Performances Evaluation of High Speed Steels for Hot Rolling by Wear and Thermal Fatigue Tests

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INTRODUCTION

In the last years a great number of hot strip mills performed trials with High Speed Steel (HSS) rolls for both rougher and finishing stands. Rolls used in the former stands are subjected to higher temperatures, slower rolling speeds and longer contact lengths between roll and stock due to the high reductions. Therefore, they require high thermal fatigue resistance. Rolls used in the latest stands usually withstand higher tribological and mechanical solicitations given by abrasion and impacts with head and tail end cobbles, respectively. The industrial experience, however, indicate that it is often very difficult to evaluate their performances by trials on hot strip mills, due to the high intrinsic variability of the industrial process. In fact, in each plant different cooling conditions interest the rolls surfaces and the rolling forces and temperatures are subjected to a large variations; moreover, the occurrence of accidents involving the temporary stop of the strip between the rolls causes consistent localized rises of the temperature in the stand and a subsequent strong thermal shock at the restarting of the mills operations. These drastic events imply a heavy damage on the surface of the roll, with a consequent severe shortening of its life. Unfortunately, these kinds of accidents are not always documented, making not reliable the interpretation of the data from the mills. Therefore, strong discrepancies of performances for the same roll grade on different hot strip mills have often been found. This has lead to the development of laboratory tests to reproduce (as well as possible) the same thermo-mechanical conditions to which work roll materials are exposed during service. Two tests have been developed with the aim of evaluating the thermal fatigue and hot wear resistance of work roll materials and in particular to study the correlation between the microstructure and the damage produced by the cited phenomena. After heat treatment, spin cast HSS rolls display a microstructure given by a tempered martensitic (martensitic/bainitic) matrix surrounded by an eutectic network of primary carbides. Thermal fatigue and hot wear resistance are strictly related to the volume percentage, type, morphology and distribution of carbides, but even to the matrix microhardness, particularly at high temperature [1]. By properly balancing the amount of carbon and alloying, it is possible to obtain a great variety of microstructures.

The results presented in this work highlight the main criteria for the selection of the best grades for a given application, depending on the relative influence of wear and thermal fatigue. It is widely accepted that it does not exist a “universal roll grade” suited for each hot strip mill, but the best choice should be made on the basis of the specific working conditions. This also implies a strict cooperation between roll producers and hot strip mills people.

MATERIALS

The requirement of rolls characterized by elevated efficiency as well as able to warrant high quality of the rolled products involves the development of high speed steels cylinders. The first HSSs rolls have began to operate from 1990 in the early finishing stands of hot strip mills. These products are characterized by a ledeburitic structure obtained by centrifugal casting. As generally these rolls don’t undergo to forging after casting, a great importance assures their chemical composition, because it deeply affects the final structure [2]. The chemical composition is similar to that of cutting tool steels, with a large amount of alloying elements to improve the hardenability and the mechanical properties at elevated temperatures; usually, Vanadium or other strong carbides formers are present too. As these elements cause a large carbon impoverishment of the matrix [3], the nominal carbon is quite elevated, generally over 1.5 wt%. According to the chemical analysis, different carbides are formed with specific composition, structure and morphology. In general, the microstructure of these steels consists in a network of eutectic carbides surrounding a metal matrix enriched of secondary carbides: the primary carbides can precipitate either in a network formed along cells boundaries (eutectic carbide type) or within the dendrite arms during solidification (proeutectic carbide type); MC carbides, promoted by the presence of Vanadium or Niobium, mainly have a round shaped form and precipitate within cells, while M_2C (generally containing Molybdenum or Tungsten) or M_7C_3 (containing Chromium) can be identified in a net surrounding the metal matrix. Since an interconnected carbides network can be regarded as a weak element because of its brittleness, it's desirable to have an amount of dissociated carbides as larger as possible; however, the centrifugal cast restricts the typology of alloys that can be poured, as the lighter phases, like Vanadium carbides, are prone to segregate faraway from the surface of the shell. In this study three high speed steels are compared to evaluate their thermal fatigue resistance and wear behaviour. The materials are all characterized by similar carbon contents and have been undergone to the same heat treatment. In Table 1 is reported the chemical composition of the steels. Variations on the alloying elements involve different amounts and typologies of carbides; at the same time, differences on the matrix grade affect the hardenability and mechanical properties.

	% C	% Cr	% W_{eq}*	% MC Formers
HSS 1	*1.7*	*5.0*	*8.0*	*4.5*
HSS 2	*1.6*	*3.0*	*4.0*	*5.5*
HSS 3	*1.8*	*3.5*	*10.0*	*6*

Table 1: Chemical composition (wt %) of the three analysed steels;
* Weq= W+ 2Mo.

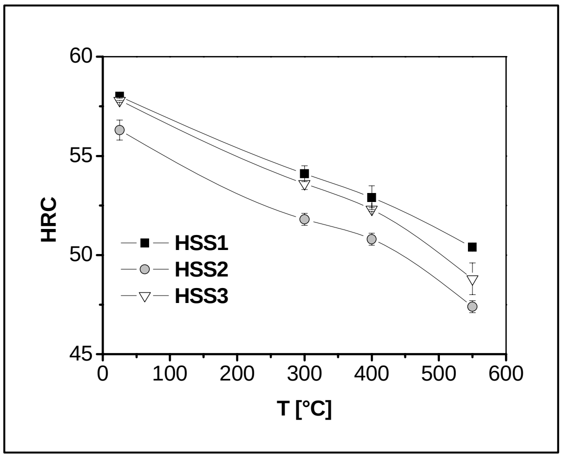


Figure 1: Hardness drop with temperature

In table 2 the amount and typology of carbides is reported for the three steel; the percentage of interconnected carbides, defined as the quantity of carbides linked in a network along the cells boundaries, is also reported. HSS1 contains a total carbide volume fraction around 10%, with a predominance of M_7C_3 on MC. In figure 2 is shown the microstructure of the analysed steels. HSS 2 is almost free from the ledeburitic network: there is a marked precipitation of proeutectic MC carbides well distributed within the dendrite arms (8% in volume). HSS3 represents the actual best compromise between the presence of proeutectic MC carbides and the level of alloying elements in the matrix. Such condition allows reaching a microstructure with a good distribution of primary carbides (approximately 10%) limiting the formation of a network of eutectic carbides; at the same time, a highly alloyed matrix warrants good hot resistance. In figure 1 the hardness versus temperature is reported. HSS1, in agreement with its elevated content of alloying elements in the matrix, has the highest hardness either at room or at elevated temperature. HSS2 shows the worst hardness among the investigated steels; this is due to the matrix relatively poor in carbon and alloying elements. An intermediate behaviour is evidenced by HSS3, with the same hardness of HSS1 at room temperatures while it shows a few higher softening at elevated temperatures.

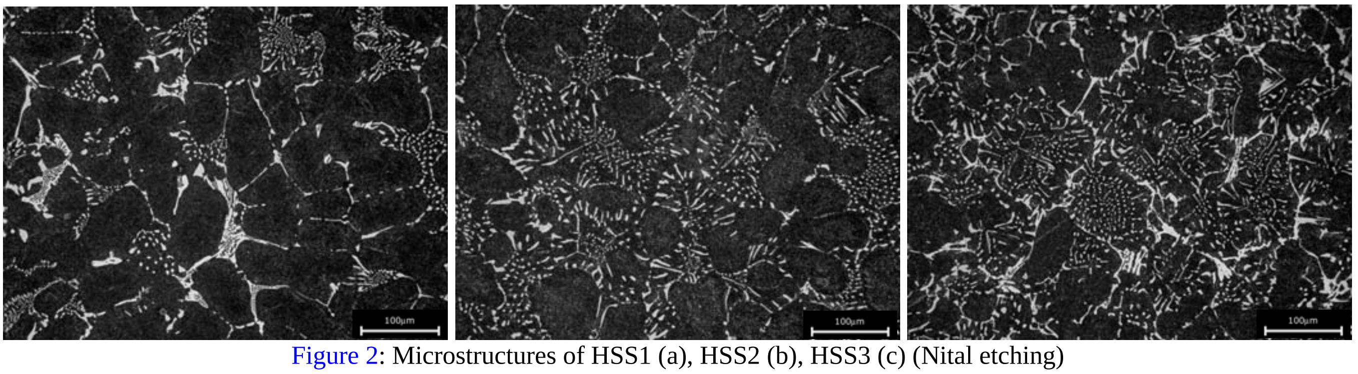


Figure 2: Microstructures of HSS1 (a), HSS2 (b), HSS3 (c) (Nital etching)

	Carbides	M_7C_3 [~1500 $HV_{0.1}$]	MC [~2500 $HV_{0.1}$]	M_2C [~1900 $HV_{0.1}$]	Interconnected Carbide	Matrix μHV [$HV_{0.1}$]	Hardness [HRC]
HSS 1	10 %	6 %	4 %	-	4.1 %	720	58
HSS 2	8 %	< 0.5 %	8 %	-	1.6 %	650	56
HSS 3	10 %	-	6 %	4 %	2.9 %	700	58

Table 2: Carbide volume percentage, interconnection level and matrix microhardness for the three examined steels.

EXPERIMENTAL PROCEDURES

In the present work, laboratory tests were carried out to characterize the three steels studied and to measure those properties which determine the roll behaviour in the mill. Microhardness tests were led on the martensic matrix by Vickers indentation with one N load. A material characterization was carried out grinding a sample for each steel; then, etching with Cogne reagent, the carbide volume percentage was determined by means of an image analyzer mounting Leica Q-Win software. The presence of M_7C_3 and M_2C carbides was evidenced by means of selective etchings, in particular Murakami's and Groebeck's reagent. The resistance to firecracking was evaluated by means of a self designed rig (Fig. 3a). A disc with a diameter of 40 mm and a width of 10 mm was forced to rotate at four rpm. During the rotation, the specimen was cyclically heated by means of an inductor up to 670°C and subsequently cooled down nearly at 80° by a water jet. The rapid and cyclic variation of temperature as well as the constraint due to the high thermal gradients between the surface and the core of the sample causes the nucleation and growth of cracks [4]. The cracks length and number have been measured every 60 cycles for total test duration of 180 cycles. The test appears very hard for this kind of steels as the specimens in few seconds undergo to an elevated temperature variation and consequently to a severe damage. This allows evaluating the firecracking resistance of the steels in very short times. It is important to underline that the appearance of the samples after the test is quite different from that showed from operating rolls. This is due to the diverse solicitations to which the specimens and the cylinders undergo; in particular, in the former case a plain stress is typically generated, while in the latter, due to the larger dimension of the rolls, a plain strain is realized. This is responsible of the formation of just radial fractures on the specimens instead of the net of cracks observed on the cylinders.

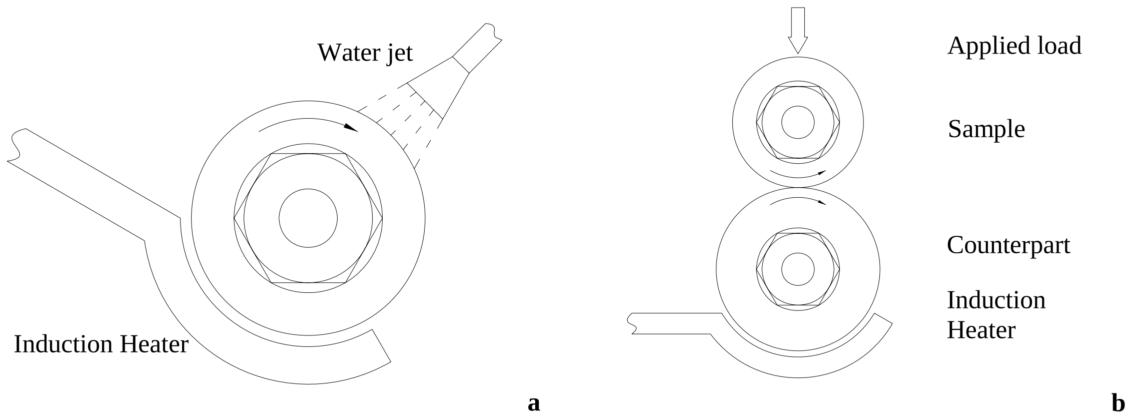


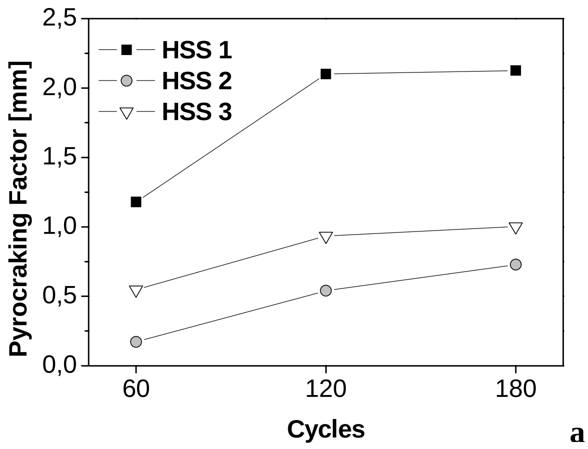
Figure 3: (a) Thermal Fatigue equipment (b) Hot wear test configuration.

For each steel wear tests were carried out in dry conditions at elevated temperatures by means of an Amsler apparatus; two discs, a sample with a diameter of 40 mm and 10 mm width and a counterpart in C40 plain carbon steel (diameter of 50 mm), were coupled and forced to run at different speeds. Both the disc and the counterpart were allowed to rotate at different speeds realizing a rolling sliding contact (Fig. 3b); in particular, the rotating speed of the counterpart was set at 200 rpm, while that of the sample was 180 rpm, thus producing a sliding speed of 0.15 m/s [5]. Considering that the rolling speed in finishing stands are nearly comprised between 1.5 and 14 m/s [6] and that the forward slip varies from 3 to 10%, the sliding speed is generally a few lower than that observed at the exit of the finishing stands. The counterpart was heated up to 700°C by means of an inductor. Before starting the test, the counterpart was led for nearly 5 minutes at the test temperature; in this interval an oxide layer grows on its surface, thus better reproducing the mill conditions, where the roll operates against an oxidized slab; the temperature of the counterpart was constantly monitored by means of an optical pyrometer. Then the sample was coupled against the counterpart and a 300 N load was applied. Regarding to the reduced elastic modulus of the heated C40 counterpart, the sample undergoes to a Hertzian stress of nearly 300 MPa, comparable to the mean pressure in the contact length of an early finishing roll. The total test length is an hour. The temperature of the counterpart is a few lower than the expected temperature of the slab in finishing stand, comprised between 1000 and 900°C [6]. The choice of these wear conditions was mostly determined by the tester limits. By means of a thermocouple touching the surface of the specimen it's possible

to evidence that its temperature during the test is always higher than 250°C; this well agrees with the temperatures of the rolls in working conditions monitored before the water jets [7]. The main difference among the wear test and the effective rolling conditions is represented by the lack of a cooling apparatus in the laboratory simulation; thus, the sample doesn't undergo to the thermal shock due to the rapid cooling after reaching its highest temperature in the contact zone. Moreover, the oxidation kinetic in a dry test is quite different from that observed in a wet environment. By means of this experimental rig it's possible to evidence the main wear mechanisms damaging the rolls during their service [8]. These represent the theoretical basis of any intervention on the microstructural features of the steels in order to improve their wear performances.
The results obtained on the thermal fatigue simulations and on the wear test have been validated by the collection of data in the mills. In detail, the examination of the surface damage on operating rolls allows to compare the damage mechanism acting with the one observed on laboratory. In particular is quite critical the estimation of the wear behaviour in operating rolls, as the pick up of data about the performances of a cylinder is deeply affected by the turning modality adopted in the mill. More specifically, although it's possible to discriminate a theoretical efficiency from an effective one, it is an exception to the rule that in a mill the rolls are dismantled just when they show a critical surface damage. On the contrary, most of the turning operations are carried out to couple the diameter of a scarcely damaged cylinder to that of a more injured one; otherwise, some rolls presenting heat checking are brought to turn before the cracks reach potentially dangerous dimensions. Thus, very often the experience and the sensitiveness of the mill operators assure a greater importance than the steel quality to determine the roll performance. In this situation, the interest on visiting a plant is mainly focused on the characterization of the surface features of the rolls rather than on the pure collection of data on their efficiency.

THERMAL FATIGUE

The operating conditions of the rolls provide their cyclical contact with the high temperature strip and then the subsequent cooling by means of water jets. This continuous variation of temperatures, from about 600°C to 100°C in an early finishing stand [7], interests just a limited portion of material under the working surface and causes its thermal expansion / contraction. The material under the surface is just weekly affected by these thermal oscillations and it is not able to accommodate the external dilatations, causing the rise of stresses that can exceed the endurance limit of the material; thus, some cracks can nucleate and propagate towards the core of the roll. This phenomenon is very dangerous for the risk that these cracks reach a critical dimension bringing the cylinder to breakage. Laboratory tests were carried out on specimens of the three examined steels inducing very high thermal stresses in a limited time. The damage shown by the samples is quite pronounced and doesn't have correspondence to the heat checking typically observed in the plants, as the rolls usually undergo to turning operations before the reaching of a so critical situation. Though, the solicitations to which the specimens are subjected are similar to those of the operating rolls. In fact, the induction heating allows keeping relatively cold the specimen heating just a limited part of its surface, as occurs to the operating rolls during the contact with the hot strip. Thus, by means the observation of the damage on the sample surface it is possible to evaluate the heat checking resistance of the steels. In Figure 4a is represented the evolution of the pyrocracking factor during the test. This parameter is defined as the product of the crack density normalized on the sample dimensions multiplied for the maximum and for the medium crack length [9]. The pyrocracking factor is considered representative of the heat checking resistance, as it contemplate a measure of the tendency of the material to nucleate cracks (i.e. the crack density) and to propagate towards the core (the maximum crack length and the medium crack length). Different behaviours for the three steels can be evidenced: HSS1 shows the worst behaviour, mainly due to the growth of a low number of very long cracks; in this sample, already after the first 60 cycles the cracks are longer than 3.5 mm (measured



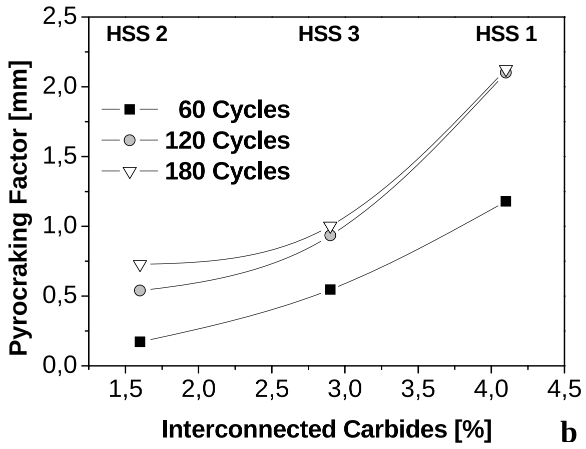


Figure 4: Pyrocracking factor versus number of cycles (a) and versus the volume percentage of interconnected carbides

perpendicularly to the surface). In the subsequent 60 cycles, there is not nucleation of other cracks, but the existing ones propagate to a depth of nearly 4.5 mm. Then, until the test end, this situation remains almost the same, with just a poor growth of the cracks. This is because the cracks have reached a dimension sufficient to accommodate the thermal strain preventing the stresses on the surface from rising. HSS2, instead, shows the best thermal fatigue resistance. In this material a large number of short cracks can be detected. The reason of the good behaviour of this steel can be adduced to the fact that, even if the amount of nucleated cracks is quite elevated, these don't increase towards the core, mainly because of the lack of a preferential path of propagation. An intermediate behaviour is evidenced by HSS3: in this material a quite large number of cracks nucleate and propagate towards the core. The mean dimensions of these cracks are greater than those observed in HSS2. Both HSS2 and HSS3, at the end of the tests, keep the dimensions of the longer cracks around 3.5 mm, quite shorter than those in HSS1.

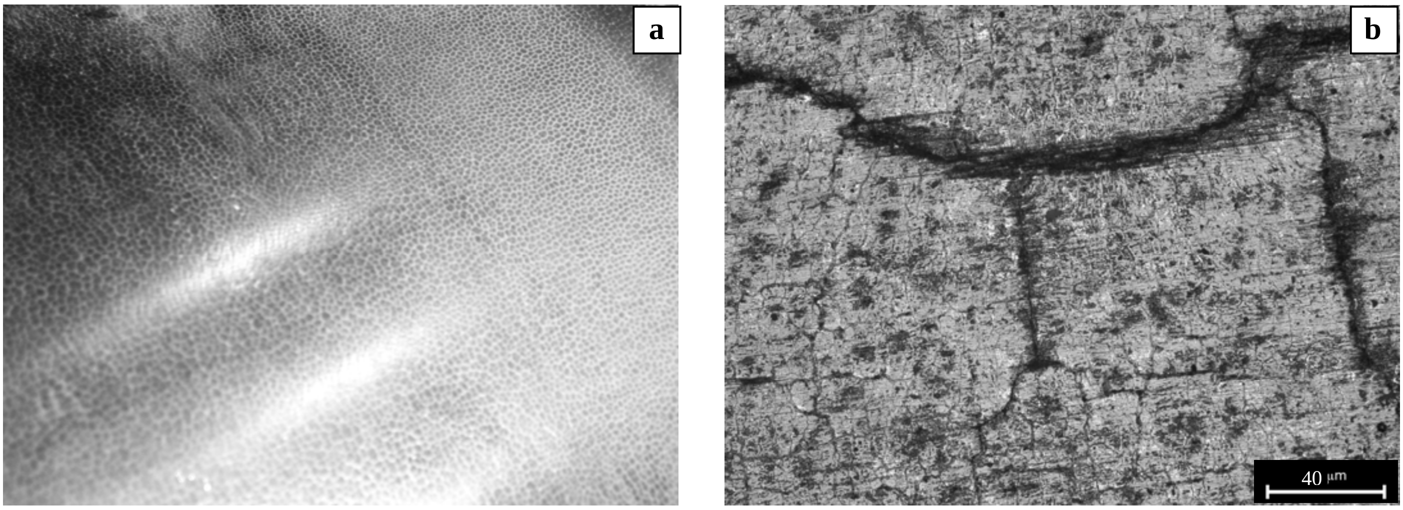


Figure 5: Network of thermal fatigue cracks on the roll surface

In previously works [10] it has been underlined the role of an interconnected carbides network in the propagation of thermal fatigue cracks. In figure 4b the pyrocracking factor is represented versus the volume percentage of interconnected carbides. It's possible to underline a strict relation between the damage of the specimens and the presence of a network of carbides. In fact, the connected carbides, due to their high brittleness, represent a preferential path to the cracks, as can be seen in figure 6a. Thus, HSS1, characterized by a consistent amount of eutectic M_7C_3 along dendrite boundaries, shows a low thermal fatigue resistance. HSS 2 and HSS3, instead, have a large amount of dissociated carbides and don't evidence a so well defined eutectic network. Thus, in these materials the cracks are forced to propagate also across the tougher matrix, dissipating a greater amount of energy. This justifies the maximum and the mean cracks lengths lower than in HSS1.

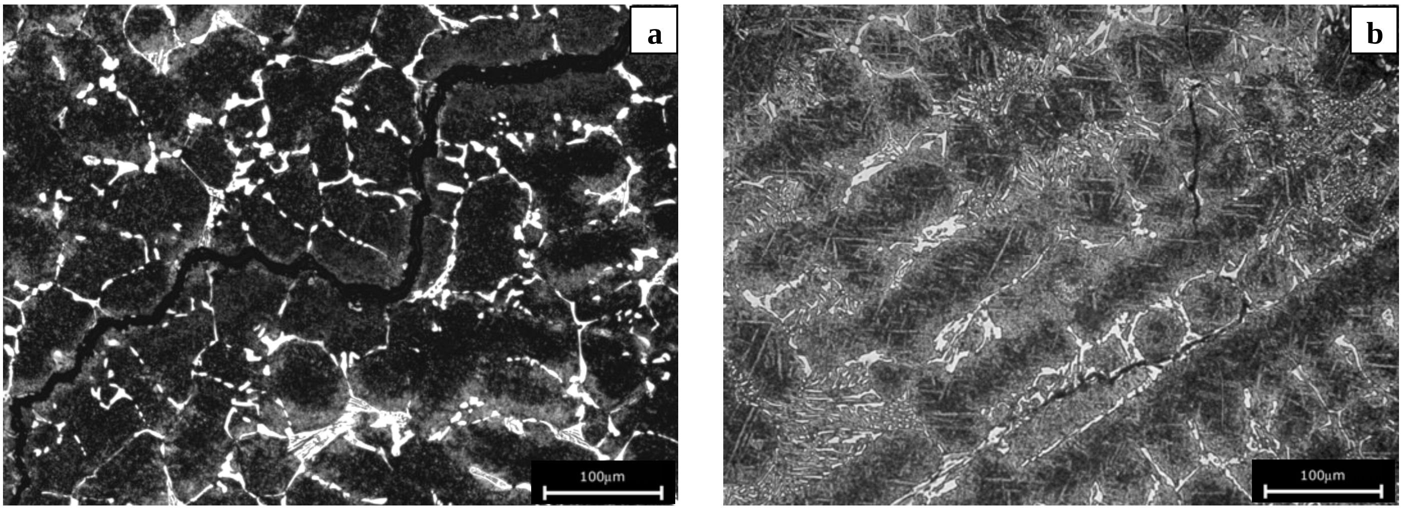


Figure 6: Thermal fatigue damage on the sample (a) and on an operating roll (b)

In figure 5a and 5b is represented a typical firecracking damage in a HSS 1 roll operating as rougher in a mini-mill. After about 6000 ton. milled, a network of cracks has developed on surface. The surface roughness was comprised between 2.0 and 2.5 μm. It's possible to recognize two kinds of cracks: there is a web of large cracks that, after having nucleated mainly at the interface carbide-matrix, rapidly has grown towards the core. An examination carried out on the section shows that these cracks have reached nearly 0.7 mm in length. Then, there is a great number of shorter cracks, having a maximum length of 200 μm, that grows along the dendrite boundaries. The former cracks are caused by a very strong thermal shock; in a first step, these cracks propagate perpendicularly to the surface and indiscriminately cross martensitic dendrites and carbides for nearly 500 μm; then, at a depth where the thermal stresses lowers, because of the accommodation to the deformation due to the presence of the breaks and the distance from the surface, the cracks began to follow the most brittle eutectic carbides network. Figure 6a and 6b are two images showing respectively the thermal fatigue damage on a sample and on the operating roll. Regarding to these pictures, the thermal fatigue test allows reproducing just the damaging mechanism concerning the secondary growth of the longer cracks. However, the results obtained in laboratory can be useful to evaluate the firecracking resistance of the rolls. It needs underline the importance of the grinding procedures, as it deeply affects the rolls performances: in the present case, after the appearance of the firecracking phenomenon, the roll was grinded for 0.3 mm on the diameter in order to remove the close network of shortest cracks and operated without incurring breakage. However, it's very common to grind the rolls until Eddy currents don't indicate the presence of any surface defect. In this eventuality, the performances of the cylinder would be deeply reduced.

WEAR

It is well known that is not possible to reproduce in laboratory the wear conditions to which rolls are subjected in the industrial practice; the main difference is represented by the mechanical solicitations: in rolling mill the contact between the roll and the lamination products is realized trough a conformal area: in this zone, the tangential forces exercised by the cylinder on the strip at the entry of the contact length have the same versus of the rolling direction, then, before reaching the exit become negative. This typology of solicitation is quite difficult to reproduce in laboratory, where the rolling sliding test is characterized just by a Hertzian contact area and by applied tangential forces having always the same versus; however, these tests allow to evidence the mechanisms acting during the wear at elevated temperature. This is indispensable in order to determine the parameters which mainly affect the surface damage of the rolls.

	Wear Rate [mm^3 / m]	**Roughness [μm]**
HSS 1	$2.99 \cdot 10^{-3}$	0.25
HSS 2	$3.13 \cdot 10^{-3}$	0.27
HSS 3	$3.24 \cdot 10^{-3}$	0.32

Table 3: Wear rate and roughness for the three steels after a test 1 hour long

In table 3 is reported the wear rate: all the steels have a quite good behaviour with very low wear rates; HSS1 behaves better than the other materials, while HSS3, despite its elevated hardness, has the highest mass loss during the test. However, the differences of mass loss for all the three steels are almost negligible. Regarding to figure 1, it's not possible to relate the wear rate to the hardness of the steels; besides, it's not evident any correlation between the mass loss and the carbide volume percentage of the samples.

The analysis of the specimens after the test evidences wear surfaces with marked oxidative phenomena (Figure 7). HSS1 and HSS2 show a quite uniform oxide layer covering the matrix, interrupted just by the carbide network, less prone to oxidize. The surface of HSS3, instead, has a less homogeneous oxide scale. The wear mechanism has to be ascribed to abrasion and triboxidation. The contact with the hot counterpart promotes the growth of an oxide scale on the specimen's surface. The wear rate is ruled by the removal of this scale due to the contact stresses. At the same time, abrasive phenomena act on the wear surfaces; in particular, the presence in the

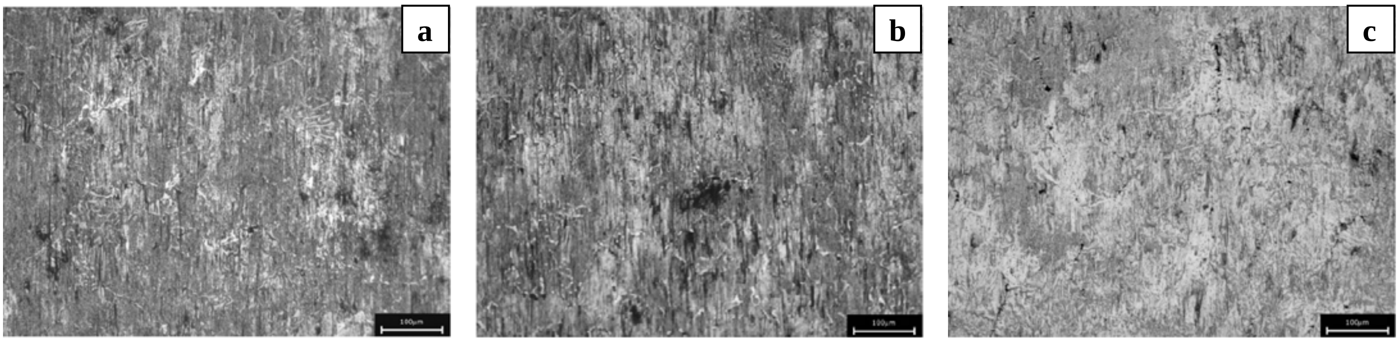


Figure 7: Wear rate surfaces of steel HSS1 (a), HSS2 (b), HSS3 (c) after a test 1 hour long

contact zone of oxide fragments and of hard carbides removed from the metal matrix gives rise to a three body abrasion, giving rise to a heavier wear. The presence of a compact oxide layer, generally, reduces the friction forces between the sample and the counterpart and, thanks to its elevated hardness, protects the wear surface [11]. Thus, the uniform oxidation observed for HSS1 accounts for its low wear rate. HSS2, theoretically, is more prone to oxidize than the other materials because of it contents a great amount of MC carbides rich in Vanadium; these carbides have a scarcely resistance to oxidation and usually form an oxide scale before than the matrix does [12]. Thus, it should expected that this steel has a better wear resistance as it can quickly constitute a protective scale on its surface; though, HSS2 has a higher mass loss than HSS1 and this can be imputed to its low hot hardness that does not allow bearing the oxide. The highest wear rate shown by HSS3, instead, has to be ascribed to its poor capability to form a compact oxide layer. Though the analysis carried out at the end of the tests indicates just little difference of roughness for all the samples, it is important underline that the lower values correspond to the specimens that have shown the higher wear resistance. More specifically, HSS1 has the smoother surface, mainly because of the small difference among the matrix and the harder carbides, but also because of the presence of an extensive and compact oxide scale; HSS3, instead, shows the rougher surface as its matrix is not able to oxidize enough to reduce surface irregularities [13].

	F2 [ton/mm]	F3 [ton/mm]	F4 [ton/mm]
HSS 1	8050	8150	7650
HSS 2	10050	9200	8150

Table 4: Performances of HSS1 and HSS2

Tonnes	Roughness [μm]	Grinding [mm]	Performances [ton/mm]
5460	1.0 – 1.5	0.50	10920
7091	1.0 –1.5	0.45	15758

Table 5: Performances variation of HSS3 in a F1 stand as a function of grinding practice

In table 4 the performances of some HSS operating in finishing stands producing nearly 2.700.000 ton/y are reported. A drop in the rolls efficiency can be evidenced for the cylinders operating in the last finishing stand, as a consequence of the higher stresses to which they undergo. Despite the results obtained in laboratory, HSS2 has better performances than HSS1. It needs underline the great variations to which the productivity of the rolls is subjected as a function of the grinding practice. In table 5 is shown the performance of HSS3 in a F1 stand. There is a difference in efficiency of more than 40% between the two cylinders considered. This can be ascribed to the fact that is a well established practice to grind periodically the rolls even if there is not a real necessity. In this case, the rolls surfaces don't show a heavily damaged surface and during the turning operations are subjected just to a minimal (sometimes superfluous) removal on the working surface; thus, on the basis of the length of the previous campaign, this procedure gives rise to very different rolls performances. It should be marked that the measured roughness before the grinding operations is quite high if compared to those observed on the laboratory specimens, indicating rather harder wear conditions in the industrial practice. A second example of the importance of grinding procedures is shown in figure 8; here the surface of two cylinders operating in a finishing stand (F2 and F3) of a mill producing austenitic stainless steel are reported. HSS1 shows a rough surface (Ra ~ 3 μm), while HSS3 has a smoother one with a roughness between 0.25 and 1 μm (comparable to those observed on laboratory tests). Though the enormous difference in the surface finishing, both the rolls had been subjected indiscriminately to grinding. This implies that, even if in this case HSS3 has shown a better wear behaviour of HSS1, its efficiency results the lowest, nearly 7880 ton/mm against 8380 ton/mm of HSS1.

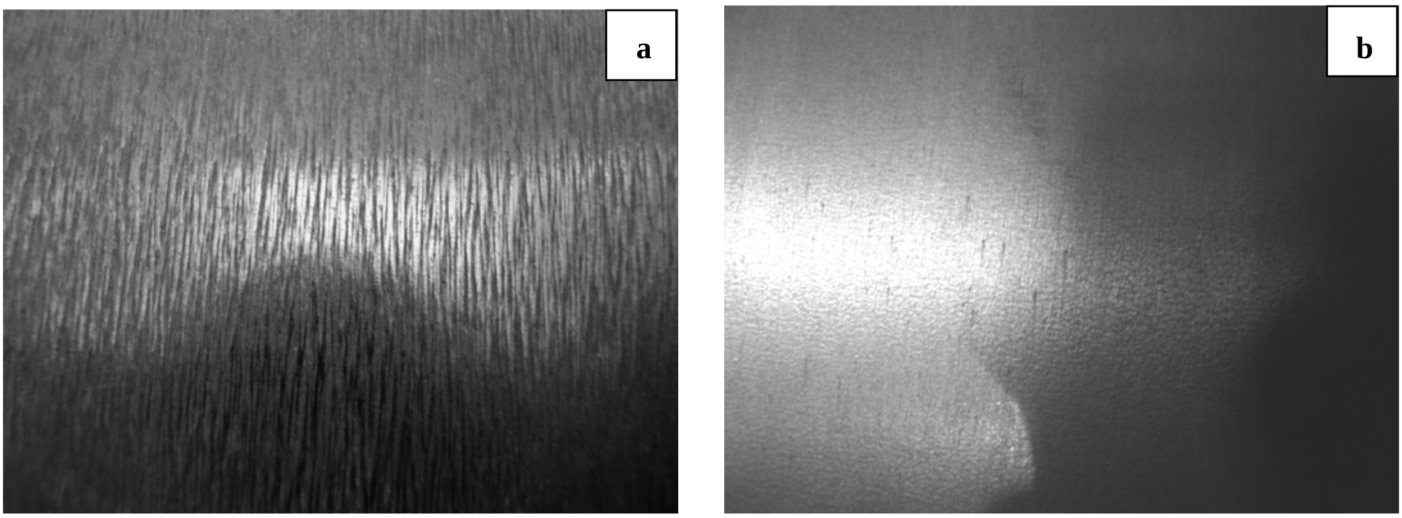


Figure 8: Working surface of HSS1 (a) and HSS3 (b) operating in a finishing stand of a mill producing stainless steel

DISCUSSION

Laboratory test can not reproduce the thermal fatigue or the tribological conditions to which rolls undergo during mill operations. In fact, the behaviour of a roll operating in a mill is affected to a sort of aberration due to the procedures of the diagnostic of the damage that make its performances not easily calculable. Considering the thermal fatigue, in the rougher stands it rules as the main factor influencing the efficiency of the cylinders. In this case, to arrange a roll withstanding to this kind of solicitations plays a strategic role in order to reach an elevated efficiency. Although, actually is not easy to estimate the thermal fatigue resistance, as it results a function of the maintenance practice of the individual mill. For example, some rolls are dismantled to be grinded at the appearance of the first surface damages, while others are kept on their positions until the Eddy currents indicate the presence of cracks long enough to constitute a real risk for a catastrophic breakage. In the former case, a materials as HSS2, that shows the nucleation of a large number of pyrocracks unable to propagate in depth, will be not appreciate for its thermal fatigue resistance, while in the last case the opposite conclusions will be drawn. In this situation, laboratory tests represent a means to evaluate the performances of the rolls. The experimental trials, on contrary, have shown well distinct behaviour for the analysed materials, offering an objective valuation on which is the better steel for operating in environment characterized by elevated thermal solicitations; moreover, it has been underlined a mechanism responsible of the damage and ruled by the brittle carbide network. Besides, the knowledge of which are the physical mechanisms on whose basis materials reply to the imposed stresses is of fundamental importance to follow the development of the steels for roll applications. In a similar manner, the evaluation of the wear in the mills is affected by the same problems: the grinding procedures, in particular when the roll material scarcely wears (hence, really in the case of High Speed Steel rolls), to extend the campaign length means deeply to affect the efficiency of the cylinder; hence, it’s very hard to draw any kind of sensate conclusion on the roll behaviour in the industrial process. In this case, the experimental tests don’t evidence remarkable differences among the investigated steels. However, it has been evidenced that the acting wear mechanism is mainly ruled by the material predisposition to form a protective oxide layer on the surface and then by the capability of the matrix and of the carbides to bear it. Therefore, from the knowledge of the oxidational behaviour and of the mechanical response at elevated temperatures of a material is possible to determine a guideline for the development of hot wear resistant roll steels.
It needs to underline, however, the importance to go on collecting data from the plants in order to receive some confirmation of the rolls behaviour deduced by laboratory tests. However, the cooperation between the mills people and roll producers has to assure a leading role in order to realize a validation of the real performances of the rolls on the basis of standardized operating procedures.

CONCLUSIONS

The thermal fatigue resistance and the wear behaviour of three different high speed steels have been investigated by means of laboratory equipments. Thermal fatigue has been evaluated on the basis of a pyrocracking parameter, representative of how thermal cracks are prone to nucleate and to propagate. The test have evidenced consistent differences between the investigated materials; more specifically, it has been highlighted an inverse relation between the heat checking resistance and the volume percentage of interconnected carbides: materials with an extensive eutectic network show the formation of a restricted number of very long cracks that compromises their thermal fatigue behaviour. Contrarily, materials with a low carbides interconnection are prone to nucleate a large amount of cracks unable to propagate. This model of fracture propagation is partially validated from the observation of the damages in operating rolls, while it is not possible to quantify the thermal fatigue resistance by the data collected in industrial mills.
The wear experiments have shown dissimilar wear rates and surface finishing at the end of the tests. All the materials evidence very limited mass losses; the wear mechanism is mainly ruled by two factors: oxidation and hot hardness. Oxidation allows the formation of a scale on the surface that acts as a protective and reduces the friction forces; hot hardness, affected by the matrix strength at elevated temperatures and by the amount and kind of present carbides, is responsible of the steel capability to bear the oxide layer. In this case, the data on the performances of operating rolls are subjected to an excessive variability depending on the mill procedures and thus, actually, it’s not possible to confirm the results of laboratory tests by industrial evidences.

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