

Some specimens exhibited, in single rub, both areas of polished and non-polished track e.g. Figure 8.11 from run 34, in a similar fashion to the appearance or non-appearance of phase transformed material in single rub, in a particular specimen. Indeed subsequent metallurgical analysis showed that a polished track could be correlated with no changes in the surface layers, whilst a non-polished track could be correlated to the formation of transformed material. It was therefore possible from track appearance alone to deduce if and where transformed material had been formed in single rub.

(a)

1mm

(b)

1mm

FIG 8.12 RUN 36 (a) GROSS TRANSFER TO T1 SPECIMEN
(b) GROSS TRANSFER FROM C73 SPECIMEN

Multiple
Rub

Single
Rub

(a)

1mm

Multiple
Rub

Single
Rub

(b)

1mm

FIG 8.13 SINGLE AND MULTIPLE RUB TRACKS (a) RUN 27
(b) RUN 40

(a)

3mm

(b)

1mm

FIG 8.14 MULTIPLE RUB TRACKS ON T1 SPECIMENS SHOWING
(a) TRANSFER TO AND (b) SEVERE DAMAGE FROM
PLAIN CARBON STEEL SPECIMENS

A (relatively rare) example of gross or macrotransfer of material from the soft C73 specimens to the hardened T1 specimens, in single rub, is illustrated in Figure 8.12a, taken from run 36. The transferred material appears "smeared" in the resultant sliding direction of the specimens. The subsequent track on the opposing surface of the C73 specimen is shown in Figure 8.12b.

A comparison between single and multiple rub tracks is indicated in Figure 8.13a and b, taken from run 27 (polished single track) and from run 40 (non-polished single track) respectively. The multiple rub tracks are larger and more severe, than in single rub, in terms of width and surface appearance. The opposing multiple rub tracks on the T1 specimens showed, in general, either adhesive transfer of material from, or, usually at higher sliding speeds, severe damage by the plain carbon specimens. An example of adhesive transfer in multiple rub is illustrated in Figure 8.14a, taken from run 35, where the transferred material appears "plate-like" but it is thought that this is due to the relative motions of the specimens, rather than a particular wear mechanism. Figure 8.14b shows a severely worn multiple rub track, taken from run 40.

8.2.5 Specimen bulk temperature

Measured rises in bulk temperature of the plain carbon specimens, for runs 1-42, are shown in Table 8.1. It can be seen that rises in bulk temperature were in general small. The maximum increase during a run was 18°C (in run 17). In all runs performed the bulk specimen temperature measured at the end of the single rub did not exceed a

max. of 44°C (run 17). The rise in temperature was primarily dependent on the length of time allowed for multiple rubbing when the specimens were initially loaded, prior to the ram traverse. Single rub had only a small influence on the total rise in bulk temperature measured during an experiment.

8.2.6 Contact size measurement

Archards (1959) flash temperature theory assumes a circular heat source of area A , where $A = \pi a^2$, a being the radius of contact. The theoretical temperature equations are expressed in terms of the contact radius and in the present tests four estimates of the contact radius have been made for each run. These are defined as:-

1. a_p , where fully plastic conditions are assumed i.e.

$$a_p = \left(\frac{W}{\pi H} \right)^{1/2}$$

2. a_e , where Hertzian contact conditions are assumed i.e.

$$a_e = 1.11 \left(\frac{WR'}{E} \right)^{1/3}$$

3. a_g , estimated from the interpolation of Figure 2.10 for contact loads between W_1 and W_3 .

4. a_a , half the average measured value of the single contact track width.

Run	Contact radius (mm)			
	a_p	a_e	a_g	a_a
1	0.185	0.24	0.255	0.30
2	0.22	0.265	0.29	0.34
3	0.25	0.29	0.33	0.42
4	0.27	0.31	0.365	0.46
5	0.30	0.33	0.39	0.425
6	0.34	0.35	0.42	0.60
11	0.22	0.265	0.29	0.35
12	0.26	0.30	0.34	0.38
14	0.32	0.34	0.41	0.50
15	0.22	0.265	0.29	0.35
16	0.26	0.30	0.34	0.40
17	0.32	0.34	0.41	0.46
18	0.32	0.34	0.41	0.50
19	0.185	0.24	0.255	0.27
20	0.22	0.265	0.29	0.34
21	0.22	0.265	0.29	0.33
22	0.22	0.265	0.29	0.32
23	0.185	0.24	0.255	0.25
24	0.185	0.24	0.255	0.28
25	0.185	0.24	0.255	0.27
26	0.185	0.24	0.255	0.31
27	0.185	0.24	0.255	0.305
28	0.145	0.20	0.21	0.23
29	0.145	0.20	0.21	0.22
30	0.145	0.20	0.21	0.215
31	0.145	0.20	0.21	0.23
32	0.30	0.325	0.375	0.47
33	0.30	0.325	0.375	0.52
34	0.30	0.325	0.375	0.50
35	0.33	0.35	0.42	0.54
36	0.33	0.35	0.42	0.51
37	0.33	0.35	0.42	0.525
38	0.33	0.35	0.42	0.505
39	0.33	0.35	0.42	0.505
40*	0.235	0.265	0.29	0.38
41*	0.235	0.265	0.29	0.36
42*	0.235	0.265	0.29	0.355

Table 8.2 Table of contact radii for four estimates of contact conditions for runs 1-42. *denotes C52 specimens.

These results are presented for runs 1-42 in Table 8.2. At least 10 readings were used to calculate a_a , both from Talysurf profiles and optical measurements. The results had a maximum scatter of ± 0.03 mm from the average value of a_a . It can be seen from Table 8.2 that, in all runs, a_a is greater than that calculated assuming elastic, a_e , or plastic, a_p , conditions and even larger than from the interpolated value of contact radius, a_g .

In general, examination of Table 8.2 shows that a_a exceed a_g by only small margins for experiments at the lower loads; the discrepancy between the two values increases with load. Track widths in multiple rub varied considerably and were typically of the order of 1.5 to 2 times the width of the single rub track. As this width in practice corresponds to an apparent area of contact, its use in flash temperature calculations would be incorrect. Therefore these measurements have not been given.

8.2.7 Flash temperature calculations

The appearance of phase transformed material in a single rub has been calculated in terms of the total temperature at the surface, θ_T , where θ_T is defined as the mean calculated flash temperature θ_m plus the bulk specimen temperature θ_{bk} . The full results may be found in Appendix 3. No reductions in the value of the mean calculated surface flash temperature were necessary as the conditions of the experiments corresponded to conditions well above the lower limits of applicability when using the high speed equations (i.e. in a typical experiment $L > 20$).

FIG. 8.16 APPEARANCE OF PHASE TRANSFORMED MATERIAL IN TERMS OF TOTAL SURFACE TEMPERATURE ASSUMING PLASTIC DEFORMATION AT CONTACT.

FIG. 8.17 APPEARANCE OF PHASE TRANSFORMED MATERIAL IN TERMS OF TOTAL SURFACE TEMPERATURE ASSUMING ELASTIC DEFORMATION AT CONTACT

FIG. 8.18 APPEARANCE OF PHASE TRANSFORMED MATERIAL IN TERMS OF TOTAL SURFACE TEMPERATURE FOR THE GRAPHICAL ESTIMATE OF CONTACT SIZE.

FIG. 8.19 APPEARANCE OF PHASE TRANSFORMED MATERIAL IN TERMS OF TOTAL SURFACE TEMPERATURE FOR THE MEASURED CONTACT SIZE

In these calculations the material properties at room temperature were used. In general the maximum value of the friction coefficient in single rub μ_s has been used, however where the single rub track exhibited an alternating appearance or where large variations in μ_s occurred, calculations were also performed with the minimum value of μ_s . The total surface temperatures are calculated in terms of the four estimates of contact radius defined previously in Section 8.2.6.

The appearance or absence of phase transformed material in single rub has been plotted graphically for runs 1-39 (Figures 8.16-19), for the four estimates of contact radius, in terms of the contact load and the calculated total surface temperature. On each graph a tentative dotted line has been drawn corresponding to the transition point. From the limited number of results it appears that for the production of transformed material in a single rub that:-

(i) Assuming plastic conditions (i.e. a_p) total surface temperatures of 1400°C or greater are required.

(ii) Assuming Hertzian elastic conditions (i.e. a_e) total surface temperatures of 1030°C or greater are required.

(iii) from the interpolated conditions of Figure 2.10 (i.e. a_g), total surface temperatures of 950°C or greater are required.

(a)

20μm

(b)

20μm

FIG 8.20 AXIAL SECTION OF C73 SPECIMEN, SHOWING PHASE TRANSFORMED MATERIAL IN (a) SINGLE RUB AND (b) MULTIPLE RUB

(iv) from measured track widths (i.e. a_d), total surface temperatures of 750°C or greater are required.

A discussion and interpretation of these conclusions will be given in Chapter 9.

For the production of transformed material in a single rub for a 0.52% plain carbon steel, runs 40-42 suggest that the same order of magnitude of total surface temperatures are required as compared to runs 1-39. It is difficult to calculate temperatures specifically for transformed material formed in the multiple rub tracks and at present they have not been performed. Several reasons have already been stated; first the large variation in μ_r , second the contact size can only be estimated and thirdly and most importantly in multiple rub transformed material has been formed by an unknown number of rubs of unknown severity.

8.2.8 Appearance of phase transformed material

Axial microsections through phase transformed material formed in single and multiple rub are shown in Figure 8.20a and b respectively, taken from run 6 (load 790 N). In this example, in single rub the depth of transformed material is approximately 10 μm , whilst in multiple rub it has a depth of approximately 20 μm . In both the transformed layer is featureless and it is possible to recognise a transition region between this and the substrate, where, in this example particularly in multiple rub some degree of sub-surface plastic deformation is apparent.

(a)

100 μm

(b)

25 μm

FIG 8.21 CHORDAL SECTION OF C73 SPECIMEN, SHOWING PHASE TRANSFORMED MATERIAL IN SINGLE RUB, LOAD 250N

(a)

200 μm

20 μm

(b)

40 μm

5 μm

FIG 8.22 CHORDAL SECTION OF C73 SPECIMEN, SHOWING PHASE TRANSFORMED MATERIAL IN SINGLE RUB, LOAD 790N

Figures 8.21 and 8.22 are chordal microsections through phase transformed material found in single rub at loads of 250 and 790 N respectively (taken from runs 25 and 38) and are representative of the features observed. Details of the transformed material and transition region are more easily visible. At low loads (Figure 8.21) transformed material is formed throughout the width of the single rub track, with the minimum depth of transformed material being at the edges and conversely the maximum depth being at the centre of the track. At higher magnifications, the featureless appearance of transformed material is revealed, having in this example, a maximum depth of 5 μm . At depths between 5-15 μm , corresponding to the transition region, it is possible to observe carbide lamellae (from the pearlite) gradually disappearing into the transformed material. No pro-eutectoid ferrite is found in the transformed layer and is only generally found at positions corresponding to the max. depth of the transition region. It is suggested that, in single rub, the transition region can be better referred to as "partially transformed region" as it appears that this region corresponds to material where the $\alpha \rightarrow \gamma \rightarrow \alpha'$ (i.e. ferrite $\rightarrow$ austenite $\rightarrow$ martensite) phase transformations have occurred but the carbide lamellae have had insufficient time to dissolve during the γ phase. Also from the state of the partially dissolved carbide lamellae little sub-surface deformation is indicated in this region. Of note in Figure 8.21b is that between the transformed layer and nickel plate some material can be seen adhering to the surface of the track. Subsequent analysis (reported later in this chapter) revealed that this material was from the hardened T1 specimens.

FIG 8.23 CHORDAL SECTION OF C73 SPECIMEN SHOWING PHASE TRANSFORMED MATERIAL IN MULTIPLE RUB, LOAD 790N

FIG 8.24 EMBEDDED PARTICLE FROM HARDENED T1 SPECIMEN, IN MULTIPLE RUB. RUN 25

At heavier loads (Figure 8.22a) the transformed material was not generally found to be formed completely across the width of the single rub track. Although in Figure 8.22a the region of transformed material is displaced towards one edge of the track, as before the maximum depth of transformed material (approximately 10 μm) corresponds roughly to the centre of the track. This displacement being a probable consequence of the motion of the specimens and the extent of the surface and sub-surface plastic deformation occurring. At higher magnifications (Figure 8.22b) the extent of the sub-surface plastic deformation is apparent, indeed the pro-eutectoid ferrite grains are now within the partially transformed region. In this example no Tl material was found adhering to the single rub track surface.

Typical chordal microsections of transformed material found in multiple rub are shown in Figure 8.23 and 24 taken from runs 38 and 25 respectively. The picture is significantly different. The depth of transformed material varies considerably, between 4 and 30 μm for example (Figure 8.23a) and is neither uniform in depth nor continuous in width with respect to the track dimensions. The boundary with the substrate is often sharp and there is considerable evidence of sub-surface plastic deformation. As in single rub, pro-eutectoid ferrite was not found in the transformed material, however several dark "striations" were observed (Figure 8.23b). These features appear to be carbides coalescing in a tempering effect due to repeated applications of frictional heat. Figure 8.24 shows transformed material in multiple rub where a particle of

FIG 8.25 CHORDAL SECTION OF C52 SPECIMEN, SHOWING PHASE TRANSFORMED MATERIAL IN SINGLE RUB, LOAD 348N

FIG 8.26 CHORDAL SECTION OF C52 SPECIMEN, SHOWING PHASE TRANSFORMED MATERIAL IN MULTIPLE RUB, LOAD 348N

(a)

30 μ m

(b)

60 μ m

FIG 8.27 RUN 22, NORMAL SECTION THROUGH TRANSFORMED MATERIAL
IN (a) SINGLE AND (b) MULTIPLE RUB

(a)

80 μm

(b)

160 μm

FIG 8.28 CHORDAL SECTION SHOWING PHASE TRANSFORMED MATERIAL FORMING AT CENTRE OF SINGLE RUB TRACK.
(a) C52, RUN 40 (b) C73, RUN 34

10μm

100μm

(a)

25μm

25μm

(b)

FIG 8.29 CHORDAL SECTION OF C73 SPECIMEN, THROUGH MICROWELD
IN SINGLE RUB, REVEALING PHASE TRANSFORMED MATERIAL
(RUN 11)

hardened T1 material is embedded. Embedded particles were often observed in specimens run at high sliding speeds (>3 m/s).

Figures 8.25 and 8.26 show chordal microsections through phase transformed material formed in single and multiple rub in specimens of C52, taken from run 40. The general features observed in Figures 8.21 to 8.23 were repeated. In single rub pro-eutectoid ferrite was again generally found only at the maximum depth of the partially transformed region, with the transformed material (of depth approximately $4\text{ }\mu\text{m}$) being featureless. Although not as clear as Figure 8.21a the maximum depth of transformed material also occurred at the centre of the track. In multiple rub the depth of transformed material varied between $2\text{--}16\text{ }\mu\text{m}$ and, as before, was neither uniform in depth nor continuous in width across the track. Similar tempering features were also observed (Figure 8.26b).

Figure 8.27 presents two normal sections through phase transformed material formed in single and multiple rub, from run 22. As one would expect forward plastic flow has occurred in both tracks, although the extent of this is considerably more severe in multiple rub.

Figure 8.28 shows single rub tracks from C52 and C73 specimens where phase transformation has just occurred. In both examples this corresponds to the centre of the track. In run 11 a peak of friction in single rub of 0.56 was measured, which was later found to correlate to a microweld in the track. A subsequent chordal microsection (Figure 8.29) shows the formation of transformed material associated with this microweld.

8.3 Identification and nature of phase transformed material

8.3.1 Etch response

As a first stage in the identification of transformed material, the effects of various etchants were examined. A review of the literature indicates little work has been done on this aspect. Tables of etchants and their effects are available (Smithells (1976), Kehl (1949)). The results are summarised in Appendix 4, the etchants being broadly classified as satisfactory or unsatisfactory depending on their response and ease of use. The general conclusion to be drawn from Appendix 4 is that phase transformed material is primarily a small grained martensitic structure with a fine dispersion of carbides. In multiple rub these carbides can coalesce and are consequently more easily etched. Indeed, reagent No 6 coloured transformed material blue, both in single and multiple rub (indicating martensite) with varying amounts of orange (indicating cementite). Pro-eutectoid ferrite was unaffected. A comparison martensitic specimen obtained from the same steel by conventional means produced a deep blue colour with this reagent without any orange. Some etches specifically designed to reveal martensite and decomposition products of martensite had little success; moreover since these also revealed little, if any, of the pearlite structure of the substrate they can be considered to be generally unsuitable.

FIG. 8.30 MICROHARDNESS / DEPTH PROFILES THROUGH TRANSFORMED MATERIAL
ON C73 SPECIMEN

FIG. 8.31 MICROHARDNESS/ DEPTH PROFILES THROUGH TRANSFORMED MATERIAL
ON C52 SPECIMEN

8.3.2 Microhardness measurements

Microhardness measurements and depth/microhardness profiles were made through several sections of transformed material in single and multiple rub from both C73 and C52 specimens. Typical depth/microhardness profiles from a C73 specimen are presented in Figure 8.30 and indicate that in single and multiple rub transformed material is extremely hard, approximately 1100-1200 HV 25g. The hardness of a sample of C73 quenched into water at 5°C to give a martensitic structure was found to be 970 HV 25g. Corresponding depth/microhardness profiles through a C52 specimen are presented in Figure 8.31 where hardness values of between 900-1150 HV 25g were measured in single and multiple rub. A conventional martensitic C52 specimen, quenched as above, was found to have a hardness of 920 HV 25g. Striations or tempering zones (Figure 8.23b) found in transformed material in multiple rub in both C73 and C52 specimens gave hardness values of between 350-600 HV 25g.

8.3.3 Tempering tests

Tempering experiments have been carried out on phase transformed material to gain further knowledge of its nature. Sections were cut from a specimen of C73, from run 36, with transformed material in both single and multiple rub and were annealed for one hour at successively higher temperatures under vacuum ($<10^{-5}$ torr). Sections were allowed to cool in the furnace, with the vacuum being maintained. Axial microsections were subsequently prepared for microhardness measurements and optical microscopy. The decrease in

FIG. 8.32 VARIATION IN MEAN HARDNESS OF TRANSFORMED MATERIAL WITH TEMPERING TEMPERATURE

(a)

(b)

(c)

20μm

FIG 8.33 AXIAL SECTIONS SHOWING TEMPERING RESPONSE OF PHASE TRANSFORMED MATERIAL IN SINGLE RUB ANNEALED AT (a) 400°C (b) 500°C (c) 700°C FOR 1 HR. RUN 36

(a)

(b)

(c)

20μm

FIG 8.34 AXIAL SECTIONS SHOWING TEMPERING RESPONSE OF
PHASE TRANSFORMED MATERIAL IN MULTIPLE RUB
ANNEALED AT (a) 400°C (b) 500°C (c) 700°C
RUN 36

(a)

(b)

(c)

20μm

FIG 8.34 AXIAL SECTIONS SHOWING TEMPERING RESPONSE OF
PHASE TRANSFORMED MATERIAL IN MULTIPLE RUB
ANNEALED AT (a) 400°C (b) 500°C (c) 700°C
RUN 36

microhardness as a function of tempering temperature is shown in Figure 8.32 where each point is the mean of four measurements. Also plotted are the results of similar tempering tests by Rogers (1969) on a hard W_2 surface phase found on scuffed cast iron diesel engine components and by Eyre and Baxter (1972) on a hard wear transformation product found on the worn surfaces of a 0.24% carbon steel after sliding. As microhardness is critically dependent on load (Buckle (1959)) and as different authors have used different microhardness loads, direct comparison between results is questionable. The tempering response of 0.8% plain carbon martensitic steel, also annealed for periods of one hour, is plotted in terms of Vickers macrohardness (Bain & Paxton (1961)). Two features seem worthy of comment from Figure 8.32. First, although microhardness values at room temperature are rather higher for phase transformed material formed in the present experiments, both in single and multiple rub the tempering response is similar to that found by Rogers (1969) and by Eyre and Baxter (1972) for their hard phases. Second, allowing for the fact that the tempering response of 0.8% martensitic plain carbon steel is plotted in terms of macrohardness, the tempering response between this and phase transformed material is not dissimilar.

The microstructure of the transformed material tempered at 400°C, 500°C and 700°C is shown in Figure 8.33 for single rub and in Figure 8.34 for multiple rub. The results are strikingly similar with apparent carbide precipitation and subsequent coarsening. No evidence of decarburisation beneath transformed material was found in these limited results.

(a)

3 μm

(b)

15 μm

FIG 8.35 AXIAL SECTIONS THROUGH PHASE TRANSFORMED MATERIAL
IN (a) SINGLE RUB (b) MULTIPLE RUB. RUN 32

Transformed material

3 μ m

FIG 8.36 AXIAL SECTION ETCHED WITH REAGENT NO 3 (APPENDIX 4)
SHOWING DEEP ETCH PITS AND FAINT STRUCTURE IN
TRANSFORMED MATERIAL

(a)

(b)

FIG 8.37 CHARACTERISTIC X-RAY SPECTRA (a) PEARLITIC C73
(b) PHASE TRANSFORMED MATERIAL IN SINGLE RUB

8.3.4 Scanning electron microscopy

Typical axial microsections of phase transformed material found in single and multiple rub, viewed by SEM are presented in Figure 8.35a and b respectively, taken from run 32. In general the contrast of images obtained using an ISI 60 SEM was poor, possible contributory effects being the highly polished condition of microsections, and that the specimens tended to be viewed at an angle of 90° to the primary beam, reducing the efficiency of emission of secondary electrons. However in both microsections, etched with picral, transformed material appears to be featureless and thus having an extremely fine grain size. Figure 8.35a shows carbide lamellae gradually disappearing or dissolving into the transformed material. In this figure the line at 45° is a scratch not completely removed in the final polishing stage. Sub-surface deformation of the carbide lamellae in multiple rub is also shown in Figure 8.35b.

In Appendix 4 on the effect of various etchants on phase transformed material, it was noted that reagent No 3 revealed some fine structure therein. An axial section etched with this reagent is presented in Figure 8.36. Although deep etch pits have occurred, some fine structure is apparent in the transformed material.

The characteristic X-ray spectra of the pearlite substrate using the EDAX system, for a specimen of C73 is presented in Figure 8.37a and can be compared against the characteristic X-ray spectra of transformed material in single rub in Figure 8.37b. For elements up to sodium in the periodic table there is no observable difference

FIG 8.38 LINE SCANS FOR CARBON, OXYGEN AND IRON ACROSS CRATER IN UNWORN C73 SPECIMEN

between the two spectra. The two peaks correspond to Fe-K α and K β . The same result occurred with the X-ray spectra from transformed material in multiple rub. At high sliding speeds adhered material was apparent on the surface of the single rub track (e.g. Figure 8.21b). The X-ray spectra of this material is also shown in Figure 8.37b plotted as white dots. The additional peaks corresponding to W-M α and L α , V-K α and Cr-K α , identifying it as material transferred from the hardened T1 specimens.

8.3.5 Auger electron spectroscopy

A crater 1.19 mm in diameter was worn into an unrubbed surface of a C73 specimen which corresponded to a crater depth of 11.8 μ m. Approximately 0.3 μ m was removed, by sputtering, from the surface before line scans 600 μ m wide were taken for carbon, iron and oxygen as shown in Figure 8.38. The line labelled SEI is one line from the secondary electron image taken at the same position on the sample as carbon, iron and oxygen lines. The SEI line was used to find the edge of the crater and was also an indication of the variation in the crater surface topography.

Analysis at two points A and B gave the following results:-

FIG 8.39 LINE SCANS FOR CARBON, OXYGEN, IRON AND NITROGEN ACROSS CRATER IN PHASE TRANSFORMED MATERIAL IN SINGLE RUB

Table 8.3

Depth (μm)	Position	Element (Wt%)		
		Fe	C	O
3.2	A	98.9	0.97	0.16
2.3	B	98.8	1.00	0.17

A slightly higher carbon concentration was recorded than that from the material analysis given in Table 7.1.

From run 38 a crater 0.62 mm in diameter, corresponding to a crater depth of 9.6 μm was worn into transformed material formed in a single rub. After sputtering, 600 μm line scans for carbon, oxygen, nitrogen and iron were taken across the width of the crater. The line scans are shown in Figure 8.39. Analyses at positions labelled A to F gave the following results:-

Table 8.4

Depth (μm)	Position	Element (Wt%)		
		Fe	C	O
8.1	A	98.9	0.95	0.16
9.5	E	98.9	0.92	0.19
9.6	B	99.0	0.79	0.20
8.1	C	99.0	0.86	0.14
6.4	F	99.2	0.69	0.14
4.8	D	99.0	0.84	0.15

FIG. 8.40 COMPOSITION/DEPTH PROFILES THROUGH TRANSFORMED MATERIAL IN SINGLE RUB

In addition to the above, analyses were carried out near the edge of the crater, giving the following results.

Table 8.5

Depth (μm)	Element (% Wt)			
	Fe	C	O	N
2.8	98.8	1.00	0.15	0.08
5.4	98.8	1.00	0.13	0

The detection limit for nitrogen is approximately 0.06% wt. A composition/depth profile for these results is shown in Figure 8.40, plotted for Fe, C and O. As nitrogen was only identified in one analysis, it has not been plotted on Figure 8.40. From these limited results it appears that, compared to the pearlite matrix, phase transformed material formed in a single rub has:-

- (i) No change in the oxygen content
- (ii) No increase in the nitrogen content
- (iii) No change in the iron content
- (iv) No significant decrease in the carbon content.

A slight increase in carbon content is suggested in Figure 8.40 towards the specimen surface, through the transformed material. However as the control sample also showed an increase in the carbon

7 μ m

70 μ m

FIG 8.41 TAPER MICROSECTION THROUGH WORN PIN

(a)

0.2 μm

(b)

FIG 8.42 PHASE TRANSFORMED MATERIAL (a) ELECTRON MICROGRAPH SHOWING VERY SMALL SUBGRAINS (b) SELECTED AREA DIFFRACTION PATTERN OF (a), CAMERA LENGTH 460mm, SHOWING COMPLETE AND PARTIALLY COMPLETE RINGS WITH SOME PREFERRED ORIENTATION

content towards the surface, it is thought that this change is not intrinsic to the formation of phase transformed material.

8.3.6 Transmission electron microscopy

A taper microsection through a worn pin, from the pin on disc experiments, is shown in Figure 8.41 and is representative of features observed. It shows a featureless layer of transformed material of varying depth, with a transition zone between this and the substrate. There is evidence of sub-surface plastic deformation in this region. The transformed material was not always found to be continuous across the surface of the worn pin. Microhardness measurements of the transformed material gave an average value 1100 HV 25g. Thus the transformed material formed on the surface is similar, in physical properties, to that formed in multiple rub on C73 specimens on the crossed cylinders machine.

Transmission electron microscopy of foils from worn pins showed that transformed material consisted of very small subgrains, (Figure 8.42a), little more than 30 nm across. The corresponding selected area diffraction pattern consisted of partially and/or wholly completed rings confirming a fine sub-grain structure was present. Line broadening and line splitting were also apparent in the selected area diffraction pattern. In general there was little evidence of carbides in the electron micrographs of transformed material, though there were occasional carbide reflections in the selected area diffraction patterns.

(a)

0.1 μm

(b)

FIG 8.43 TRANSITION REGION (a) ELECTRON MICROGRAPH SHOWING TRANSFORMED MATERIAL AND DISTORTED CARBIDES (b) SELECTED AREA DIFFRACTION PATTERN OF (a), CAMERA LENGTH 460mm, SHOWING 'SPOTTY' CARBIDE REFLECTIONS AND INCOMPLETE RINGS

1 μ m

FIG 8.44 ELECTRON MICROGRAPH OF CONVENTIONAL MARTENSITE
SHOWING ACICULAR STRUCTURE

Electron micrographs were also obtained as shown in Figure 8.43, it seems probable that they represent the transition region between the transformed material and substrate. Both broken and distorted carbide lamellae and transformed material may be recognised. The corresponding selected area diffraction patterns were not as "ring-like" as Figure 8.42b but tended to be "spotty" and containing many carbide reflections. The ring patterns were usually only partially complete (Figure 8.43b).

As the structures found in phase transformed material do not appear to be characteristic of those normally associated with martensite, for a comparison a pin was heated to 850°C for 1 hr and quenched into water at 5°C, to produce a martensitic structure. Examination by electron microscopy revealed an acicular structure, characteristic of martensite (Figure 8.44). This structure appears different to that observed in transformed material. However it must be realised that if transformed material is martensitic in structure, it would be significantly altered/modified in appearance by the effects of the very small grain size and/or the extent of plastic deformation.

Analysis of SAD patterns

Phase transformed material will have a body centred cubic structure if they are plastically deformed layers, whereas their structure will be body centred tetragonal if they are martensitic. The presence of martensite also results in the broadening of the spots

Fig 8.45 X-ray diffraction patterns of material abraded from (a) an unworn C73

specimen (b) & (c) phase transformed material formed in multiple rub.

or rings in selected area diffraction patterns and in some cases splitting occurs (Cullity (1959)).

The diffraction patterns were indexed in a manner described by Edington (1975), the details of which are given in Appendix 5. Figure 8.42b of the transformed material was subsequently indexed as a tetragonal structure i.e. martensitic. Some rings were not found, however in practice patterns do not provide a complete set of rings because of systematic (structural) or accidental extinctions or because of the weakness of certain lines (Vainshtein (1964)). No analysis of Figure 8.43b was performed as only partially complete rings were found.

8.3.7 X-ray analysis

X-ray diffraction patterns, for material abraded from an unworn surface (a) and for material abraded from multiple rub tracks containing phase transformed material from run 32(b) and from run 38(c), are shown in Figure 8.45. It can be seen that the diffraction patterns of (b) and (c) are slightly different from that of the unworn surface (a). Two peaks could be identified from the diffraction pattern of (a) at $2\theta = 65.7^\circ$, corresponding to $\alpha\text{Fe}(200)$, and at $2\theta = 44.7^\circ$, corresponding to $\alpha\text{Fe}(110)$ and $\text{Fe}_3\text{C}(022)$. As stated previously phase transformed material will have a BCC structure if it is plastically deformed material, whereas its structure will be BCT if it is martensitic. For material abraded from multiple rub tracks an additional peak at $2\theta = 63^\circ$ occurs indicating the presence of α' (martensite) $\text{Fe}(002)$ whilst

line broadening occurs at $2\theta \approx 64^\circ$, possibly indicating the presence of $\alpha'\text{Fe}(200)$, and at $2\theta \approx 47^\circ$ suggesting the additional presence of $\alpha'\text{Fe}(110)$.

While a martensitic BCT structure will produce line broadening, it is also caused by plastic deformation of a crystalline matrix. Due to this and the fact the additional peak is weak, and difficult to positively confirm, the identification of a BCT structure in multiple rub tracks cannot be stated with certainty, however from these limited results its presence can be inferred as the most likely conclusion.

8.4 Conclusion

An extended discussion of the results presented here will be given in the following Chapter 9.

CHAPTER 9

DISCUSSION AND CONCLUSIONS

9.1 Introduction

This thesis has concerned the study of phase transformed material formed in the rubbing of steels and considered its appearance in terms of the frictional flash temperatures generated at the interface. The initial observations which form the starting point for the present work were made by Welsh (1957) (1965) and Archard (1959) who found that the appearance of the material, amongst other features, corresponded with mean calculated flash temperatures at the interface of approximately 750°C or more. They considered this material to have a structure akin to the martensitic structure of conventional ferrous metallurgy and is similar to the so-called white layers. To simplify the approach the experiments have been confined to the unlubricated rubbing of plain carbon steels and to the phenomena associated therewith. The extensive, yet selective, review of Chapter 2 has emphasised the diversity of conditions under which such layers are found. The author is well aware that other considerations almost certainly enter into the generation of white layers under other conditions.

To interpret the appearance of phase transformed material in terms of the transient surface temperature Chapter 2 has considered the relative merits of flash temperature theory and measurement techniques. It has been shown that practically all methods of estimating the surface temperature give, at best, the bulk interfacial temperature which is

generally considerably lower than the transient frictional flash temperature. Infra-red techniques, under limited material combinations, have, however, been able to measure flash temperatures. Comparisons between values derived using the latter method coupled with certain physical changes occurring at the rubbing surface have given credence to some flash temperature theories, in particular to that of Archard (1959). Since the model used in this theory i.e. a single contact spot (or at worst a compact cluster of contact spots which act as one) is rarely achieved it can only be used to illustrate the possible flash temperatures generated at the interface. Likewise in practically all reported rubbing experiments the surfaces have been subjected to an unknown number of rubs of unknown severity. The application of Archard's flash temperature model to this condition must be subject to question. To avoid these problems a crossed cylinders machine has been used and exploratory experiments were performed with soft plain carbon steels rubbing against themselves. The crossed cylinders machine, described in Chapter 3, allows both single and, to a lesser extent, multiple rubbing to occur under well defined conditions. Using smooth specimens of identical diameter in single rub, a single circular contact spot is expected to be produced.

The major and significant conclusion to be drawn from these exploratory experiments, reported in Chapter 4, is that phase transformed material appears as a result of a single traversal of the surface only under conditions so severe that the calculated flash temperatures are far in excess of the values which would be expected from the earlier work. Typically transformed material appears in a single traversal only when the calculated flash temperatures have values of the order of 1250°C or

more. Using the sub-surface temperature analysis derived in Chapter 2 the temperature will be reduced to 1100°C at a depth of $4\mu\text{m}$ which is the typical depth of phase transformed material found in a single traversal. However there were some problems in the finer details of the interpretation of these exploratory experiments. One difficulty being the formation and breaking of welds, in and extending beyond, the single rub track. Since there will be energy dissipation by plastic deformation within the body of a weld the validity of flash temperature calculations becomes less certain. Moreover at that stage it was unknown whether macrowelding itself was a prerequisite for phase transformed material.

In the light of this evidence the interpretation of the earlier work required re-examination, indeed the fundamental metallurgical considerations were re-considered. In particular it seemed appropriate to ask whether conventional ferrous metallurgical theory is applicable to small volumes of material subject to a combination of high temperatures and pressures etc, for times of the order of one millisecond or less. Chapter 5 concerned a discussion of the metallurgical considerations involved. It was found that in the temperature range $723-910^{\circ}\text{C}$ times required to complete the transformation to homogeneous austenite (being the first stage in the production of martensite) were some 3-6 orders of magnitude greater than those found in a typical tribological contact and those experienced in the crossed cylinders machine experiments. Above 910°C the situation is significantly changed. The austenitisation of a normalised plain carbon steel will now consist of two stages:

(i) the allotropic α - γ transformation

(ii) the dissolution of carbide with surplus carbon diffusing into the γ phase.

The first stage will be rapid and it therefore seems likely that the overall rate determining process will be the diffusion of carbon in austenite. A simple estimate of the diffusion of carbon in austenite over a representative distance of 0.5 μm gave values well in excess of 910°C.

A simple diffusion theory for lamellar pearlite has been provided. From this theory it has been found to obtain phase transformed material in a single rub requires temperatures of the order of 1250°C. This temperature is comparable to that required by the order of magnitude calculation and those found for the formation of phase transformed material in a single rub from Chapter 4. More detailed analyses of the diffusion of carbon in lamellar pearlite have served to reinforce the conclusions derived from the simple theory. However of greater significance was a consideration of the temperature/time history (or more correctly the diffusion rate/time history) of the contact. To date the interpretation of phase transformed material had been based upon a mean surface flash temperature and a representative contact time. It was found that the use of such conditions grossly underestimates the chances of diffusion. This is principally due to the exponential dependence of the diffusion coefficient D on the temperature θ ; it follows that diffusion (and therefore the formation of phase transformed material)

will be determined by the maximum temperature which exists during the contact history rather than the mean.

A consideration of the coincidental effects of high contact pressures, shear stresses and other related mechanical effects has argued that while at first sight, these factors may influence the phase changes occurring (e.g. by increased defect concentration enhancing diffusion or by lowering the critical points of the Fe-C equilibrium diagram) such factors may be much less marked under the experimental conditions of this thesis. In the present analysis such complications have been left aside.

Chapter 6 concerned a discussion of the probable significance of this revised metallurgical theory in explaining the wear transitions of Welsh (1965) which still constitutes the major source of experimental evidence of significance on the subject of this thesis. In the interpretations of the patterns of wear reported in Welsh's work it has been appropriate to use the well known theory of adhesive wear (Archard 1953)).

$$\frac{U}{L'} = K \frac{W}{H} \quad (6.2)$$

$\frac{U}{L'}$ is the worn volume (U) per unit sliding distance (L'). W is the load and H the hardness of the softer of the two contacting surfaces. The wear coefficient K is usually taken to indicate the proportion of all rubbing contacts which results in the production of a worn particle. Alternatively, in a continuous wear experiment $1/K$ represents the average number of times each part of the rubbing surface is rubbed before it is worn. It was found that it was possible to reinterpret Welsh's T_2 and T_3 transitions and the region between these transitions if total

transient surface temperatures $\theta_T (= \theta_f + \theta_{bk})$ were considered. The influence of carbon content, hardness and externally induced bulk temperatures were also shown to be in broad accord with this reassessment. The consequential interpretation of the K factor as representing the number of temperature cycles to which an element on the rubbing surface provides the essential link between the single rub experiments of this thesis and the multiple contact conditions of a typical wear experiment such as those of Welsh (1965).

The main experiments of Chapter 8 will now be considered. These were performed to avoid ambiguities found in the detailed analysis of the exploratory experiments and after the development of the theories of Chapters 5 and 6. The discussion of the main experimental results will be split into two parts. The first will consider the results of the crossed cylinders machine experiments. The second will discuss tests performed to indicate the nature of phase transformed material.

9.2 Discussion of main experiments

9.2.1 Preliminary runs 1-19

The effect of a small degree of specimen eccentricity ($\approx \pm 0.025\text{mm}$), in runs 1-6, was found to impose a cyclic variation on the frictional record (Figure 8.1). Though the overall pattern of friction was probably unaffected, i.e. $\mu_r \approx 2 \mu_s$, in comparison to runs 11-16, the absolute values were additionally increased. The effect was resolved by having each set of specimens, mounted on mandrels, ground to size and concentric to the running centres prior to each run. This method, however, was both laborious and inefficient in machine time and workshop labour. In subsequent work it would be desirable to manufacture stiffer mandrels and key both specimens and mandrels to allow accurate relocation.

A "Distac" non contacting eddy current probe was used in runs 7-10 to measure the frictional displacements of the strain rod. Although this method provided a high frequency response the friction records for these runs were remarkably different between themselves and from the overall pattern established in runs 1-6. In subsequent runs (11 onwards) a Talymin probe was used to measure the displacements of the strain rod. Repeatable and fairly constant friction values in single rub were obtained. The use of a Talymin probe (or Talysurf head) will inevitably give a time averaged value of μ determined by the time constant of the device. It would appear that attempts to use devices with a short time constant, which would have also revealed short time variations in friction, have also measured undesirable transients such as vibrations. The influence of vibration on metal to metal contact and on kinetic friction has

been studied by Godfrey (1967). The effect of vibration on these parameters was found to be complex, though in general contact area and kinetic friction were reduced with vibration. In the subsequent discussion of the paper it was pointed out that some of the observed effects could be due to the dynamic response of the experimental apparatus itself. The essential conclusion for the present work is that vibration may indeed cause both periodic variations and alter the absolute value of μ . The use of time averaged values is justified therefore for the following reasons. Firstly, flash temperature theory assumes uniform heat dissipation i.e. a uniform value of μ over the contact; thus the time constant need only be comparable with the time of passage of a point on the surface through the contact. Secondly, metallurgical examination is made at a number of different points on the track to examine typical, representative, behaviour during the single traversal. Thus typical average values of μ seem appropriate. Where examination of the single tracks on the same specimens revealed both the presence and absence of phase transformed material, the maximum and minimum values of μ_s were used. Transient values have generally been ignored. Finally this approach seems justified by the general agreement between flash temperature theory, on one hand, and the results of the metallurgical examination on the other.

Runs 11-16 had been performed using similar conditions to runs 1-6 with an identical specimen cleaning procedure. In the later runs friction values were low and phase transformed material was not found in single rub. In an effort to increase μ_s a more severe

cleaning procedure was adopted involving ultrasonic techniques. Runs 17-19 were performed using this method and high values of μ_s obtained. However subsequent examination of the specimen surfaces revealed gross or macrotransfer of material from the soft carbon steel specimen to the hardened tool steel specimen, in both single and multiple rub. A condition to which the applicability of flash temperature theory is questionable. This feature illustrates the balance of surface cleanliness required to produce a reasonably high and consistent value of μ_s without gross transfer. In the initial proving of the machine and specimen geometry, a set of uncleaned specimens as supplied from workshops were used. In the single rub track values of $\mu = 0.05$ were measured with, to the naked eye, no transfer of material.

9.2.2 Main runs, 20-42

With the solution of the earlier experimental problems and adoption of standardised experimental procedure the main results of runs 20-42 will now be considered.

9.2.3 Friction records

Figure 8.4 shows typical friction traces obtained. The friction coefficient in single rub, μ_s , varies over a small range of values and is generally less than the coefficient of friction in multiple rub (μ_r). The variations in μ_r are much larger than in μ_s and approach peak values of up to 1.0. A feature commonly observed and illustrated in Figure 8.4 is for μ_r to increase both in its mean

value and its fluctuations prior to the ram traverse to produce the single rub track. Two contributory reasons for this effect are as follows.

(i) Impact loading of the specimens was minimised by slowly lowering the loaded upper specimen using a scissor jack arrangement. μ_r will tend to rise until the load is fully supported at the contact between the specimens. This probably accounts for the steady increase in μ_r at the start of the run.

(ii) Track roughening, running-in, the formation of phase transformed material, material removal etc, may all cause μ_r to increase further together with the variations observed. The balance of surface cleanliness proposed earlier broke down at high contact loads (790 N). Runs 36, 38 and 39 all gave high values of μ_s (up to 0.67) with gross material transfer in single and multiple rub. This balance also failed in run 30 where specimen cleaning was accidentally allowed to continue for a longer period, gross transfer being obtained in single rub in the subsequent experimental run.

9.2.4 Track appearance

The track appearance in single rub has been defined as either polished or non-polished and was found to be related to the absence or presence of phase transformed material. The differences in track appearance were also reflected in the appearance of the hardened counter surfaces and with Talysurf traces across the tracks. The features of the surfaces where no phase transformed material was

found can be explained as follows: at loads greater than W_1 (Figure 2.10) the hardened surface acts to a limited extent as a "plough" or indenter in the soft steel specimens producing generally a smooth surface in the latter. Little surface alteration can be expected on the hardened surface though a small amount of microwelding might be expected and is indeed observed.

The formation of phase transformed material in non-polished tracks is associated with the higher values of μ_s and therefore higher surface temperatures. It is suggested that this factor coupled with the complex stress distributions beneath the rubbing surfaces and the small change in volume occurring at the surface layers with α - γ transformation results in greater adherence between the surfaces. The similarity between the appearance of a microweld in a polished track (Figure 8.5) and that of a non-polished track implies that continuous microwelding process occurs at the surfaces of non-polished tracks. On separation of the surfaces the adherence causes the track to roughen. The position of the new interface seems frequently to occur in the hardened specimen. Not only has material been removed in the single rub track of the hardened specimen but also metallurgical analysis has shown (Figures 8.21 and 8.25) that small amounts of hardened T1 material are found in many cases to be adhering to the phase transformed material.

Figure 8.11 shows from run 34 alternating polished and non-polished tracks in the same single rub track. The interface is quite sharp. there is evidence of microwelding initiating the higher values of μ_s (and thus higher surface temperatures) and causing the non-polished

track. The process would seem to end when surface contaminants reduce μ_s for the transformation to be arrested.

Gross or macrotransfer between specimens causes the track appearances to be quite different (Figure 8.12).

The multiple rub tracks on the plain carbon steel specimens (Figure 8.13) in practically all cases were found to contain phase transformed material. Their respective track appearances were usually rough and torn. This figure also illustrates the differences between the defined track conditions in single rub to those of the more complex conditions of multiple track which are typical of many rubbing experiments.

The corresponding multiple rub tracks on the hardened T1 specimens also show conditions difficult to analyse. Material transfer from the carbon steel specimens occurred under mild conditions. Under more severe conditions the carbon steel specimens could damage and remove material from the T1 specimens (Figure 8.14); this would appear to illustrate well the wear resistant nature of phase transformed material.

Evidence drawn from the appearance of the tracks may be summarised as follows:

- (1) Multiple rub tracks were difficult to analyse because of the complexity of conditions .

(ii) A few single rub tracks under heavy loads showed gross damage and these results were ignored.

(iii) The remaining single rub tracks showed a sharp division between two types of behaviour - that producing a polished track and that producing a non-polished track. There was a clear correlation between these two types of appearance and other features both when the track showed a single character and when both types of behaviour occurred on a single track.

(iv) Polished tracks showed low friction, low calculated temperatures and no phase transformed material.

(v) Non-polished tracks showed higher friction, higher calculated temperatures and the presence of phase transformed material.

(vi) It is concluded that the transition to non-polished behaviour is associated with the generation of temperatures sufficient to cause the α - γ transformation in a layer of carbon steel material of significant depth.

9.2.5 Bulk temperature

Bulk temperature rises were low, the maximum bulk temperature observed was 44°C. The influence of the single rub portion was found to have only a small effect on the overall increase. Small bulk temperature rises were to be expected considering the size of

FIG. 9.1 GRAPHICAL COMPARISON BETWEEN ESTIMATES OF CONTACT RADIUS a , FOR RUNS 1-42. BARS REPRESENT RANGE OF VALUES OF a_a , AT A PARTICULAR LOAD.

both specimens and mandrels together with the short duration of each run. What these results do demonstrate, however, is the importance, in each particular experimental study, of the specimen geometry and test duration on the phenomena observed. For instance in this work it has been proposed that one feature of Welsh's T_3 transition is due to phase transformed material being formed on the pin first (rather than the ring) because of its higher bulk temperature. The crucial role of bulk temperature has been emphasised, for example by Story et al (1980) in a reassessment of scuffing failure data of discs (with concomitant white layer formation). In their view disc and gear tests for scuffing, have little value for critical examination unless the bulk temperature has been continuously monitored.

9.2.6 Contact size

It has been found (Table 8.2) that the measured contact sizes expressed as the half-track-width (a_a) have larger values than those derived assuming either elastic or plastic deformation (a_e and a_p) or from the values derived using the graphical method (a_g) for the intermediate regime. For convenience the essential part of Figure 2.10 is replotted here in Figure 9.1 and the measured values of a_a are also plotted here. The high values of a_a are caused by the fact that they arise from a sliding contact whereas the plotted values of a_e , a_p and a_g assume static loading.

The experiments of Chapter 8 were conducted under conditions which are geometrically similar to a hard sphere sliding on a soft flat

FIG. 9.2 EQUIVALENT CONTACT CONDITIONS TO
HARD TOOL STEEL CYLINDER IN CONTACT WITH
SOFT CARBON STEEL CYLINDER

(Figure 9.2). The rotation of the hardened cylindrical specimen, designed to ensure a moving contact on the hardened steel surface, is equivalent to the rotation of the sphere on the axis shown in Figure 9.2. Since this rotation involves surface speeds $\ll V$, this has no major influence on the sliding speed. The contact conditions may differ from those of a static contact under the same load (a) because the loading conditions lie in the intermediate elastic/plastic range ($W_1 < W < W_3$), and (b) because the surface layers at the rear of the contact will be heated to temperatures which significantly lower their mechanical strength. For these reasons there may be relaxation of the load carrying capacity in the rear half of the contact, therefore the consequences of these factors need further discussion.

Figure 9.1 shows that even the lowest load (152 N) used in these experiments exceeds the value of W_1 , the load for the onset of plastic yielding under static loading. When the concomitant effects of tangential surface stresses are considered (see Section 2.8) it is clear that in all experiments some plastic yielding will occur in the forward half of the contact at depths between $0.5a$ and the surface and that the extent of this yielding will increase with increasing loads. This yielding and the consequential deformation is observed in sections (particularly normal sections) of rubbed surfaces. Under the extreme conditions of full plasticity ($W > W_3$), which is not reached in our experiments, there would be negligible elastic recovery in the rear half of the contact, the load would be borne on a semi-circular contact (Bowden and Tabor (1950)) and

$$\frac{1}{2} \pi (a_a)^2 H = W \quad (9.1)$$

or $a_a = \sqrt{a_p}$ (since $\pi (a_p)^2 H = W$). Thus in the range $W_1 < W < W_3$ it would be expected that $a_g < a_a < \sqrt{2} a_p$, where a_g is the theoretical value derived from a full elastic/plastic plot for static loading.

This conclusion is largely justified by the measured values of a_a listed in Table 8.2 and plotted in Figure 9.1. This figure shows that to a good approximation the values of a_a lie between the limits defined above. Values of a_a from runs 1-6 and from runs where gross transfer occurred in single rub are not included in Figure 9.1. Further consideration of these results suggests that they are primarily dependent on load and that sliding velocity has little influence. It may be concluded therefore that the apparent increase in contact size is not strongly correlated with the generated temperatures at the interface. This conclusion is supported by the lack of correlation between the ratio a_a/a_g and the calculated maximum surface temperature.

A similar conclusion can be derived from the following approach. Plastic flow occurs in the front half of the contact. In the elastic/plastic regime there will be plastic flow in the front half and elastic recovery in the rear. It is in the rear half of the contact that the maximum temperature occurs, since for high speed conditions $\theta \propto t^{1/2} \propto x^{1/2}$. Thus this means that in the experiments of Chapter 8 (at loads not greatly in excess of W_1) the plastic deformation is in the colder region of the contact and the major

FIG 9.3 PROPOSED CONTACT CONDITIONS
AT LOADS BETWEEN W_1 & W_3

temperature rises are confined to surface layers in the trailing half of the contact where the load is supported by elastic stresses at much greater depths than those subject to heating. Whilst the above arguments provide a plausible explanation for the apparent increase in the measured contact radius a_a , the effect this will have on the calculated flash temperatures will now be considered.

In the elastic/plastic load range, for reasons given earlier, a suitable model for the contact might be as follows:

(i) A semi-circular area of radius a_a in the front of the contact.

(ii) A semi-elliptical area of semi-axes a_a and b (Figure 9.3.) at the rear.

Thus for $W_1 < W < W_3$, $a_a > b > 0$ where $b \rightarrow 0$ for heavier loads.

The effect of tangential stresses will move this elastic/plastic regime to lighter loads but this is ignored below. Consider, now, the effect of the above assumptions upon the values of flash temperatures which have been used to justify the revised metallurgical theory. For mathematical simplicity the effect on $\theta_{\max}$ will be examined. The arguments are broadly similar for other parts of the contact. Recalling that

$$\theta_{\max} = 1.64 \theta_m \quad (2.41)$$

values of $\theta_{\max}$ are calculated assuming a known rate of heat generation, Q , and postulating a circular contact of radius a_a .

Thus using the high speed equation

$$\theta = \frac{2qt^{1/2}}{(\pi k \rho c)^{1/2}} \quad (2.6)$$

for the calculation of $\theta_{a \max}$

$$q = Q/\pi(a_a)^2 \quad (9.2a)$$

$$t = 2a_a/V \quad (9.2b)$$

For the conditions assumed in Figure 9.3 we have a new value of

$$\theta_{\max}, (\theta_{\max} = \theta'_{\max})$$

where in equation (2.6)

$$q = \frac{Q}{\frac{1}{2}\pi(a_a)^2 + \frac{1}{2}\pi a_a b} = \frac{Q}{\frac{1}{2}\pi a_a (a_a + b)} \quad (9.3a)$$

$$t = (a_a + b)/V \quad (9.3b)$$

Using equations 9.2 - 9.3 and 2.6 the following equation is obtained

$$\theta'_{\max} = \theta_{a \max} \left(\frac{2a_a}{a_a + b} \right)^{\frac{1}{2}} \quad (9.4)$$

In equation 9.4 the factor $\left(\frac{2a_a}{a_a + b} \right)^{\frac{1}{2}}$ has values between 1, for $b = a_a$, at light loads and $\sqrt{2}$ as $b \rightarrow 0$, at heavy loads. Thus equation 9.4 shows that the effect of load relaxation in the rear half of the contact (and the consequential increase in the track half-width, a_a , over the theoretical value of a_g) is to increase the values of the calculated flash temperatures by factors of between 1 and $\sqrt{2}$. The effect is expected to increase, in our experiments, as the loads are increased. This factor must therefore be taken into account when calculated values of the maximum surface temperature are considered.

The plastic deformation in the front half of the contact will obviously make a major contribution to the pile-up of material at the edges of the tracks. It is possible that this pile-up may be enhanced by movement of the surface layers softened by high temperatures at the rear. ("mud on ice" to reverse Bowden's well-known analogy for the sliding of hard oxide layers on soft metals). A minor effect may be noted here. The effect of slow rotation of the hard tool steel cylinder would be to bias any such movement of material and could produce the tilt or asymmetry of the resultant crescent of phase transformed material which was observed in chordal sections such as Figs 8.21a, 8.22a and 8.25a.

Track widths in multiple rub were found to be typically of the order of 1.5 to 2 times the width of the single rub tracks; because of their dubious value in relation to flash temperature calculations in multiple rub they have not been included.

9.2.7 Flash temperature calculations

Before proceeding to a discussion of the overall pattern of results it is appropriate to consider, in greater detail, the results obtained from a single experiment. For this analysis we select Run 24 for which the conditions of significance were as follows.

Load 250 N, sliding speed 4.6 m/s

Speed of hardened cylinder 0.23 m/s

Coefficient of friction in single rub $\mu_s = 0.34 - 0.39$

Bulk temperature rise of soft steel specimen $\theta_{bk} = 25 - 31^\circ\text{C}$.

Half measured back width, $a_a = 0.28 \text{ mm}$

Theoretical contact radius (graphical interpolation) $a_g =$
0.255mm

Theoretical mean contact flash temperatures:

from measured track width (θ_{m_a}) = 842°C

from theoretical (graphical) contact size (θ_{m_g}) = 967°C

This experiment has been selected for discussion since it represents a marginal result, a run in which regions with and without phase transformed material were observed upon the single rub track. It also represents an experiment at the lower end of the load range

used in the experiments. Using the mean contact flash temperatures the deduced values do not provide immediate confirmation for the theory of Chapter 5. In listing the results for this experiment the only values of contact size which have been quoted are the theoretical contact radius, a_g , deduced from graphical interpolation in the elastic/plastic range and the half track width as measured. In the light of the discussion of Section 9.2.6 these seem the only two values worthy of further consideration. The theoretical value a_g may underestimate the true contact size because, as explained in Section 9.2.6 there will be some relaxation of the load in the rear half of the contact; moreover this will be somewhat greater than is indicated by the theoretical line of Figure 9.1 because the action of superimposed tangential forces will increase the extent of plastic flow. The measured track half-width, a_a , may overestimate the size of the contact because it may include material piled up at the edges which does not contribute fully to load support. Moreover use of the measured half track-widths will certainly result in an underestimation of the flash temperatures because of the effects illustrated in equation (9.4) and the associated discussion. It is also possible that a similar underestimation would arise from the use of the theoretical contact radius a_g . For these reasons in what follows we shall only be concerned with values of flash temperature deduced using values a_a and a_g .

Before proceeding to a discussion of the metallographic evidence it is necessary to examine critically the method used in the

calculation of the mean value of the flash temperature. Following the procedure discussed in Chapter 2 we first calculate the mean values of the flash temperatures (θ_B and θ_C) which would be attained if all the generated heat were supplied to the body in question. Thus for Run 24 we obtain the following result using a_g .

For the moving heat source (high speed), soft steel:

$$\theta_C = 1138^\circ\text{C}$$

For the stationary (slow speed) heat source, hard steel:

$$\theta_B = 6420^\circ\text{C}$$

Thus the mean flash temperature (θ_m) at the contact is given by

$$\frac{1}{\theta_m} = \frac{1}{\theta_B} + \frac{1}{\theta_C}, \quad \theta_m = 967^\circ\text{C} \quad (9.5)$$

It may be noted that this calculation by-passes the calculation of the way in which the generated heat is divided between the two bodies; in fact, it implies that 85% is supplied to the high speed heat source (soft steel) and 15% to the slow speed (stationary) heat source (hard steel).

The following assumptions are made in these calculations.

(i) Heat is divided between the two bodies in such a way as to make the interface temperature calculated for each of the two bodies the same.

(ii) Complete matching of the temperature over the whole of the interface has not been attempted; all that has been done is to divide the heat so as to equalise the mean values of the contact temperatures for the two bodies.

Since it has been shown that in order to equalise these mean temperatures 85% of the generated heat flows to the high speed surface it follows that the requirements of heat conduction in this body must be the primary consideration in determining the overall temperature distribution at the interface. In physical terms, the faster moving body is constantly presenting fresh cool material to the vicinity of the interface and this is present for such short periods of time that only a thin layer of its material is heated. On the other hand the stationary (or slowly moving) body is present for such long periods that any required steady state temperature distribution can be established within it. It follows that in determining the temperature distribution at the interface the requirements of heat conduction in the high speed body must dominate and the temperature of the interface will have a distribution close to that given by the equations for a high speed heat source.

How then can conditions in the stationary (or slow moving) body be modified to meet the conditions dominated by these in the opposing body? The conventional heat conduction equations for the stationary

FIG. 9.4 a ISOTHERMS FOR CONVENTIONAL STATIONARY HEAT SOURCE
WITH UNIFORM HEAT GENERATION. THE ISOTHERMS SHOW TEMPERATURES
AS A PROPORTION OF THE MAXIMUM VALUE

FIG. 9.4 b HYPOTHETICAL ISOTHERMS FOR STATIONARY CONTACT WITH SURFACE TEMPERATURES HAVING A CLOSE APPROXIMATION TO THE HIGH SPEED EQUATION TEMPERATURE DISTRIBUTION. THE ISOTHERMS SHOW THE TEMPERATURES AS A PROPORTION OF THE MAXIMUM TEMPERATURE WHICH OCCURS AT THE BACK EDGE OF THE CONTACT

(or slow moving) body assume a heat source with uniform generation of heat and as a result the heat source (the interface) is an isotherm at the maximum temperature; the steady state heat conduction isotherms are somewhat as shown in Figure 9.4a. In practice the interface temperature must, as we have argued, have a temperature distribution similar to that for a high speed heat source, the steady state heat conduction isotherms will then take the form somewhat as shown schematically in Figure 9.4b and the supply of heat to this body (i.e. the proportion of the total heat generated) will vary over the interface so as to produce a close approximation to the required result. One obvious consequence of this will be that, in the leading half of the contact, the heat flow (the proportion of the total generated heat) to the stationary (or slow moving) body which will be required to attain the required (lower) values of the interface temperature will be less than those assumed for the conventional steady state temperature distribution of Figure 9.4a. Similarly the heat supply in the rear half of the contact will be greater than those assumed in Figure 9.4a. It will be recalled that our first approximate calculation, upon which the value of θ_m in equation (9.5) is based, shows that approximately 85% of the generated heat flows to the high speed body and 15% to the stationary (or slow speed) body. Thus, in order to produce equal temperatures at the surfaces of both bodies over the interface these proportions will vary; however it remains true that almost all the heat will go to the high speed body and the temperature distribution will be very similar to that already assumed. The foregoing argument would suggest that the temperatures will rise slightly more rapidly than forecast in the leading half of the

contact and somewhat less rapidly than expected in the rear half of the contact. A most important consequence of this argument for what follows is that we would expect that the effect of these considerations upon the maximum contact temperature will be fairly small.

There is one minor effect of these considerations which may be added here, as it were, in parenthesis. The fact that the hard steel surface is not stationary but is moving slowly normal to the main sliding direction will mean that this body will carry its temperature history with it causing slightly higher temperatures in its direction of motion. This will be an additional cause for the asymmetry of the crescent-shaped phase transformed material which has been noted in the discussion of Figures 8.21a, 8.22a and 8.25a at the end of Section 9.2.6.

If therefore we assume as a reasonable approximation that the temperature distribution at the interface is that given by the high speed equations then using the result $\theta_{\max} = 1.64 \theta_m$ we obtain for the maximum values of the flash temperature deduced for Run 24 the following results.

FIG 9.5 TYPICAL TAPER MICROSECTION THROUGH PHASE TRANSFORMED MATERIAL, IN SINGLE RUB. RUN 24

(a) Contact Geometry

(b) Maximum temperatures at different positions in the track

FIG 9.6 MAXIMUM SURFACE TEMPERATURES

From the measured track width ($\theta_{\max_a}$) = 1371°C

From the theoretical (graphical) contact size ($\theta_{\max_g}$) = 1586°C

Bearing in mind the fact that the result using a_a may underestimate the flash temperatures because of the influence of recovery in the rear half of the contact, as represented by equation (9.4) and the associated discussion, it can be concluded that these two results are in reasonable accord and forecast a value of the maximum flash temperature ($\theta_{\max}$), say, in the range 1450–1550°C; i.e. values of the maximum surface temperature (with allowance for the bulk temperature) in the range 1480°C–1580°C.

We can turn now to a consideration of the metallographic evidence drawn from Run 24. Figure 9.5 shows a chordal microsection with a taper magnification of $\times 7$ taken from the single rub track of this experiment. The theoretical contact diameter $2a_g$ and the mean measured track width, $2a_a$, for this experiment are also shown. In the following calculations we will use the theoretical value of the contact size.

The maximum flash temperature to which any point on the surface has been subjected will vary across the track. Figure 9.6a shows the contact geometry. At a distance x from the centre of the track the surface will have been subjected to heating for a time $(2y/V)$ whilst at the centre of the track the surface will have been heated for a time $(2a/V)$. Bearing in mind the fact the the high speed equations are based upon the equation

$$\theta = 2qt^{1/2}/(\pi kpc)^{1/2}$$

and that from Figure 9.6a

$$y^2 = a^2 - x^2,$$

the relationship between the maximum temperature $(\theta_{\max})_x$ at a distance x from the centre of the track and the maximum temperature $(\theta_{\max})_0$ at the centre can be deduced as follows.

$$\frac{(\theta_{\max})_x}{(\theta_{\max})_0} = \frac{(2y)^{1/2}}{(2a)^{1/2}} = (1 - x^2/a^2)^{1/4} \quad (9.6)$$

This result is shown graphically in Figure 9.6b.

The maximum sub-surface temperature at any position can be found in terms of the corresponding value of the maximum surface temperature by the use of Figure 2.7. Thus, for the centre of the track, using $L = 42.5$ and $a = a_g = 0.255$ mm, we obtain Table 9.1.

Table 9.1

Run 24 Maximum temperatures at a depth z (θ_z) below the centre of the track

$\theta_z/\theta_o (= \theta_{\max})$	z/a	z (μm)
1	0	0
0.9	0.013	3.3
0.7	0.043	10.9
0.6	0.06	15.3
0.5	0.08	20.4

A similar argument can be applied to any point below the surface but at, say, a distance x from the centre of the track. For example, from Fig 9.6b at a distance $x = 0.58a = 0.148$ mm from the centre of the track the maximum temperature to which the surface is subjected is $0.9\theta_{\max}$ where $\theta_{\max}$ is the maximum flash temperature for the whole contact. Then, again applying Fig 2.7, one can obtain the maximum temperature to which the material is subjected at different depths, the results being shown in Table 9.2.

Table 9.2

Run 24: Maximum values of the temperature θ_z at a distance 0.148mm from the centre of the track and at different depths z . The temperatures are expressed as a ratio of $\theta_{\max}$ for the whole contact.

$\theta_z/\theta_{\max}$	θ_z/θ_o	z/a	z (μm)
0.9	1	0	0
0.8	0.89	0.014	3.5
0.7	0.78	0.03	7.6
0.6	0.67	0.05	12.7
0.5	0.56	0.07	18

The results of Figure 9.6 and Tables 9.1 and 9.2 are plotted on

Figure 9.5 and with these points the sub-surface isotherms have been drawn with sufficient accuracy for the present analysis. The most important conclusion drawn from this result is that the boundary of the fully transformed material corresponds remarkably closely to the temperature isotherm of $0.8 \theta_{\max}$.

Taking our previous conservative estimate of the maximum flash temperature for the whole contact in the range $1450-1550^{\circ}\text{C}$ this gives the boundary of the phase transformed material as corresponding to a maximum flash temperature in the range $1160-1240^{\circ}\text{C}$. Allowing for a bulk temperature of, say, 30°C the maximum temperature to which phase transformed material has been subjected lies in the range $1190-1270^{\circ}\text{C}$.

Three conclusions can be drawn from this more detailed analysis of the results of Run 24.

(i) The shape of the boundary of the phase transformed material is in very good agreement with the more detailed analysis of flash temperature theory which has been produced in this thesis.

(ii) The position of the boundary of the phase transformed material corresponds well with the values of the transient temperature (say $\sim 1250^{\circ}\text{C}$) which emerged both from the earlier exploratory experiments described in Chapter 4 and, more importantly, from the metallurgical theory of Chapter 5.

• PHASE TRANSFORMED MATERIAL

○ NO TRANSFORMATION

FIG 9.7a APPEARANCE OF PHASE TRANSFORMED MATERIAL IN TERMS OF THE MAXIMUM TOTAL TEMPERATURE $\theta_T (= \theta_{max} + \theta_{bk})$ FOR THE GRAPHICAL CONTACT SIZE a_g .

FIG.9.7b APPEARANCE OF PHASE TRANSFORMED MATERIAL IN TERMS OF THE MAXIMUM TOTAL TEMPERATURE $\theta_T (= \theta_{\max} + \theta_{bk})$ FOR THE HALF TRACK WIDTH a_a

(iii) It is clear from the analysis, as also from the earlier discussion, that in arriving at an estimate of the likelihood of the production of phase transformed material the most significant parameter is the maximum temperature to which the material in the contact has been subjected.

With this last conclusion in mind the flash temperature results of Chapter 8 (tabulated in Appendix 3) have been recalculated to obtain theoretical estimates of the maximum values of the total surface temperatures to which the rubbing surfaces have been subjected in the single rub tracks, $\theta_T (= \theta_{\max} + \theta_{bk})$. For reasons given earlier these calculations are confined to those based upon the theoretical contact radius, based upon graphical interpolation, a_g , and those based upon the measured value of the track width, $2a_a$. These results are fully tabulated in Appendix 6 and are presented here graphically in Figures 9.7 a and b. It will be observed from Figure 9.7a that using the theoretical value of the contact size suggests that phase transformed material appears when the maximum surface temperature exceeds a value of 1500°C . Similarly when using the measured value of the track width, Figure 9.7b suggests that a maximum surface temperature of the order of 1230°C is required for the production of phase transformed material.

In considering the results of Fig 9.7a and b and the conclusions drawn from them we can take the following factors into account:

(i) The use of the measured track width ($2a_a$) for calculation of flash temperature will underestimate the maximum value of the flash temperature because of recovery in the rear half of the contact as illustrated in Figure 9.2 and the associated discussion.

(ii) The use of the theoretical contact size for the size of the contact is a little more complex. A small overestimation of the flash temperature may arise because the superposition of tangential surface tractions will increase the contact size over the theoretical value. However, under these conditions there will exist the opposite influence of the effects of recovery in the rear half of the contact which have already been discussed.

(iii) To produce a finite volume of phase transformed material the lower flash temperatures at other parts of the track must be considered. However as the discussion of Run 24 and the micrograph of Figure 9.5 shows, it would appear that a flash temperature of the order of $0.8 \theta_{\max}$, where $\theta_{\max}$ is the maximum value at the centre of the track seems appropriate.

With these considerations in mind we can draw the conclusion that the analysis of the results of the main experiments of Chapter 8 show that temperatures in the range 1100-1200°C are required for the

production of phase transformed material in a controlled single rub experiment. This conclusion is in very satisfactory broad agreement with the metallurgical theory of Chapter 5.

One final conclusion may be drawn from this analysis of the main experiments of this thesis. It has been shown that the production of phase transformed material is best considered in terms of the maximum value of the flash temperature $\theta_{\max}$. It would be seen that the critical value of the temperature is in the order of $0.8\theta_{\max}$. Under high speed flash temperature theory conditions, $\theta_{\max} = 1.64 \theta_m$ where θ_m is the mean value of the flash temperature calculated using the Archard (1959) paper. Thus we are now arguing that under these conditions the critical value of the flash temperature is of the order of $1.3 \theta_m$. Since the earlier work of Archard (1959) and Welsh (1965) used values of θ_m , the arguments of this paragraph gives additional support to a reassessment of Welsh's work which has been given in Chapter 6.

9.2.8 Appearance of phase transformed material

Although axial and normal metallographic sections through single and multiple tracks are useful in indicating the presence of phase transformed material and the extent of sub-surface deformation, chordal sectioning techniques have been found to reveal information most relevant to the theory of phase transformations and the related thermal analysis.

It has been found in single rub for both 0.52% and 0.73% plain carbon steels that

(i) The crescent-shaped cross-sectional appearance of phase transformed material is consistent with the generated temperature distribution for a contact at high sliding speeds.

(ii) The transition region between phase transformed material and substrate clearly corresponds to a partially transformed region. Carbide lamellae, from the pearlite, can be observed gradually dissolving into the transformed material as the surface is approached. This is of particular importance in providing evidence which supports the metallurgical theory of Chapter 5.

(iii) Particularly at low loads the carbide lamellae are continuous with the substrate and reveal little sub-surface deformation.

(iv) At low loads there is no evidence of pro-eutectoid ferrite in fully or partially transformed material. At higher loads some entrainment of pro-eutectoid ferrite in the partially transformed region does, however, occur; it seems likely that sub-surface deformation at these conditions is responsible. The lack of pro-eutectoid ferrite in transformed material is further evidence of surface temperatures in excess of 910°C .

In multiple rub many of these features are significantly altered. There is generally no symmetric appearance of phase transformed material in relation to the multiple rub track, moreover the depth of both fully and partially transformed regions vary considerably from point to point within the track. Extensive sub-surface deformation is indicated, indeed particles of transformed material or particles from the hardened tool steel specimens or both have been found embedded in the multiple rub tracks. "Striations" are commonly observed which it is suggested are carbides coalescing in a tempering effect due to repeated applications of frictional heat. The complex conditions which exist during multiple rubbing contacts for the production of phase transformed material are evident and the difficulty of applying flash temperature calculations to such conditions is apparent.

Normal sections (Figure 8.27) indicate the relative severity of forward flow between single and multiple rub conditions. Some of the microsections (Figure 8.28) have revealed phase transformed material just starting to form in the centre of the tracks i.e. in the region associated with the highest flash temperatures during the duration of the contact. This evidence, though, must be weighed with the sharp transition between the appearance or absence of phase transformed material at positions near the centre of tracks e.g. Figure 8.11. The appearance of microwelds in the non-polished tracks could often be correlated with a friction peak in the single rub portion of the friction trace. Chordal microsections through such microwelds have typically found phase transformed material (Figure 8.29).

9.3 Identification and nature of phase transformed material

At the completion of the crossed cylinders experiments some supplementary tests were performed to indicate the nature of phase transformed material. A relatively wide range of analytical methods was attempted for identification purposes, since no single technique can be relied on to supply the overall picture. These tests have indicated that phase transformed material is of martensitic form. It is evident that detailed analysis of the material must form part of future work if the results of the present experiments are to gain greater significance.

As specific etches are available for martensite and because of the predominance in the literature of the use of nital or picral etchants, the effect of a range of etchants on phase transformed material was studied first. The general conclusion from this study was that phase transformed material is a fine grained martensitic structure with a fine dispersion of carbides. Etch No 6 from Appendix 4 seems worthy of further examination because of its selective coloured attack on martensite.

Microhardness measurements and profiles through phase transformed material formed in single and multiple rub show it to be harder than a conventional martensite of the same composition, though the difference is somewhat less than commonly stated. In the present work microhardness values were typically 10% greater than those of conventionally formed martensite. This difference may be readily accounted for by the assumption of a fine grain size. There is no significant indication in the results obtained for different values of hardness for material

formed in single or multiple rub. The steady decrease of microhardness with depth is thought to be associated with a partially transformed region and, particularly in multiple rub, with sub-surface work hardening. Values obtained in striated regions in multiple rub are of the order of those expected if tempering had occurred.

The results of tempering tests, performed in a high vacuum thus minimising effects of decarburisation and oxidation, were assessed by axial metallographic sections and microhardness measurements. Graphical comparison of the tempering response with that of the W_2 phase of Rogers (1970) and of the wear transformation product of Eyre and Baxter (1972) indicates a good degree of correlation (Figure 8.32). The major discrepancy is in the relative values of microhardness prior to tempering. Since this latter value is affected primarily by the conditions during formation, as illustrated by the range of microhardness values quoted for white layers, these differences are of probably lesser significance than might be implied from Figure 8.32. Also plotted in this figure is the tempering response of a conventional 0.8% carbon martensite, in terms of macrohardness. If some allowance is made for the different type of hardness measurement, the overall trend between a conventional martensite and the various hardened phases is similar.

Axial microsections through both single and multiple rub transformed material at various tempering temperatures are alike in their appearance and show precipitation, coarsening and eventual spherodisation of the carbide. However these microsections do not reveal whether the carbide is derived during martensitic decomposition or is present as a fine

dispersed phase. No evidence of decarburisation or oxidative effects at the surface or sub-surface were observed.

Scanning electron microscopy of axial microsections gave some limited indication of structure in transformed material, though in general the contrast and definition of images obtained was disappointing.

Mixed nitric and hydrochloric acids in glycerol (etch number 3 Appendix 4), in addition to developing deep etch pits, gave some indication of structure in phase transformed material (Figure 8.36). From these features a grain size of $<0.5 \mu\text{m}$ is implied for transformed material.

Comparisons between the X-ray spectra of the initial C73 material and that of phase transformed material in single and multiple rub using an EDAX system linked to the SEM revealed no discernable differences in composition for elements up to Sodium in the period table. EDAX enabled the identification of the material adhering to phase transformed material, as found in many of the microsections, as originating from the hardened tool steel counter rubbing surface.

To reveal any changes in the low atomic number elements, particularly carbon, nitrogen and oxygen between phase transformed material and substrate an Auger electron spectrometer was employed. It has been summarised in Chapter 2 how the addition or incorporation of these elements has been postulated to account for the properties of white layers. The technique of elemental depth profiling relies, in this method, on the fashioning of a well defined spherical crater on the surface using a hardened steel ball and diamond paste. This method for

phase transformed material was found to be laborious in the time taken for acceptable crater formation. Within the detection limits of this technique (typically $\pm 0.06\%$ wt) comparison with an unworn surface indicated no differences in oxygen, nitrogen, carbon or iron content. A small increase in carbon content in phase transformed material was also observed in the unworn surface. Limitations of machine time and cost did not allow examination of material formed in multiple rub. In the light of the results of Chapter 8 little intrinsic change would be expected with material in multiple rub.

The last two paragraphs have discussed results essentially negative in nature because they demonstrate the identical elemental composition of phase transformed material with that of the substrate or of the original unworn material. However this does not detract from their value. In the light of the approach and theory of this thesis a positive difference would require explanation.

Comparative X-ray diffraction patterns were obtained for material carefully abraded from an unworn surface and from material from single and multiple tracks. The X-ray spectra of the unworn surface revealed a large peak corresponding to α Fe (110) and Fe_3C (022) and a second smaller peak corresponding to αFe (200). Both diffraction patterns from transformed material revealed line broadening and some limited evidence of an additional peak corresponding to $\alpha'\text{Fe}$ (200) i.e. martensite. There is some question as to the accuracy of this last result both in the relative interpretation of the extra peak and the line broadening; therefore identification of a BCT structure indicating martensite, although a likely conclusion cannot be stated with certainty.

The formation of phase transformed material on pins, manufactured from C73, rubbing against a hardened T1 disc under dry conditions, allowed easier TEM foil preparation and obviated anticipated problems with foil preparation from the curved surfaces of the crossed cylinders specimens. Taper microsections and microhardness measurements serve to illustrate the similarity of phase transformed material produced with that found in multiple rubbing of the crossed cylinders experiments. Since the only difference between these experiments is the specimen geometry the simplifying assumption of similarity has been made. Even if this were not so, the analysis can be considered in terms of material formed on a pin on disc machine alone.

Transmission electron micrographs illustrate the differences between a conventional martensite, fully transformed material and the transition region. The electron micrograph of fully transformed material reveals a very small grain size of 3 nm or less while, in confirmation, selected area diffraction patterns show partial and complete rings with some preferred orientation. These rings were indexed as a BCT structure indicating the transformed material to be martensite. There is some line broadening which is caused by a small particle size and presence of microstrains associated with a distorted tetragonal lattice. This may also account for the lack of perfect correlation between the ring pattern and that of a standard BCT structure. The transition region appears to consist of broken carbides interspersed with original and transformed material. This is indicated by the part ring-like part spotty SAD pattern, again with some preferred orientation. In the foil examined there was no evidence of a partially transformed region, however since optical microsections show considerable deformation at this region and

and little evidence of partial transformation, this effect is not unexpected.

The analytical techniques which have been discussed above form only a limited portion of the total work described in this thesis.

Nevertheless, with this reservation, the burden of evidence thus provided indicates that phase transformed material is a fine grained martensitic structure. In multiple rubbing conditions the repeated temperature and deformation cycles can modify or distort the initial transformed structure. Differentiation between a BCC or BCT structure so formed and typical of most rubbing experiments may therefore become subject to interpretation.

9.4 Conclusions

The work described in this thesis may be summarised as follows:

1. The occurrence of hardened, apparently structureless and etch resistant material formed, typically, on sliding ferrous surfaces has been shown to take place over a wide range of engineering situations and metallurgical conditions. Terms used to describe this material include white layers, white phase and phase transformed material.
2. Various theories have been put forward to account for the nature and formation of this material, though none has gained universal acceptance. The major and significant theories can

be grouped under the headings of metallurgical phase changes, deformation structures and the effects of external environment.

3. The approach of this thesis has been to consider the formation of this material as a surface phase change induced by the transient frictional flash temperatures during a single well defined rub.

This would therefore be expected to have a structure akin to the martensitic structure of conventional ferrous metallurgy.

4. As yet reliable measurements of flash temperatures between contacting ferrous surfaces are limited and theoretical calculations have been used. Archard's (1959) flash temperature theory has been employed because of its mathematical simplicity and of the range of evidence to support its validity.

5. Previous work by Archard (1959) and Welsh (1965) using existing theory based on the application of conventional ferrous metallurgy suggested that temperatures of approximately 750°C are needed to produce phase transformed material on dry rubbing steel surfaces.

6. Exploratory single rub experiments on a crossed cylinders machine using soft plain carbon steels rubbing against themselves under unlubricated conditions have suggested significantly higher temperatures of the order of 1100°C , are needed to produce phase transformed material.

7. A re-examination of the metallurgy of steels indicates that in the range $723-910^{\circ}\text{C}$, the times needed to complete the transformation

to homogeneous austenite (and on quenching martensite) are higher than typical flash temperature durations by 3 to 6 orders of magnitude.

8. This re-examination of the metallurgy of steels suggests that above 910°C the transformation is more rapid and the probable rate determining process is the diffusion of carbon in austenite. A theoretical analysis of the diffusion in a pearlite structure suggests that this transformation is not completed in a single rub unless the temperature is raised to the order of 1250°C .

9. Above 910°C (say 1000°C) the transformation can be completed in a number of multiple rubs (say 150) that is likely to apply in a wear experiment. Welsh's T_2 , T_3 and the regime between these transitions can be reinterpreted in terms of this metallurgical theory.

10. More detailed experiments on the crossed cylinders machine using soft plain carbon steels rubbing against a hardened tool steel under revised sliding conditions avoided many of the problems associated with the earlier experiments.

11. These latter results confirmed the magnitude of the flash temperatures required to form phase transformed material in a single rub. Since diffusion is biased towards the maximum rather than the mean temperature occurring in a contact and that the vast proportion of points within a circular sliding contact will at some time experience temperatures in excess of the mean flash temperature

during the duration of the contact, it is proposed that formation of phase transformed material can best be considered in terms of the maximum temperature.

12. A sub-surface temperature theory has been formulated which when applied to a typical microsection also indicates that the formation of phase transformed material is biased towards the maximum temperature.

13. A review of the effects of pressure, deformation and shear stresses on the metallurgy of steels and their possible influence on phase transformations in rubbing indicates that these effects may be much less marked on the present experiments than might appear at first sight.

14. The burden of evidence from a range metallurgical and analytical techniques denotes that phase transformed material to be a fine grained martensitic structure.

15. Thus the metallurgy of phase transformations which occur to produce white layers needs to take into account the short duration of the high temperatures involved.

9.5 Suggestions for further work

Various features of this thesis require additional work if the theories and results contained are to gain wider acceptance.

There is a need for further development and correlation of flash temperature theory with measurement. Direct measurement of these temperatures would be of considerable advantage. Infra-red techniques would seem to offer the most promising approach. Unambiguous measurement of flash temperatures at contacting ferrous surfaces remains to be resolved.

A sub-surface temperature theory has been proposed which clearly requires further examples of applicability before full confidence can be placed in its conclusions.

In Chapter 5 a new metallurgical theory has been proposed where the formation of homogeneous austenite has been considered to occur in two temperature ranges i.e. 723-910°C and >910°C. This theory is based upon a range of observations and clearly a co-ordinated study of the effect of short heating times (say 1 millisecond or less) on the austenitisation of steels would clarify the situation. Laser heating techniques have been recently employed to achieve short heating periods, the results of which have largely confirmed the new metallurgical theory proposed. However the observation of the A_{c1} temperature not exceeding approximately 800°C for carbon steels by Orlich (1974) requires further study.

The importance of diffusion of carbon in austenite on the formation of phase transformed material could be demonstrated by altering the distribution of carbon within the original structure. This could be achieved by producing carbide lamellae of varying interlamellar distances in pearlite by suitable heat treatment or by hardening and tempering where the carbide would be distributed as roughly spherical particles in

a matrix of ferrite. In the latter case the diffusion theory presented in Chapter 5 for homogenisation is not directly applicable because the geometric configuration is rather more complex.

The main experiments have been confined to transformations occurring in two plain carbon steels 0.52% and 0.73%C rubbing against a hardened tool steel over a limited range of loads and sliding speeds. Clearly the pattern established should be confirmed over the full range of plain carbon steels, again as further indication of the importance of carbon diffusion. To extend the range of loads and speeds will be difficult with the present machine. At the higher sliding speeds, approximately 5.5 m/s, the speed available for the ram to traverse the lower specimen and produce the single contact helical track was only just sufficient to ensure no overlapping arose, while heavier loads would only be achieved by using solid cylindrical specimens. At the high loads (790 N) gross transfer was also found to occur in single rub, with the standardised cleaning procedure.

The interpretation of results has mainly been confined to the formation of phase transformed material in single rub. The difficulty of extending this analysis to multiple rubbing conditions has been discussed, for example, in terms of the variable coefficient of friction and subdivision of contact. Chapter 6 has discussed the production of phase transformed material in a number of multiple contacts typical of a wearing surface. The next stage of this work must involve an analysis of the effect of increasing the number of contacts at a rubbing surface and its effect on the cycle of production of phase transformed material.

Some limited metallurgical and physical analytical techniques have been attempted to identify the nature of phase transformed material. Whilst the burden of evidence favoured a fine grained martensitic structure, extensive further work is possible and is required for unambiguous identification of this material. In particular, it has been suggested (private communication) that reflection Mossbauer spectroscopy is a useful method in the examination of surface layers. This method involves the radiation, absorption and scattering of X-rays without recoil. It is characterised by an unusually high resolution and has allowed the distinction between retained austenite, ferrite and martensite at scuffed surfaces.

The work described in this thesis took as its starting point the correlation between the formation of phase transformed material with the reversion from severe to mild wear and at calculated flash temperatures of approximately 750°C. Exploratory experiments using a crossed cylinders machine to produce the material, under well defined conditions, in single rub found considerably higher temperatures were necessary. Therefore the major objectives of this work have been to account for this apparent discrepancy by formulation of a new metallurgical theory for phase transformations in steels for times of the order of 1 millisecond or less and to demonstrate how the theory can be used to re-interpret Welsh's classical T₂ and T₃ wear transitions. Later improved experiments have confirmed the magnitude of the required temperatures.

Improvements both to the theories and experiments will no doubt be made in the future. However it is thought that the work described in this thesis is the first serious attempt to understand the magnitude of the

temperatures required for the formation of phase transformed material in a single rub and to suggest its significance in the cycle of production in multiple rubbing conditions.

APPENDIX 1

Outline solution of diffusion equation with limited solubility

We require a solution of

$$\frac{\partial F}{\partial t} = D \frac{\partial^2 F}{\partial y^2} \quad (5.9)$$

with boundary conditions

$$\frac{\partial F}{\partial y} = 0 \text{ at } y = (l-l^*) \quad t > 0 \quad (A1.1)$$

$$F = 0 \text{ at } y = 0, 2(l-l^*) \quad t > 0 \quad (A1.2)$$

Solve by separation of variables.

Solution is of form $F(y,t) = Y(y)T(t)$

$$\text{Therefore } \frac{1}{DT} \frac{\partial T}{\partial t} = \frac{1}{Y} \frac{\partial^2 Y}{\partial y^2} = -w^2$$

$$\text{giving } F = (P \cos wy + B \sin wy) e^{-Dw^2 t} \quad (A1.3)$$

Substituting equation A1.1 gives

$$F = B \sin wy \cdot e^{-Dw^2 t}$$

Substituting equation A1.2 gives

$$\omega = (2n+1) \frac{\pi}{2} \frac{1}{(\ell-\ell^*)}$$

Resubstituting in equation A1.3 gives

$$F(y,t) = \sum_{n=0}^{\infty} B_n \sin (2n+1) \frac{\pi}{2} \frac{y}{(\ell-\ell^*)} \cdot \exp -Dt \left[(2n+1) \frac{\pi}{2} \frac{1}{(\ell-\ell^*)} \right]^2$$

(A1.4)

To evaluate coefficients of B_n

ie $F = 0$ $y = 0$

$F = C'$ $0 < y < 2(\ell-\ell^*)$

$F = 0$ $y = 2(\ell-\ell^*)$

Odd function, half range period $p = 2(\ell-\ell^*)$

$$f(y) = \sum_{n=1}^{\infty} b_n \sin \frac{n\pi y}{p}, \quad b_n = \frac{2}{p} \int_0^p f(y) \sin \frac{n\pi y}{p} dy \quad (A1.5)$$

Substituting boundary conditions into equation A1.5

$$b_n = \frac{1}{l-l^*} \int_0^{2(l-l^*)} C' \sin \frac{n\pi}{2(l-l^*)} y dy$$

$$= \frac{2C'}{\pi} \left[-\cos \frac{n\pi y}{2(l-l^*)} \right]_0^{2(l-l^*)}$$

$$= \frac{2C'}{\pi} [1 - \cos n\pi]$$

(A1.6)

$$\therefore b_n = \frac{4C'}{\pi} \sum_{n=0}^{\infty} \frac{1}{(2n+1)}$$

Substitute in equation A1.5 gives

$$f(y) = \frac{4C'}{\pi} \sum_{n=0}^{\infty} \frac{1}{(2n+1)} \sin \frac{(2n+1)\pi y}{2(l-l^*)} \quad (A1.7)$$

Comparing equation A1.7 with A1.4 and substituting $C = C' - F$ gives

$$C(y,t) = C' \left[1 - \frac{4}{\pi} \sum_{n=0}^{\infty} \frac{1}{(2n+1)} \sin \frac{(2n+1)\pi y}{2(l-l^*)} \exp -Dt \left[\frac{(2n+1)\pi}{2(l-l^*)} \right]^2 \right]^2$$

(A1.8)

in the region $0 < y < 2(l-l^*)$

APPENDIX 2

Solution of $[Dt]_{\text{eff}}$

We are looking for a solution of the integral

$$[Dt]_{\text{eff}} = \int_0^{t^*} D_0 \exp\left(\frac{-Q'}{S(\theta_{bk} + K^* t^{\frac{1}{2}})}\right) dt \quad (\text{A2.1})$$

Simplifying and inserting the values

$$1. \quad K = Q'/SK^*$$

$$2. \quad A = \theta_{bk}/K^*$$

Equation (A2.1) becomes

$$[Dt]_{\text{eff}} = D_0 \int_0^{t^*} \exp\left(\frac{-K}{A + t^{\frac{1}{2}}}\right) dt \quad (\text{A2.2})$$

Let $1/Z = A + t^{\frac{1}{2}}$ Differentiation w.r.t. Z gives: $dt = - \frac{2(1-A Z)}{Z^3} dZ$

with limits (A2.3)

$$t = 0, \quad Z = 1/A$$

$$\begin{aligned} t = t^*, \quad Z &= 1/(A + t^{*\frac{1}{2}}) \quad (\text{let } t^* = t) \\ &= 1/(A + t^{\frac{1}{2}}) \end{aligned}$$

Substituting limits and equation (A2.3) into equation (A2.2) and rearranging gives:

$$[Dt]_{\text{eff}} = 2 \int_{\frac{1}{A+t^{1/2}}}^{\frac{1}{A}} \frac{(1-AZ) \cdot e^{-KZ}}{Z^3} dZ \quad (\text{A2.4})$$

Integrating by parts the individual terms.

$$\int \frac{1}{Z^3} e^{-KZ} dZ = \frac{-e^{-KZ}}{2Z^2} - \int K e^{-KZ} \cdot \frac{1}{2Z^2} dZ \quad (\text{A2.5})$$

$$\int \frac{A}{Z^2} e^{-KZ} dZ = \frac{-A e^{-KZ}}{Z} - \int K e^{-KZ} \cdot \frac{A}{Z} dZ \quad (\text{A2.6})$$

Substituting equation (A2.6) into equation (A2.5) and dividing through by A gives

$$\begin{aligned} \int \frac{1}{Z^3} e^{-KZ} dZ &= \frac{-e^{-KZ}}{2Z^2} - \frac{K}{2} \left(\frac{-e^{-KZ}}{Z} - \int \frac{K e^{-KZ}}{Z} dZ \right) \\ &= \frac{-e^{-KZ}}{2Z^2} + \frac{K}{2} \left(\frac{e^{-KZ}}{Z} + \int \frac{K e^{-KZ}}{Z} dZ \right) \end{aligned} \quad (\text{A2.7})$$

Subtracting equation (A2.6) from equation (A2.7) gives the integral

$$\int_{\frac{1}{A+t^{1/2}}}^{\frac{1}{A}} \frac{(1-AZ) e^{-KZ}}{Z^3} dZ = \left[\frac{e^{-KZ}}{2Z^2} (2A + K) - \frac{1}{Z} \right]_{\frac{1}{A+t^{1/2}}}^{\frac{1}{A}} + K(A + K/2) \int_{\frac{1}{A+t^{1/2}}}^{\frac{1}{A}} \frac{e^{-KZ}}{Z} dZ \quad (\text{A2.8})$$

The first half of equation (A2.8) can now be found.

Consider the second half. Multiply top and bottom by K .

$$= K (A + K / 2) \int_{1/A + t^{1/2}}^{1/A} \frac{K e^{-K Z}}{K Z} dZ$$

Let $K Z = y$, $dZ = dy/K$

Substituting

$$= K (A + K / 2) \int_{K/A + t^{1/2}}^{K/A} \frac{e^{-y}}{y} dy \quad (A2.9)$$

This can be written as

$$\int_{K/A + t^{1/2}}^{\infty} \frac{e^{-y}}{y} dy - \int_{K/A}^{\infty} \frac{e^{-y}}{y} dy$$

Kreyszig (1972) gives for incomplete γ functions - special case

Exponential Integral.

$$\int_x^{\infty} \frac{e^{-t}}{t} dt = E(1)(x) \approx e^{-x} \left(\frac{1}{x} - \frac{1!}{x^2} \right) + \left(\frac{2!}{x^3} - \frac{3!}{x^4} \right) + \dots (A2.10)$$

t, x are any variables.

Thus by evaluating the series for a value of $\frac{K}{A+t^{\frac{1}{2}}}$ and for $\frac{K}{A}$ and subtracting, the integral can be solved. Therefore

$$\int_{1/A+t^{\frac{1}{2}}}^{1/A} \frac{(1-AZ)e^{-KZ}}{Z^3} dZ = \left[\frac{e^{-KZ}}{2Z} \left(2A + K - \frac{1}{Z} \right) \right]_{1/A+t^{\frac{1}{2}}}^{1/A} \\ + K(A+K/2) \left[e^{-y} \left\{ \frac{1}{y} - \frac{1!}{y^2} + \dots \right\} \right]_{K/A}^{K/A+t^{\frac{1}{2}}} \quad (A2.11)$$

Simplifying this becomes

$$[Dt]_{\text{eff}} = 2 \text{ Do } \left\{ \frac{K}{2y} e^{-y} \left[(2A + K) \left\{ \frac{(1-1)}{y} + \frac{2!}{y^2} \left(\frac{1-3}{y} \right) + \dots \right\} \right. \right. \\ \left. \left. - \left(2A + K \left(1 - \frac{1}{y} \right) \right) \right] \right\} \frac{K}{A+t^{\frac{1}{2}}} - \frac{K}{A} \quad (A2.12)$$

which is the full solution.

APPENDIX 3

Run	Total Temperature °C				Single rub P.T.M.
	θ_{Tp}	θ_{Te}	θ_{Tg}	θ_{Ta}	
1	746	524	483	386	No
2	1055	812	715	572	No
3	1274	1029	856	608	No
4	1717	1412	1114	798	Yes
5	544	450	355	262	No
	1501	1311	1031	910	Yes
6	715	626	496	439	No
	1621	1549	1190	712	Yes
11	825	789	609	370	No
	608	471	416	321	No
12	615	505	425	363	No
14	608	561	432	330	No
15	272	214	191	152	No
16	379	311	263	213	No
17	1461	1340	1022	869	Yes
18	2424	2228	1691	1267	Yes
19	748	690	528	401	No
	1046	750	691	669	No
20	1204	922	812	650	No
21	1492	1140	1001	850	Yes
22	1105	847	745	634	Yes
	1631	1249	1114	966	Yes
23	1207	927	827	718	No
	1460	1051	985	981	Yes
24	1518	1090	998	873	Yes
25	1327	954	874	765	No
	1901	1359	1246	1149	Yes
26	1201	871	797	736	No
	645	469	432	332	No
27	995	716	657	511	No

Appearance of phase transformed material in runs 1-42 in terms of total temperature at surface θ_T (flash temperature + bulk temperature).

Run	Total Temperature °C				Single rub P.T.M.
	θ_{Tp}	θ_{Te}	θ_{Tg}	θ_{Ta}	
28	1639	1027	959	842	Yes
	1333	838	782	688	No
29	1596	998	929	869	Yes
30	2051	1277	1188	1149	Yes
31	1135	713	664	583	No
32	1316	1171	952	706	Yes
	1075	957	780	579	No
33	1721	1533	1242	781	Yes
	1173	1045	850	539	No
34	1804	1591	1300	858	Yes
	1280	1130	925	614	No
35	829	767	592	416	No
36	2089	1913	1466	1106	Yes
	1986	1817	1393	1052	Yes
37	933	854	659	481	No
38	1819	1658	1270	982	Yes
	1728	1575	1206	934	Yes
39	1742	1605	1230	954	Yes
	1486	1369	1050	816	Yes
40*	1667	1387	1220	825	Yes
	1027	856	754	514	No
41*	1126	940	824	603	No
42*	1059	886	780	723	No

Appearance of phase transformed material in runs 1-42 in terms of total temperature at the surface θ_T (flash temperature + bulk temperature) * denotes C52 specimens.

APPENDIX 4

Results of Etching Tests

Satisfactory etches

Etchant Composition	Use	Remarks
1. HNO_3 2ml CH_3OH 100ml	General reagent for carbon steels	No structure revealed in transformed material. In multiple rub tempering effects and sub-surface distortion are exposed. Rapid rate of etching. Ferrite and transformed material different in their response.
2. Picric Acid 4g CH_3OH 100 ml	As 1	No structure revealed in transformed material. Exposes carbide lamellae in transition zone and tempering effects. Easier to control rate of etching. Ferrite and transformed material different in their response.

Etchant Composition	Use	Remarks
3. HNO_3 10 ml HCL 30 ml Glycerol 30 ml	As 1	With prolonged etching, some fine structure revealed in transformed material. Forms deep etch pits, particularly in transformed material and in transition region.
4. HNO_3 4 ml Glycerol 100 ml	Used to distinguish ferrite & martensite	No structure revealed in transformed material, differing response, for ferrite. Exposes carbide lamellae in transition zone.
5. HCL 5 ml Picric Acid 1g CH_3OH 100 ml	To etch up retained austenite	Similar to 4, also reveals sub-surface deformation.
6. Picric Acid 4% in alcohol - 2 secs. HNO_3 5% in Amyl alcohol - 10 secs	Hardened Steels colours marten- site blue, cementite orange, banite shades of brown	Cementite in pearlite coloured orange, transformed material coloured blue with varying amounts of orange. Conventional martensite gave deep blue colour.

Unsatisfactory etches.

Etchant Composition	Use	Remarks
7. NaOH 25g Picric Acid 2g H ₂ O 100 ml	General reagent used boiling	Faint structure revealed in multiple rub transformed material after repeated swabbing.
8. K ₃ Fe(CN) ₆ 4g KOH 10 g H ₂ O 100 ml	As 7	As 7, pearlite mildly etched. Tends to form etch pits.
9. NaOH 10% 20 ml H ₂ O ₂ 10 ml	General reagent, attacks carbides	No structure revealed in transformed material.
10. Meta nitro- benzene Sulphonic Acid 5% in CH ₃ OH	Colours austenite brown, pearlite blue. Attacks martensite.	Difficult to dissolve. Little effect found using swabbing method.
11. Orthonitro- phenol saturated in C ₂ H ₅ OH 10 ml HCL in Amyl alcohol 20 ml	Attacks martensite and decomposition products	Faint structure revealed in multiple rub after continuous swabbing. Tends to form etch pits.

APPENDIX 5

Indexing of Figure 8.42b

If the structure is BCC
$$\frac{1}{d^2} = \frac{h^2 + k^2 + l^2}{a^2} = \frac{N}{a^2} \quad (\text{A5.1})$$

where d is the

interplanar spacing of the plane (hkl), a is the lattice parameter.

Possible values of h, k & l for reflection (up to 20) are: N = 2, 4, 6, 8, 10, 12, 14, 16, 18, 20.

Likewise if the structure is BCT
$$\frac{1}{d^2} = \frac{h^2 + k^2}{a^2} + \frac{l^2}{c^2} \quad (\text{A5.2})$$

c & a are lattice parameters. Possible values of h, k & l for reflection (up to 20) are: $h^2 + k^2 = 1, 2, 4, 5, 8, 9, 10, 13, 16, 17, 18, 20$.

From Figure 8.42b

Ring No	Ring Diameter (mm)	Ring ratio	$h^2 + k^2$
1	12	$12^2/12^2 = 1$	1
2	15.8	$15.8^2/12^2 = 1.73$	2
3	23.8	etc = 3.9	4
4	27	= 5.06	5
5	35	= 8.5	8
6	39	= 10.5	10
7	43	= 12.8	13
8	49	= 16.7	17
9	53	= 19.5	20

Although 3 rings missing, ring pattern can be indexed as tetragonal in structure.

Each ring is from a definite plane. Using equation A5.2 it is possible to find (h,k,l) for each ring, assuming that the diameter of rings is known and the camera length is constant.

Now

$$d = \frac{2\lambda L}{D} \quad (A5.3)$$

where D = diameter of rings

λ = wavelength = $0.037 \times 10^{-10} \text{m}$ (100 keV)

L = camera length = 0.46m

Therefore camera constant $\lambda L = 1.702 \times 10^{-12} \text{ m}^2$

Substituting equation A5.3 into A5.2 gives

$$\lambda = c \left[\frac{D^2}{4(\lambda L)^2} - \left(\frac{h^2 + k^2}{a^2} \right) \right]^{\frac{1}{2}}$$

For a 0.73%C martensite $a = 2.861 \times 10^{-10} \text{ m}$

$$c = 2.956 \times 10^{-10} \text{ m}$$

(Roberts (1953))

Thus

Ring Diameter (mm)	$h^2 + k^2$	1	Possible planes
12	1	0	{100} {010}
15.8	2	0	{110}
23.8	4	0	{200} {020}
27	5	0	{210} {120}
35	8	1	{221}
39	10	1	{311} {131}
43	13	0	{320} {230}
49	17	0	{410} {140}
53	20	0	{420} {240}

APPENDIX 6

Run	Maximum Total Temperature		Single rub P.T.M.
	θ_{Tg} °C	θ_{Ta} °C	
1	772	613	No
2	1156	922	No
3	1386	979	No
4	1811	1293	Yes
	566	414	No
5	1691	1475	Yes
	813	703	No
6	1934	1150	Yes
	981	589	No
11	666	510	No
12	679	577	No
14	687	520	No
15	294	230	No
16	413	331	No
17	1651	1397	Yes
18	2754	2059	Yes
	846	638	No
19	1115	1079	No
20	1308	1042	No
21	1621	1374	Yes
	1201	1019	No
22	1808	1565	Yes
	1337	1158	No
23	1590	1600	Yes
24	1617	1412	Yes
	1414	1235	No
25	2021	1862	Yes
	1285	1185	No
26	683	519	No
27	1057	818	No

Appearance of phase transformed material in runs 1 -42 in terms of the maximum total temperature at the surface θ_T (maximum flash temperature + bulk temperature).

Run	Maximum Total Temperature		Single rub P.T.M.
	θ_{Tg} °C	θ_{Ta} °C	
28	1552	1360	Yes
	1262	1108	No
29	1506	1407	Yes
30	1935	1871	Yes
31	1072	939	No
32	1541	1138	Yes
33	1259	230	No
	2013	1257	Yes
34	1369	860	No
	2109	1385	Yes
35	1495	985	No
	953	682	No
36	2383	1793	Yes
37	2263	1704	Yes
	1064	772	No
38	2066	1593	Yes
39	1961	1515	Yes
	2000	1548	Yes
40*	1705	1322	Yes
	1982	1339	Yes
41*	1217	824	No
	1337	975	No
42*	1259	1330	No

Appearance of phase transformed material in runs 1 -42 in terms of the maximum total temperature at the surface θ_T (maximum flash temperature + bulk temperature). *denotes C52 specimens.

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ABSTRACT

Metallurgical phase transformations in the rubbing of steels

By R A Rowntree

The formation of phase transformed material in the unlubricated rubbing of plain carbon steels using a crossed cylinders machine, in single and multiple rubs, has been investigated.

Previous studies of the dry wear of steels had suggested that if the frictional flash temperatures reached values of approximately 750°C , then a phase transformed material, akin to the martensitic structure of conventional ferrous metallurgy, appeared upon the rubbing surfaces.

Exploratory experiments indicated that phase transformed material appears as a result of a single rub of the surface only under conditions so severe that the calculated flash temperatures are of the order of 1100°C or more.

The fundamental metallurgical considerations applicable to small volumes of material subject to a combination of high temperatures and pressures of short duration which are the special characteristics of flash temperatures, have therefore been re-examined.

The formation of homogeneous austenite, which is the first stage in the production of transformed material, has been considered to occur in two temperature ranges $723 - 910^{\circ}\text{C}$ and $>910^{\circ}\text{C}$. Only above 910°C can the transformation be completed in a typical flash temperature duration.

Diffusion of carbon in austenite has been proposed as the rate determining process. Theoretical temperatures of approximately 1250°C are calculated for the formation of homogeneous austenite from lamellar pearlite.

It has been found that the pioneering results of Welsh (1965), in particular his T_2 and T_3 transitions can be given satisfactory interpretations in terms of the new metallurgical theory. An important element in these interpretations is the number of rubs to which any element of the surface is subjected. This factor has been deduced from wear theory.

Further detailed experiments have confirmed the magnitude of flash temperatures required for the production of phase transformed material. The influence of the maximum surface temperature has been discussed.

Metallurgical and physical analysis indicates phase transformed material to be a fine grained martensitic structure.