The abrasion and abrasion–corrosion properties of a 9% chromium steel

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Abstract

A 9Cr2Ni0.7Mo steel was assessed for its dry abrasion and abrasion–corrosion performance using a laboratory test which simulates the conditions experienced by rock conveyors in South African gold mines. The alloy is found to exhibit abrasion and abrasion–corrosion properties comparable with most and better than several higher chromium containing alloys designed for mining applications. The good dry abrasion performance of the alloy is attributed to the good combination of high hardness, strength and toughness imparted to the alloy by its duplex microstructure of lath martensite and thin film of interlath retained austenite. The good abrasion–corrosion results are a product of the good dry abrasion performance coupled with the excellent corrosion properties in the simulated mine water. The favourable corrosion properties are linked to the effect of the Ni and Mo additions on suppressing pitting tendencies thereby imparting corrosion resistance equivalent to higher chromium containing steels. Experimental evidence however shows that the alloy is operating at the limit of its passivity for the 46 h corrosion period employed in the corrosion–abrasion test. Exposure for periods much in excess of this figure will lead to breakdown and poor corrosion–abrasion performance relative to higher chromium containing steels. The production variables of prior cold working and plate thickness are found to exert negligible influence on dry abrasion and abrasion–corrosion properties.

1. Introduction

The high cost associated with the deterioration of equipment under the action of the harsh abrasive–corrosive conditions prevalent in South African gold mines [1] together with the realisation of the unavailability of suitable corrosion–abrasion resistant alloys [2, 3] prompted an alloy development programme.

This work is related to the mechanisation of operations and handling systems which demand more reliable and durable materials. The requirements for a general purpose steel were defined as: resistance to corrosion and abrasion, sufficient toughness for structural applications, formability, weldability, machineability and flame cutability [3]. Naturally the material must be economically attractive. Several different approaches were adopted by independent parties in an attempt to achieve these desired properties. These approaches included the development of metastable austenitic [4, 5], low carbon lath martensites [6, 7] and modification of the duplex ferritic–martensitic 3CR12[8].

The manufacturer adopted the rationale of reducing manufacturing costs by producing an as-rolled steel requiring no further heat treatment. They considered local production facilities and opted for a martensitic type steel [9]. The alloy formulation was based on that proposed by Thomas *et al.* [10–12] which consists of a duplex microstructure of dislocated, fine lath martensite with thin films of interlath retained austenite, the essential difference being higher chromium content for increased corrosion resistance. A range of laboratory vacuum melts containing 0.1–0.3% and 4–12% Cr together with a variety of elemental additions were produced. Four basic alloy classes were identified, *viz.* Cr–Ni, Cr–Mo, Cr–Mo–Cu and Cr–Ni–Mo. The melts were forged, austentised at 1050 °C and oil quenched.

Salt spray corrosion tests showed the Cr–Ni–Mo steels to exhibit superior corrosion properties to the other classes. Potentiodynamic tests in mild synthetic mine water revealed that 8–10% Cr–Ni–Mo showed better passivation than a plain 12% Cr steel. This led Grobler *et al.* [9] to conclude that a decrease in Cr content was adequately compensated for by Ni and Mo additions.

Corrosion–abrasion tests simulating underground mining conditions performed on the four classes of laboratory vacuum melts yielded values comparing favourably with steels generally in use in the mines [13]. On the basis of the abrasion and abrasion–corrosion

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performance and compromising as regards the requirements for a corrosion–abrasion resistant steel for general use, a micro-alloyed 9Cr–2Ni–0.7Mo steel was produced in a 160 ton arc furnace melt, rolled and air cooled in line. This alloy, designated 927, forms the subject of this paper. The corrosion–abrasion performance of the alloy is examined in detail and compared with that of other corrosion–abrasion development alloys.

2. Experimental materials and procedure

The chemical composition of the 927 alloy is given in Table 1. Also shown in the Table are compositions of variations of the alloy used to investigate the affects of changing the sulphur and carbon content on corrosion–abrasion properties. Typical mechanical properties of the alloy in the as-rolled condition include a yield strength of 1000 MPa, UTS of 1500 MPa, elongation of 14%, hardness of 500 HV30 and a Charpy toughness (TL) of 60–80J at 20 °C [9]. The microstructure consists of fine, lath martensite (Fig. 1) surrounded by a thin film of retained austenite [9].

2.1. Experimental methods

A laboratory dry abrasion and an intermittent abrasion plus corrosion test have been developed which rank materials in the same order and produce similar wear morphology to an *in situ* test conducted on a rock conveyor [14]. The dry abrasion test is performed on a pin-on-belt abrasion testing machine. The specimens ($10\times10\times25$ mm^3) are abraded end-on on a bonded 80 grit abrasive belt under a load of 32.1 N and a velocity of 0.260 m s^{-1}. The test rig is so designed that the specimen always abrades against unworn particles. The specimens are initially run in, cleaned ultrasonically, dried and weighed to an accuracy of 0.01 mg. The specimens are then abraded, cleaned and weighed at the end of each 3 m abrasion cycle for a total path length of 12 m. The relative dry abrasion resistance or RAR is then computed as the ratio of material loss between mild steel O70M20 and the material.

TABLE 1. Compositions of the 927 test alloys

Steel	Chemical composition (wt.%)						
	C	Cr	Ni	Mn	Mo	S	Al
927	0.16	8.88	2.01	0.62	0.68	0.002	0.019
927–5	0.15	9.05	2.36	0.70	0.83	0.0024	0.026
927–6	0.15	9.04	2.36	0.74	0.85	0.0170	0.095
927–7	0.17	9.18	2.32	0.69	0.84	0.0023	0.059
927–8	0.17	9.18	2.39	0.72	0.84	0.0110	0.066

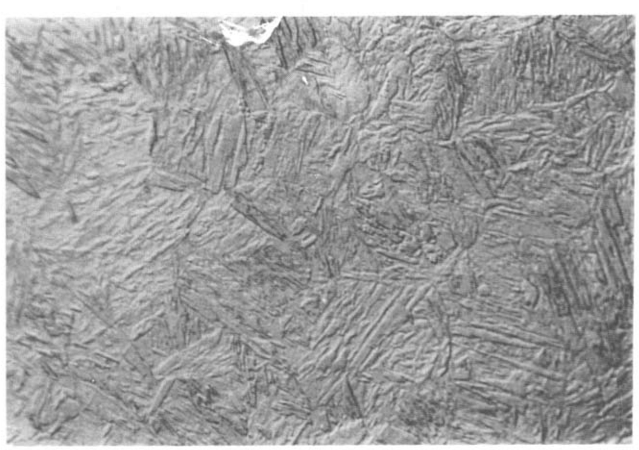

Fig. 1. The fine grained lath martensitic structure of the 927 material (micrograph taken using Normarski interference facility).

For the abrasion–corrosion test identical specimens are run in on the abrasion rig, weighed to 0.01 mg, coated on the five unabraded sides with a lacquer and mounted with the abraded surface flush with a perspex test plate. The plate is set at an angle of 30° and synthetic mine water maintained at 30 °C, is circulated over it. The simulated mine water comprises of 345 p.p.m. sulphate and 175 p.p.m. chloride at a pH of 6.8. At intervals the specimens are removed from the rig, cleaned of laquer and oxide product by ultrasonic agitation in a buffered mild acid solution and the corrosion weight loss determined. The specimens are then abraded a short distance, cleaned, re-weighed and replaced into the corrosion rig. This corrosion–abrasion procedure is repeated for four cycles. The relative wear resistance laboratory (RWRL) value is then computed as the ratio of total volume loss of mild steel to the material under investigation.

Two versions of the test have been devised which differ in terms of the duration and extent of corrosion and abrasion cycles of the test. The RWRL 1 test involves four cycles of 0.25 m abrasion and 46 h corrosion whereas the RWRL 2 test involves four cycles of 1 m abrasion and 22 h corrosion. The two tests thus place different emphasis on the abrasion and corrosion properties of a material. The RWRL 1 test with its longer corrosion time and shorter abrasion path length places more emphasis on the corrosion properties of a material. The converse is true for the RWRL 2 test.

In addition to the above-mentioned standard test, another test was conducted in which several identical specimens were prepared as for the corrosion–abrasion tests and placed in the corrosion rig. At periodic intervals over a period of two weeks, a specimen was removed and the volume loss determined. All the tests were complemented with stereo and scanning electron microscopy of the abraded and abraded–corroded surfaces in order to gain insight into the mechanisms of corrosion and abrasion operating.

3. Results

The average of the test results for the dry abrasion tests (RAR), low frequency laboratory corrosion–abrasion tests (RWRL 1) and the high frequency (RWRL 2) tests are shown in Table 2. Also shown in the Table for comparison purposes are the wear indices for several other development alloys devised for corrosive–abrasive applications.

Both the relative dry abrasion performance (RAR) and the corrosion–abrasion performance (RWRL 1 and 2) compare extremely favourably with the other development alloys included in the corrosion–abrasion test programme (Figs. 2 and 3).

3.1. Dry abrasion

Scanning electron microscopy of the dry abraded 927 surfaces (Fig. 4) revealed very shallow, thin, straight and clean wear tracks with very thin side ridges. The thinness of the side ridges is related to the limited ductility of the material and is symptomatic of the dominance of the cutting process in the abrasion of these types of materials.

Tapered sections through the abraded surface revealed a small (2–5%) difference in hardness between the bulk and the abraded surface. This is a reflection of the low work hardening ability and limited ductility of the material.

3.2. Corrosion–abrasion testing

Corrosion–abrasion indices compare well with those of higher chromium containing steels like 12/10 and 3CR12. The alloy suffers small corrosion volume losses in both the RWRL 1 and RWRL 2 tests, being in the range 4–10% and 2–5% respectively.

TABLE 2. Abrasion and abrasion–corrosion wear indices of 927 and other steels in the alloy development programme

Material	Hardness HV30	RAR	RWRL 1	RWRL 2	Micro-structure
927	490	1.55	10.45	2.59	M
12/10	329	1.45	12.13	2.96	A+M
3CR12	183	1.03	8.15	2.05	F+M
304L	206	1.30	11.00	2.30	A
Abrasalloy	433	1.36	1.06	1.21	M
Wearalloy [4]	461	1.34	1.16	–	M

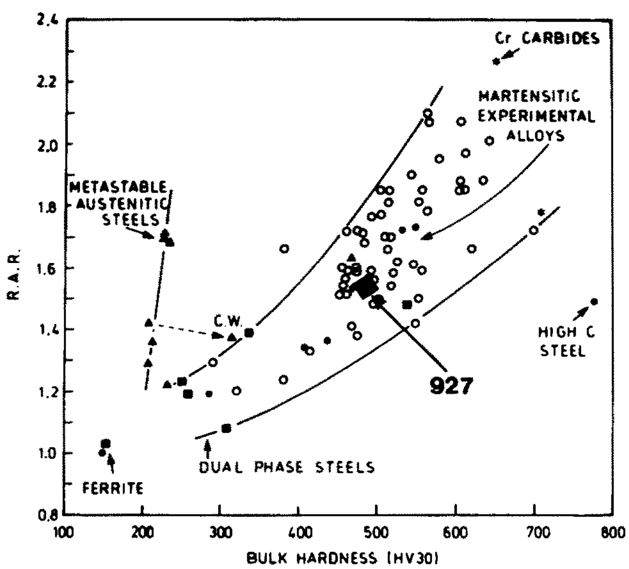

Fig. 2. Relationship between abrasion resistance (RAR) and bulk hardness. The position of 927 is clearly marked [19].

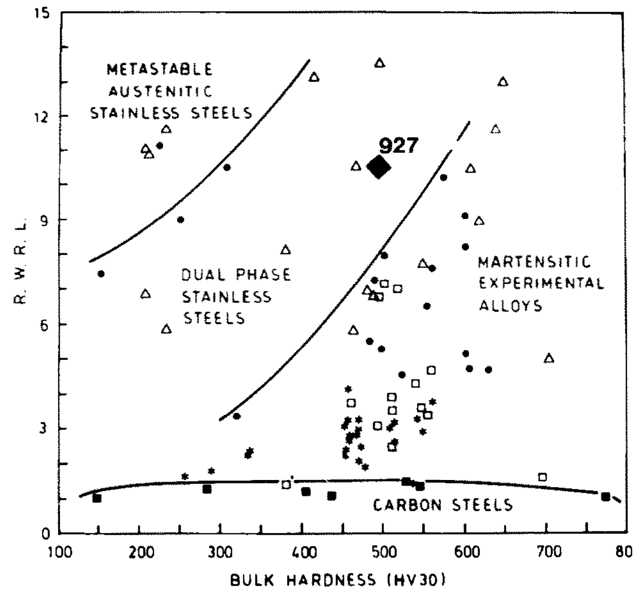

Fig. 3. Relative wear resistance RWRL 1 as a function of bulk hardness for the four groups of steels containing increasing amounts of chromium [19]. Again the position of 927 is clearly marked.

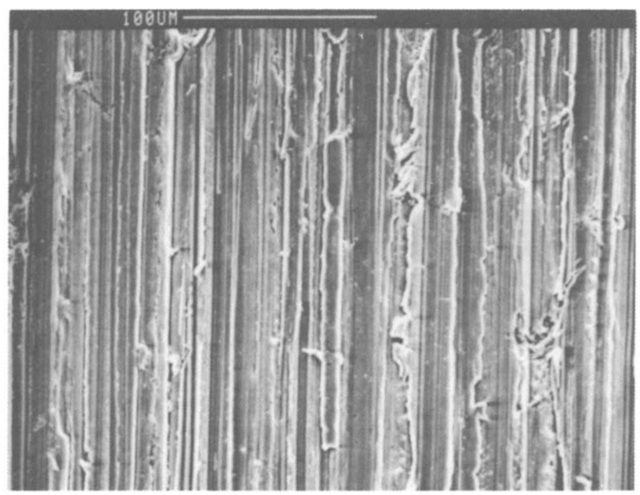

Fig. 4. The abraded surface of 927 showing clean shallow wear tracks with thin side ridges.

In a plot comparing the corrosion volume loss as a function of chromium content for the different development alloys (Fig. 5) 927 can be seen to have experienced a small but similar corrosion volume loss to some of the higher chromium containing alloys. However, with its chromium content of 9% it is in the critical region with respect to the chromium content for sustained passivity in the simulated mine water solution.

Visual examination of the abraded–corroded surface after 46 h corrosion revealed trails of a reddish-brown corrosion product on the surface eminating largely from several well developed pits. Subsequent SEM examination of the abraded–corroded surface showed that the corrosion product was randomly distributed on the abraded surface, occurring both in the bottom of wear tracks and along ridges. Removal of this corrosion product by ultrasonic agitation in a buffered mild acid solution and subsequent examination with the SEM disclosed the existence of numerous small corrosion pits in the bottom of wear tracks (Fig. 6) and the occasional well-developed pit (Fig. 7). The inherent chemical stability and mechanical strength of the oxide film became apparent by the difficulty in removing the film by ultrasonic agitation in the mild acid solution. The removal of the oxide film on steels containing 9% chromium but no nickel or molybdenum was relatively easy.

3.3. Corrosion testing

Potentiodynamic scans performed on the alloy according to ASTM G5-78 (ASTM, 1980) at a scan rate

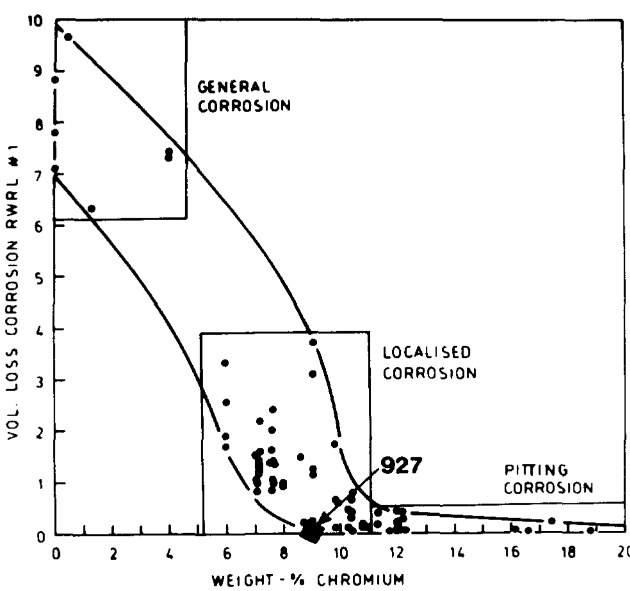

Fig. 5. Volume loss by corrosion of an abraded surface under RWRL 1 test conditions as a function of chromium content [19]. Three types of chromium are indicated. The position of 927 is clearly marked.

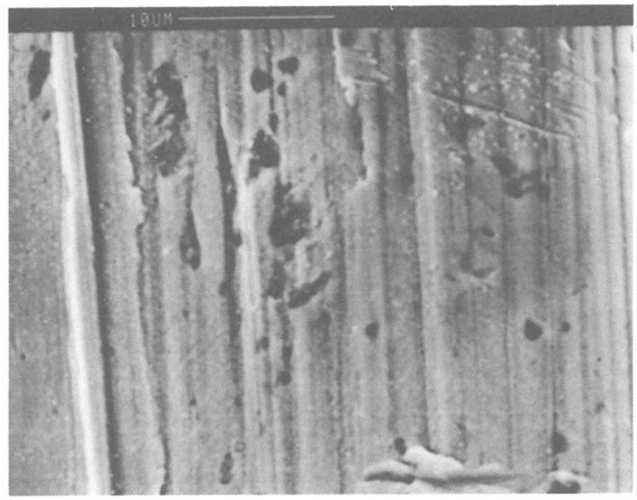
Fig. 6. The alloy on the verge of breaking down with very small corrosion pits forming in wear tracks.

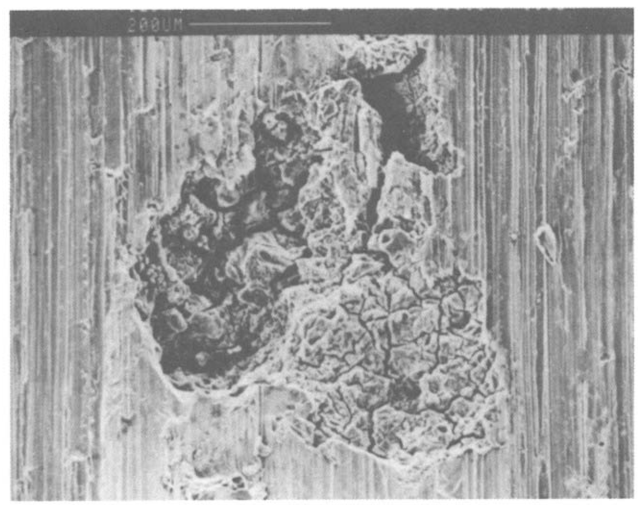
Fig. 7. A typical example of a well-developed pit which was found randomly on the surface of several specimens.

of 0.14 mV s^{-1} revealed that the alloy is passive in the mine water used in the investigation, exhibiting a passive potential range of 500 mV, passive current density of 9.5×10^{-1} μA cm^{-2} and a free corrosion potential of -300 mV (*vs.* SCE).

4. The effect of production variables on abrasion and abrasion–corrosion properties

Once initial tests had proved the feasibility of 927 for corrosive–abrasive applications, a need was identified for investigating the affect of production variables like plate thickness and prior cold working (caused by subsequent forming operations) on abrasive and abrasive–corrosive properties.

4.1. Plate thickness

The 927 material was produced in a 160 ton arc furnace melt and was rolled into 1700–2000 mm wide plate of 4–25 mm thick and air cooled. The results of the RAR, RWRL 1 and RWRL 2 tests as a function of plate thickness are presented in Table 3 together with the bulk hardness values (Vickers 30 kg) for the respective test samples. The results of the tests are presented graphically in Fig. 8 together with a hardness plot.

There is no significant influence of production plate thickness on either the dry abrasion or corrosion–abrasion performance. It is also evident from the figure that the hardness decreases only marginally with increasing plate thickness, exhibiting a 10% decrease in hardness from the thinnest to the thickest plate. SEM examination also failed to highlight any difference between the appearance of the abraded and abraded–corroded surfaces of the various plate thickness samples.

4.2. Cold work

Strain hardening can be defined as an increase in the hardness and strength caused by plastic deformation at temperatures below the recrystallisation temperature. Cold work is the operation of producing such deformation and resulting permanent strain. The different degrees of cold working were achieved by cold reducing of the 25 mm plate.

TABLE 3. The abrasion and abrasion–corrosion wear indices as a function of plate thickness. Also shown is the percentage of total volume loss attributed to corrosion in the RWRL 1 and RWRL 2 tests

Plt. Thk. (mm)	Hardness (HV30)	RAR	RWRL 1	Wt. % Corr.	RWRL 2	Wt. % Corr.
M.S.	193.6	1.00	1.00	89.96	1.00	42.71
6	487	1.56	15.23	3.77	2.71	1.61
8	478	1.50	14.92	4.52	2.56	6.82
10	467	1.50	14.67	9.09	2.70	2.41
17	449	1.58	13.45	0.00	3.08	3.21
25	441	1.57	13.23	6.56	2.78	2.07

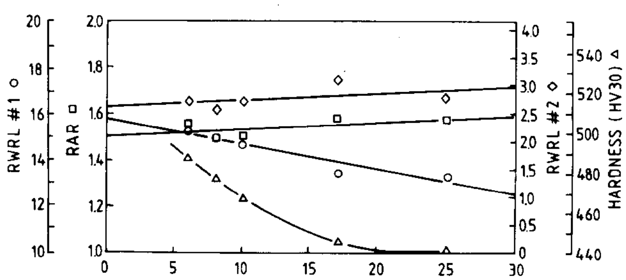

Fig. 8. A plot of the abrasion and abrasion–corrosion wear indices as a function of plate thickness.

The RAR, RWRL 1 and RWRL 2 wear indices are presented in Table 4 together with the hardness values and are represented graphically in Fig. 9. Although there is a 100 HV30 (20%) hardness difference between undeformed and maximum deformed (57%) samples, the degree of prior cold work is seen to exert negligible influence on dry abrasion and abrasion–corrosion properties. Cold working is reputed to increase the corrosion kinetics but no detectable influence was found in this examination either by difference in corrosion volume losses or appearance of the corrosion surface under SEM (*i.e.* morphology and distribution of pits).

4.3. Influence of carbon and sulphur content

Four different 927 samples containing variations in carbon and sulphur content (Table 1) were tested for the influence of these constituents on the wear performance. The detrimental effect of sulphur on corrosion resistance by the preferred initiation of pits at manganese sulphide inclusions is well documented. Consequently, one might anticipate poorer abrasion–corrosion performance by the higher sulphur–containing alloys, especially in the RWRL 1 test with its longer corrosion period. This expectation was not borne out by the test results as corrosion volume losses in the RWRL 1 and RWRL 2 tests did not differ significantly between the lower and higher sulphur containing samples (Table 5). This was confirmed by SEM examination of the abraded–corroded surfaces which showed no higher incidence of corrosion pits in the higher sulphur-containing samples.

5. Discussion

Zum-Gahr [15] identified the properties of a material influencing abrasive wear as yield strength under restraint or hardness, work hardening, ductility, crystal anisotropy and mechanical instability. This is because resistance to elastic/plastic strain or hardness together with work hardening determine the contact area between an abrasive particle and the material.

It is important to realise that during abrasion, material deformation and eventually fracture occur at very localised areas under conditions of high hydrostatic pressure (leading to high surface strains) and high strain rates. Thus material properties that can delay the early satisfaction of fracture criteria, *e.g.* increased ductility or strain accomodating ability (as often described by the strain hardening exponent), increased macro- and micro-toughness or properties that reduce the amount of material removed or cold worked, *e.g.* high hardness, will increase the abrasion performance of a material.

One would expect 927 with its almost exclusively martensitic microstructure (thin film of retained aus-

TABLE 4. The abrasion and abrasion–corrosion wear indices as a function of prior cold working. Also shown is the percentage volume loss attributed to corrosion in the RWRL 1 and RWRL 2 tests

% Cold Wk.	Hardness (HV30)	RAR	RWRL 1	Wt. % Corr.	RWRL 2	Wt. % Corr.
M.S.	193.6	1.00	1.00	88.29	1.00	41.13
0	490	1.63	10.44	10.71	2.45	1.95
7	473	1.59	10.21	9.63	2.34	2.66
20	511	1.52	10.82	7.41	2.44	1.57
21	–	1.56	10.63	5.45	2.45	1.26
29	550	1.56	10.63	9.09	2.73	1.95
43	558	1.52	9.74	10.00	2.75	1.52
48	583	1.48	11.03	5.66	2.80	1.59
57	585	1.64	10.08	3.45	2.75	1.18

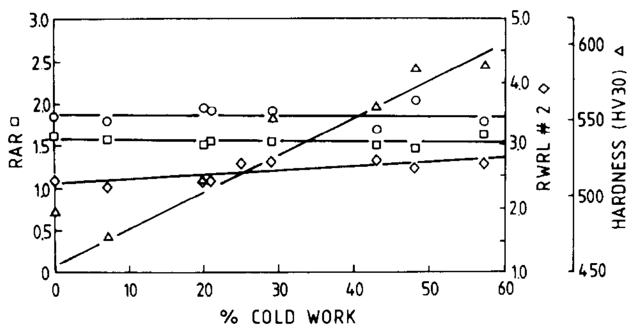

Fig. 9. Abrasion and abrasion–corrosion indices as a function of prior cold working.

TABLE 5. Strain hardening exponents of the development alloys [13]

Material	Strain hardening exponent	Microstructure
927	0.13	Martensitic
AISI 304	0.45–0.50	Austenitic
3CR12	0.173	Ferritic/Martensitic

tenite around laths [9] and its associated high dislocation densities [Honeycombe [16] reports initial dislocation densities for lath martensite of 10^{11} to 10^{12} cm^{-2}] is likely to afford the material little strain-hardening ability. The strain hardening exponent derived from a tensile test is given in Table 5 together with those of other development alloys. Although 927 has a low strain hardening exponent and hence low fracture strain, this detrimental influence is offset by the high yield and tensile strengths which result in a smaller portion of the abrasive strikes [16] being able to produce plastic deformation and resultant surface fracture events. This coupled with the reduction in abrasive penetration owing to the high hardness means that less material will be removed during abrasion. Also the thin layer of inter-lath retained austenite is reported to increase the toughness of the alloy (Charpy (TL) = 80–90 J at 20 °C) and is likely to play a role in reducing the frequency of surface fracture events (in comparison with a fully martensitic structure).

The alloy passivates in the simulated mine water solution and exhibits similarly small corrosion volume losses to the higher chromium containing steels like 3CR12, 12/10, despite its nominal chromium content of 9%. These good corrosion properties are linked to the effect of the 2% Ni and especially the 0.7% Mo on suppressing pitting tendencies [9, 18]. The good dry abrasion properties coupled with the low corrosion volume losses in the RWRL 1 and RWRL 2 (46 h and 22 h corrosion respectively) yield very good corrosion–abrasion indices. They are comparable with those of the higher chromium containing steels and in some instances better. Tests performed on the alloy using a more aggressive mine water solution and corrosion periods of five days resulted in the production of a thin, black oxide layer. This oxide layer was found to be chemically stable and robust, properties which are highly desirable for corrosive–abrasive applications.

The existence of numerous small pits in the bottom of wear tracks after 46 h exposure in the RWRL 1 test (Fig. 6) indicates that the passivity is in the process of being lost. A plot of corrosion rate *vs.* time (Fig. 10) shows that the corrosion rate virtually doubles after 80 h, showing that corrosion exposure in excess of the period used in the RWRL 1 test will lead to a significant corrosion contribution to volume loss and considerably poorer abrasion–corrosion performance relative to the higher chromium-containing steels. Barker and Ball [19] illustrated that steels with 8–10% chromium are borderline with respect to adequate corrosion protection in mining environments.

Three categories of corrosion behaviour at an abraded surface were identified [19]. The first is displayed by low alloy steels which exhibit general corrosion immediately after abrasion and have no induction period (Type I). Type II steels have a short induction period prior to localised corrosion. The highly alloyed stainless steels occupy Type III behaviour and passivate rapidly,

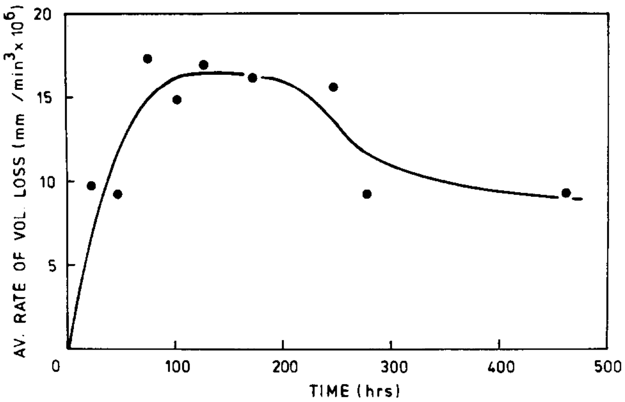

Fig. 10. Rate of corrosion volume loss as a function of time for the alloy.

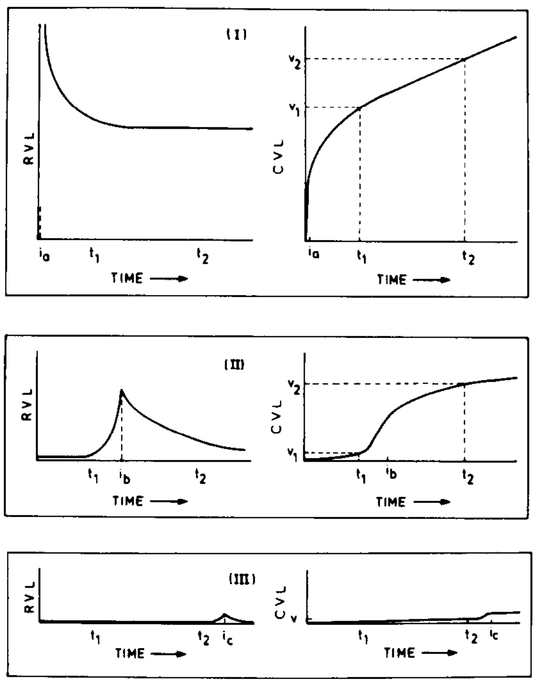

Fig. 11. Schematic representation of the rate of volume loss (RVL) by corrosion + cumulative volume loss (CVL) as a function of time for the three types of steel in the abraded condition [19].

exhibiting a long induction period and lose little material other than by pitting corrosion. The three types are summarised in Fig. 11. They argue that if the frequency of corrosive–abrasive cycles is such that the corrosive activity is interrupted before the induction period of breakdown, then the material will perform well in abrasive–corrosive conditions.

The alloy 927 falls into category II and evidence is such that the frequency of the RWRL 1 test (46 h corrosion) is marginally short of the induction period for breakdown. If the corrosion–abrasion frequency were further decreased, the performance of alloy 927 relative to the higher chromium containing abrasive–corrosive resistant alloys will depreciate.

The apparent non-effect of plate thickness on dry abrasion properties is linked to the negligible influence of as-finished plate thickness on mechanical properties (as reflected by the small 10% change in bulk hardness between the 6 and 25 mm plate). The abrasion–corrosion trend essentially follows that of the dry abrasion because of the small contribution of corrosion to volume loss in both the RWRL 1 and RWRL 2 corrosion–abrasion tests.

Intuitively one would expect prior cold work to exert an influence on dry abrasion resistance because the work hardening of the surface is inherent in the abrasion process. The effect of cold work is to increase the hardness and decrease the ductility of the material. The dilemma is essentially whether the positive effects of increased hardness and strength on the abrasion-resisting properties of the material are offset by the decrease in ductility and associated macro- and micro-toughness.

Barker and Ball [19] report that a 30% increase in hardness produced a 10% decrease in the RAR of mild steel. This is related to the marked effect of cold work on the macro- and micro-mechanical properties of mild steel and hence on abrasion resistance. In the case of the 927 alloy, the RAR remains essentially unaffected by cold work, although there is a 20% hardness difference between underformed and 57% deformed material. This is again explained in terms of the limited work hardening ability of the martensitic microstructure of the alloy and the associated small influence of prior cold working on the micro- and macro-properties of the material relevant to abrasion. Again, minimal corrosion volume loss in the corrosion–abrasion test means that these trends essentially follow the dry abrasion trends.

Overall, these differences in carbon and sulphur content (Table 1) are not large enough to exert a significant influence on corrosion–abrasion behaviour. If one corroded for longer periods than 46 h, differences in sulphur content may become significant. Barker [13] reports that although increasing the carbon content leads to an increase in hardness (interstitial hardening) and a general increase in RAR, the relationship between RAR and percentage carbon is not very strong for the alloyed martensitic steels. High levels of carbon will have an adverse affect on the corrosion resistance via sensitisation and hence on corrosion–abrasion performance.

6. Conclusions

The 927 alloy performs very well under the conditions of the laboratory dry abrasion and abrasion–corrosion tests. It outperforms several higher chromium-containing alloys. The favourable dry abrasion performance can be credited to the good combination of high strength and toughness given to the alloy by its microstructure of lath martensite and thin layer of interlath retained austenite. The good corrosion–abrasion performance is attributed to the excellent corrosion resistance imparted to the alloy by the Ni and Mo additions and the effect of these elements on suppressing localised corrosion and pitting.

Plate thickness and prior cold working exhibit little influence on the mechanical and corrosion properties of the material and hence on the abrasion–corrosion performance.

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