The Dry Wear of Steels II. Interpretation and Special Features

N. C. Welsh

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THE DRY WEAR OF STEELS
II. INTERPRETATION AND SPECIAL FEATURES

By N. C. WELSH,* Ph.D.

Research Laboratory, Associated Electrical Industries, Aldermaston Court, Aldermaston, Berkshire

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[Plate 1]

The pattern of wear outlined in part I is interpreted in the light of further experiments which reveal that the change from severe wear to mild is governed by the hardness and state of oxidation of the surfaces. At light loads ($< T_1$) severe wear is inhibited by the combined effects of strain hardening and oxidation. At higher loads ($> T_2$) mild wear recurs primarily as a consequence of a change of phase induced by frictional heating. The hardness accompanying the phase change is great enough, initially, to suppress severe wear without the intervention of an oxide film. At loads immediately above T_2, however, the hardness tends to fall if rubbing is prolonged and oxidation is again essential to preserve the mild wear state. Sustained phase-hardening does not occur until a higher load, roughly coinciding with the T_3 transition, is attained and this finding has an important bearing on the influence of inert atmospheres. The onset of permanent hardening is not responsible for the divergent pin and ring wear rates at T_3, though the phenomena may be linked by the magnitude of the temperatures required to cause phase-hardening; the T_3 transition and the trend at higher loads have been identified as special effects associated with the thermal asymmetry of the rubbing system.

1. Introduction

It was established in part I of this paper that when soft steels rub together the equilibrium wear process is either of the severe type producing rough surfaces and coarse metallic débris, or of the mild type, producing relatively smooth surfaces and fine, oxidized débris. The circumstances in which these forms of wear develop have been traced and the transition loads and speeds welded into a general pattern. An attempt must now be made to interpret the main aspects of this pattern.

It may be recalled that three distinct hypotheses, based on softening, hardening and oxidation of the surfaces, have been advanced to account for the change from severe wear to mild.

The concept of thermal softening was proposed by Kragelskii & Shvetsova (1955) to explain the transition at a critical value of sliding speed. This speed was assumed to represent the point at which frictional heating increases the plasticity of the contact regions to such an extent that damage is restricted to a thin, superficial layer and the wear rate falls in consequence. This concept was not based on direct evidence; its main attraction seems to have been the fact that it allowed the authors to adumbrate an explanation

* Now at: Research Department, Gillette Industries Limited, Basingstoke Road, Reading, Berkshire.

covering the behaviour of certain other metals, especially the finding that the very soft metal, tin, gave a mild form of wear at all speeds tested.

Kragelskii & Shvetsova did not discuss surface hardening as a possible cause of the transition and this omission is rather curious since they found, like Kehl & Siebel (1939) in earlier work, that if the bulk hardness was increased sufficiently, mild wear developed at all speeds. This finding would seem to provide a firm basis for inferring that the transition in soft steels is promoted by spontaneous *hardening* of the points of contact. Subsequent observations by the present author (Welsh 1957) that the change from severe wear to mild with sliding distance is accompanied by intensive surface hardening, and the conclusion of Archard (1959) that the theoretical 'flash' temperatures corresponding to the critical speed are great enough to induce quench hardening, add further, considerable weight to the hypothesis of surface hardening.

Although the mild wear state on surfaces rubbed together in air is characterized by obvious signs of oxidation, evidence regarding the role of oxidation is mainly negative in its import, since the severe to mild wear transition has been reproduced in nominally inert atmospheres (Kehl & Siebel 1939; Kragelskii & Shvetsova 1955; Welsh 1957). However, observations of this kind lose much of their force in the light of Rosenberg & Jordan's conclusion (1934) that oxidation still occurs in nitrogen or hydrogen of high purity and these authors considered that, so long as the bulk hardness of the steel is sufficiently high, mild wear develops in such atmospheres *because* of the protective action of the oxide film.

Rosenberg & Jordan believed that the oxide film formed in air prevents severe wear whatever the hardness of the steel but this patently cannot hold true for all rubbing conditions and, following the arguments advanced in part I, it is possible to reconstrue the evidence as signifying that the hardness required to support an oxide film in air is less than that required in atmospheres of low oxidizing potential (see figure 18, part I). The hypotheses of surface hardening and oxidation can then be reconciled by postulating that an increase of hardness during the rubbing of soft steels provides the substrate hardness required to support the oxide film. Detailed evidence will in fact be advanced in this paper to validate the dual hypothesis as an explanation of the severe to mild wear transition. The problem has, however, many ramifications when the wear rate pattern is considered as a whole.

2. Surface hardening

In the author's earlier work (1957) intense surface hardening was observed on soft steels of all carbon contents (0·12 to 0·78 %) in the mild wear state and the hardening was accompanied by a definite metallographic change. The high hardness values were always associated with regions in which the ferrite–pearlite structure had changed to one of single-phase appearance (see figures 1 and 2, plate 1). It was assumed that this hard phase represented material transformed to austenite by local temperatures above the α-γ transition point and then converted to martensite by rapid cooling. In some respects, however, the product differed from normal martensite and to avoid issues which may be contentious, the transformation will here be referred to simply as 'phase-hardening'; the metallurgical problems will be discussed in detail elsewhere.

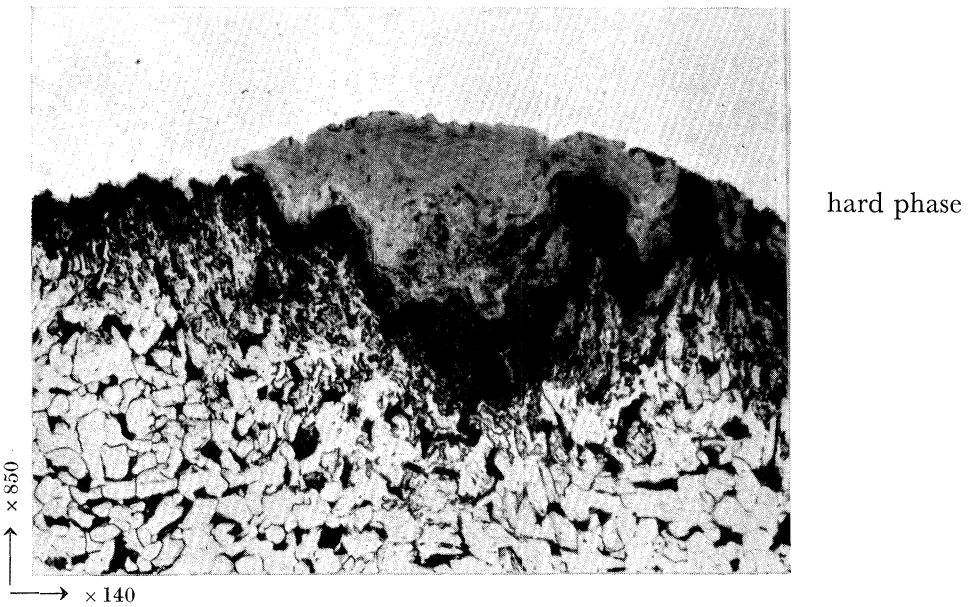

FIGURE 1. Taper section of 0·12 % C ring showing hard transformed material.

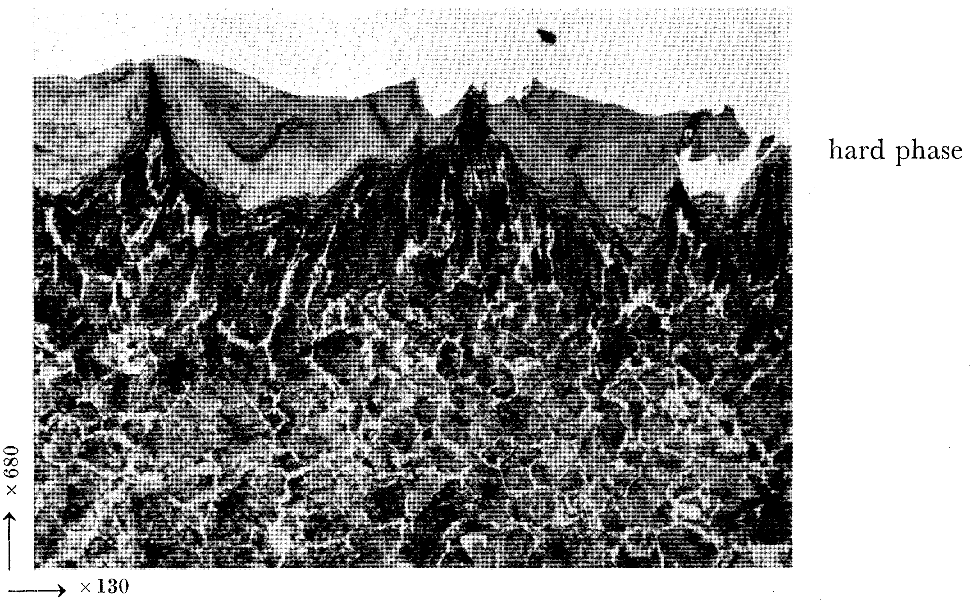

FIGURE 2. Taper section of 0·54 % C ring showing hard, transformed material.

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It is now known that the loads and speeds selected (arbitrarily) in this early work would correspond to points above the T_2 transition for each steel and the results need not be germane to the mild wear state below the T_1 transition. In the present study the hardness changes were investigated in a much more comprehensive fashion. The impressions were made by means of a micro-indenter, with a load of 50 or 100 g.* The values given below are averages of at least ten impressions but, owing to the roughness of the surfaces, high accuracy cannot be claimed. Moreover, the hardness quoted is not the average hardness of the whole surface but only of those areas which appeared to be the contact regions at the moment rubbing ceased. In the mild wear state these usually consisted of burnished patches on a comparatively rough (and softer) background.

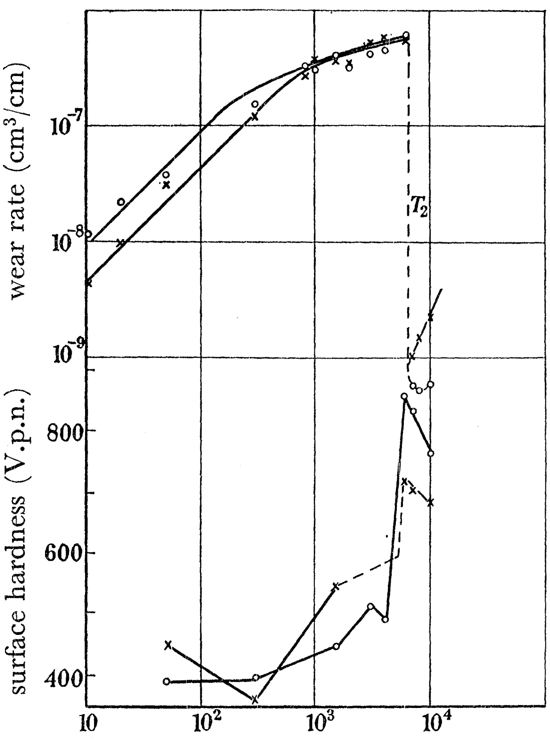

FIGURE 3. Surface hardness changes associated with the T_2 transition for the 0·12 % C steel. Sliding speed 200 cm/s. ×, Pin; ○, ring.

(a) Changes with load

Figure 3 shows the hardness change associated with one of the simplest forms of graph of wear rate against load, i.e. involving only the transition from severe to mild wear (T_2).

At low loads in the severe wear range the surface hardness is about 400 V.p.n. A slight rise occurs at intermediate loads, but immediately preceding the T_2 transition the hardness rises to over 800 V.p.n. on the ring and 700 V.p.n. on the pin.

Figure 3, which is characteristic of all steels rubbed under conditions producing a similar form of wear–load graph, is entirely compatible with the postulate that surface hardening by frictional heating is the cause of the transition from severe to mild wear.

* In comparative tests on bulk hardened samples, the micro-hardness values agreed quite closely with conventional Vickers hardness measurements.

It may be argued that the hardness level obtaining at low loads is merely the result of plastic deformation, i.e. strain-hardening, and that, as the frictional temperatures rise with increased load, a point is reached at which the α-γ transformation point is exceeded

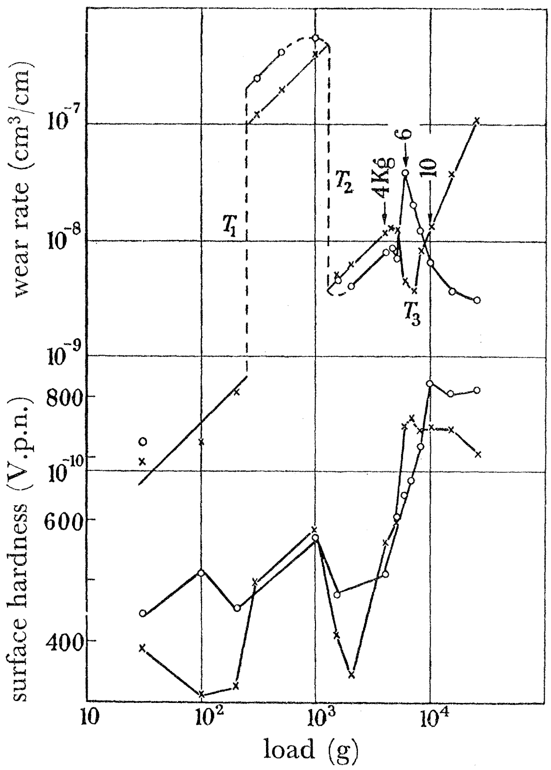

FIGURE 4. Surface hardness changes associated with the T_1, T_2 and T_3 transitions for the 0·63 % C steel. Sliding speed 100 cm/s. ×, Pin; ○, ring.

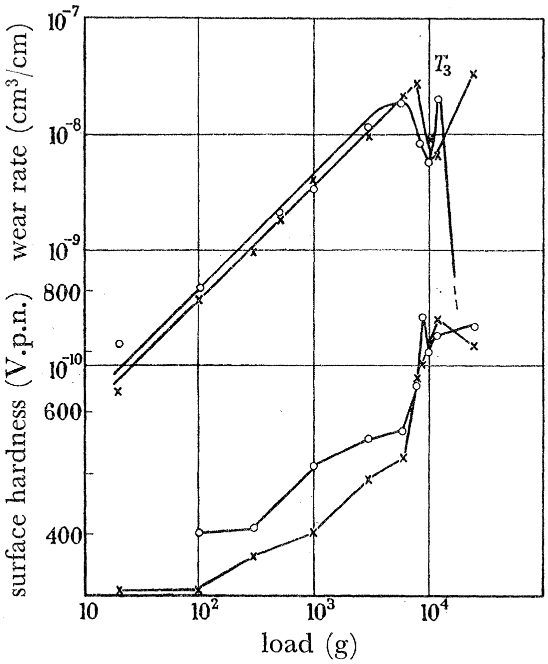

FIGURE 5. Surface hardness changes associated with the T_3 transition for the 0·78 % C steel. Sliding speed 67 cm/s. ×, Pin; ○, ring.

locally and phase-hardening of the points of contact begins. The fact that the hardness rises to high values slightly before T_2 is not surprising on this basis, as a state of dynamic equilibrium may be visualized in which the rate of production of hardened material is offset by its loss through wear and the transition will not occur until a sufficiently large proportion of the points of contact have been hardened (Welsh 1957).

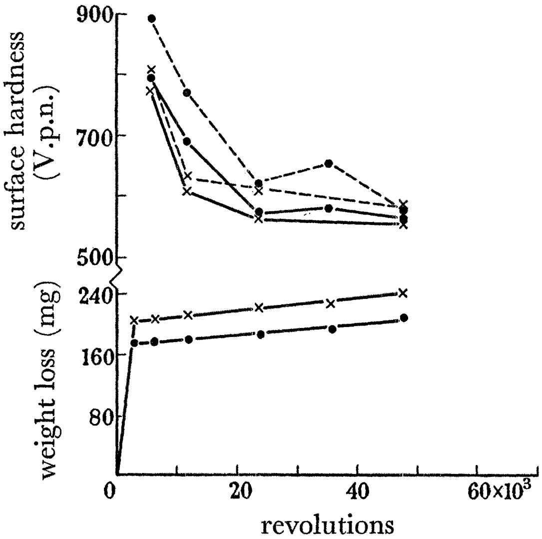

FIGURE 6. Change of hardness with sliding distance for the 0·63 % C steel at a load of 4 Kg. Sliding speed 100 cm/s. – – –, Peak hardness; ——, mean hardness; ×, pin; ●, ring.

This hypothesis receives a serious rebuff when the results in figures 4 and 5 are considered. Figure 4 reveals the hardness developed by the 0·63 % C steel in circumstances where the wear rate–load graphs show the three transitions, T_1, T_2 and T_3. In the mild wear range below T_1 the hardness is moderate (300 to 500 V.p.n.) and little change occurs in passing into the severe wear range between T_1 and T_2. However, on crossing the T_2 transition and returning to mild wear the hardness tends to fall rather than increase, and intensive hardening does not occur until the T_3 transition is attained. In figure 5, for the 0·78 % C steel, the wear rate–load graph shows only the T_3 transition. The surface hardness is seen to rise slowly with load until, in the region of this transition, it reaches values which might be expected from phase-hardening.

(*b*) *Changes with sliding distance*

From the results in figures 4 and 5 it might be concluded that intensive surface hardening is neither the cause nor, in general, a concomitant of the mild wear state and that if phase-hardening is responsible for any transition it is the comparatively minor change in wear rate at T_3. It is, however, essential to note that these hardness tests were made at the end of the considerable period of rubbing required to determine the equilibrium wear rate. A new approach to the problem is provided by figures 6 to 8 in which the average and peak surface hardnesses for the 0·63 % C steel are plotted as a function of sliding distance. In figure 6 the load of 4 Kg represents a point between T_2 and T_3, as reference to the wear rate–load graph in figure 4 makes clear. In the early stages of rubbing the pin and ring tracks are extremely hard (mean 775 to 800 V.p.n.) but the hardness diminishes rapidly to a steady value less than 600 V.p.n., i.e. similar to that shown in figure 4. In figure 7, the load of 6 Kg coincides with the T_3 transition. The pin hardness diminishes at first in a manner similar to figure 6, but then rises again to a high value; the ring hardness fluctuates in a similar, though less pronounced fashion. In figure 8, the load of 10 Kg lies above the T_3 transition and high hardness is now preserved throughout the rubbing period.

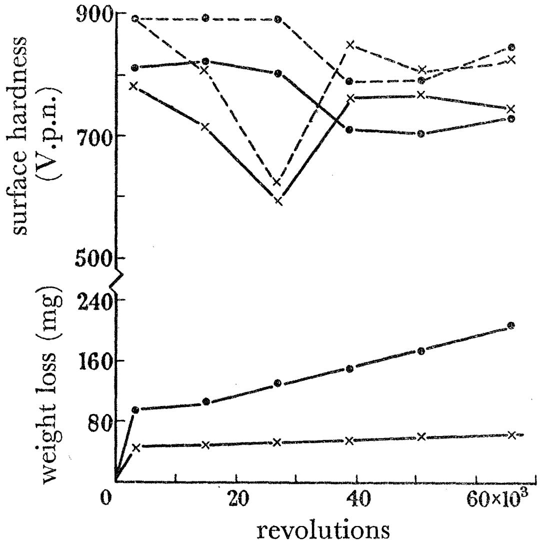

FIGURE 7. Change of hardness with sliding distance for the 0·63 % C steel at a load of 6 Kg. Sliding speed 100 cm/s. ---, Peak hardness; ——, mean hardness; ×, pin; ●, ring.

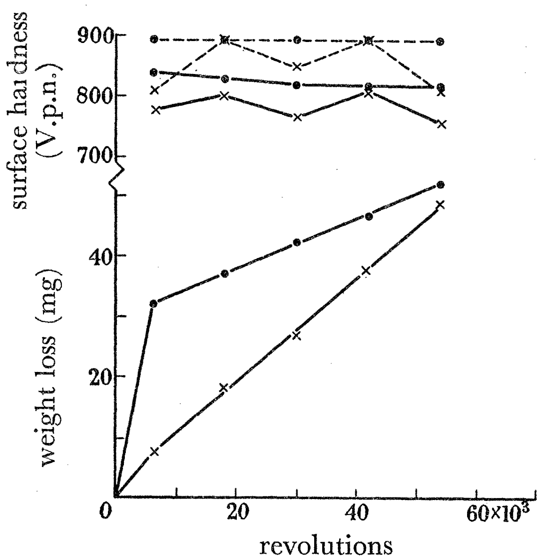

FIGURE 8. Change of hardness with sliding distance for the 0·63 % C steel at a load of 10 Kg. Sliding speed 100 cm/s. ---, Peak hardness; ——, mean hardness; ×, pin; ●, ring.

It is evident that hardness measurements at equilibrium, as in figures 3 to 5, are in some respects deceptive. Intensive hardening does take place at loads between T_2 and T_3 but this hardness is persistent only when the T_3 transition is exceeded. (Figure 3 represents a special case in which the T_3 transition is missing, but the results support the contention in part I that for the lowest carbon steels T_2 overlaps T_3; see figure 16, part I.) It may therefore be deduced that T_3 represents the minimum load at which frictional heating is maintained at a level high enough to cause phase-hardening and that at lower loads these temperatures are developed only in the early stages of rubbing. Since the change from severe to mild wear takes place in these early stages (see the wear curves in figures 6 to 8) it can still be maintained that this change is initiated by phase-hardening; it must now be conceded, however, that very high hardness is not essential to *perpetuate* the mild wear state. The significance of this conclusion will become apparent when the role of oxidation is considered.

3. Influence of oxidation on the T_2 transition

(*a*) *Effect of etching*

The equivocal nature of wear tests carried out in nominally inert atmospheres has already been referred to, but the results of the following experiments, in which the effects of removing the oxide skin were studied, seem to be relatively free from ambiguity.

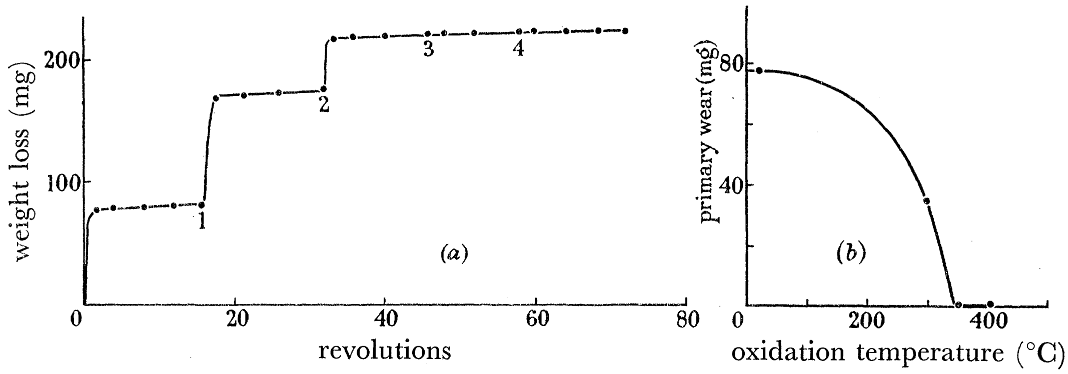

Figure 9. Influence of etching and oxidation on the wear of 0·78 % C steel at a load (4 Kg) between T_2 and T_3. Sliding speed 100 cm/s. Point 1, etched; 2, etched and oxidized at 300 °C; 3, etched and oxidized at 350 °C; 4, etched and oxidized at 400 °C.

A standard wear test was conducted with annealed 0·78 % C steel (198 V.p.n.) until an equilibrium wear rate had been attained. The load, 4 Kg, and speed, 100 cm, were selected to give a point between T_2 and T_3 (see figure 13, part I). The track surfaces on the pin and ring were then lightly rubbed with Selvyt cloth moistened with 50 % hydrochloric acid. The acid rapidly removed all visible signs of oxidation without causing appreciable loss of metal. The run was then continued with the effect shown in figure 9 (*a*). An immediate increase in wear rate occurred, followed by a fall to the equilibrium mild wear rate, i.e. the initial wear behaviour was reproduced. (Only the ring wear is illustrated, but the pin results followed the same course.)

This experiment indicated that in the mild-wear state a tendency for the system to revert to severe wear was suppressed by the presence of an oxide film. Confirmation was provided in the following way. The sequence was repeated (on the same pin and ring) but, after etching, the surfaces were re-oxidized by heating the pin and ring in air; these results are also shown in figure 9 (*a*). After etching and oxidizing for 30 min at 300 °C, the primary period of severe wear recurred, though the amount of initial wear was somewhat reduced. On repeating the sequence, but re-oxidizing at 350 °C, no severe wear was observed, indicating that at this temperature a protective film had formed. As would be expected, the same result was obtained by re-oxidizing at 400 °C. The trend is shown more clearly in figure 9 (*b*). It may be noted that microbalance measurements of the change in weight produced by heating an unrubbed sample of the same steel for 30 min at 400 °C corresponded to a film thickness of about 2×10^{-5} cm. While the oxidation rate of the rubbed tracks may not have been the same, it would appear that an oxide film of this order of thickness can completely inhibit severe wear.

These tests signify that, in these particular circumstances, oxidation plays an important role in the severe to mild wear transition. The results must, however, be considered in conjunction with the following experiments.

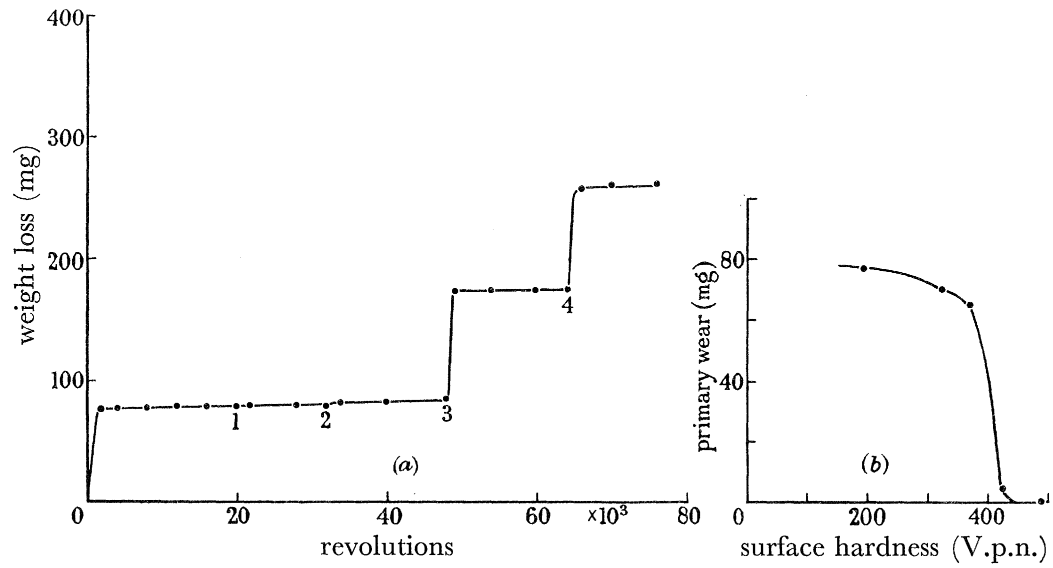

FIGURE 10. Influence of tempering on the wear of 0·78 % C steel at a load (4 Kg) between T_2 and T_3. Sliding speed 100 cm/s. The numbered points refer to heating in argon at the following temperatures: 1, 450 °C; 2, 500 °C; 3, 550 °C; 4, 600 °C.

(*b*) *Effect of heat treatment*

The same general sequence was repeated but, instead of etching the track surfaces, the pin and ring were heated to various temperatures in an atmosphere of high-purity argon. The purpose of the heat treatment was to soften the tracks without disturbing the oxide skin. As the temperatures employed lay well below the temperature of the prior annealing process (850 °C), the bulk hardness of the steel would not be modified. After each heat treatment the wear test was continued. The results are shown in figure 10 (*a*). Heating for 30 min periods at temperatures up to 500 °C did not affect the subsequent course of the wear rate. At higher temperatures, however, a primary stage of severe wear developed. The superficial hardness of the ring track after each heat treatment was determined and the curve in figure 10 (*b*) shows the relationship between the amount of primary ring wear and the track hardness. It is apparent that at some hardness between 375 and 425 V.p.n. the primary wear diminishes very sharply.

The results of the oxidation and tempering experiments are compatible if it is assumed that severe wear is inhibited by an oxide film when the substrate exceeds a critical value. The following experiment is crucial. A pin and ring were run into the mild-wear state and then heated in argon at 600 °C. This temperature lay well above the minimum required to initiate substantial primary wear (see figure 10*a*). The track surfaces were then etched in HCl and re-oxidized by heating in air at 350 °C. As recorded in figure 9*a* the latter treatment, when applied to the as-rubbed surface, completely inhibited the primary wear stage. On the pre-softened tracks, however, the full primary wear developed (see figure 11), demonstrating that the oxide film had now, through lack of adequate support, lost its protective value.

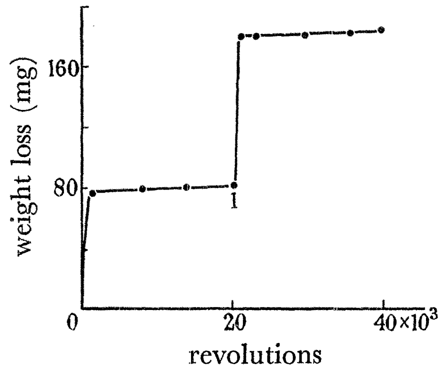

FIGURE 11. Influence of tempering, etching and oxidation on the wear of 0·78 % C steel; load 4 Kg. Sliding speed 100 cm/s. Point 1, heated in argon at 600 °C; etched and oxidized at 350 °C.

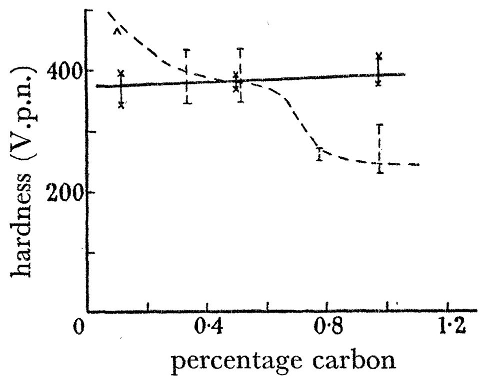

FIGURE 12. Change in critical hardness with carbon content. ⊢—⊣, Critical bulk hardness; ×——×, critical surface hardness.

Similar analyses for the 0·12 % C and 0·52 % C steels gave the same general results. In tempering experiments equivalent to those illustrated in figure 10 severe wear did not recur so long as the surface hardness remained greater than 345 to 395 V.p.n. for the 0·12 % C steel and 369 to 385 V.p.n. for the 0·52 % C steel (measurements made on the ring track). The critical surface hardness values for the three steels are thus strikingly similar; as revealed in figure 12, they also tally well with the critical bulk-hardness values established for quenched and tempered steels in part I, though the critical surface hardnesses appear to be relatively insensitive to carbon content. Exact parity in these values would indeed be rather surprising when their different derivation is considered. The critical bulk hardness is the substrate hardness required to arrest severe wear (the surface hardness may be considerably greater), while the critical surface hardness is the hardness required to maintain mild wear when the substrate hardness is low. Since the interplay of surface and substrate hardness is uncertain, it is difficult to gauge how closely these critical values should correspond.

(*c*) *Effect of sliding distance*

The etching tests described above were carried out after considerable periods of rubbing. It has, however, been demonstrated that in the mild-wear state, at loads between T_2 and T_3, the surface hardness tends to fall quite rapidly (see figure 6). A new and important feature is revealed by figure 13. Figure 13 (*a*) illustrates the results of an etching test on 0·63 % C steel when rubbing at a load (3 Kg) between T_2 and T_3. After a long

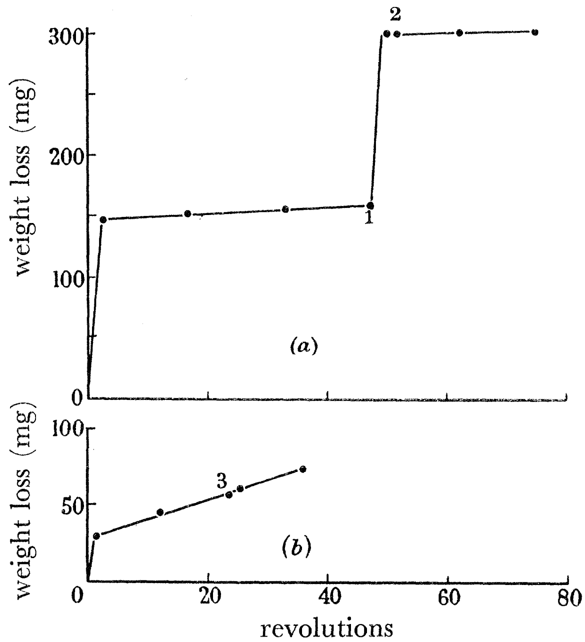

FIGURE 13. Influence of etching the 0·63 % C steel, (*a*) after prolonged and brief rubbing at a load between T_2 and T_3, (*b*) after prolonged rubbing at a load above T_3. Sliding speed 100 cm/s. Point 1, etched, 553 V.p.n.; 2, etched, 802 V.p.n.; 3, etched, 775 V.p.n.

rubbing period the hardness of the ring track had fallen to 553 V.p.n. When the surfaces were etched a substantial primary wear stage was observed in accord with the results in figure 9. The etching was, however, repeated after a further very short rubbing period when the hardness was still high (802 V.p.n.). No primary wear stage was observed, indicating that when the surfaces are very hard they are immune to the effects of etching. This conclusion is reinforced by the results in figure 13*b* for the same steel rubbed at a load (8 Kg) exceeding T_3. At this load intense hardening is persistent (see figures 7 and 8) and after a prolonged rubbing period a hardness of 775 V.p.n. was recorded on the ring track. After etching no primary wear stage was observed.

These results seem to signify that if the surface hardness exceeds a certain value, lying between the limits 553 and 775 V.p.n., severe wear can be inhibited without the intervention of an oxide film (or at least of a thick oxide film). This conclusion has important implications (see Discussion).

4. Influence of oxidation on the T_1 transition

In the light of the foregoing analysis it is natural to infer that the mild-wear state at loads below T_1 is also maintained by a combination of oxidation and surface hardening. A direct attempt to confirm this point by the technique of etching and heat treatment proved to be abortive as the amount of primary severe wear at the low loads was too small, and irreproducible. It has, however, been possible to resolve the issue by a modified procedure which introduces a special feature of the wear pattern.

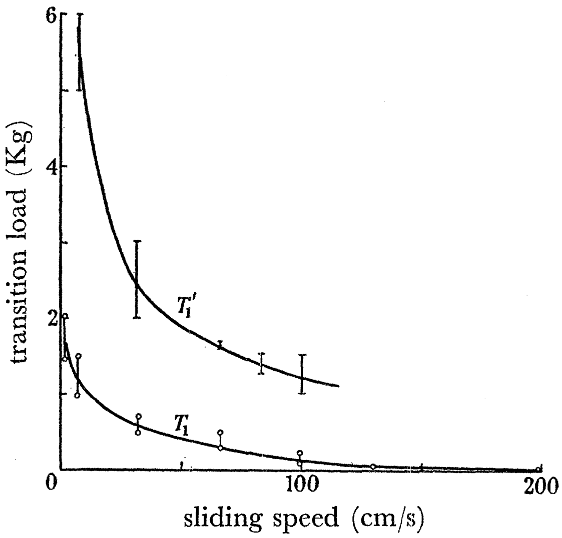

Figure 14. Change in the T_1 transition load produced by incremental loading. 0·52 % C steel. ⊢⊣, Incremental loading; o—o, full load applied immediately.

In all the wear tests described previously the full load was applied immediately, but by incremental loading the low load, mild-wear régime could be considerably extended. Figure 14 shows the effect of progressive loading on the T_1 transition for the 0·52 % C steel at various sliding speeds. In each case the initial load applied was well within the normal mild-wear range and the load was then increased in increments of 100 g with a period of 10 min at each load. The new T_1 value, which will be designated T_1', exceeded the normal value by a factor of 3 to 5. Some surface change evidently occurred at low loads which tended to inhibit severe wear at higher loads. It was apparent that if this inhibiting factor could be nullified while rubbing at some load between T_1 and T_1', the system would revert to continuous severe wear and this change would obviously be more definite than the cursory, severe wear stage below T_1.

Etching the oxide from the surfaces at loads between T_1 and T_1' caused an immediate reversion to severe wear. In view of the results of etching tracks in the mild-wear state above T_2, this behaviour was not unexpected. It was, however, observed that much of the oxide produced within the range T_1–T_1' was loosely adherent and could easily be removed by rubbing with a tissue. An experiment was carried out in which the wear rate was allowed to stabilize at a load between T_1 and T_1' (300 g, 100 cm/s; see figure 14) and the tracks were then rubbed with a tissue while stationary. On continuing the run, severe wear took place immediately. The range T_1–T_1' would therefore seem to be a metastable state perpetuated by the presence of loosely adhering debris on the tracks. It may be noted that rubbing the tracks with tissue at loads below T_1 (or above T_2) did not initiate severe wear.

The effect of the hardness of the substrate on the stability of the oxide skin was studied in the following way. At a sliding speed of 100 cm/s, T_1 for the 0·52 % C steel lay between 100 and 200 g and T_1' between 1000 and 1500 g. A load of 100 g was applied and mild wear allowed to stabilize. The load was now increased to 500 g and mild wear continued. The pin and ring were then tempered in argon at 600 °C for 20 min, reducing the hardness (ring track) from 449 to 268 V.p.n. On continuing the run at 500 g severe wear occurred immediately. The load was then reduced to 100 g to restore the mild-wear state, increased again to 500 g and then tempered at a lower temperature. By repeating the sequence, as illustrated in figure 15, it was established that at surface hardnesses above 377 to 408 V.p.n. mild wear would continue with the load of 500 g.

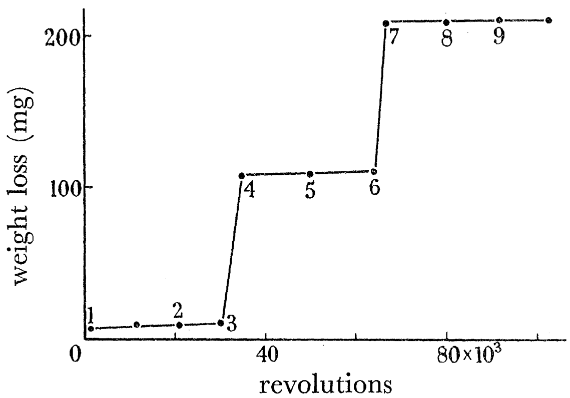

FIGURE 15. Influence of softening the tracks at a load between T_1 and T_1'. 0·52 % C steel. Sliding speed 100 cm/s. Point 1, initial load 100 g; 2, 500 g applied; 3, tempered at 600 °C, 268 V.p.n.; 4, 100 g re-applied; 5, 500 g applied; 6, tempered at 500 °C, 377 V.p.n.; 7, 100 g re-applied; 8, 500 g applied; 9, tempered at 400 °C, 408 V.p.n.

In a similar set of experiments with loads cycled between 100 and 1000 g, a critical hardness of 381 to 417 V.p.n. was established and it appears that this value is not markedly dependent on load. Indeed, when these results are considered in conjunction with the critical hardness determined for this steel at a load above T_2 (369 to 385 V.p.n. at 10 Kg), it would seem that the hardness required to support the oxide film is independent of load, at least in the range 500 g to 10 Kg.

5. Influence of argon atmosphere

The etching and tempering experiments described indicate that surface oxidation is essential to preserve the mild-wear state at loads below T_1 and between T_2 and T_3, but that intensive, permanent hardening which develops at higher loads is sufficient defence against severe wear. With these concepts in mind it is interesting to compare the wear rate–load patterns for the 0·63 % C steel in air and in argon. The argon atmosphere was provided by constructing a Perspex shroud around the pin and ring assembly. This was flushed with argon (99·95 %) for long periods before rubbing began and a constant flow was maintained throughout each run to reduce air leakage.

It is apparent in figure 16 that the severe wear range is much more extensive in argon than in air; T_1 has been reduced from 200–300 g to 30–100 g and T_2 raised from 1–$1\frac{1}{2}$ Kg to 4–5 Kg. The latter point is now almost coincident with the T_3 transition which is itself little affected by the change of atmosphere.

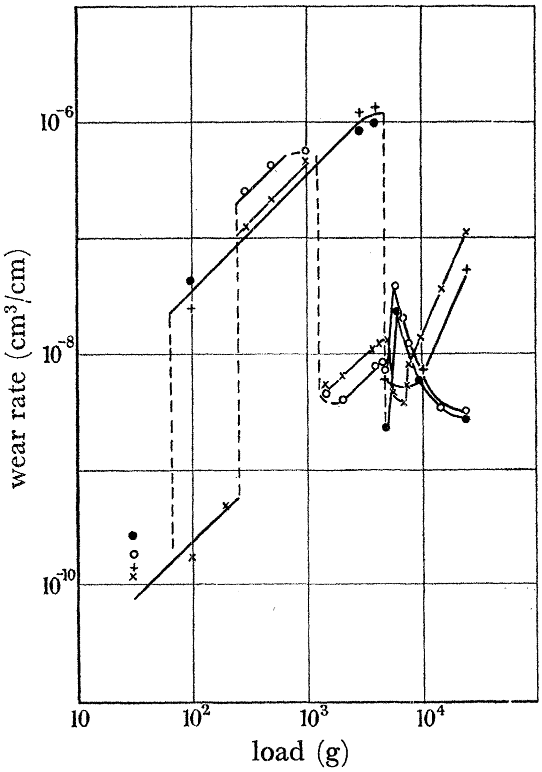

FIGURE 16. Influence of an argon atmosphere on the wear rate–load graph for the 0·63 % C steel. Sliding speed 100 cm/s. × (pin), ○ (ring), air; + (pin), ● (ring), argon.

The track surfaces below T_1 and above T_2 still showed definite signs of oxidation. It may, however, be assumed that oxidation would occur less readily than in air and it is significant, in view of the postulated role of oxidation, that the mild-wear régime below T_1 has been reduced and T_2 displaced towards the point of permanent hardening. It is probable that this latter point had actually been attained as the surface hardness remained high during prolonged rubbing at the new T_2 load.

6. Significance of the T_3 transition

While there is now evidence that the T_3 transition coincides at least approximately with the minimum load required to cause permanent surface hardening, the salient feature of this transition is the fact that the ring wear rate rises to a peak while the pin wear rate usually (though not invariably) falls to a minimum. At higher loads the wear rates again diverge, but in the opposite sense and both features must be in some way associated with the inherent asymmetry of the rubbing system.

An obvious consequence of the disparate size of the rubbing components is illustrated in figure 17 which shows the temperatures attained, at equilibrium, at points $\frac{1}{8}$ in. below the centre of the tracks on the pin and ring. In the high-load range the difference in the temperature of the two components is substantial. Furthermore, if allowance is made for the fact that higher temperatures will prevail closer to the rubbing interface, the absolute values are great enough to raise the suspicion that they may influence the wear rate. (The results apply, of course, only to one sliding speed, but one at which the effects in question were very pronounced.)

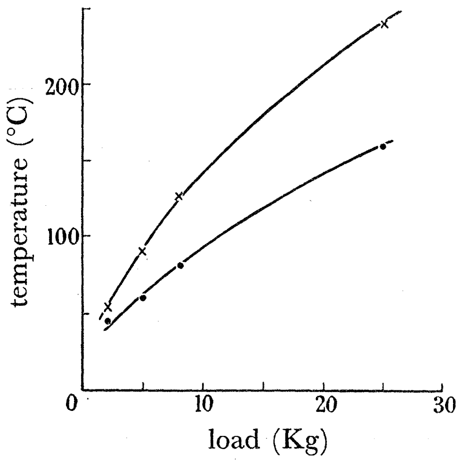

FIGURE 17. Pin and ring temperatures during rubbing at 100 cm/s. Measurements made with fine wire thermocouples in holes $\frac{1}{8}$ in. below centre of tracks. ×, Pin; ●, ring.

It is possible to explain the behaviour at high loads simply as a consequence of thermal softening. Although the surface hardness measurements in figures 3 to 5 do not intimate that the pin softens at high loads, these are final hardnesses. The hardening process involves (it is postulated) local heating to high temperatures and at such temperatures the steel will be very soft. Although, on thermodynamic grounds, the interface temperature must be the same for both components, the thermal gradient in the pin will be less abrupt and the effects of thermal softening therefore deeper-seated. Moreover, since an asperity on the pin can remain in continuous contact with the ring, the hot spots on the pin will tend to be of relatively long duration. It is accordingly easy to visualize that the pin wear may be accelerated by a tendency for the ring to wipe off thermally softened material. If a proportion of this material adheres to the ring its wear rate will diminish (or a gain in weight occur, as often recorded at the highest loads). This view is strongly supported by the fact that at loads and speeds exceeding only slightly the range adopted in this work, the pin wear rate increases dramatically and transfer takes place on a gross scale, macroscopic patches of pin material adhering to the ring.

It would, at first sight, seem to be impossible to invoke the relative softness of the pin to interpret the converse effect, viz. a relatively high ring wear rate at the T_3 point. It is, however, instructive to refer to figure 18 which shows the wear rate–load graph for the 0·63 % C steel when the ring had been pre-hardened by quenching. Results are presented for tests in air and argon and in both instances there is a very marked drop in the pin wear rate in the T_3 region; this drop is indeed more pronounced than for the soft/soft combination (cf. figure 4). In argon there is a corresponding increase in the ring wear rate: this effect is not so well defined in air. The remarkable fact emerges that within a narrow range of load a nominally soft pin can wear away a hardened ring much faster than it wears away itself (approximately $\times 10$ in air, and $\times 20$ in argon). The pin track, of course, was not soft; as the micro-hardness values included in figure 18 reveal, hardening had taken place in the vital range of load. However, the surface hardness never exceeded the bulk hardness of the ring and it is unlikely that this self-hardening could *cause* the differential wear rates in the T_3 region.

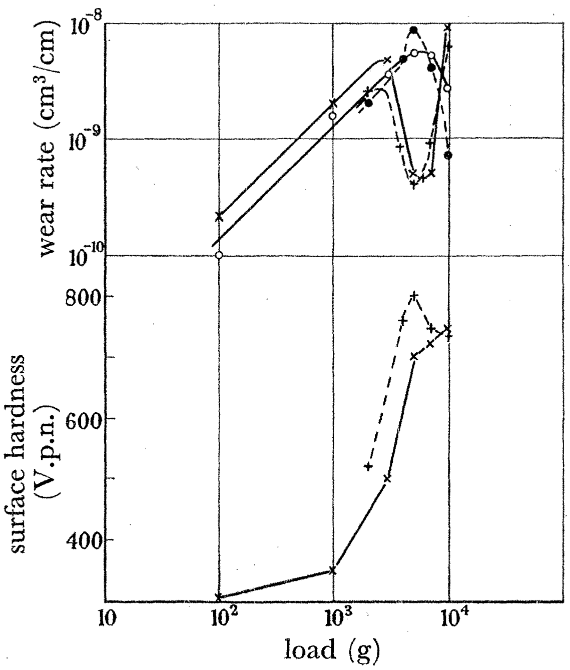

FIGURE 18. Wear rate–load graph for soft 0·63 % C pin on hard ring in air and argon. Sliding speed 100 cm/s. × (pin, 260 V.p.n.), ○ (ring, 780 V.p.n.), air; + (pin, 260 V.p.n.), ● (ring, 780 V.p.n.), argon.

The key to this problem may be found in the results of certain independent experiments shown in figure 19. In these experiments a pin of copper (110 V.p.n.) was rubbed on a ring of hardened tool steel (900 V.p.n.). The loads (30 g to 20 Kg) and speed (100 cm/s) were similar to those employed in the soft steel/hard steel tests illustrated in figure 18. At low loads the copper wore rapidly and the ring wear rate was too low to be measured. At higher loads, however, the hard tool steel wore more rapidly than the soft copper ($\times 3$ to $\times 13$). In this high load range the copper pin acquired a dark, oxidized surface and the explanation for the surprisingly high wear rate of the ring is almost certainly the fact that the copper surface became impregnated with iron oxide, the effective hardness of which exceeded that of hardened steel. A somewhat similar effect has been observed by Lancaster (1963) when rubbing brass on tool steel. It is, of course, well known that oxides formed during rubbing can serve a dual role, inhibiting severe damage but causing abrasion (see, for example, Dies 1943). It is indeed probable that the mutual abrasive action of the oxidized surfaces is a dominant factor in the mild wear process when steel rubs on steel. If then, for some reason, the pin acquires in the T_3 range a more abrasive surface than the ring, the divergent wear rates are explicable. It is conceivable that the disparity in the temperatures of the pin and ring engenders some difference in the amount or type of oxide on the surfaces. However, the important factor may simply be that the relatively pronounced thermal softening of the pin allows its surface to become impregnated with oxide and that oxide mechanically entrained in this way, in the manner of a lap, is more abrasive in character than the normal superficial oxide. If this is so, the fact that the transition in wear rate coincides with the load at which the surfaces are permanently hardened may not be fortuitous, since this is, paradoxically, also the point at which thermal softening is likely to become pronounced.

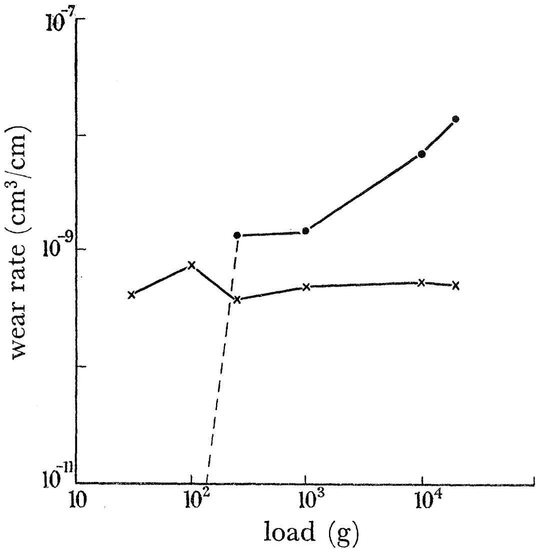

FIGURE 19. Wear rate–load graph for copper pin on high-speed tool steel ring. Sliding speed 100 cm/s. ×, Copper pin; ●, tool-steel ring.

Attention must be drawn to one additional feature of the T_3 transition. According to the foregoing theory, the onset of permanent hardening is not responsible for the divergent wear rates at the T_3 transition. However, close inspection of the wear rate–load graphs in figures 4 and 5 reveals that prior to the T_3 peak the ring wear rate tends to fall in sympathy with that of the pin. This effect, which is quite definite in figure 5 (see also figures 13 and 14, part I), probably does reflect the increased hardness at equilibrium, i.e. the wear rate of both components tends to fall to a slightly lower value characteristic of intensely hardened steel. If so, the T_3 transition must be regarded as the composite effect of three consecutive and partly overlapping phenomena: (1) a decrease in the wear rate of both components due to increased surface hardness, (2) decreased pin wear and increased ring wear due to impregnation of the pin with oxide, (3) increased pin wear and decreased ring wear due to thermal softening of the pin, this becoming the dominant factor at all higher loads.

7. Surface temperature

It is now necessary to consider whether the surface temperatures are great enough to produce the changes postulated. The temperature of the hot spots produced by rubbing like metal pairs can only be estimated and in table 1 the graphical method devised by Archard (1959) has been used to compute the temperatures corresponding to the various transition loads and speeds. The following points must be noted: (1) θ_m is the highest possible temperature, i.e. that which should be generated if all the load is borne momentarily on a single asperity; relatively few contacts will attain this peak value. Moreover,

θ_m is the temperature of the rubbing interface and not that of a finite volume of metal. (2) θ_m varies with the hardness of the contacting asperities. In the compilation of table 1, the bulk hardness of the steel has been adopted but considerable errors may arise in this way since, on the one hand, the surface hardness is increased by strain-hardening and on the other, thermal softening may reduce the effective hardness below that of the bulk steel.

Table 1. Calculated values of maximum hot spot temperatures corresponding to the transitions

			T_1		T_2		T_3	
steel	V.p.n.	sliding speed (cm/s)	load (Kg)	θ_m (°C)	load (Kg)	θ_m (°C)	load (Kg)	θ_m (°C)
0·12 % C	141	20	0·625	70	—	—	—	—
		67	0·150	100	—	—	—	—
		100	0·040	90	18	900	—	—
		133	—	—	9·5	860	—	—
		167	—	—	8·5	950	—	—
		200	—	—	6·5	990	—	—
		266	—	—	4·5	1030	—	—
0·34 % C	205	1·73	2·75	20	—	—	—	—
		33·3	0·5	120	—	—	—	—
		67	0·275	170	—	—	—	—
		100	0·1	160	7·5	850	10	950
		133	0·02	100	5·5	920	—	—
		200	—	—	1·3	750	—	—
0·52 % C	236	1·73	1·75	15	—	—	—	—
		6·7	1·25	50	—	—	—	—
		33	0·6	140	—	—	—	—
		67	0·4	220	20	1030	—	—
		100	0·15	210	5·5	860	7	950
	(A) 212	100	0·15	200	5·5	810	8	920
	236	133	—	—	2·5	810	—	—
		200	—	—	0·125	350	—	—
0·63 % C	260	6·7	3·5	80	—	—	—	—
		100	0·25	260	1·25	310	5·5	900
0·78 % C	278	20	11·75	360	—	—	—	—
		33	9·0	500	—	—	—	—
		47	11·75	720	22·5	920	—	—
		67	—	—	—	—	12	950
		100	—	—	—	—	4·5	900
	(A) 258	100	0·15	200	0·65	400	5	870
	(A) 197	100	—	—	1·6	490	8	870
0·98 % C	319	33	22·5	760	37·5	900	—	—
		47	—	—	—	—	20	940
		100	—	—	—	—	3	830
	(A) 216	100	—	—	1·5	630	5	790

(A), annealed steel.

(3) θ_m is the rise above ambient temperature, i.e. above the general surface temperature. As the results in figure 17 indicate, the latter value may already be substantial. (4) θ_m is proportional to the coefficient of friction μ. Friction measurements during the standard wear tests in the present work gave average values for μ varying, in most circumstances, between 0·8 and 1·0. A value of 0·9 has been used in table 1 but it is important to stress the fact that this is the average coefficient. The peak values of μ were certainly greater, especially in the severe wear stage when the frictional force fluctuated rapidly and widely.

As the calculations involve so many uncertain features, the temperatures recorded must be regarded circumspectly. It is apparent, however, that the θ_m values corresponding to the T_3 transition consistently exceed the eutectoid temperature, about 725 °C, which is the minimum temperature required to initiate the α-γ transformation (neglecting any lowering of the transformation point which may result from high local pressures during rubbing). The θ_m values therefore justify both the assumption that this transformation is the source of phase-hardening and the deduction that T_3 represents the load at which the frictional temperatures at equilibrium sustain the hardening process. The θ_m values corresponding to T_2 vary widely and the lowest values are well below the eutectoid temperature. In no instance, however, would it be necessary to postulate a coefficient of friction much greater than 2·0 to bring θ_m into the hardening range and local values of this magnitude could easily occur in the primary severe wear stage, accounting for the intense but transitory hardening observed at loads between T_2 and T_3. θ_m for the T_1 transition varies very widely indeed, ranging from 20 to 760 °C. There is, of course, no reason to suppose that this transition is governed by the temperature of the hot spots, but θ_m ranges so widely that phase-hardening might, sometimes, occur in this régime. It is relevant therefore to note that minute, intensely hardened patches have often been detected on high carbon steels in the mild wear state below T_1 (though on a very small scale compared with loads above T_2).

8. Discussion

There can be little doubt that the hardness and state of oxidation of the surfaces are the principal factors controlling the wear rate pattern and the major problem in this work has been to disentangle their effects. The vital conclusion is that oxidation can only inhibit severe wear if the substrate hardness exceeds a critical value and hardness per se is only effective if a second, higher value is exceeded.

By softening worn tracks the hardness required to support an oxide film has been identified as falling within the limits 340 to 425 V.p.n. (figure 12). In their softest state, the hardness of all steels is less than this value and severe wear is inevitable in the early stages of rubbing. During rubbing the surface hardness increases but it does not follow that mild wear will inevitably develop once the critical hardness is exceeded. The critical value is the hardness required to maintain mild wear when the surfaces are oxidized and run-in, i.e. the load is distributed over many points of contact. It will, of course, be much more onerous to establish an oxide film during severe wear, when the local stresses are greater and heavy plastic deformation is occurring; the fact that the mild-wear régime below T_1 can be extended by incremental loading (figure 14) testifies to this point. The critical hardness must, therefore, be regarded as a *minimum* criterion for mild wear. The same is true of the second critical hardness identified, by removing the oxide from worn tracks, as falling between the limits 553 to 775 V.p.n. (figure 13); this is the minimum hardness required for mild wear without the intervention of an oxide film.

In the mild-wear range below T_1 the requisite surface hardness is provided mainly by strain-hardening. The T_1 transition may therefore be regarded as the load at which the combined rates of oxidation and strain-hardening are no longer able to outweigh the increasing scale and vigour of the severe wear process. The importance of the rate of oxidation at this stage is confirmed by the influence of an argon atmosphere on this transition (figure 16). It is also relevant to note that the rapid increase in the T_1 load observed at low sliding speeds (figure 6, part I) may stem from the fact that as the speed decreases the time available for the surfaces to oxidize (between repeated contacts) increases; this concept was advanced to explain a similar trend in a mild to severe wear transition observed when rubbing brass on tool steel (Lancaster 1963).

At higher loads frictional heating produces a new and more intensive form of hardening. Since the hardness associated with the phase change exceeds the second critical value it might seem to be unnecessary to postulate a role for oxidation at this stage. It could simply be inferred that once phase-hardened regions are forming with sufficient frequency, severe wear is arrested and mild wear restored. This view would not, however, be consistent with the observed displacement of the T_2 transition in an argon atmosphere (figure 16). The clue to this problem seems to be the finding that, in air, sustained phase-hardening does not occur until some higher load than T_2 is attained, i.e. T_3. At loads between T_2 and T_3 the hardness tends to fall and if rubbing is prolonged the preservation of the mild wear state becomes once more dependent on a combination of strain-hardening and oxidation. If the rate of oxidation is reduced, as by rubbing in argon, it will clearly be more difficult for this final stage to develop and T_2 will be displaced to higher loads. In the range between T_2 and T_3, the phase change can accordingly be regarded as the primary factor tending to suppress severe wear, with oxidation acting in a secondary, consolidatory role. In this range a change in either rate process will affect the load at which the transition occurs, but by varying the rate of oxidation T_2 cannot be made to exceed the point at which permanent phase-hardening occurs, namely T_3.

On this basis it is possible to see why the critical speed, analogous to the T_2 transition, is increased by changing to inert gas atmospheres (Kehl & Siebel 1939; Kragelskii & Shvetsova 1955) and to deduce that there will be a limiting critical speed, representing the point at which permanent, intensive surface hardening occurs. It is also now apparent why, for quenched and tempered steels, the bulk hardness required to guarantee mild wear is greater in inert atmospheres than in air. In Rosenberg & Jordan's tests in hydrogen (1934) the critical bulk hardness for 0·43 % C and 0·81 % C steels fell within the limits 535 to 750 V.p.n. (see figure 18, part I), values which compare well with the critical surface hardness identified in the present work by etching worn tracks on 0·63 % C steel (553 to 775 V.p.n.). It is tempting to infer that these figures reflect a limiting value of bulk or surface hardness above which severe wear cannot develop in any atmosphere, since it does seem improbable, *a priori*, that very hard steels can ever undergo the gross welding and tearing that characterize severe wear. However, oxidation still occurred in Rosenberg & Jordan's experiments despite the high purity of the hydrogen atmosphere and it must, of course, be conceded that a thin oxide film would reform after etching in the present work. It would, therefore, be unwise to extrapolate to surfaces completely devoid of oxide and it can only be maintained that the arguments above should be valid for all 'inert' atmospheres of normal purity. In this context it may be remarked that although this work has been devoted to dry rubbing, there is no obvious reason why the T_1 and T_2 transitions should not recur in the presence of a lubricant, provided that metal–metal contact occurs and that oxygen dissolved or entrained in the oil can cause oxidation. If oxidation is precluded then, following the argument for inert atmospheres, the severe to mild wear transition should still take place (and might even be facilitated by carbon absorbed from the oil).

The change in the transition loads with composition of the steel remains to be considered. It was shown in part I (figure 16) that there is a genuine increase in the T_2 and T_3 loads as the carbon content diminishes. Following the interpretation of the general pattern, the change in T_2 can be explained by assuming that with decreased carbon content the surfaces harden less readily or oxidize less readily, though only the first factor can be invoked for the change in T_3. Regarding oxidation, published information does not indicate a simple relationship with carbon content and measurements of the oxidation rate of filings of the steels in question, at temperatures up to 800 °C, failed to establish any consistent trend. It is, moreover, very uncertain whether such measurements are germane to oxidation during frictional rubbing and this issue must at present be left open. Regarding the hardening propensity, it is common knowledge that with decreasing carbon content the response to thermal treatment diminishes and in the limit, steels of very low carbon content (like the Armco iron studied in this work) are virtually unhardenable. This is, admittedly, over-simplifying the situation since low carbon steels harden more intensively during rubbing than during normal heat-treatment. Nevertheless, it is reasonable to infer that, as the carbon content diminishes it will become progressively more difficult for an effective quantity of phase-hardened material to develop on the surfaces and the transition loads will rise in consequence.

On the dual basis of surface hardening and oxidation it is thus possible to construct consistent explanations for the main facets of the wear rate pattern and to incorporate in the general scheme the principal findings of previous, kindred studies. The special effects associated with thermal asymmetry require no further comment except to remark that wear tests are often conducted in circumstances in which these effects are likely to prevail; unless the wear rates of both rubbing components are specified, misleading conclusions may be drawn.

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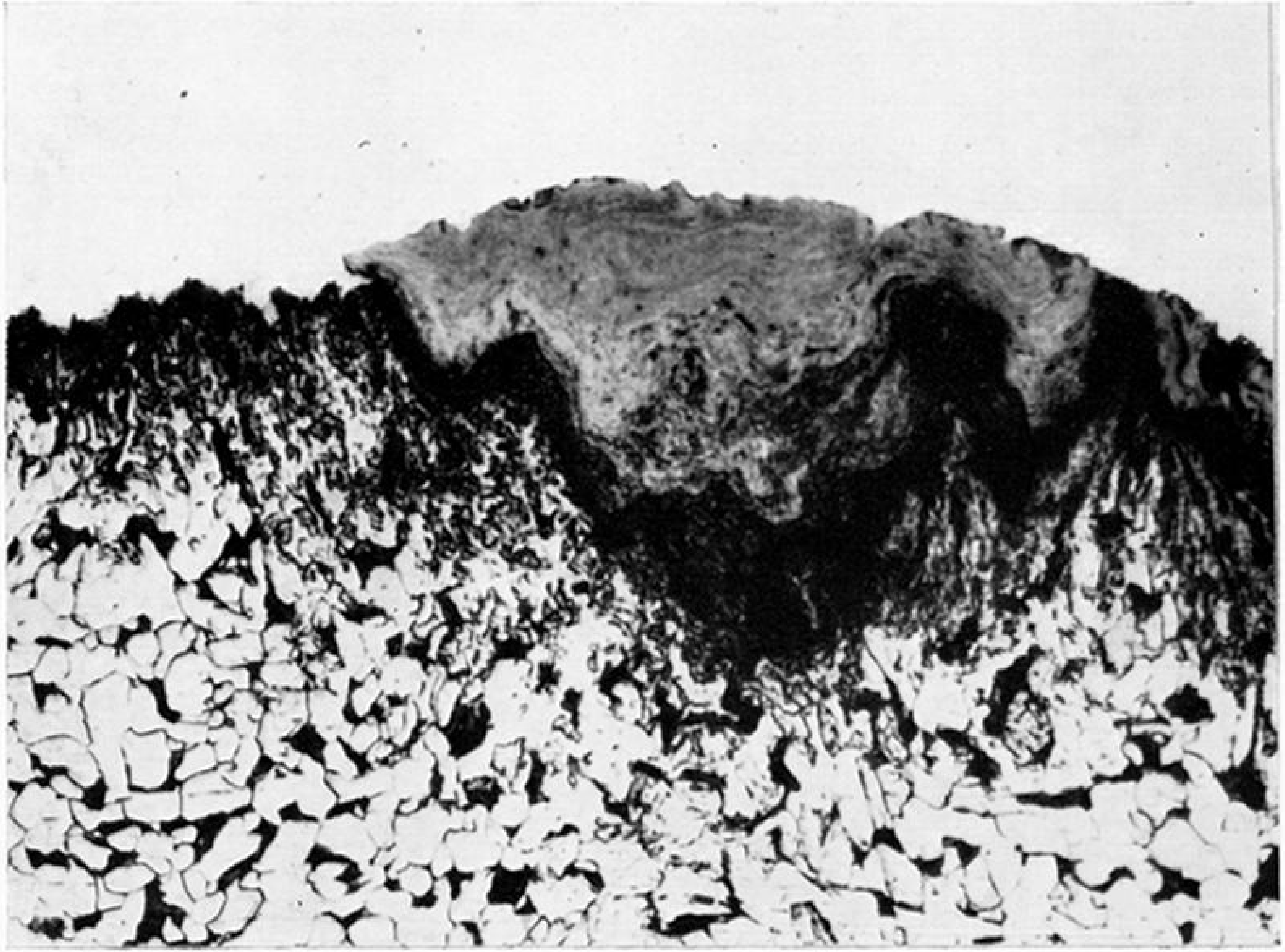

FIGURE 1. Taper section of 0·12 % C ring showing hard transformed material.

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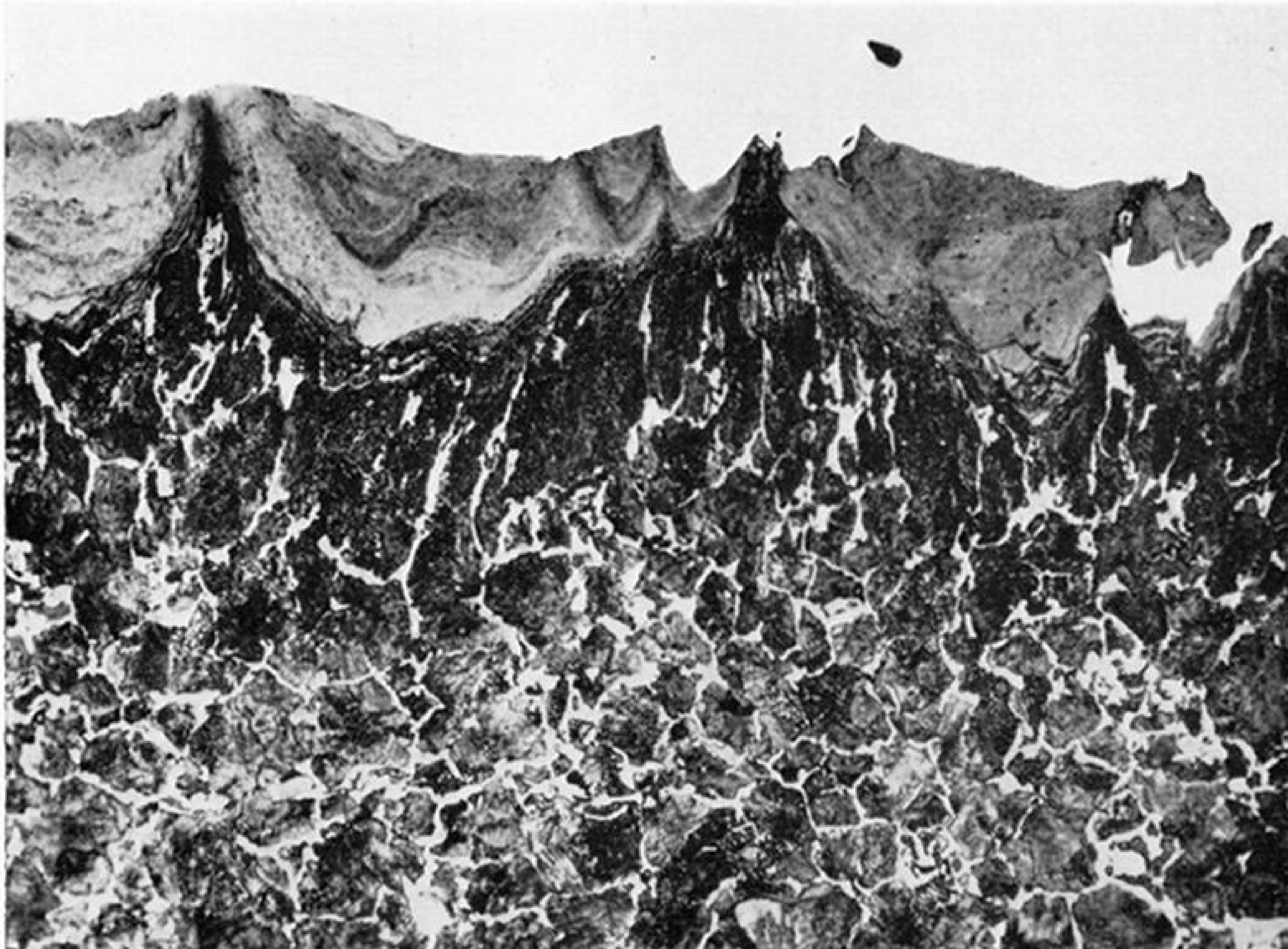

FIGURE 2. Taper section of 0·54 % C ring showing hard, transformed material.

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