Delamination behaviour of high speed steels for hot rolls

A. Molinari, M. Pellizzari, A. Biggi, G. Corbo, A. Tremea

In this paper, the results of a research project aimed at the study of the delamination behaviour of High Speed Steels used for the production of finishing hot rolls are presented. Laboratory wear tests have been carried out on several materials with different composition, and the results are discussed in correlation with their microstructure and mechanical properties. The delamination resistance is strongly related to the matrix microhardness, while the primary carbides have a negative effect.

Parole chiave: acciaio, laminazione, tribologia

INTRODUCTION

In the production of the finishing hot rolls, High Speed Steels (HSS) are receiving an increasing interest since they may achieve better performance than High Chromium Iron /1/. This is due to superior mechanical properties at high temperature, provided by the high content of carbon (1.5-2.5%) and alloying elements, such as V, Nb, Cr, Mo and W which form hard and stable carbides /2/. In addition, HSS have a better wear resistance than Indefinite Chill Iron and they could replace these materials in the last stands if the resistance to accidents would be increased. The combination of the chemical composition and the heat treatment parameters is potentially capable of producing a range of technological properties wide enough to match the specific requirements of the different stands and of the different hot mills.
The roll consumption depends on the wear behaviour of the material. Roll wear is a rather complex phenomenon determined by the combination of the contact stresses, abrasion, surface oxidation and thermal fatigue. During rolling a smooth wear profile must be maintained in order to ensure the regular contact between the roll and the strip, which minimizes the contact stresses, and to reduce re-machining of the roll as much as possible.
The formation during rolling of a mechanically stable and wear resistant oxide layer on the roll surface contributes to the development of such as smooth profile. In contrast, tribological phenomena as delamination, thermal fatigue and thermal fatigue cause excessive wear and worsen the regularity of the roll profile. From this point of view the knowledge of the behaviour of the roll materials with respect to these phenomena is very important for the development of the applications of HSS in hot rolling.
This paper is focused on delamination. Delamination is caused by the nucleation of a subsurface crack whose propagation parallel to the surface produces large plate-like metallic fragments /3/. Crack nucleation is caused by the plastic deformation of the surface and subsurface layers, which can find a favourable situation during rolling owing to the high temperature reached by the roll surface.
In order to study the delamination behaviour of spincast HSS, specific laboratory tests were carried out on several materials with different composition. The tests were carried out on an Amsler trybometer by setting-up the test parameters in order to induce subsurface plastic deformation in the specimens. The aim of this work is the interpretation of the wear behaviour on a metallurgical basis, to individuate the key metallurgical factors which influence the resistance to delamination. The results of the tests are therefore here presented and discussed in relation to the microstructural characteristics of the materials investigated.

A. Molinari, M. Pellizzari, *University of Trento, Italy*

A. Biggi, G. Corbo, A. Tremea, *INNSE Cilindri, Brescia, Italy*

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EXPERIMENTAL PROCEDURE

The materials investigated were produced by spincasting and the chemical composition was varied in a wide range as reported in Tab. I.

C	Cr	Mo	V	W	Si	Mn	Ni
1.0-2.4	3.5-9	2-5	1-6	0-4	0.4-1.3	0.5-1.4	0.6-1.5

Tab. I: Chemical composition range of the HSS investigated (wt. percentages).

Tab. I: Intervalli della composizione chimica degli acciai HSS studiati (percentuali in peso).

The specimens were quenched from 1020-1080 °C and tempered at 480-560 °C in order to develop the secondary hardening for each specific composition.
The microstructural analysis of the materials was carried out at the Optical Microscope (OM) and the Scanning Electron Microscope (SEM), and the volume fraction of the primary carbides (proeutectic and eutectic) was determined by Image Analysis on deeply etched metallographic specimens. Twenty images at 200x were analysed for each material. Both hardness HRc and matrix microhardness HV0.1 were measured.
Wear tests were carried out on an Amsler tribotester by using disks of 40 mm diameter and 10 mm height. Dry rolling-sliding tests were carried out. The application of a force of 500N to the two disks gives rise to a maximum herzian pressure of 430 MPa which, combined with the frictional force caused by the absence of lubrication may induce plastic deformation in the surface and subsurface layers of the material /4/. The rotation speed of the two disks was 400 rpm and 360 rpm respectively, giving rise to a 0.0628 m/sec sliding rate component. Under these conditions the frictional heating causes surface oxidation of the two specimens. The experiments are then capable to induce delamination and oxidative wear together.
The wear curves were obtained by interrupting the tests

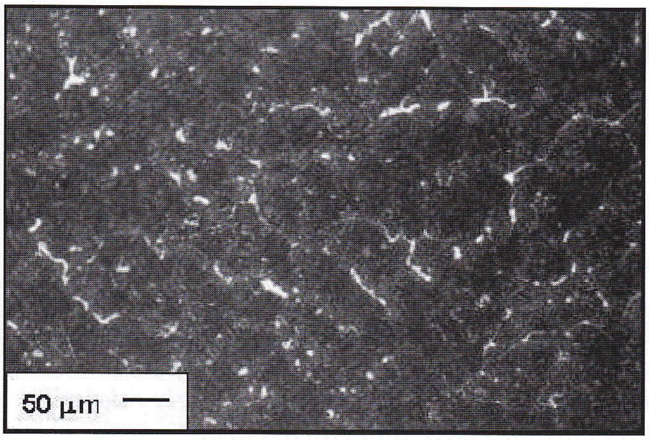

Fig. 1: Microstructure of HSS containing 1.0% C.
Fig. 1: microstruttura dell'HSS contenente 1% C.

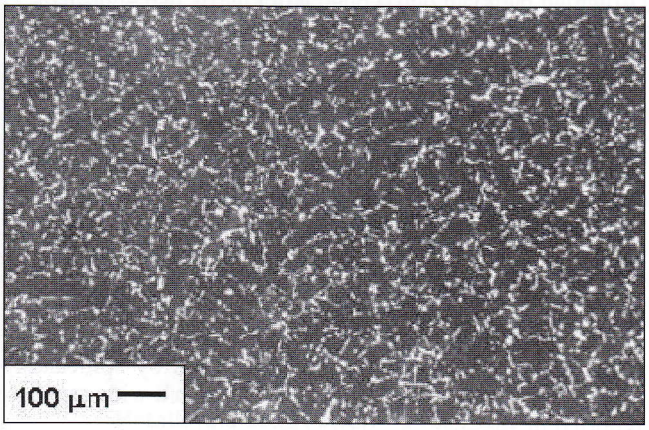

Fig. 2: Microstructure of HSS containing 1.7% C.
Fig. 2: microstruttura dell'HSS contenente 1.7% C.

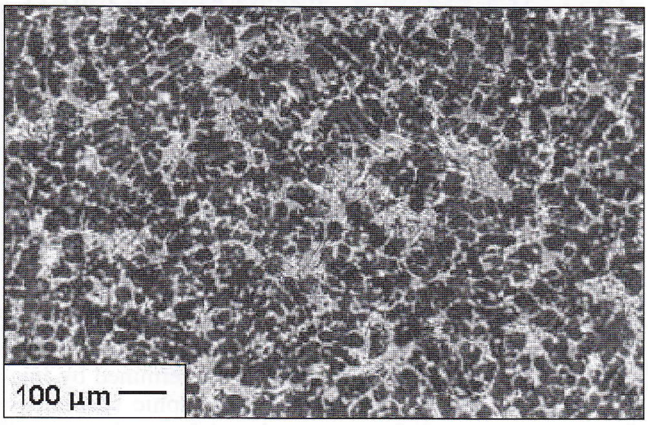

Fig. 3: Microstructure of HSS containing 2.3% C.
Fig. 3: microstruttura dell'HSS contenente 2.3% C.

every 60 minutes, weighing the two specimens with a precision balance (0.0001 g) after acetone cleaning and plotting the cumulative mass loss as a function of the rolling distance. The wear debris was collected and analysed by X-Ray Diffractometry (XRD) and by SEM to confirm their constitution. Moreover, at the end of each test the worn specimens were analysed to observe the surface and subsurface damage.

RESULTS AND DISCUSSION

Figures 1, 2 and 3 show the microstructure of three spincast HSS with different carbon contents (1%, 1.7% and 2.3% re-

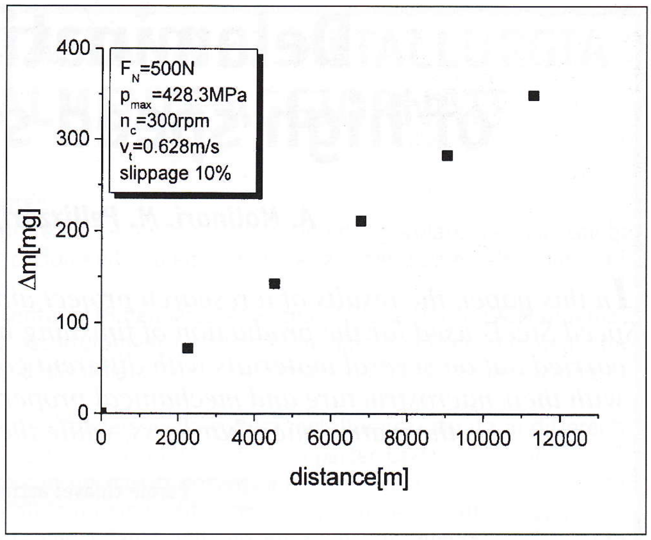

Fig. 4: Wear curve of the low C HSS (1,0% C).
Fig. 4: Curva di usura dell'HSS contenente 1% C.

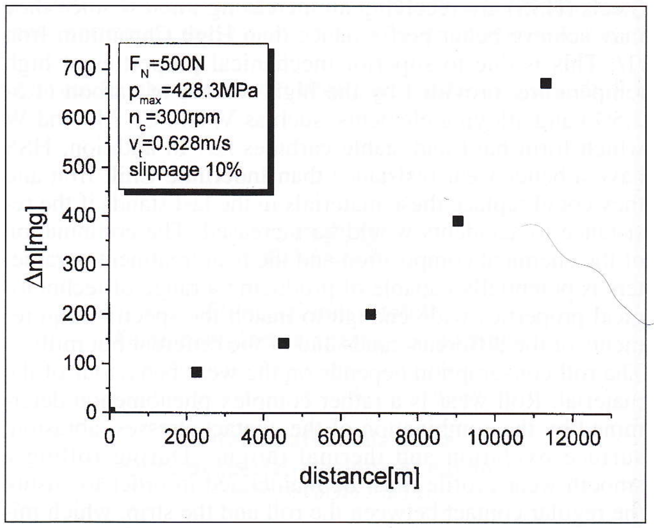

Fig. 5: Wear curve of the high C HSS (2.3% C).
Fig. 5: Curva di usura dell'HSS contenente 2.3% C.

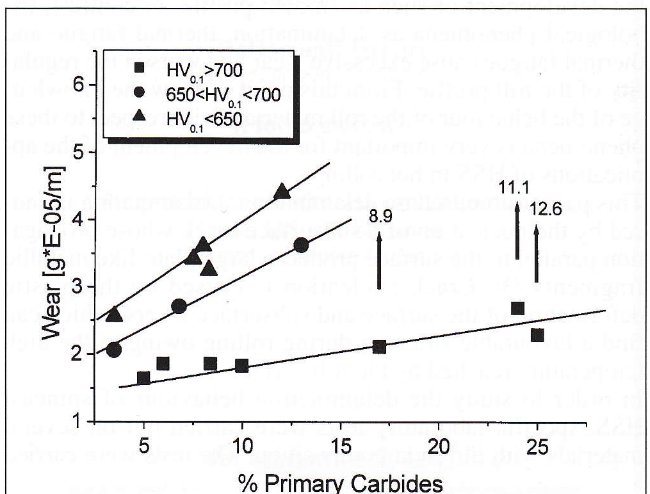

Fig. 6: Wear rate vs. primary carbides percentage for steels with different matrix microhardness
Fig. 6: Velocità di usura in funzione della percentuale di carburi di acciai con diversa microdurezza di matrice.

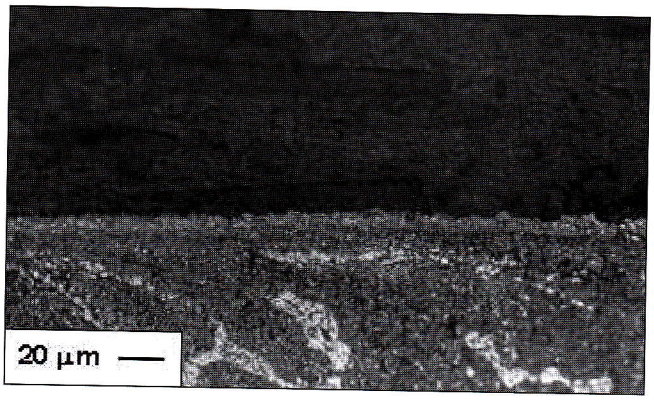

Fig. 7: Cross section of the worn specimen in medium C HSS.

Fig. 7: Sezione metallografica del campione usurato dell'acciaio HSS a contenuto di carbonio intermedio.

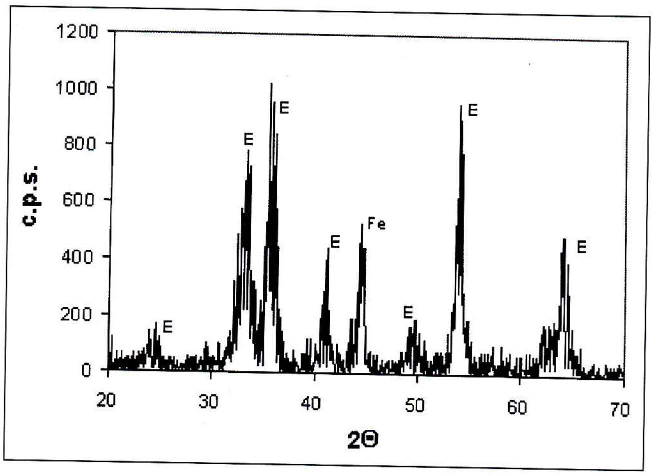

Fig. 8: XRD spectrum indicating metallic particles in the wear debris of medium C HSS.

Fig. 8: Diffrattogramma-X dei frammenti di usura dell'acciaio con contenuto intermedio di carbonio.

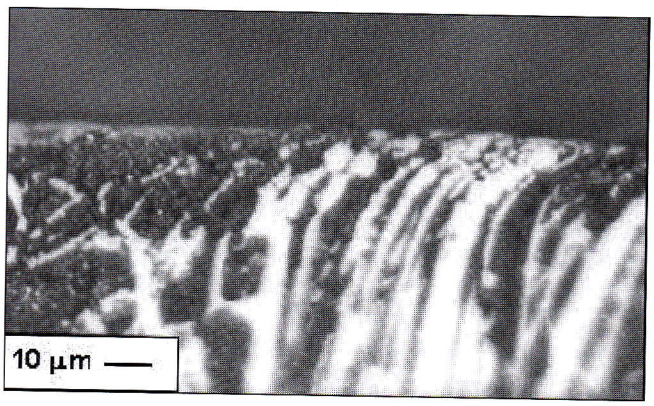

Fig. 9: Cross section of the worn specimen in high C HSS.

Fig. 9: Sezione metallografica del campione usurato dell'acciaio HSS con il contenuto di carbonio più elevato.

spectively). The microstructure comprises a dendritic matrix with a significant amount of primary carbides, both proeutectic (the V-rich MC and M_4C_3) and eutectic.
The carbides precipitating during solidification depend on the chemical composition of the steel. In principle V and Nb in combination with a high carbon content tend to form proeutectic MC and M_4C_3 /2,5/. With a lower carbon content, the formation of the proeutectic carbides is reduced and mainly eutectic carbides form. In this case MC particles tend to dissolve Cr and Mo /6/, differently from the proeutectic ones. Concerning the other carbides, the combination of the Cr, Mo and W contents determines the relative amount of M_7C_3, M_2C and M_6C /7/. In the steels here considered, the M_7C_3 carbide tends to predominate on the others. The amount of primary carbides ranges between 4 and 25 % (in volume), and increases with the carbon content, as clearly shown by figures 1, 2 and 3. The dendritic matrix has a tempered martensite microstructure. Secondary carbides are very fine and they cannot be detected at OM. For this reason the determination of the volume fraction of carbides carried out by Image Analysis does not take into account these carbides. The microhardness of the matrix ranges between 550 and 900 HV0.1. Figure 4 shows, as an example, the wear curve of a low carbon (1% C) and therefore low carbide volume fraction HSS. The experimental points can be interpolated by a straight line, which indicates the constancy of the wear mechanism during the whole test. The slope of the straight line is the wear rate. For comparison, figure 5 shows the wear curve of a high carbon HSS (2.4% C). It presents two steps, both of them being characterised by the constancy of the wear mechanism. A transition from a low wear rate step to a high wear rate step occurs at about 7000 m rolling distance. This kind of curve occurs in all the materials with a carbon content higher than 2%, i.e. with a content of primary carbides higher than 15%.
Figure 6 plots the wear rate versus the content of primary carbides. To interpret the results, the experimental points are drawn with different symbols depending on the matrix microhardness class, as explained in the label. For the materials with the transition in the wear curve, the wear rate of the first step is reported as an experimental point and that of the second step is indicated by the actual value. A general trend may be observed: the increase in the carbide content tends to increase the wear rate on three levels which depend on the matrix microhardness. At the same carbide content, the higher the matrix microhardness, the lower the wear rate. A general conclusion may be drawn: the wear rate decreases by increasing the matrix micorhardness and by decreasing the content of primary carbides.
The effect of carbides is confirmed by the materials with the highest carbide content which display the wear transition; the wear rate in the second step is significantly higher than the wear rate of all the other HSS, even if the matrix microhardness is very high. Figure 7 shows, as an example, the microstructure of the cross section of the worn specimen in the case of a material with an intermediate carbon content. It clearly shows the subsurface plastic deformation which involves the carbide network. It has to be noted that in the outer part of the strained layer the ledeburitic carbides are not detected. They are in fact fragmented and dispersed in the deformed matrix. The wear profile is rather regular, even if wear debris contains metallic particles as demonstrated by XRD analysis, whose results are shown in figure 8. Wear results from the combination of oxidative wear and delamination.
Figure 9 shows the microstructure of the cross section of the worn specimen in the case of a material with high carbon content. Here the fragmentation of the carbides network is very evident and, in addition, the wear profile is rather irregular (fig. 10 - section normal to the sliding direction), clearly indicating the detachment of large metallic particles. The micrograph was taken on a specimen after the wear transition has occurred, so that the damage mechanism shown can be easily correlated to the very high wear rate. Even in this case, wear results from the combination of oxidative wear and delamination, but the contribution of the latter phenomenon predominates on that of the former. Finally the steels with the lower carbon content produce almost only oxide fragments, as shown by the XRD spectrum of figure 11. In this case the contribution of delamination is negligible.

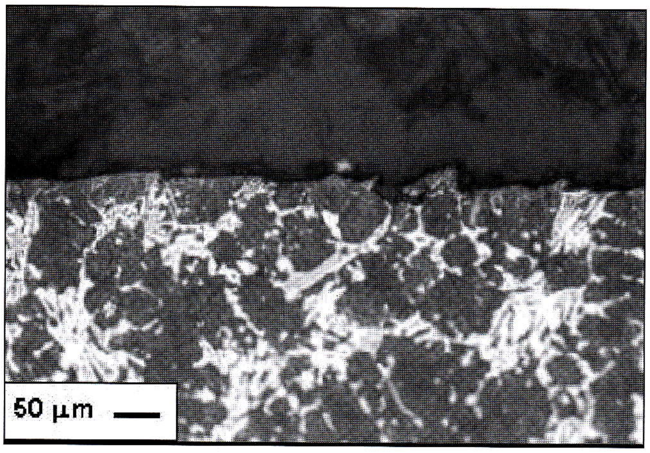

Fig. 10: Cross section, normal to the sliding direction, of the worn in high C HSS specimen.

Fig. 10: Sezione metallografica (normale alla direzione di scorrimento) del campione usurato dell'acciaio HSS con il contenuto di carbonio più elevato.

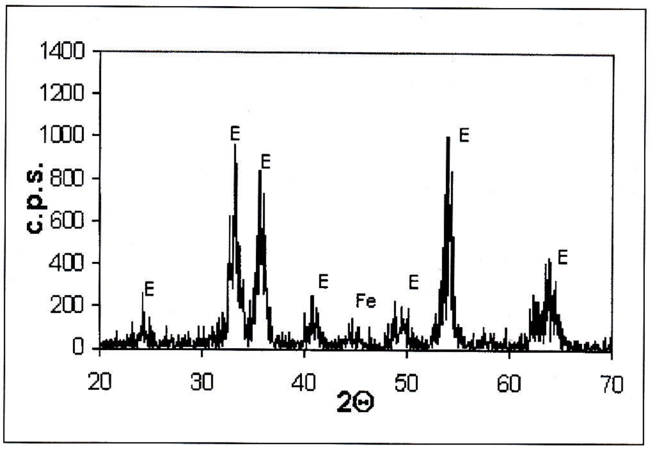

Fig. 11: XRD spectrum indicating metallic particles in the wear debris of low C HSS.

Fig. 11: Diffrattogramma-X dei frammenti di usura dell'acciaio con il contenuto di carbonio più elevato.

To interpret the results, we have to consider that when wear results from the combination of oxidative wear and delamination, the predominant mechanism determines the wear rate. The ability of the material to resist plastic deformation or, alternatively to develop a mechanically stable surface supports the oxide layer, giving rise to a low wear rate. This ability depends on the material hardness which, in turn, depends on the matrix microhardness /8/. The higher the matrix microhardness, the higher the material hardness and, in turn, the resistance to plastic deformation. Therefore in the materials here considered, wear rate decreases by increasing the matrix microhardness. However, the primary carbides have an important role in the yield behaviour of the materials, since they cause stress accumulation at the matrix-carbide interface, which enhances plastic deformation. From this point of view, carbides have a negative influence on the ability of the material to develop a mechanically stable surface during wear. Moreover, given the intrinsic brittleness, carbides tend to fail during plastic deformation of the surface layers, when it involves the whole volume. Crack can propagate in the matrix, giving rise to the formation of plate-like metallic fragments, which increase the wear rate. Even from this point of view, carbides have a negative influence on the delamination resistance of the materials. The first step of the wear curves of the materials with the highest carbide content seems to indicate that a high carbide content could be compensated by a high matrix microhardness. In fact the first step wear rate is of the same level of that of the materials with a lower carbide content with a similar, or lower, matrix microhardness. However, in this material, an additional mechanism becomes active during the wear process, causing the observed wear transition. It is caused by the strain accumulation /9/ which creates the conditions for an accelerated delamination process, which greatly increases the wear rate. Therefore, even when the matrix has a very high microhardness, the presence of a great amount of primary carbides tends to worsen the delamination resistance of spin cast HSS.

CONCLUSIONS

The delamination behaviour of spincast High Speed Steels used for the production of finishing hot rolls was studied in order to correlate the wear resistance to the microstructural characteristics and the mechanical properties of the materials. In the adopted experimental conditions, wear occurs by a combination of delamination and oxidative wear, and the wear rate increases when delamination predominates on oxidative wear. The controlling parameter is the matrix microhardness, since it increases the load bearing capacity of the material, reducing the plastic strain of the surface and subsurface layers where delamination crack nucleates. Primary carbides tend to favour the occurrence of plastic deformation, because of the stress accumulation at the matrix-carbide interface. Moreover, carbides fail in the strained layers, because of their brittleness, and this assists the nucleation of the delamination crack. From this point of view, primary carbides have a negative effect on the delamination resistance.

The proper combination of matrix microhardness and content of primary carbides results in an optimized delamination resistance. In this condition, HSS have a better wear resistance than HCrI and ICI, and this supports their continuously growing application in the production of finishing hot rolls.

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ABSTRACT

COMPORTAMENTO A DELAMINAZIONE DI ACCIAI HSS PER CILINDRI DI LAMINAZIONE A CALDO

Il consumo dei cilindri di laminazione dipende dal comportamento tribologico dei materiali con i quali sono fabbricati nelle complesse condizioni di utilizzo dei cilindri stessi. E' essenziale che il cilindro mantenga un profilo di usura il più regolare possibile per minimizzare gli sforzi di contatto e ridurre la quantità di materiale asportato nelle rettifiche periodiche e non danneggiare la qualità superficiale del materiale laminato.
L'usura delaminativa è un fenomeno che tende a peggiorare il profilo di usura, ostacolando la regolare formazione di uno stabile strato di ossidazione superficiale, che minimizzi l'usura e mantenga basso il coefficiente di attrito. Il presente lavoro studia il comportamento a delaminazione di alcuni acciai HSS prodotti per colata centrifuga, con l'obiettivo di correlare la resistenza del materiale alla microstruttura e in particolare alla percentuale di carburi primari.
La tabella I riporta gli intervalli di variazione del contenuto degli elementi in lega dei materiali studiati. Gli acciai sono stati temprati da 1020-1080 °C e rinvenuti nell'intervallo fra 480 e 560 °C alla temperatura del massimo indurimento secondario. Le prove di usura sono state condotte con un tribometro Amsler, in condizioni di rotolamento-strisciamento (10%) a secco e con una forza di 500N applicata ai due dischi contrapposti.
Le figure 1, 2 e 3 riportano la microstruttura di tre HSS con diversi contenuti di carbonio (1%, 1.7% e 2.3% rispettivamente). La microstruttura è costituita da una matrice dendritica o cellulare di martensite rinvenuta, con una quantità di carburi primari (determinata con l'Analizzatore d'immagine al microscopio ottico) che aumenta 4 a 25% all'aumentare della percentuale di carbonio. La microdurezza della matrice varia da 550 a 900 HV0.1.
La figura 4 mostra la curva di usura di un acciaio con 1% C (bassa percentuale di carburi). Essa presenta una variazione costante della perdita in peso all'aumentare della distanza di rotolamento e la pendenza della retta interpolante i dati sperimentali è la velocità di usura. La figura 5 mostra la curva di usura dell'acciaio con 2.4% C; essa presenta due stadi, il secondo con una velocità di usura maggiore del primo. Tutti gli acciai con contenuto di carbonio maggiore di 2%, e una percentuale di carburi maggiore del 15%, presentano questa transizione.
La figura 6 riporta la velocità di usura in funzione della percentuale di carburi, distinguendo tre classi di materiali sulla base della microdurezza di matrice. Per i materiali che mostrano la transizione sopra descritta, nella figura è riportata, come punto sperimentale, solo la velocità di usura del primo stadio; quella del secondo è indicata con il valore specifico.
Dalla figura si osserva che all'aumento della percentuale di carburi aumenta la velocità di usura, per tutte le classi di microdurezza sperimentate. Inoltre, alla stessa percentuale di carburi, all'aumento della microdurezza corrisponde un calo della velocità di usura.
La figura 7 mostra la sezione metallografica del campione usurato dell'acciaio con un contenuto di carbonio intermedio fra quelli studiati. Essa mostra la deformazione plastica subsuperficiale, che coinvolge i carburi primari. Questi, nello strato più esterno, sono frantumati e dispersi nella matrice deformata, e pertanto non sono visibili nella micrografia. In questo materiale si ha un'usura delaminativa-ossidativa, come confermato dall'analisi XRD dei frammenti (figura 8).
La figura 9 mostra la sezione metallografica del campione usurato dell'acciaio con il maggiore contenuto di carbonio. La frammentazione dei carburi è molto evidente e il profilo di usura è molto irregolare, come mostrato dalla figura 10. Il meccanismo è simile al caso precedente, ma con un maggiore contributo dell'usura delaminativa, responsabile del distacco di particelle metalliche che, a sua volta, causa l'irregolarità del profilo mostrata. Gli acciai con il minore contenuto di carbonio subiscono ancora usura delaminativa-ossidativa, ma con il contributo prevalente di quest'ultima. La figura 11 mostra l'analisi XRD dei frammenti, con il picco del ferro di modesta intensità.
In presenza di tale meccanismo di usura, la capacità del materiale di resistere alla deformazione plastica (responsabile della delaminazione) e quindi di supportare lo sviluppo dell'ossidazione superficiale (per il meccanismo ossidativo), determina bassi valori della velocità di usura. Più alta è la microdurezza di matrice, maggiore è la resistenza alla deformazione plastica, e questo ne giustifica l'effetto positivo sulla resistenza all'usura. Anche i carburi hanno un effetto importante, in quanto causano accentuazione locale delle sollecitazioni, all'interfaccia con la matrice. Questo facilita la deformazione plastica. Inoltre, data la loro fragilità, i carburi possono rompersi innescando la cricca di delaminazione.
Un'elevata percentuale di carburi potrebbe essere compensata da un'elevata microdurezza di matrice. Tuttavia, la velocità di usura degli acciai con elevata percentuale di carburi rimane bassa solamente durante il primo tratto della curva, dopo di che si ha un aumento notevole. Questo è attribuito ad un fenomeno di accumulo della deformazione (durante il primo stadio), che crea le condizioni per un pesante danneggiamento e la conseguente transizione al secondo stadio. Pertanto, anche con una matrice molto dura, un'elevata percentuale di carburi ha un effetto negativo sulla resistenza alla delaminazione. Il migliore comportamento a delaminazione si ha pertanto con una giusta combinazione di microdurezza di matrice e di percentuale di carburi primari. Gli HSS hanno comunque una resistenza all'usura maggiore sia delle ghise ad alto cromo che delle ghise a tempra indefinita, e pertanto il loro impiego nella produzione di cilindri finitori a caldo trova sempre maggiore diffu-