

APPLICATION OF VN ALLOY IN HIGH SPEED STEEL

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APPLICATION OF VN ALLOY IN HIGH SPEED STEEL

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1 INTRODUCTION

Research into low alloy high speed (HS) steels has been carried out in several countries with the objectives of improving the utilization of alloying elements and of reducing costs, thus improving the economics of use of such steels. In Russia, the annual production of low alloy HS steels accounts for some 16% of the total production of HS steels. In Sweden, half of the HS steel saw blades for machines are made of steel D953. In the principal HS steel plant in Austria, the annual production of low alloy HS steel constitutes 6.3% of the total production^[1]. Some low alloy HS steels, such as W3Mo2Cr4VSi and W4Mo3Cr4VSi have also been developed and applied in China.

Adding nitrogen is one of the approaches for improving the structure of HS steels. Nitrogen added to these steels can increase both their hardness and thermal stability in the tempered condition. Up to 800ppm nitrogen can be absorbed in HS steels, where it can form complex carbo-nitrides with vanadium, tungsten and molybdenum. The propensity of vanadium to combine with nitrogen is greater than that of either tungsten or molybdenum.

The most important criterion in the development of modern HS steels has been their ability to utilize lower levels of expensive alloying additions, while maintaining or improving their cutting properties. To this end, the potential reductions in molybdenum and tungsten levels which may be achieved by increasing the nitrogen level of vanadium-containing HS steels could have significant technical and economic benefits. It is the purpose of this paper to explore these benefits.

2 EXPERIMENTAL MATERIAL AND METHOD

2.1 Materials

On the basis of the compositions of current HS steels, five high nitrogen steels with varying carbon, nitrogen and vanadium contents were used, while tungsten and molybdenum contents were kept to low levels. Three lower nitrogen steels were also used for comparison.

The liquid steel chemical compositions are given as Steels 1# and 2# and Steels G1 to G4 in Table 1. These steels were melted in an intermediate frequency induction furnace with a capacity of 150kg. The molten steels were cast as 50kg ingots and 70kg (80mm diameter) electrode bars. The electrode bars were remelted as 60kg ingots, of diameter 160mm, in an electroslag remelting furnace. The ingots were annealed at 880°C for 4h before forging. The 50kg ingots and the ESR ingots were forged into 20mm diameter bars or 30mmx30mm square bars, respectively. For forging, the steels were slowly reheated to 500°C, held for one hour and they were then heated to 1150°C-1200°C and held for 1.5 hours. Forging commenced at 1150°C and was completed above a temperature of 880°C. After forging, the bars were annealed for one hour at 830°C and cooled at less than 50°C/hr until 500°C was reached, after which they were air cooled to room temperature. All specimens were machined from the annealed bars.

Two industrial melts were also made in an intermediate frequency induction furnace with a capacity of one ton. Their chemical compositions are shown as steel V3 and steel V3N in Table 1. The steels were cast as 160mm diameter consumable electrode bar and then were remelted in an ESR furnace. The electrode bars were annealed and forged to 50mmx50mm square billet and then they were annealed again. The square billets were rolled into 18mm diameter bars and annealed. Specimens were machined from the annealed bars. The schedule for preparation of all of these specimens is shown, diagrammatically, in Fig. 1

Table 1 Chemical composition of the steels, wt%

Steel	C	Si	Mn	P	S	Cr	Mo	W	V	N	Alloyed
1#	0.70	0.18	0.29	0.028	0.023	4.37	3.34	2.00	2.35	0.1040	VN
2#	0.65	0.19	0.28	0.027	0.023	4.45	3.28	1.95	3.78	0.1660	VN
G1	0.76	0.44	0.40	0.025	0.020	3.86	2.98	1.97	1.73	0.0084	FeV
G2	1.34	0.45	0.46	0.036	0.022	3.96	2.93	1.91	3.42	0.0120	FeV
G3	0.83	0.35	0.44	0.026	0.021	3.99	2.97	2.02	1.74	0.1290	VN
G4	1.28	0.34	0.36	0.030	0.023	4.19	3.08	2.05	3.40	0.1720	VN
V3	1.23	0.22	0.31	0.027	0.007	3.82	2.81	11.29	2.57	0.0150	FeV
V3N	1.23	0.20	0.33	0.026	0.004	3.86	3.00	11.48	2.61	0.0820	VN

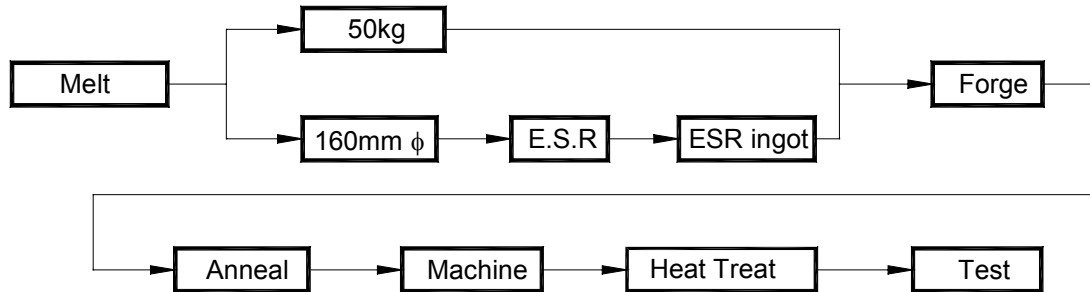


Fig. 1 Specimen preparation (diagrammatic).

2.2 Experimental Methods

2.2.1 Heat Treatment

Specimens, of dimensions 20mm diameter×15mm long, were heated in an electric resistance furnace, quenched in an oil bath (~30°C) and tempered in an electric resistance furnace. The heat treatment schedule was as follows:-

Quenching: The reheating temperature was varied between 1150°C and 1280°C and the heating time was approximately 40 to 60 min. After heating the steels were quenched into an oil bath.

Tempering: Tempering was carried out at 560°C for 1 hour and repeated three times.

Red hardness heat treatment^[2]: For determination of Red hardness heating was carried out at 625°C for 1 hour and repeated four times.

2.2.2 Microstructural Examination

The microstructures of the steels were examined by optical microscopy and scanning electron microscopy (SEM). Prior austenite grain size was measured in the steels after quenching. The amount of retained austenite was determined using a D/max-rc X-ray diffractometer (XRD). Chemical electrolysis was used to extract carbo-nitrides from the specimens. The morphology and composition of the carbo-nitrides were examined and analysed using a JSM-500LV (SEM), equipped with an INCA energy dispersive spectroscopy (EDS). The crystal structure and amount of carbo-nitrides in the steels were determined by means of X-ray diffraction and chemical phase analysis.

2.3 Mechanical Properties

2.3.1 Hardness Test

Hardness testing was carried out on the specimens after the heat treatments according to Chinese standard GB/T230-91 (Rockwell Hardness Test Procedure for Metals). After heat treatment the specimen surface was ground to remove any oxidized or decarburised layer and was then polished. Before testing the machine was calibrated following the procedure in GB/T230-91. The mean hardness value was determined from three indentations which were made on the central area of the specimen.

2.3.2 Tensile and Impact Tests

Tensile and bending tests were carried out on the steels after heat treatment, according to Chinese Standards of GB/T228—2002 and GB/T14452-1993, respectively.

The strength in bending and reduction of area of the steels were determined using a universal tester.

2.3.3 Cutting Test

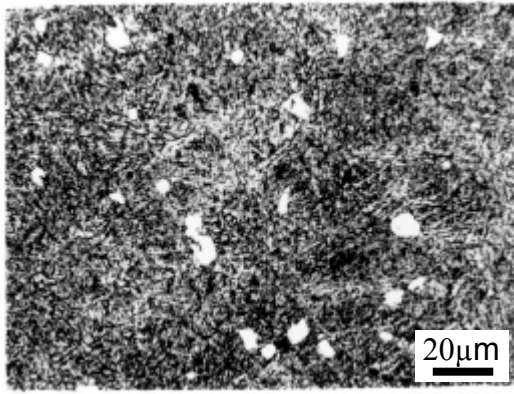
Machinability tests were carried out to evaluate the performance of cutting tools manufactured from Steels G1, G3, V3 and V3N.

3 RESULTS AND DISCUSSION

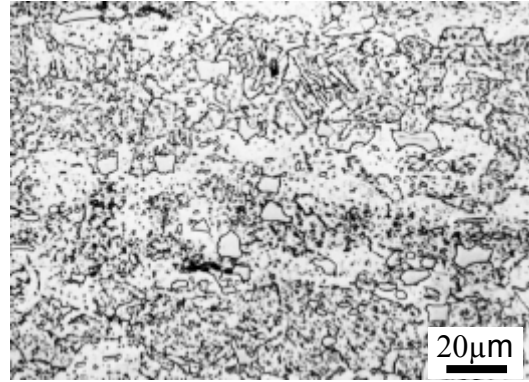
3.1 Microstructures

3.1.1 As-Annealed

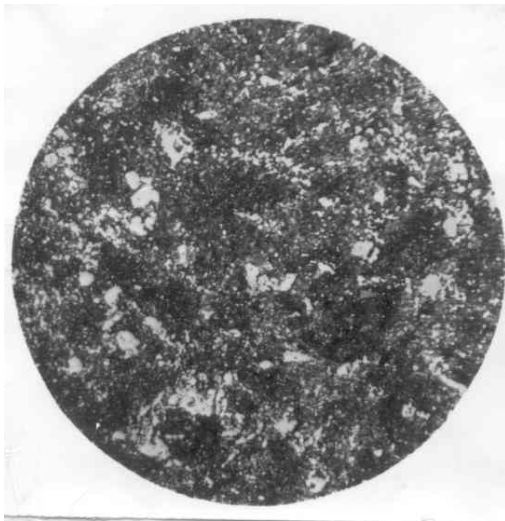
Typical as-annealed microstructures of Steels 1#, 2#, V3 and V3N are shown in Fig. 2. The matrix was sorbite while the white massive objects were primary eutectic carbo-nitrides and the fine particles were secondary carbo-nitrides.



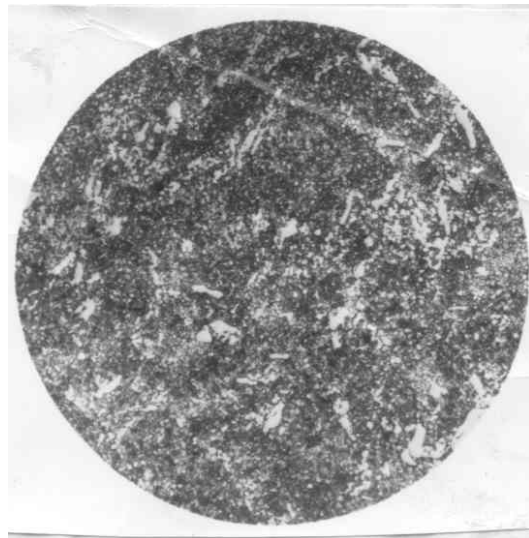
Steel 1#



Steel 2#

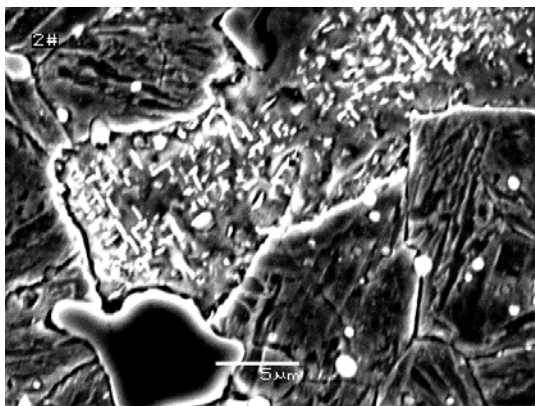


Steel V3N

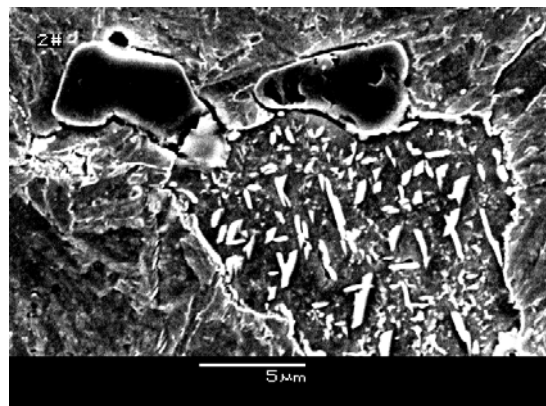


Steel V3

Fig. 2 The as-annealed microstructures of steels 1#,2#,V3N and V3.



(a) As-quenched



(b) As-quenched and tempered

Fig 3 SEM micrograph showing the black structure in Steel 2#.

3.1.2 As-Quenched and Tempered

Typical microstructures were obtained in the steels after quenching and after tempering when using certain austenitising temperatures. However, a large amount of black structure (Fig. 3a),

which is not normally seen in high speed steels after quenching, was also found in Steel 2#[4]. The black structure was identified as carbide phase, which had precipitated in δ -ferrite during quenching and which remained in the steel after tempering (Fig. 3b). It is suggested that the very low carbon content in Steel 2# resulted in the formation of the black structure.

3.1.3 Austenite Grain Size

Typical as-quenched microstructures, in which the variation in austenite grain size of Steels 1# and 2# and Steels G1-G4 can be seen, are shown in Fig. 4. The microstructure in the quenched condition consisted of martensite, carbo-nitrides and retained austenite. The relationships between the austenitising temperature and the austenite grain size for these steels are shown in Fig. 5. The austenite grain size depended largely on the composition and austenitising temperature and it increased with the austenitising temperature.

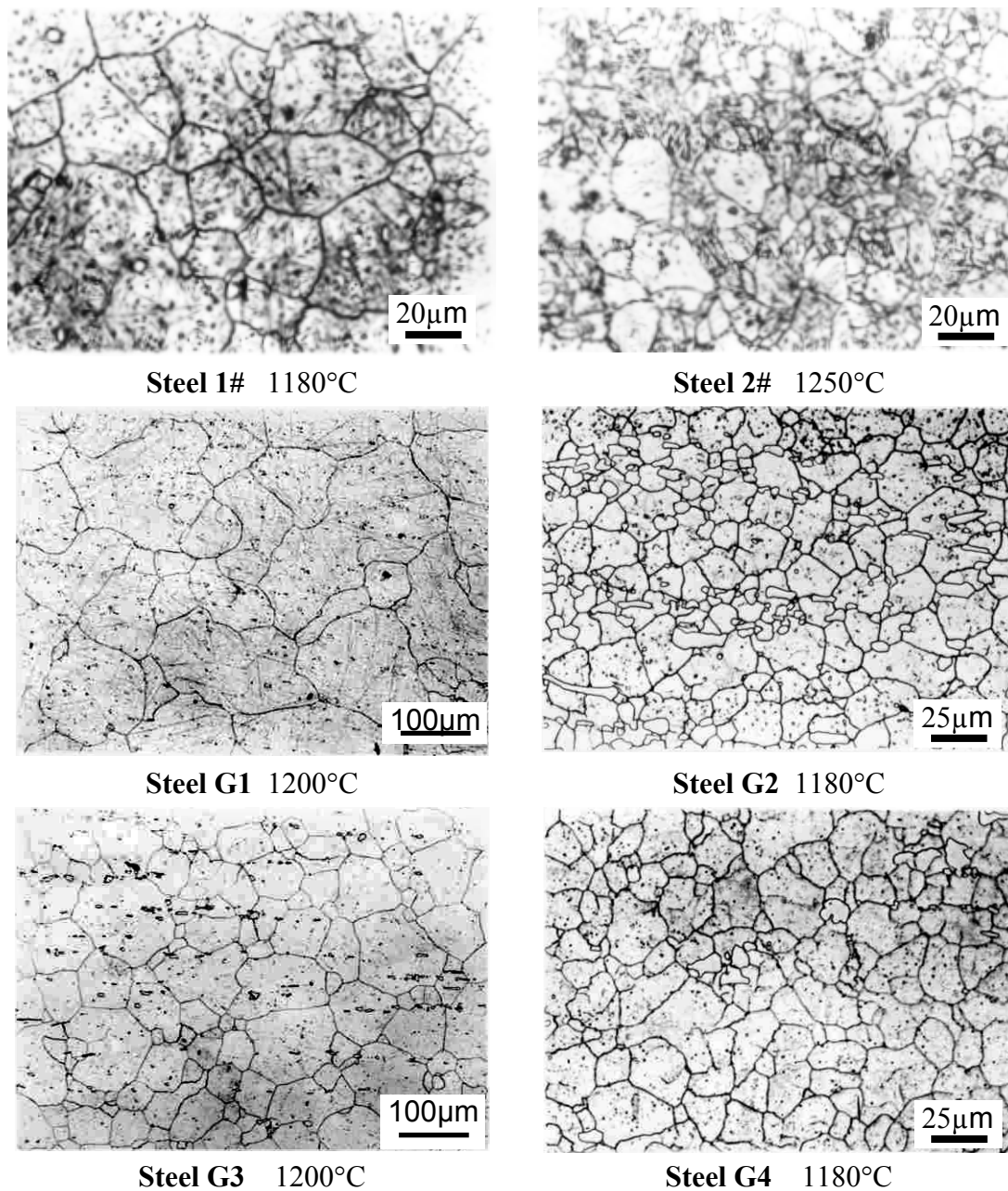


Fig. 4 The austenite grain structure at typical austenitising temperatures for each steel.

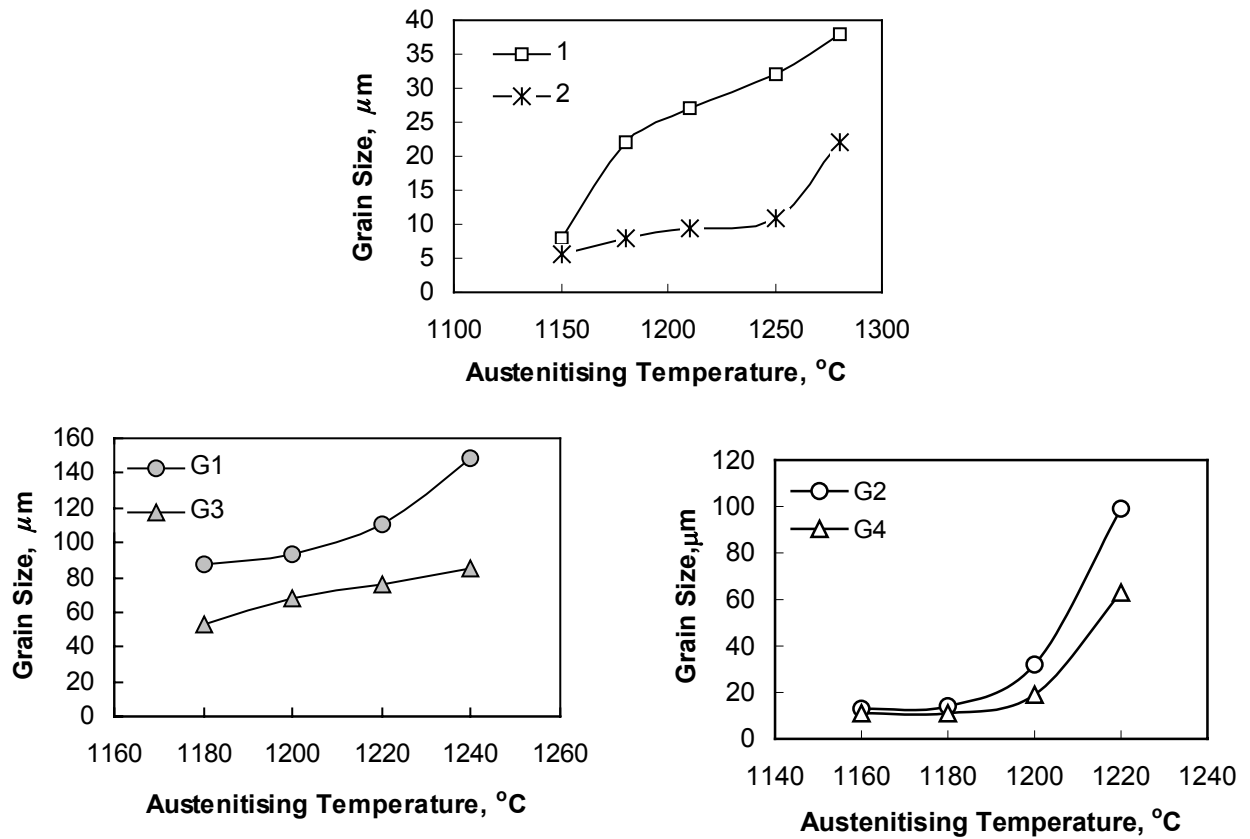


Fig. 5 The effect of austenitising temperature on the austenite grain size.

The austenite grain size of Steels G2 and G4 with, high carbon and high vanadium levels, was obviously finer than that of Steels G1 and G3, with lower carbon and vanadium levels. Moreover, at equal carbon and vanadium content, the austenite grain size of Steels G3 and G4, with higher nitrogen level, was finer than that of Steels G1 and G2, with lower nitrogen content. This clearly demonstrates that increasing the nitrogen level had a significant effect on austenite grain refinement.

It is well known that, in addition to factors such as austenitising temperature and holding time, the austenite grain size will be affected by the composition of the steel as well as the number, distribution, size and shape of any undissolved carbides. Austenite grain growth is dependent upon the step-by-step movement of grain boundaries. Second phase particles, including carbo-nitrides and non-metallic inclusions can pin the grain boundary, thus inhibiting grain growth^[3].

High carbon, high vanadium, steels have more MC-type carbides than low carbon, low vanadium, steels. As a result of their higher dissolution temperatures and better resistance to agglomeration, MC carbides can inhibit austenite grain growth more effectively than M_6C and $M_{23}C_6$ carbides, leading to a finer grain size.

Increasing the nitrogen content of complex carbo-nitrides normally results in an increase in their dissolution temperature. Particles rich in nitrogen are, therefore, more effective at inhibiting austenite grain growth than particles lean in nitrogen. At the same austenitising temperature, the austenite grain size of the steels with higher nitrogen levels was finer than it

was with lower nitrogen levels. Thus, a higher austenitising temperature can be adopted for higher nitrogen steels, without necessarily resulting in austenite grain growth.

3.1.4 Retained Austenite

The effects of austenitising temperature on the amount of retained austenite in the as-quenched microstructure are shown in Fig. 6.

The level of retained austenite in Steel 1# was approximately 10% and was less than 5% for Steel 2#. The amount of retained austenite was high in steels G1 to G4 (8%-46%) and there was a tendency for it to increase with both carbon content and austenitising temperature. The amount of austenite retained in the steels with higher nitrogen levels (8%-35%) was much less than that (26-46%) in the lower nitrogen steels. This suggests that HS steels with an elevated nitrogen level will require fewer tempering cycles to reduce the level of retained austenite than would be required in lower nitrogen steels. However, in Steels 1# and 2#, of lower carbon content, the stability of the austenite was reduced and austenite decomposition was accelerated, leading to a lower level of retained austenite. Consequently, single tempering of low carbon, HS steels could be adopted for such steels.

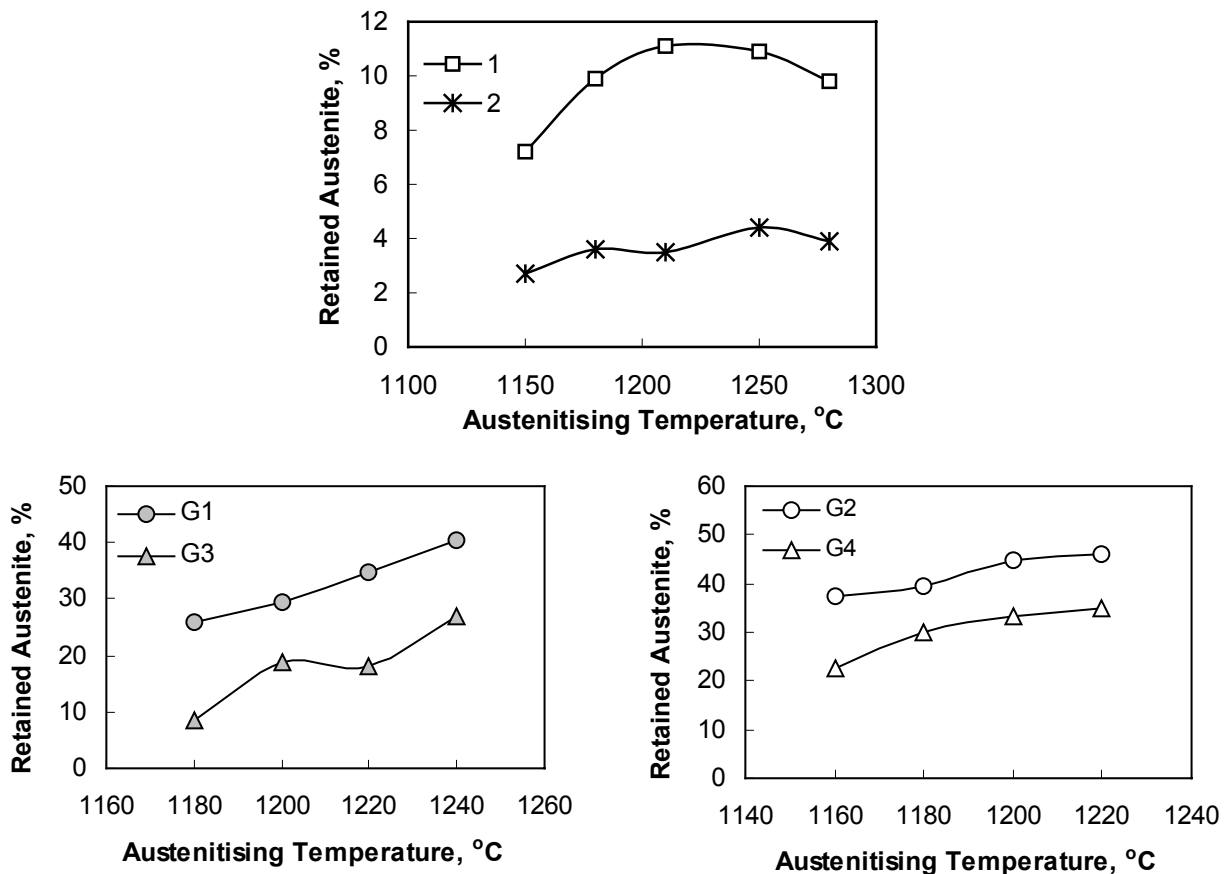
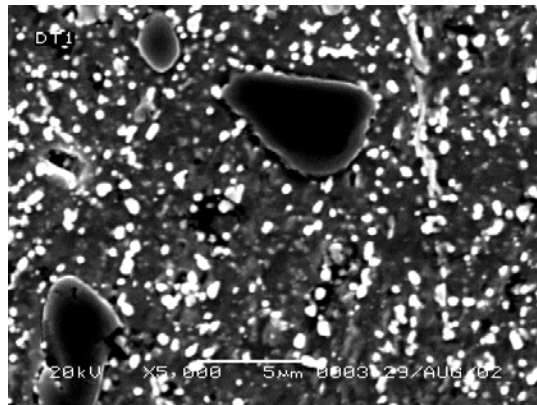


Fig. 6 The effect of austenitising temperature on the amount of retained austenite.

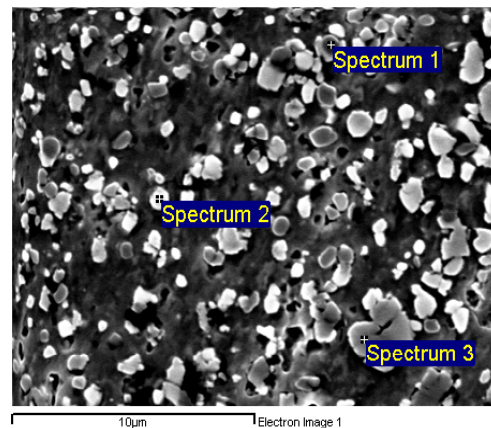
3.2 Carbo-nitrides

3.2.1 Morphology

SEM micrographs of carbo-nitrides, in the as-annealed condition, are shown in Fig. 7. SEM and EDS analysis indicates that the particles consisted of a few large M(C,N) particles, which contained mainly vanadium, along with a large number of smaller particles which contained higher levels of Cr, Mo and W and some V. Furthermore, some of the alloying elements were dissolved in the matrix with the level of Cr being highest, followed some way behind by V and then W. There was almost no Mo dissolved in the matrix.

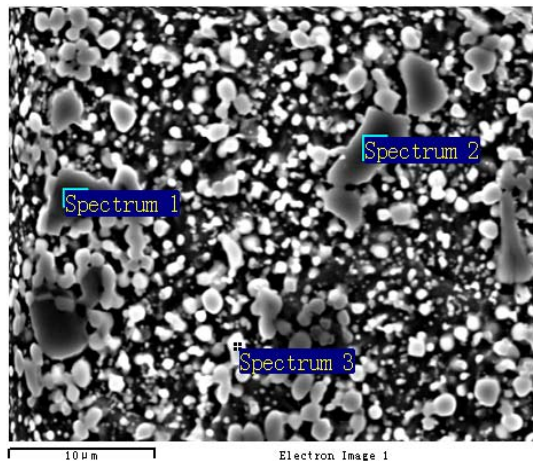


Steel 1#



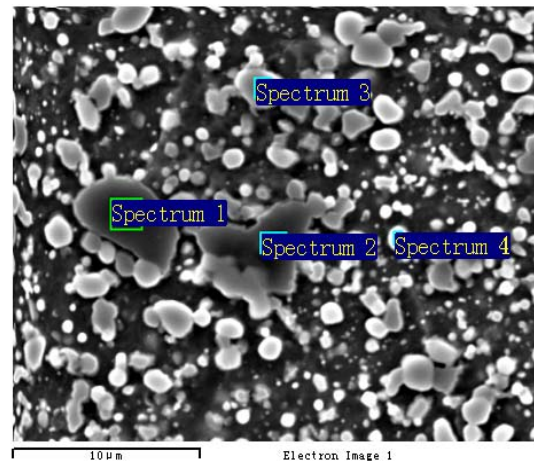
Steel 2#

Carbide	Composition of carbide, wt%				
	V	Cr	Fe	Mo	W
1	45.21	3.49	44.89	3.18	3.23
2	6.24	4.27	63.86	12.38	13.25
3	4.56	3.95	72.10	9.19	10.20



Steel V3

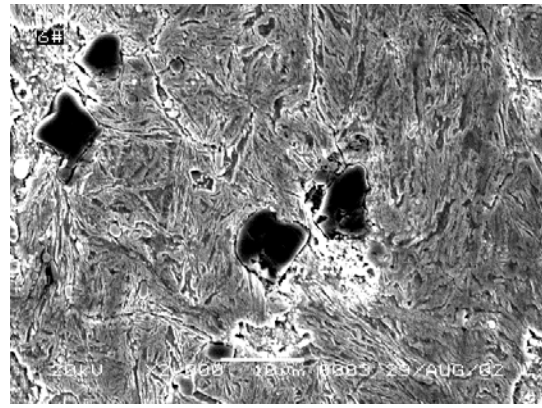
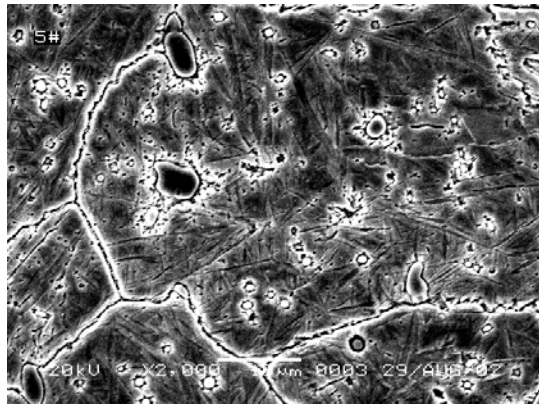
Carbide	Composition of carbide, wt%				
	V	Cr	Fe	Mo	W
1	43.33	3.59	7.97	6.94	38.17
2	49.16	3.85	5.93	6.17	34.89
3	2.46	4.02	33.91	10.44	49.18



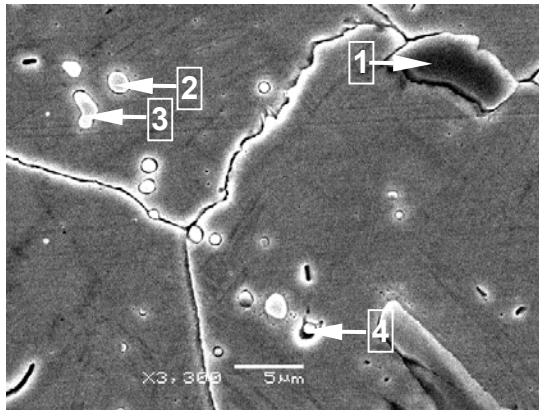
Steel V3N

Carbide	Composition of carbide, wt%				
	V	Cr	Fe	Mo	W
1	59.38	5.30	3.88	5.92	25.52
2	39.05	4.75	28.51	5.53	22.14
3	2.52	3.88	35.31	10.21	48.08
4	1.80	4.43	49.04	6.92	37.80

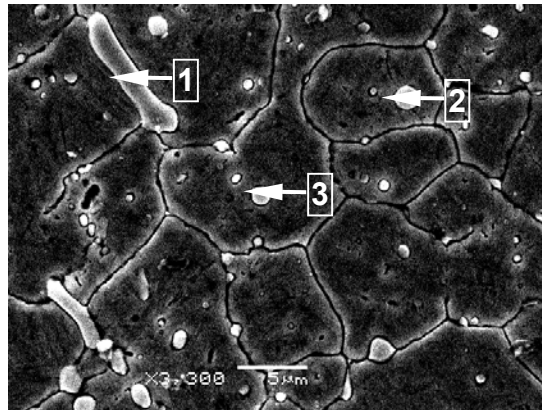
Fig. 7 The morphology and chemical composition of the as-annealed carbo-nitrides



Steel 1#



Steel 2#

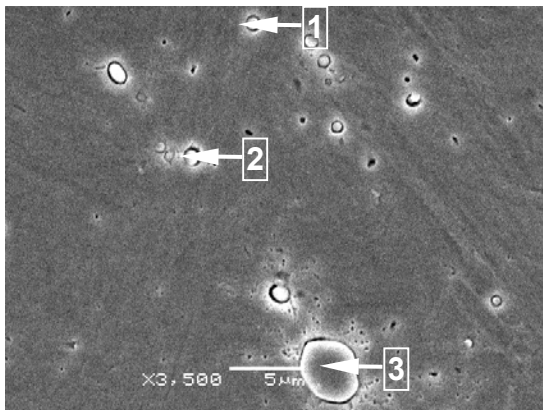


Steel G1

Carbide	Composition of carbide, wt%				
	V	Cr	Fe	Mo	W
1	50.93	5.14	25.85	9.87	8.21
2	37.80	5.14	42.27	7.96	6.83
3	50.28	6.02	27.45	9.81	6.43
4	31.72	5.57	48.69	7.08	6.95

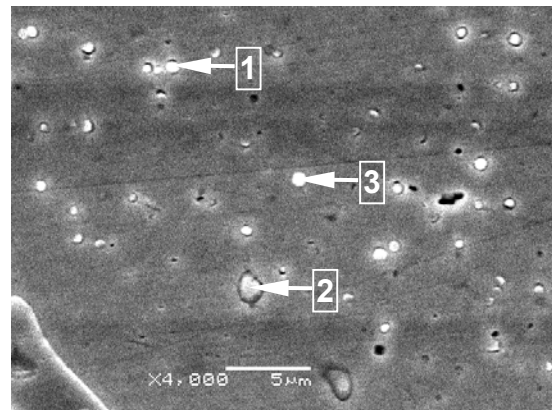
Steel G2

Carbide	Composition of carbide, wt%				
	V	Cr	Fe	Mo	W
1	59.80	4.27	8.36	12.72	14.85
2	28.48	5.14	34.42	17.05	14.90
3	32.48	4.28	42.52	9.11	11.61



Steel G3

Carbide	Composition of carbide, wt%				
	V	Cr	Fe	Mo	W
1	23.87	4.18	52.13	9.47	10.35
2	27.69	6.05	39.36	14.23	12.67
3	50.90	3.99	/	23.78	21.33



Steel G4

Carbide	Composition of carbide, wt%				
	V	Cr	Fe	Mo	W
1	26.18	4.11	57.13	6.62	5.96
2	62.65	5.19	9.25	12.81	10.10
3	16.05	3.33	69.70	6.58	4.33

Fig. 8 The morphology and composition of the carbo-nitrides in the as-quenched condition.

SEM micrographs of carbo-nitrides in the as-quenched condition are shown in Fig. 8. Most of the particles had dissolved in the austenite at the austenitising temperature. Only some large, refractory, carbo-nitrides remained out of solution. The compositions of the undissolved particles were similar to those observed in the annealed condition. The particles remaining out of solution in the higher nitrogen steels appeared to be finer than those in the lower nitrogen steels.

The numbers of carbo-nitrides were greatest in the as-annealed condition, where they could achieve close to full precipitation. Most of the carbo-nitrides were dissolved in austenite at the austenitising temperature and their numbers were significantly reduced after quenching. After tempering they increased again as the secondary carbides precipitated from martensite and retained austenite.

3.2.2. Precipitate Types

The types of carbo-nitrides observed in Steels 1# and 2# are given in Table 2.

Table 2 The phases present in Steels1# and 2#.

Steel	State	Phase	Lattice Constant	Crystal System
1#	Annealed	M ₆ C	a ₀ =11.04~11.05	F
		VC	a ₀ =4.14~4.15	F
		V(CN)	a ₀ =4.15~4.16	F
		M ₂₃ C ₆	a ₀ =10.58~10.60	F
		CrN	a ₀ =4.14	F
		(CrFeMoW) ₇	a ₀ =13.982, c ₀ =4.506	H
		C ₃	a ₀ =5.224	C
		α-MnS	a ₀ =4.163	C
	Quenched	V(CN)	a ₀ =4.11~4.12	F
		VC	a ₀ =4.15~4.16	F
		V ₂ (CN)	a ₀ =2.910~2.980, c ₀ =4.598~4.708	CH
		α-MnS	a ₀ =5.224	C
		CrN	a ₀ =4.14	F
2#	Annealed	VC	a ₀ =4.15~4.16	F
		V(CN)	a ₀ =4.12~4.13	F
		V ₂ (CN)	a ₀ =2.910~2.980, c ₀ =4.598~4.708	CH
		α-MnS	a ₀ =5.224	C
		CrN	a ₀ =4.14	F
	Quenched	M ₆ C	a ₀ =11.04~11.05	F
		VC	a ₀ =4.14~4.15	F
		V(CN)	a ₀ =4.11~4.12	F
		CrN	a ₀ =4.14	F
		α-MnS	a ₀ =5.224	C
	Tempered	V(CN)	a ₀ =4.11~4.12	F
		VC	a ₀ =4.15~4.16	F
		α-MnS	a ₀ =5.224	C
		CrN	a ₀ =4.14	F
		VC	a ₀ =4.16~4.17	F
		V(CN)	a ₀ =4.13~4.14	F
		α-MnS	a ₀ =5.224	C
		CrN	a ₀ =4.14	F

Note: F = face-centered cubic lattice system, C = cubic system, H = hexagonal system, CH = close-packed hexagonal system.

After annealing the range of carbo-nitride types present was very complex. The main particles were M_6C , VC, V(CN), CrN, Mo_2N , $(CrFeMoW)_7C_3$ and $M_{23}C_6$. After quenching and quenching and tempering only VC, V(CN), $V_2(CN)$ and CrN remained. M_6C , M_7C_3 and $M_{23}C_6$ dissolved in austenite during reheating for quenching and these precipitates were not, subsequently, observed. On the other hand, vanadium carbo-nitrides, as well as chromium nitride, have lower solubility in austenite and remained in the steels after quenching and after tempering. These nitrogen-containing phases play an important role in improving both wear resistance and high temperature hardness. Furthermore, MnS was observed in both steels.

The lattice constant of V(CN) was observed to be slightly less than that of VC, indicating that nitrogen dissolved in VC reduces the lattice constant of VC. The vanadium-containing precipitate in the high nitrogen steel was practically all V(CN). It is also worth noting that the lattice parameters of CrN and VC are almost identical.

Finally, $V_2(CN)$ was observed after tempering in Steel 1#, but not in Steel 2#. $V_2(CN)$ has a smaller lattice constant than V(CN) and it belongs to the close-packed hexagonal system which is different from face-centered cubic V(CN).

3.2.3. Amount of Carbo-nitride

For Steels 1# and 2#, the amounts of individual alloying element combined as carbo-nitride, as well as the sum of all of the alloying elements combined as carbo-nitride, are shown in Table 3.

Table 3 The amount of alloying element combined as carbo-nitride, wt%.

Steel	Cond.	Total	Fe	Cr	Mn	W	Mo	V	N	[N] solution
1#	A	11.633	2.928	1.433	0.050	2.227	2.610	2.290	0.095	0.009
	Q	3.701	0.117	0.234	0.021	0.831	0.664	1.749	0.085	0.019
	T	5.265	0.145	0.585	0.021	1.124	1.392	1.909	0.089	0.015
2#	A	8.269	1.117	0.304	0.025	1.852	1.581	3.248	0.142	0.024
	Q	4.266	0.054	0.122	0.024	0.943	0.364	2.625	0.134	0.032
	T	4.551	0.064	0.147	0.025	0.955	0.371	2.851	0.138	0.028

Notes: A = annealed, Q = quenched, T = quenched and tempered.

In both steels, the annealed condition had the greatest amount of alloying element combined as carbo-nitrides of all the three conditions. The amount decreased after quenching and increased again after tempering. After annealing or after tempering it was greater in Steel 1# than in Steel 2#. However, the trend was similar in both steels and this behaviour is typical of HS steel.

The changes in composition of the carbo-nitrides were more marked in the cases of W, Mo and Cr than they were for V. Furthermore, the change in the W and Mo content of the precipitates between the quenched and quenched and tempered conditions was more marked than the change in Cr or V content. This suggests that W and Mo played a more significant role in secondary hardening, especially in the case of Steel1#.

The nitrogen content of the carbo-nitrides was stable and did not change much with heat treatment. 80-90% of the total nitrogen in the steel precipitated as carbo-nitrides and the remainder was in solution. It is worth noting that, as the nitrogen content increased, both the amount of nitrogen in the carbo-nitrides and that in solution also increased.

3.3 Hardness

The effects of austenitising temperature and of tempering on Rockwell hardness, for all of the steels, are shown Fig. 9 (a-h).

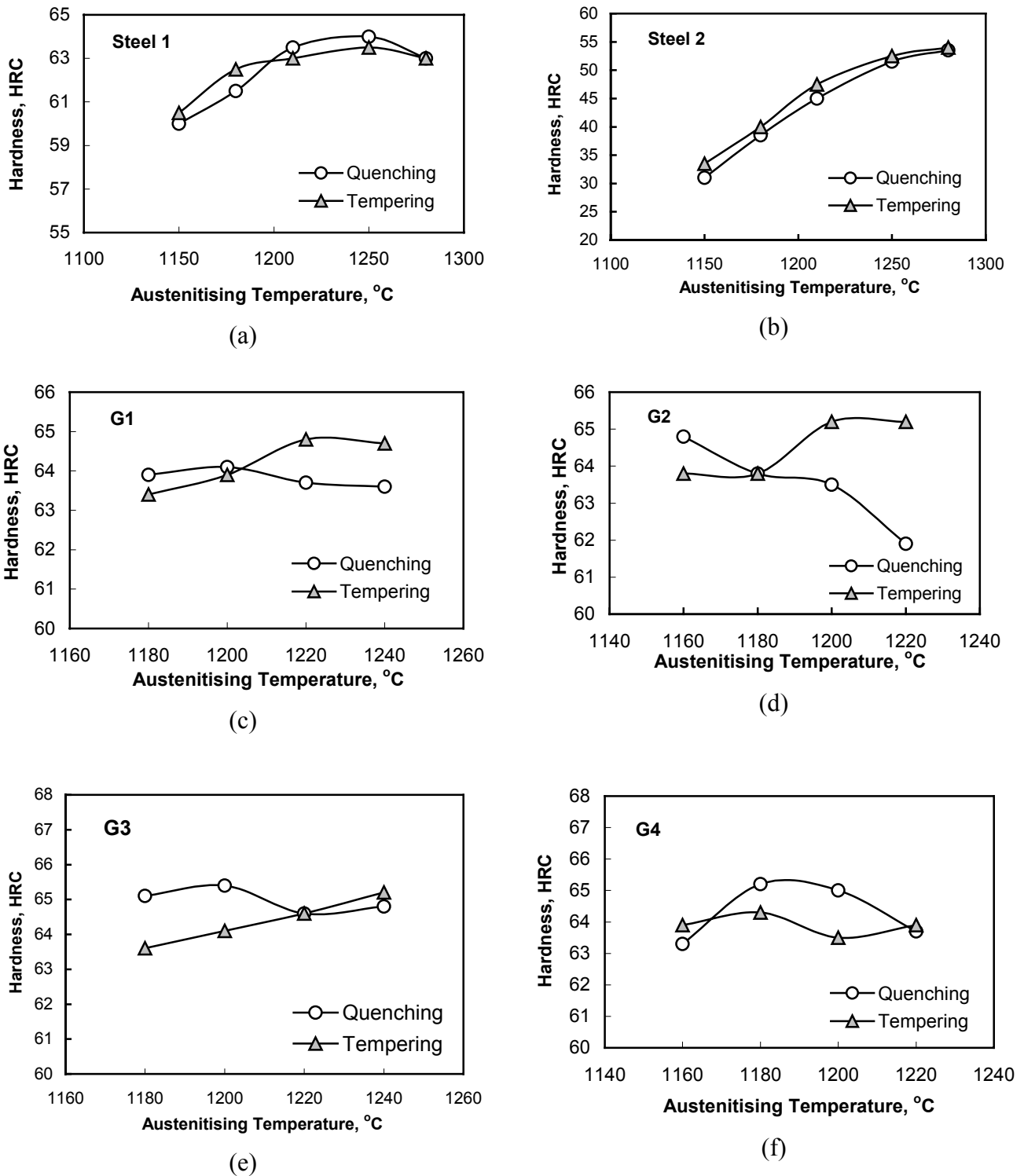


Fig. 9 (a-f) The effects of austenitising temperature and of tempering on the hardness of the steels.

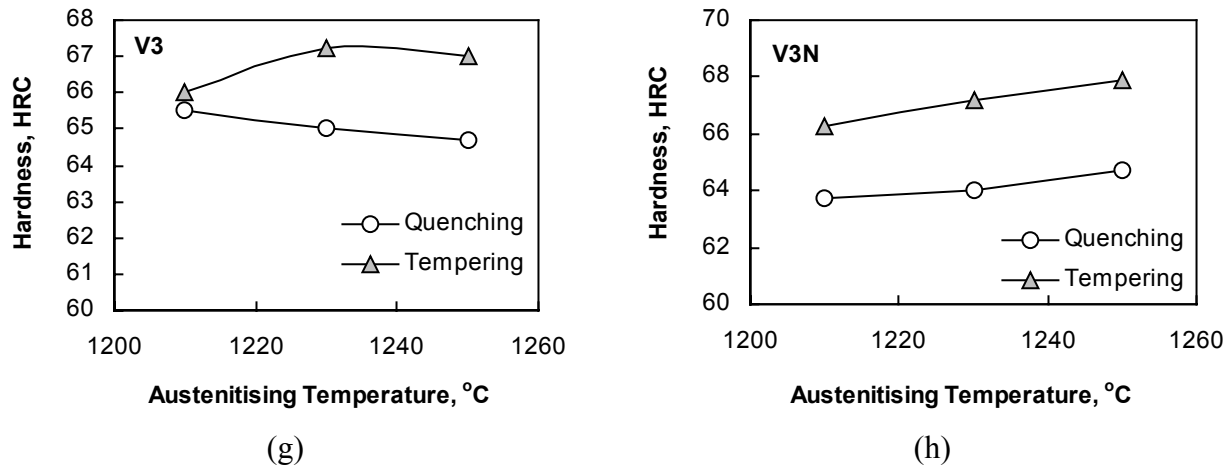


Fig. 9 (g-h) The effects of austenitising temperature and of tempering on the hardness of the steels.

For Steel 1# (Fig. 9(a)), it can be seen that, in both the quenched and quenched and tempered conditions, the hardness exhibited a peak at approximately 1250°C. The as-quenched hardness was above 60HRC for all of the austenitising temperatures examined and the maximum as-quenched hardness for this steel was 64HRC. After tempering, the hardness was also over 60HRC and the maximum was 63.5HRC.

Conversely, the hardness of Steel 2# (Fig. 9(b)) was relatively low and its maximum as-quenched hardness was only 53.5HRC, while its hardness after tempering was 54HRC. The hardness curves for Steel 2# were different from those of typical HS steel and are probably related to its chemical composition. Generally, HS steel has high carbon content and increasing the austenitising temperature increases the amounts of carbon and alloying element in solution, in austenite. This results in a rapid increase in the stability of the austenite, with the consequence that a large volume fraction of austenite (30%-40%) is normally retained after quenching in HS steels. Steel 2# had low carbon content and its high nitrogen content would ensure that dissolution of carbo-nitrides was displaced to higher temperatures. Both of these led to the low level (<5%) of retained austenite observed in this steel after quenching and to the absence of a hardness peak after quenching and tempering.

From Fig. 9 (c-f), it can be seen that the quenched hardness of Steels G1 and G3 reached a peak at an austenitising temperature of approximately 1200°C. Further increase in austenitising temperature resulted in a decrease in quenched hardness. For Steels G2 and G4 the quenched hardness decreased with increase in austenitising temperature and there were no hardness peaks such as those observed in Steels G1 and G3. Furthermore, the quenched hardness was around 1.5HRC higher in the high nitrogen steels (G3 and G4) than it was in the low nitrogen steels (G1 and G2) and the quenched hardness decreased more slowly with increase in austenitising temperature in the high nitrogen steels. Thus, when quenching from a temperature below 1200°C the quenched hardness of Steels G3 and G4 was maintained above 65HRC, while that of Steels G1 and G2 was typically around 64HRC.

The appearance of a hardness peak in the as-quenched condition is probably due to a change in the amount of austenite retained after quenching^[2]. Increasing the austenitising temperature normally results in an increase in the degree of dissolution of the carbo-nitrides, giving rise to an increase in both the carbon and alloying elements in solution. This will be reflected in an

increase in the carbon content of martensite forming on quenching and will cause an increase in hardness. However, increasing the carbon content and the amount of alloying elements in solution will tend to stabilize austenite and lower the Ms temperature and, in turn, result in an increase in the amount of retained austenite, which reduces the as-quenched hardness. When the retained austenite reaches a critical amount, i.e. the effect of retained austenite preponderates over the effect of martensite, the as-quenched hardness will begin to decrease, resulting in a peak in hardness. This is clearly reflected in the as-quenched hardness of Steels G1 and G3. However, the amount of retained austenite was significantly higher in Steels G2 and G4 and this dominated the as-quenched hardness, resulting in a decrease in hardness with increasing austenitising temperature and absence of a hardness peak. Additionally, the level of retained austenite was higher in Steel G2 (low N) than it was in Steel G4 (high N), consequently, the as-quenched hardness decreased more rapidly in the former steel. A similar, but less obvious, effect was observed in Steels G1 and G3.

The hardness after tempering also tended to increase with increasing austenitising temperature and hardness peaks were observed in some of the steels at some temperatures. Steels G1 and G2 exhibited reasonably well defined peaks after tempering but those in the other steels were less distinct. When quenched from low temperature, the hardness after tempering tended to be lower than that after quenching. Conversely, when quenched from higher temperature the reverse was the case and this was most obvious in the low nitrogen steels, G1 and G2. Additionally, in steels G1 and G2 the maximum hardness after tempering was higher than that after quenching while in steels G3 and G4 it was lower. These results can be related to the high levels of retained austenite observed in steels G1-G4 and to the transformation of this austenite, with the accompanying precipitation of carbo-nitrides, during tempering

Fig. 9 (g) and (h) show the effect of austenitising temperature on quenched and quenched and tempered hardness for steels V3 and V3N. As the austenitising temperature increased from 1210°C to 1250°C, the quenched hardness of steel V3 decreased by about 1HRC while that of steel V3N increased by a similar amount. After tempering, the hardness of both steels increased by 1-2 HRC with the above increase in austenitising temperature. Furthermore, the tempered hardness of both steels V3 and V3N lay in the range 66-68HRC, some 1-4HRC above those in the quenched condition with, if anything, steel V3N being a little harder than steel V3. Consequently, it would appear that the austenitising temperature of steel V3N could be 10°C-20°C higher than that of steel V3.

One parameter which is known to significantly affect the hardness of HS steels is the available carbon, defined as^[5]:-

$$C_p \text{ (available carbon)} = 0.033W + 0.063Mo + 0.06Cr + 0.2V.$$

This compares with the total carbon, defined as:-

$$C_s = C + N$$

A comparison of these parameters for Steels 1# and 2# is given in table 4

Table 4 Carbon-equivalent of Steels 1# and 2#

Steel	C _p	C _s	ΔC=C _p -C _s	A=C _s /C _p
1#	1.06	0.70+0.104=0.804	0.26	0.75
2#	1.40	0.65+0.166=0.816	0.58	0.58

From the results, it can be seen that the values of ΔC and A of Steel 1# are close to the levels of AISI M2. Its peak hardness after tempering reached 63.5HRC which is near the level of AISI M3. This suggests that, at low carbon level, along with lower levels of tungsten and molybdenum, high hardness can be obtained by increasing nitrogen level.

For Steel 2# carbon the content was lower and the vanadium and nitrogen contents were much higher than those of Steel 1#, resulting in a higher value of ΔC and lower value of A. Consequently, δ -Fe was observed in Steel 2# and this decreased the hardness of the steel.

3.4 Properties

Some of the properties of the steels are summarised in Table 4.

Table 4 The properties of the steels.

Type	Reduction of Area, %	Tempered Hardness HRC	Red Hardness HRC	Bending Strength MPa
1#	/	63.5	56.5	/
2#	/	54.0	46.0	/
G1	/	61.8	51.6	2540
G2	/	64.5	53.5	1560
G3	/	63.4	52.3	2300
V3N	16.4	67.9	63.9	2280
V3	21.1	67.2	61.8	3130

The Red hardness for both Steels 1# and 2# was not high, the maximum values being 56.5HRC and 46HRC, respectively.

Steel G1 and G3, with lower carbon and vanadium, had higher bending strength than the Steel G2, with higher carbon and vanadium. Comparing Steels G1 and G3, there appeared to be no obvious effect of nitrogen on bending strength, the performance of these two steels being almost identical.

The tempered hardness and red hardness of Steel V3N were both higher than those of Steel V3. However, the reduction of area and bending strength of Steel V3N were lower than those observed in Steel V3. It would appear, therefore, that increasing the nitrogen content of HS steel could possibly increase the hardness and wear resistance, albeit with a loss of some ductility and bending strength.

3.5 Machinability

The results of cutting tests carried out using tools manufactured from Steels G1, G3, V3N and V3 are shown in Table 5.

Under the same cutting procedure, Steels G3 and V3N (high N) had a lower level of wear than Steels G1 and V3 (low N). The results can be explained by the fact that the carbides in the high nitrogen steels were finer and more stable. SEM and DES analyses also indicated that the particles remaining in the steels after quenching and tempering were mainly V(C,N) and CrN. Thus, increasing the nitrogen content of HS steel can improve the hardness and service life of cutting tools.

Table 5 The results of the cutting tests.

Steel	Material Machined	Procedure	Cutting Parameter			Cutting Time min	Amount of Wear mm
			v	t	s		
G1	Steel 45, 224HB	Turning	42	2	0.276	35	0.35
G3			42	2	0.276	35	0.16
V3N	Steel GCr15, 37HRC		13	0.5	0.15	30	0.30
V3			13	0.5	0.15	30	0.40

Note: v-cutting speed, m/min, t-cutting depth, mm, s-feed rate, mm/r.

4 CONCLUSIONS

- (1) High speed steels with high vanadium and high nitrogen levels can be made by vanadium-nitrogen alloying. The recovery of vanadium was stable and greater than 80%. However, the recovery of nitrogen was affected by the amount of nitrogen added and the melting technique. The maximum of recovery of nitrogen was about 70%, but steels containing 0.17%N could still be manufactured.
- (2) High speed steel with high nitrogen content has a fine austenite grain size. Consequently, the temperature for quenching can be raised and the amount of retained austenite is lower than that of low nitrogen steels.
- (3) Nitrogen can form vanadium carbo-nitride and chromium nitride in the steel. These precipitates were very stable and remained in the final product and were beneficial to wear resistance and red hardness. Nitrogen dissolved in VC reduces the lattice constant of VC.
- (4) The nitrogen content in the carbo-nitrides was stable and did not vary with heat treatment conditions. The carbo-nitrides contained 80-90% of the total nitrogen in the steels.
- (5) With an appropriate vanadium level, along with lower tungsten and molybdenum levels, high nitrogen HS steel can exhibit high hardness, which approaches that of conventional HS steel, even at lower carbon levels
- (6) Under the same cutting conditions, the resistance to wear of vanadium-nitrogen alloyed HS steels was better than that of similar HS steels with a low nitrogen level, albeit at a loss of some ductility and bending strength.

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