

THE ABRASIVE-CORROSIVE WEAR OF STAINLESS STEELS*

C. ALLEN and A. BALL

Department of Metallurgy and Materials Science, University of Cape Town, Cape Town (South Africa)

B. E. PROTHEROE

Mining Technology Laboratory, Chamber of Mines Research Organization, Johannesburg (South Africa)

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Summary

Laboratory abrasive and abrasive-corrosive testing has been carried out on a range of ferritic, austenitic and martensitic stainless steels and the results compared with the testing of similar materials *in situ* in the abrasive-corrosive conditions of a gold mine. All grades were found to have better abrasive-corrosive resistance than proprietary abrasion-resistant alloys.

The austenitic grades derive their outstanding properties from their capacity to resist unstable fracture of microshear lips on the abraded surface. This is due to the strain capacity afforded by the mechanical induction of the martensitic phase transformations and the high work-hardening characteristics of the transformation product. The influence of this transformation has been studied as a function of prior cold work and velocity of abrasion on a range of austenitic grades of stainless steels.

Notwithstanding these advantages of the austenitic grades, the ferritic grades are superior in terms of cost per unit volume lost and the new duplex ferritic-martensitic steel designated 3CR12 has potential as an abrasion-resisting material in corrosive environments.

1. Introduction

Stainless steels are not generally regarded as abrasion-resistant materials despite the fact that many such alloys can be cold worked or heat treated to strength and hardness levels comparable with those of commercial abrasion-resistant materials. Furthermore, the corrosion resistance of such alloys

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would appear to make them potential candidates for many situations involving mixed corrosive–abrasive wear.

A typical application is in the systems now being utilized for the transportation of quartzite rock from the face of the gold mines in South Africa [1]. The hard quartzite rock slides over metal surfaces and copious amounts of acidic water together with the temperature and humidity levels give rise to an extreme condition of corrosive–abrasive wear. This results in high replacement costs for conventional materials.

An attempt to assess the relative abrasion–corrosion resistance of a variety of stainless steels and to correlate this resistance with microstructural parameters is made in this paper. The work forms part of a total programme aimed at reducing the cost of abrasive wear in a mining environment and other workers are referred to a companion paper [2] in which our approach to the selection of materials is described.

2. Experimental programme

2.1. Materials

The grades and chemical compositions of the steels are shown in Table 1. Mild steel and two typical abrasion-resistant materials were included for comparison purposes.

2.2. Abrasion testing

Abrasion testing was carried out on an extensively modified Rockwell disc–belt sanding machine; this embodies the principle of abrading a dead loaded cylindrical specimen (8.9 mm in diameter \times 20 mm in length) against the surface of a bonded abrasive belt. The continuous belt runs horizontally at a constant velocity while the specimen is made to transverse normally to the direction of belt movement. Thus the specimen at all times abrades against unworn particles.

Before testing, the specimen surfaces are run-in by abrading for a length of 3.66 m of fresh abrasive cloth. The specimens were then weighed to an accuracy of 0.1 mg and abraded for a distance of 14.64 m. The loss in mass for each specimen was converted to volume loss and the wear resistance was calculated relative to mild steel by using the reciprocal of volume loss, *i.e.*

$$\text{wear resistance} = \frac{\text{volume loss of mild steel}}{\text{volume loss of specimen}}$$

The reproducibility of the test was found to be within $\pm 4\%$.

The actual conditions which were employed for the majority of the testing programme were as follows: load, 20 N; length of abrasive path, 14.64 m; mean diameter of abrasive particles, 300 μm ; type of abrasive, Al_2O_3 ; velocity of the belt, 260 mm s^{-1} .

TABLE 1
Chemical analysis of the materials used

Material	Composition (%)											
	C	Mn	S	Cr	Mo	Ni	P	Si	V	Al	Ti	Nb
Austenitic grades												
316L	0.035	1.32		16.60	2.20	10.40		0.55	0.08		0.02	
304L	0.035	1.38		18.80	0.11	8.40		0.42	0.09		0.02	
301	0.074	1.52	0.009	17.1	0.049	7.22	0.021	0.56		0.02	0.01	0.024
201	0.074	5.17	0.014	17.7	0.070	3.91	0.022	0.54		0.033	0.008	0.049
Ferritic grades												
430	0.06	0.65		17.40	0.05	0.27		0.37	0.09		0.02	
3CR12	0.03	0.81		10.70	0.07	0.53		0.36	0.08		0.03	
431	0.15	0.95		16.4		1.8						
Mild steel	0.15	0.44		0.02	0.01	0.03		0.08	0.01		0.01	
Abrasion-resistant steels												
A	0.21	0.56		0.05	0.61	0.13		0.29	0.01		0.02	
B	0.3	0.5	0.02	0.72	0.2							

2.3. Abrasion-corrosion testing

It was clear that the testing of materials for dry abrasion resistance alone would probably not provide a totally satisfactory guide in the selection of materials for aggressive abrasive-corrosive conditions. Consequently, a technique of laboratory testing involving fixed periods of corrosion followed by the abrasion of the steel samples was developed. The samples were similar to those used in dry abrasion testing. The specimens were initially run in on the abrasion machine, weighed and then mounted in a Perspex holder-plate previously drilled with a series of holes 8.9 mm in diameter. The holder was mounted at 45° inside a closed Perspex box and water at a temperature of 30 °C and at pH 5.5 containing chloride and sulphate ions was allowed to flow over the previously abraded surfaces. The specimens were mounted in such a way that only the abraded surface was exposed to the corrosive environment; the other surfaces were covered with an antirust compound. The specimens were removed after a predetermined time in the rig and were cleaned, abraded and weighed. In this way the relative amounts of abrasion and corrosion were closely controlled.

A more detailed description of the methods used and the results obtained for the laboratory testing of a wide variety of low alloy and stainless steels together with results from *in situ* testing of similar materials on quartzite-carrying shaker conveyors can be found elsewhere [2].

2.4. Structural examination

Both optical microscopy and scanning electron microscopy were carried out on the abraded specimens. A selection of abraded specimens was prepared for transmission electron microscopy (TEM) by abrading one surface of annealed discs of material which were then thinned electrolytically to foils in a solution of 10% perchloric acid in methanol from the opposite side only of the disc.

X-ray analysis using both copper and molybdenum radiation was carried out with a Philips diffractometer. Retained austenite measurements were made in accordance with the recommended procedure of the manufacturer. Bulk hardness measurements were made on a Vickers hardness machine under a load of 20 kgf.

3. Results

3.1. Dry abrasion

A summary of the results of dry abrasion testing is shown in Table 2. It is clear that the abrasion resistances of all the stainless grades, whether based on a ferritic, austenitic or martensitic matrix, are superior to that of mild steel but inferior to that of the abrasion-resistant steels A and B. Furthermore, the dry abrasion resistances of the stainless grades are seen to vary with both structure and composition; the f.c.c. austenitic is generally better than the b.c.c. ferritic grades. The 301 and 201 grades and the

TABLE 2
Stainless steels

Grade ^a	Dry abrasion wear resistance relative to mild steel at 260 mm s ⁻¹	Hardness (HV 20)	Austenite stability factor	M _s (°C) ^b	M _{d30} (°C) ^c
316L (HR ann.)	1.22	148	25.82	-180	-20
304L (HR ann.)	1.32	159	23.19	-55	+30
301 (HR ann.)	1.59	181	21.99	-26	+57
201 (HR ann.)	1.49	193	21.1	+18	+85
430 (HR ann.)	1.25	180			
3CR12 (HR ann.)	1.11	132			
431 (hardened)	1.56	387			
Mild steel (normalized)	1.00	107			
Abrasion-resistant steel A (RQ temp.)	1.71	415			
Abrasion-resistant steel B (RQ temp.)	1.44	388			

^a HR ann., in the hot-rolled and annealed condition; RQ temp., in the roller-quenched and tempered condition.

^b M_s, the temperature at which martensite starts to form spontaneously.

^c M_{d30}, the temperature at which 50% martensite is produced by a true strain of 0.30 (the formulae for M_s and M_{d30} are taken from ref. 10).

hardened and tempered 431 grades appear to have the best wear resistance with the new 12% Cr grade 3CR12 [3] exhibiting a dry wear resistance only slightly superior to that of mild steel.

The effects of cold working the annealed austenitic grades by 24% cold rolling before dry abrasion testing are shown in Figs. 1 and 2 while Fig. 3 illustrates the greatly improved hardness levels caused by cold work. For 304L, 301 and 201 grades such prior cold working has the effect of *lowering* the abrasion resistance by a small but significant amount despite the large increase in bulk hardness of about 100% - 150%. The 316L grade is unique in this group of austenitic stainless steels since the opposite effect occurs.

The effects of abrasion velocity on the volume loss of material for selected grades of stainless steels are shown in Fig. 4. It is clear that the effect of increasing the speed of abrasion from 20 to 450 mm s⁻¹ results in an increased loss of material and thus a decrease in the abrasion resistance of all the alloys tested. This effect is most apparent at the lower velocities but diminishes for velocities of abrasion greater than 300 mm s⁻¹.

The ferritic grades, however, show a much smaller rate of abrasion volume loss with increase in abrasion velocity than the austenitic steels. The volume loss for a ferritic 3CR12 increases by 20% - 25% compared with an increase of 100% for the austenitic steel 201 over the speed range used.

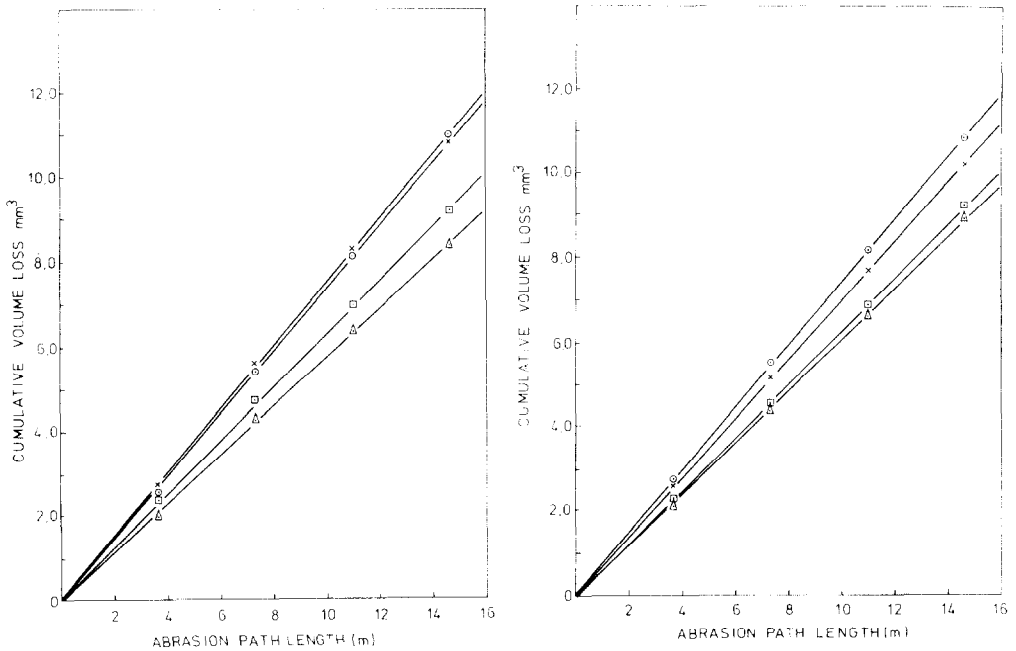


Fig. 1. The effect of prior cold working on the dry abrasive wear of 316L and 301 grades of stainless steel: x, 316L; ○, 24% cold-worked 316L; □, 24% cold-worked 301; △, 301.

Fig. 2. The effect of prior cold working on the dry abrasive wear of 304L and 201 grades of stainless steel: ○, 24% cold-worked 304L; x, 304L; □, 24% cold-worked 201; △, 201.

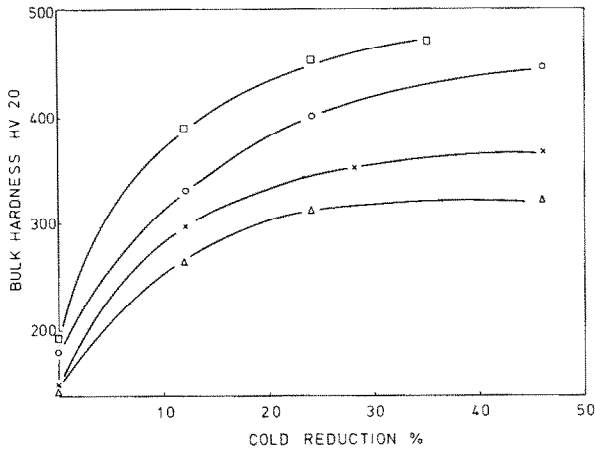


Fig. 3. The effect of cold working on the bulk hardness of austenitic stainless steels: Δ , 316 L; \times , 304L; \circ , 301; \square , 201.

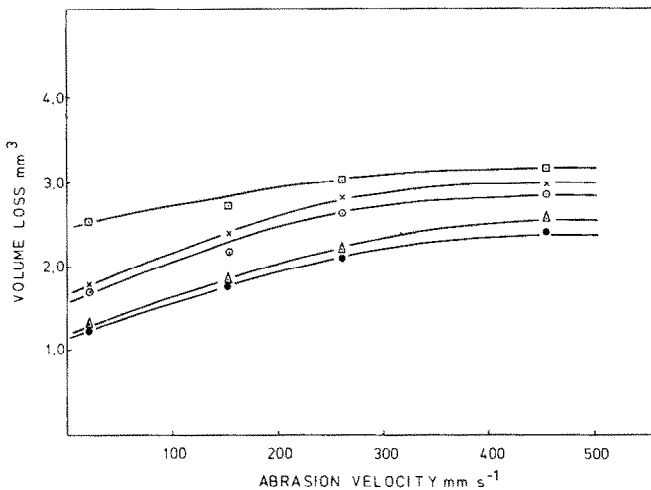


Fig. 4. The effect of the velocity on the dry abrasive wear of stainless steels, mild steel and a proprietary abrasion-resistant steel: \square , 3CR12; \times , 316L; \circ , 304L; Δ , 201; \bullet , 301.

3.2. Abrasion-corrosion

Abrasion-corrosion test results are summarized in Table 3 together with the results of some *in situ* testing of stainless steel and other steels.

With increasing amounts of corrosion relative to abrasion, the wear resistance of the stainless steel is seen to improve dramatically with respect to mild steel or other grades of proprietary abrasion-resistant materials. Figure 5 illustrates the way in which corrosion affects the relative wear rate for selected steels. Although the wear-resistant alloys are generally better in dry abrasion they become markedly worse than all the stainless steel grades when significant corrosion is present. Table 3 also shows that the *in situ*

TABLE 3
Abrasion-corrosion test results

Material	Relative wear resistance ^a in the following situations					Relative cost allowing for wear rate ^c
	Abrasion for 1.2 m at 100 mm s ⁻¹	Corrosion for 24 h and abrasion for 500 mm at 100 mm s ⁻¹	Corrosion for 48 h and abrasion for 500 mm at 100 mm s ⁻¹	In situ in a shaker conveyor	Cost kg ⁻¹ b	
316L	1.45	2.5	3.13	3.85	R3.45	1.89
304L	1.89	3.33	3.70	3.45	R1.90	1.22
430	1.25	1.96	2.00	4.17	R1.40	0.71
3CR12	1.19	2.13	2.27	3.13	R1.02	0.72
431	1.49	2.33	3.03	2.56		
Mild steel	1.00	1.00	1.00	1.00	R0.46	1.00
Abrasion-resistant steel A	1.49	1.32	1.20	1.23	R1.23	2.1
Abrasion-resistant steel B	1.64	1.52	1.39	0.98	R1.00 ^d	2.27

^a Relative wear resistance = (volume loss of mild steel)/(volume loss of specimen).

^b Cost based on 100 ton of 4.5 mm plate (South African prices, January 1980).

^c The wear rate of mild steel was set equal to unity.

^d Not known; R1.00 used.

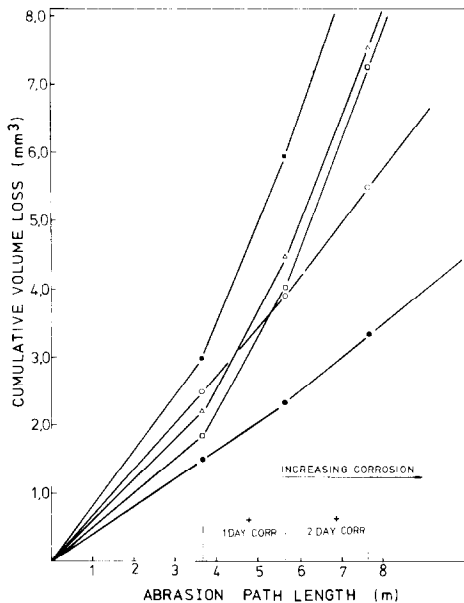


Fig. 5. The effect of intermittent corrosion on the abrasive wear of steels: ■, mild steel; △, abrasion-resistant steel A; □, abrasion-resistant steel B; ○, 3CR12; ●, 304L.

testing results for materials on a shaker conveyor are similar to those obtained in the laboratory which had been corroded for 2 days followed by abrasion for 500 mm. The practical significance of this result is discussed by Allen *et al.* [2].

3.3. X-ray diffraction

X-ray diffraction was used to detect the presence of transformation products in the cold-rolled and abraded specimens of the austenitic stainless steels. Figures 6 - 9 show the results obtained.

Figure 6 illustrates the effect of increasing amounts of cold-rolling deformation on the structure of an AISI 301 steel. In the annealed state only the f.c.c. phase could be detected. After 12% cold reduction, hexagonal peaks were observed; these corresponded to the formation of the ϵ martensite phase. There was also some evidence for a b.c.c. (110) peak; this was most probably due to the formation of α^1 martensite. With increasing amounts of deformation, the hexagonal peaks slowly disappear to be replaced by the b.c.c. peaks of the α^1 martensite phase. The amount of α^1 martensite increased with increasing amounts of cold reduction as shown in Fig. 7. In the computation of the amount of α^1 martensite, no allowance was made for the ϵ martensite in the lattice. Nevertheless, the values obtained agree closely with previous findings [4].

This transition from a wholly austenitic phase to a phase containing substantial quantities of α^1 martensite via the formation of ϵ martensite was only found in the three alloys 304L, 301 and 201 after cold working. The

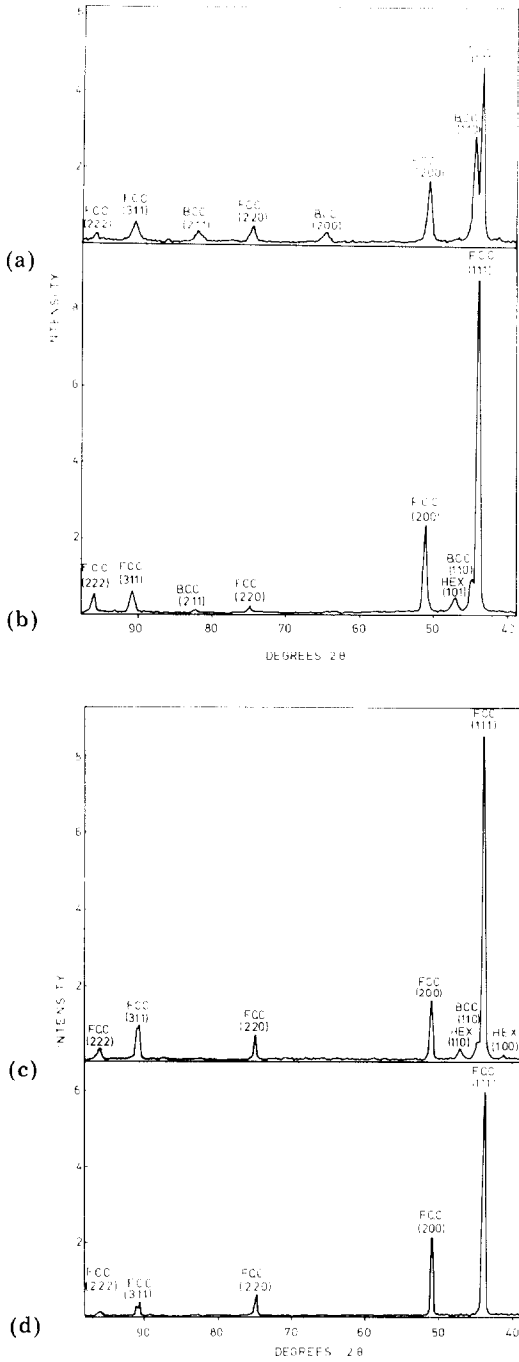


Fig. 6. The X-ray diffraction patterns for a 301 grade stainless steel after various amounts of cold working: (a) 46% cold reduced; (b) 24% cold reduced; (c) 12% cold reduced; (d) annealed at 1050 °C.

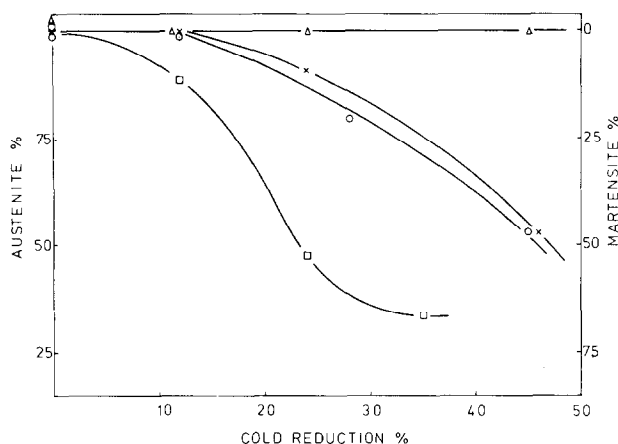


Fig. 7. The effect of prior cold working on the amount of martensite present in the austenitic grades of stainless steels: Δ , 316L; \circ , 304L; \times , 301; \square , 201.

201 grade was found to be the most metastable alloy and exhibited greater amounts of α^1 martensite for any given amount of deformation than either the 304L and 301 grades. No martensite was detected in the 316L grade even after 45% deformation.

The presence of α^1 martensite was detected in the abraded surfaces of all the austenitic stainless steels and the results for the 316L grade are shown in Fig. 8. At the lower abrasion speed of 20 mm s^{-1} the abraded surface of this steel contained approximately 25% α^1 martensite in the austenitic matrix. As the abrasion velocity increased, the amount of α^1 martensite in the surface layers decreased (Fig. 9). A similar trend of decreasing amounts of α^1 martensite in the abraded surface layers of the other austenitic grades (304L, 301 and 201) with increasing abrasion velocity was also found (Fig. 9).

A comparison of the diffraction traces for the 304L and 201 steels in the annealed, annealed and abraded and cold-worked states was made. We noted that the hexagonal peaks previously found in all these steels after cold reduction corresponding to the formation of the ϵ martensite phase were not observed at any stage in the surfaces of the abraded specimens.

3.4. Scanning electron microscopy

The wear tracks observed on the abraded surfaces of all the specimens after abrasion were typical of those normally seen on ductile materials in which extensive plastic deformation occurs simultaneously with cutting and shearing. The abrasive particles penetrate the surface and plough out furrows or grooves in the course of which significant lateral flow of material occurs. Figures 10 - 13 show typical grooves and ridges on the abraded surfaces of martensitic, austenitic and ferritic stainless steels. The associated strain markings and fractures in the deformed materials and adjacent matrix can be seen to be different.

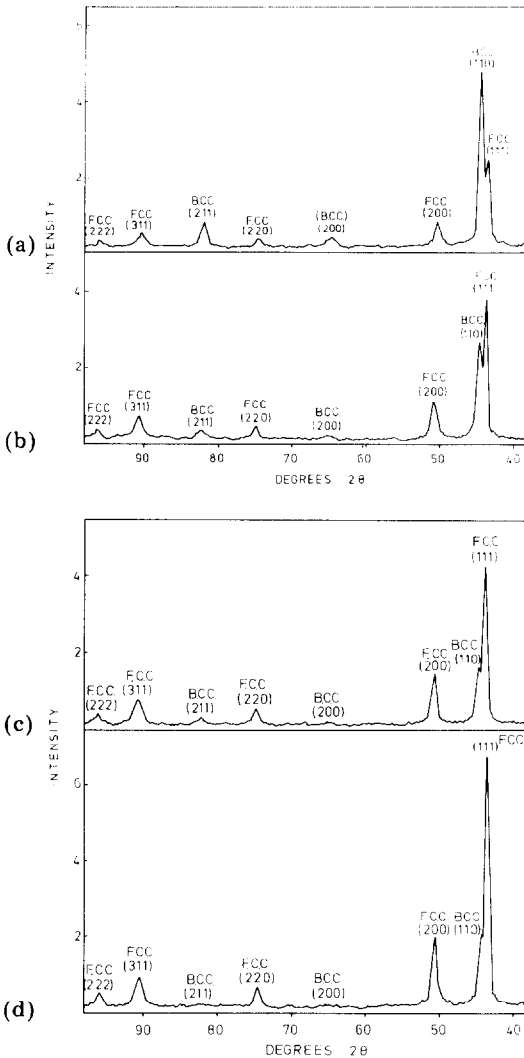


Fig. 8. The X-ray diffraction patterns for a 316L grade of stainless steel after dry abrasive wear at different velocities: (a) 20 mm s⁻¹; (b) 150 mm s⁻¹; (c) 260 mm s⁻¹; (d) 450 mm s⁻¹.

For the harder martensitic steels (Fig. 10) extensive plastic flow is not apparent and material is lost by a series of brittle cracking events along the sides of the wear tracks. The more ductile austenitic and ferritic materials show plastic flow activity in the regions adjacent to the wear tracks and material is lost by ductile flow of shear lips (Figs. 11 - 13). It appears that fragments are lost when the ability of the metal to resist shear is exhausted. A comparison of the characteristics of the slip lines in the regions adjacent to the wear tracks is informative. In the austenitic steels (Fig. 11), the flow markings are straight and continuous and this is consistent with persistent

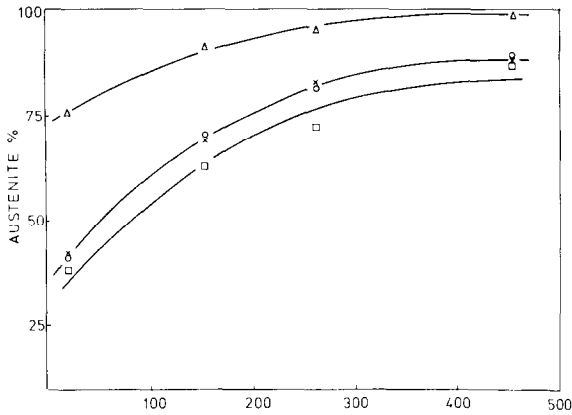


Fig. 9. The effect of abrasion velocity on the degree of phase transformation induced by dry abrasive wear (specimen diameter, 9.8 mm; load, 20 N; Al_2O_3 grit size, 300 μm): Δ , 316L; \times , 304L; \circ , 301; \square , 201.

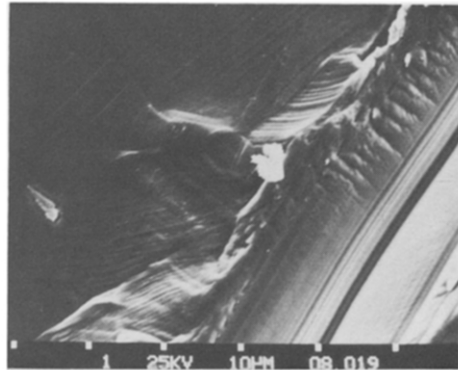
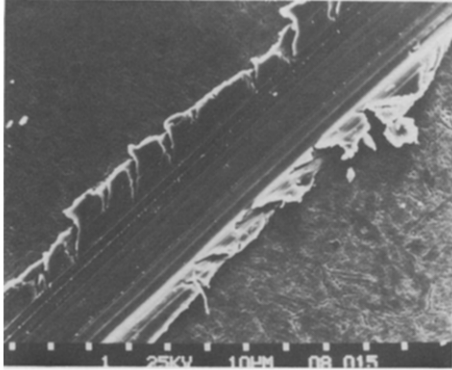


Fig. 10. A wear track produced by the dry abrasion of stainless steel 431.

Fig. 11. A wear track and associated deformation produced by the dry abrasion of stainless steel 316L.

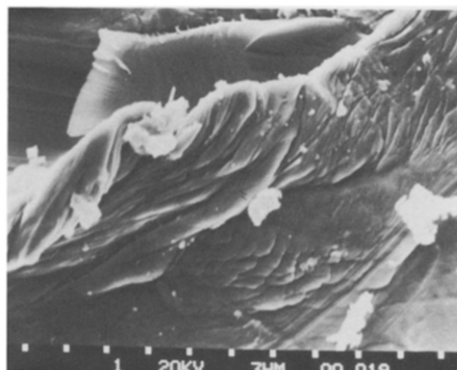
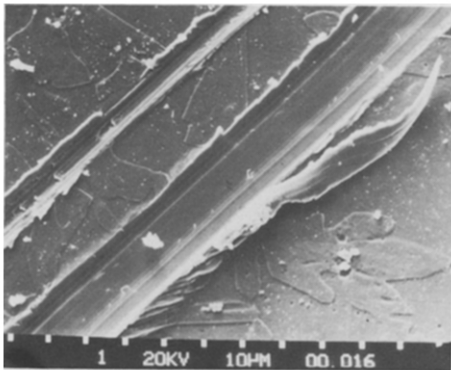


Fig. 12. A wear track produced by the dry abrasion of stainless steel 3CR12.

Fig. 13. A wear track and associated deformation produced by the dry abrasion of stainless steel 430.

slip bands, twinning and strain-induced martensite which are all known to be a consequence of the low stacking fault energies of these metals. The wear tracks in the ferritic steels are accompanied by many varied and multidirectional flow markings (Fig. 13). Numerous cross-slip events have taken place and the slip flexibility characteristic of b.c.c. metals is evident.

From Fig. 12 it can be seen that the scratching damage is unaffected by the two-phase structure of the 3CR12 steel in that the different hardnesses of the two phases do not affect the scratch profile.

3.5. Transmission electron microscopy

The deformation structures produced in the near-surface regions of abraded specimens of 304L and 430 stainless steels are illustrated in Figs. 14 and 15 respectively. The austenitic steel clearly shows the effects of a low stacking fault energy. Dislocations are confined to distinct slip planes and microtwins have been produced. There are numerous examples of inter-lapping stacking faults and pile-up configurations on neighbouring slip planes which, according to Brooks *et al.* [5], are the nuclei for transformation to ϵ and α^1 martensites. Evidence for gross transformation to martensite was not obtained and this, in conjunction with the X-ray diffraction work, strongly suggests that the stress-induced transformation only occurs extensively at the abraded surface. This roughened surface had to be removed by electropolishing to obtain suitable uniformly thin samples. The present results do, however, demonstrate that the wear deformation does not produce a dislocation cell structure such as that produced in austenitic steels by fatigue deformation [6] but rather a structure more characteristic of that produced by shock loading [7].

The dislocation structures formed near the abraded surfaces of the ferritic 430 stainless steel (Fig. 15) are not similar to structures produced by either cold work or the fatigue of b.c.c. metals. The presence of straight dis-

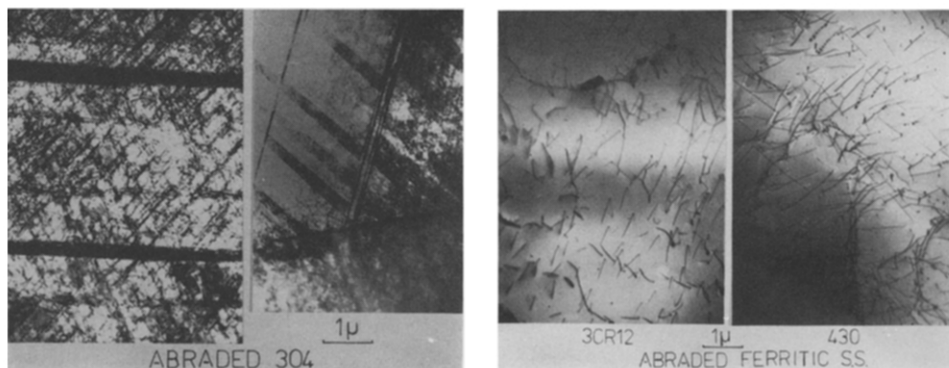


Fig. 14. Typical dislocation structures produced in the near-surface regions by the dry abrasion of stainless steel 304L.

Fig. 15. Typical dislocation structures produced in the near-surface regions by the dry abrasion of stainless steel 430.

locations and three-dimensional networks suggests that recovery processes are simultaneous to the abrasion. It must be concluded therefore that local instantaneous temperatures produced by frictional heating are sufficient to allow a rearrangement of the dislocation configurations.

4. Discussion

It is clear from the results of the present laboratory and *in situ* testing that the superior corrosion resistance of stainless steels as a group makes them very attractive materials for situations involving an abrasive and a corrosive environment. Examination of Table 3 demonstrates that all the stainless steel grades, austenitic, ferritic and martensitic, outperform proprietary abrasion-resistant materials in the abrasive–corrosive conditions experienced by ore-handling equipment in a mining environment. However, it is important for the optimization of these stainless steels that the microstructural parameters which contribute to their dry abrasive resistance are understood. Certainly, no simple mechanical property such as initial bulk hardness appears sufficient to predict their wear resistance (see Table 2).

The abrasive condition considered in the present paper involves the movement of very hard particles of alumina or quartz under high normal loads along the surface of relatively soft metallic surfaces. As such, the magnitude of the stress generated in the metal surface must always be well in excess of the yield strength. It follows therefore that the loss of metal by abrasion cannot be combated by moderate increases in the yield strength of the ferrous metal but rather by an improvement in the ability of the metal to accommodate the shear strains associated with the abrasion. Figures 10 - 13 show clearly that material is lost by rupture of the shear lips formed along the wear grooves. The ability of these lips to accommodate the high shear strains without fracture must be related to the resistance to unstable ductile fracture which in turn is related to the work-hardening characteristics of the metal. In the analysis of the ideal true stress σ –true strain ϵ behaviour of a sample tested in tension, the strain to unstable “necking” of the ϵ neck can be shown [8] to be equal to the magnitude of the work-hardening exponent of the equation $\sigma = K\epsilon^n$ which represents the plastic flow curve. The metastable austenitic stainless steels have distinct advantages when this is considered since the strain to necking is enhanced by both the strain afforded by the induced martensitic transformation and the high work-hardening capacity of the transformation product. Figure 16 shows a schematic representation of this advantage. The maximum advantage will be realised provided that no transformation has occurred before deformation and that an optimum amount of martensite forms during deformation to give a high work-hardening rate. The maximum strain has been found by Lecroisey and Pineau [9] to be dependent on temperature within the range between the temperature M_s for spontaneous transformation and the maximum temperature M_d at which the transformation can be mechanically

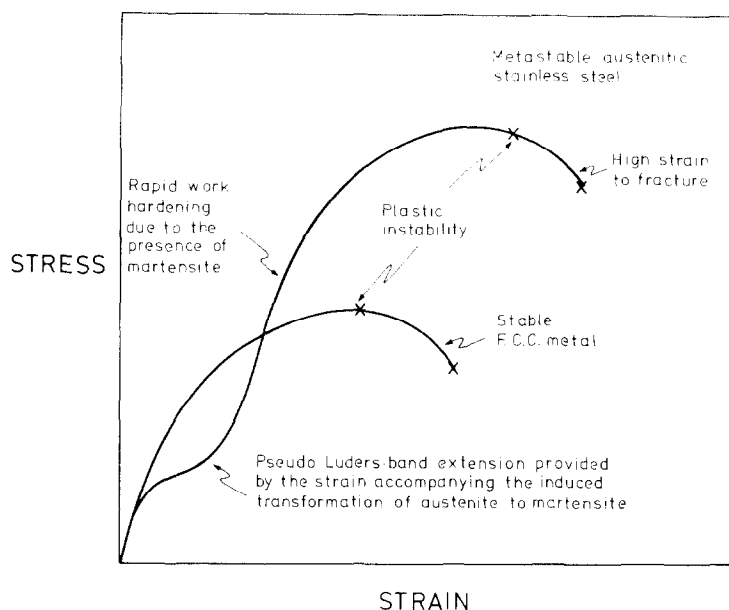


Fig. 16. Schematic representation of the stress-strain curves for stable and unstable austenitic stainless steels demonstrating the advantages provided by the stress-induced transformation to martensite.

induced. This optimum temperature and the value of the optimum strain are controlled by the actual composition of the stainless steel. An attempt will now be made to explain the dry abrasion behaviour of the various austenitic stainless steels tested in the present investigation as a function of prior cold work and speed of abrasion in terms of the rationale presented above.

It is well documented that the metastable austenitic matrix in the alloys 304L, 301 and 201 can undergo a stress-induced transformation to α^1 martensite during deformation. X-ray diffraction and TEM results have shown that such a transformation occurs for *all* the austenitic grades investigated during the dry abrasive testing. This is a somewhat surprising result since the austenite of the 316L grade is normally considered to be stable and has an M_{d30} temperature of -20°C (Table 2). Transformation has, however, been observed during sliding wear tests on this type of alloy [9]. The dynamic nature of the shear which accompanies abrasion and the associated high strain rates together with a "lean" composition are probable explanations of the observations.

The instability is thought to influence the wear process by resisting the fracture of shear lips in the manner described above and thus it might be expected that the most unstable grade 201 should have shown a superior abrasion resistance to the slightly more stable 301 grade. This was not found to be the case (see Table 2). This result can be attributed to the probability that some transformation has occurred spontaneously in the 201 before abrasion whereas the 301 grade will be totally austenitic before testing. This

follows from the values of the M_s temperatures, which are 18 °C and -26 °C for grades 201 and 301 respectively. The presence of this martensite before abrasion testing will raise the flow stress of the metal relative to the work-hardening rate and thus decrease the capacity of the material to accommodate high strains.

The introduction of the martensite phase into the specimens of 304L, 301 and 201 by prior cold work will similarly reduce the strain to fracture of these alloys and thus decrease their abrasion resistance, as shown in Figs. 1 and 2. The grade 316L, unlike the other austenitic grades, has a subambient M_d temperature and thus cold work before abrasion does not reduce its resistance (Fig. 1).

The experimental results (see Fig. 4) show that the velocity of abrasion in the range 20 - 450 mm s⁻¹ has an important influence on the abrasion resistance of the austenitic steels. As the speed of abrasion was decreased from 450 to 20 mm s⁻¹ the wear resistance of the 316L and 304L, 301 and 201 grades improved by approximately 100%. At the very low speeds the abrasion resistance of these grades was superior to that for the abrasion-resistant steel. No such dramatic effect was found for the other grades of stainless steels, mild steel or the abrasion-resistant steels. Only a slight decrease in wear resistance was observed as the speed of abrasion was increased from 20 to 300 mm s⁻¹. The abrasion resistance for each material became almost constant above a velocity of 300 mm s⁻¹.

It is expected that a change in the velocity would result in (a) an increase in the temperature of the abraded surface leading to easy slip, deformation and higher mass losses and (b) an increase in strain rate which increases the flow stress of the material, while making deformation more difficult with a consequent improvement in abrasion resistance. Any change in volume loss of material with abrasion velocity should therefore reflect a balance between these two factors. For the more strain-rate-sensitive and temperature-sensitive b.c.c. ferritic alloys these two factors appear to be almost balanced for the range of abrasion velocities investigated and because of this very little change in volume loss resulted.

For the austenitic steels, however, it appears that the surface temperature term is dominant, since an increase in the abrasion rate leads to a reduction in the amount of deformation-induced α^1 martensite in the austenitic matrix of each steel (Figs. 8 and 9). As the temperature is raised by frictional heating and approaches the M_d temperature for deformation-induced α^1 martensite for the alloy concerned, there is a decrease in the amount of martensite produced in the austenitic matrix. This will have a detrimental effect on the abrasion resistance since the reduced ability to form martensite will decrease the strain capacity of the abraded surface via a reduction in the strain associated with the transformation and the work-hardening rate. In a real situation the presence of mine water, albeit corrosive in nature, would offset this decrease in abrasive properties by acting as a coolant and lubricant and so reducing the frictional heating.

Examination of Table 3 reveals the distinct cost advantages of the cheaper ferritic 430 and duplex 3CR12 steels over the more abrasion-corrosion resis-

tant austenitic grades. However, in view of the inherent wear properties of the metastable austenitic steels described herein, it seems appropriate that research effort should be directed towards the design of an inexpensive steel of this type or, alternatively, of a duplex ferrite–austenitic steel in which the austenitic phase is unstable with respect to the formation of martensite on abrasion.

5. Conclusions

(1) The abrasion resistance of various ferritic, martensitic and austenitic stainless steels was shown to be superior to that of mild steel. Their abrasive–corrosive properties are very much superior to those of most proprietary abrasion-resistant alloys.

(2) Metastable austenitic alloys appear to have the best overall abrasion resistance within the stainless steel group and under slow rates of strain their dry abrasion resistance is superior to those of typical abrasion-resistant alloys. Such abrasion resistance coupled with toughness and good corrosion resistance would clearly allow these alloys to compete successfully in many situations involving abrasive–corrosive wear.

(3) The origin of the high abrasion resistance of these metastable steels can be rationalized in terms of the high strains which can be accommodated before material is lost from the abraded surface. This capacity is provided by the strain associated with the stress-induced transformation and the high work-hardening rate of the martensitic product. This proposal is consistent with the effects which prior cold work and speed of abrasion have on the wear properties and the X-ray diffraction and electron microscopy examinations of the abraded surfaces.

(4) It must be remembered, however, that on the basis of cost the ferritic or martensite grades are very attractive. A cost analysis for the *in situ* testing has shown that ferritic grade 430 or duplex 3CR12 are much cheaper to use per unit of material lost than commercial austenitic grades. The optimum stainless steel would appear to be either an inexpensive metastable austenitic grade or a duplex stainless steel consisting of a fine-grained mixture of ferrite and unstable austenite.

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