

The physical metallurgy of vanadium steels

J. H. Woodhead

(University of Sheffield, England)

Vanadium is a very strong ferrite stabilizer with high solubility in pure iron. However, it has a great affinity for carbon and nitrogen so that vanadium steels are largely controlled by solubility relationships. Vanadium in solution in austenite has a direct effect on both hardenability and recrystallization during hot working. The amount of vanadium in solution at elevated temperatures controls the extent of precipitation during cooling to give ferrite-pearlite structures or during the tempering of martensite and bainite. Because the solubilities of vanadium carbide and vanadium nitride are very different, the relative amounts of carbon and nitrogen in vanadium steels are important. So is the presence of other elements which may compete for nitrogen, such as aluminum.

Vanadium (atomic weight, 50.942) has a b.c.c. structure with lattice parameter 3.0231 Å at 25°C. This value relates to a material of estimated purity 99.998%. Higher values of the lattice parameter are frequently quoted but these relate to material of lower purity. The melting point has not been established with high accuracy but is close to 1900°C (3452°F).

As might be expected from the crystallographic data, vanadium and iron are completely miscible at high temperatures. The equilibrium diagram for the iron-vanadium system is well established; details may be found in the Metals Handbook. The most interesting feature of the diagram, in relation to the use of vanadium in steels, is the existence of a γ -loop which limits the existence of the γ -phase to alloys containing less than 1.5% vanadium. Clearly vanadium is an α -stabilizer. However, additions of up to 0.2% vanadium lower the A_3 temperature, i.e. the γ -loop displays a minimum similar to that observed in the iron-chromium system; thus, in small quantities, vanadium acts as a γ -stabilizer.

Interdiffusion coefficients in dilute solutions of vanadium in both α and γ iron are reasonably well established. The most reliable data are those of Haworth and Morton^{2,3} who obtained the results shown in Table I for the parameters in the equation:

$$D = D_0 \exp\left[-\frac{Q}{RT}\right], \quad \text{Equation 1}$$

Table I. Interdiffusion data for vanadium in α and γ iron*

	Q , kJ/mol	D_0 , mm ² /sec
α	242 ± 2	398 ± 75
γ	293 ± 4	365 ± 140

* The errors quoted are one standard deviation

from their own electron probe microanalysis data and selected data from earlier work. These results indicate that the diffusion of vanadium is relatively rapid, particularly in γ -iron as shown by the relative diffusion coefficients in Table II.

The interdiffusion coefficient for iron-vanadium in austenite is markedly increased by the presence of carbon in the alloys. The diffusion coefficient is increased by a factor of 5 by the addition of 1% carbon. The effect of smaller amounts of carbon may be obtained easily, as the logarithm of the multiplying factor is proportional to the carbon concentration^{2,3}.

In the ternary iron-carbon-vanadium system the major feature, relevant to steels, of the equilibrium diagram is a marked reduction in the solubility of carbon in austenite as a result of the ready formation of vanadium carbide, VC.

CARBIDES AND NITRIDES

The high affinity of vanadium for carbon and nitrogen and the low solubility of the carbide and nitride are responsible for the major effects of vanadium on the properties of steels. Both the carbide and the nitride are responsible for the major effects of vanadium on the properties of steels. Both the carbide and the nitride have f.c.c. NaCl type structures, so ideally they should be representable as VC and VN. However both compounds have defect lattices in which some of the sites for carbon or nitrogen are unfilled. In both cases the stoichiometry can range from about $VX_{0.75}$ to VX , where X is either carbon or nitrogen. In steels, the composition of the carbide tends towards the bottom of the range and, consequently, it is often referred to as V_4C_3 ; it is probably best to symbolise the compound as VC_{1-x} but in this paper VC will be used so as to emphasise its crystallography. The nitride in steels tends towards the upper limit of the range and is therefore generally described as VN. The two compounds show complete mutual solubility so that carbonitrides VC_xN_y of variable composition, but with the restriction that $x+y$ falls in the range 0.75 to 1, are found in steels.

The lattice parameters of the various compounds have been established by Brauer and Schnell.^{1,4} The lattice parameter of the carbide varies from 4.1285 Å for $VC_{0.75}$ to 4.1686 Å for $VC_{0.92}$; that of the nitride from 4.066 Å for $VN_{0.72}$ to 4.1398 Å for VN. As would be expected, the carbonitrides show parameters anywhere between the lowest and highest of the values given above, depending on the

Table II. Diffusion coefficients relative to self-diffusion for various elements in iron

	V	Cr	Mn	Fe	Ni
γ at 1000°C (1830°F)	2.7	2.1	1.8	1	0.3
α at 800°C (1470°F)	2.0	1.7	2.3	1	0.9

Table III. Solubility product data for vanadium carbide and vanadium

	P	Q
VC _{0.75} in austenite ⁵	11.33	17660
VC _{0.75} in ferrite ⁶	13.02	21500
VN in austenite ⁷	5.23	16280
VN in austenite ⁸ with 1.5% Mn	8.24	19180
VN in ferrite ⁷	5.64	18030

composition of the compound. Inspection of these lattice parameter values shows clearly that it is quite impossible to categorize a compound as a carbide or a nitride from selected area electron diffraction photographs.

The solubilities of the compounds in both austenite and ferrite may be expressed by the solubility product relationship which may be written, for a compound AB_n, as:

$$1_n[A][B]^n = P - \frac{Q}{T} \quad \text{Equation 2}$$

where P and Q are constants for any pair of solutes in a given solvent and T is the absolute temperature.

For the carbide, assuming the stoichiometry VC_{0.75} and the nitride assuming the stoichiometry VN and, hence, the solubility products are [V][C]^{0.75} and [V][N], the values of the constants P and Q are given in Table III.

The variation of the solubility products with temperature is shown in Fig. 1. The solubility of the carbide is much greater than that of the nitride in both austenite and ferrite. The two sets of data for the solubility of VN in austenite show that manganese increases the solubility of VN in austenite. Fig. 1 makes plain a fact which is not often appreciated, i.e. that the solubility of VN in austenite is less than that of VC in ferrite.

As the solubility product is constant for any one temperature, the solubility data can be plotted as solubility isotherms giving hyperbolic curves when plotted on linear axes. Such diagrams are very useful in predicting matrix compositions and amounts of compounds undissolved at any temperature^{9,10}. If the diagrams are to be used simply for assessing solubility limits it is more convenient to plot them

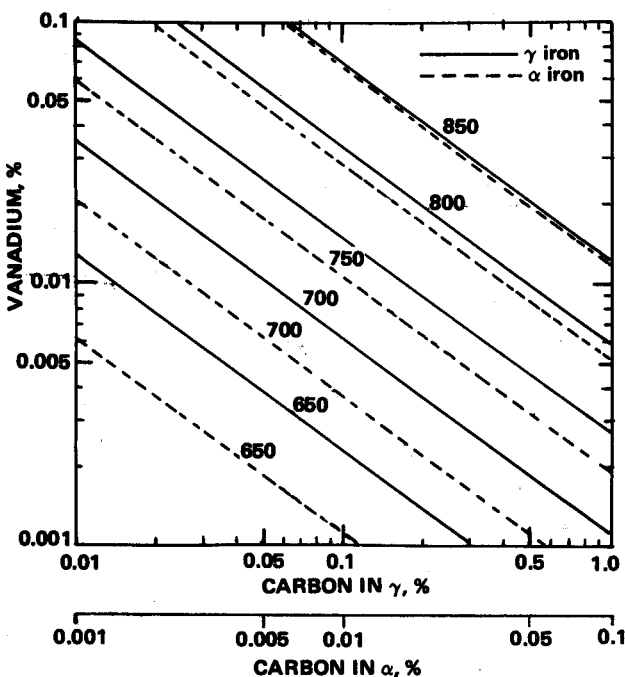


Fig.2 Solubility isotherms for vanadium carbide in α and γ iron

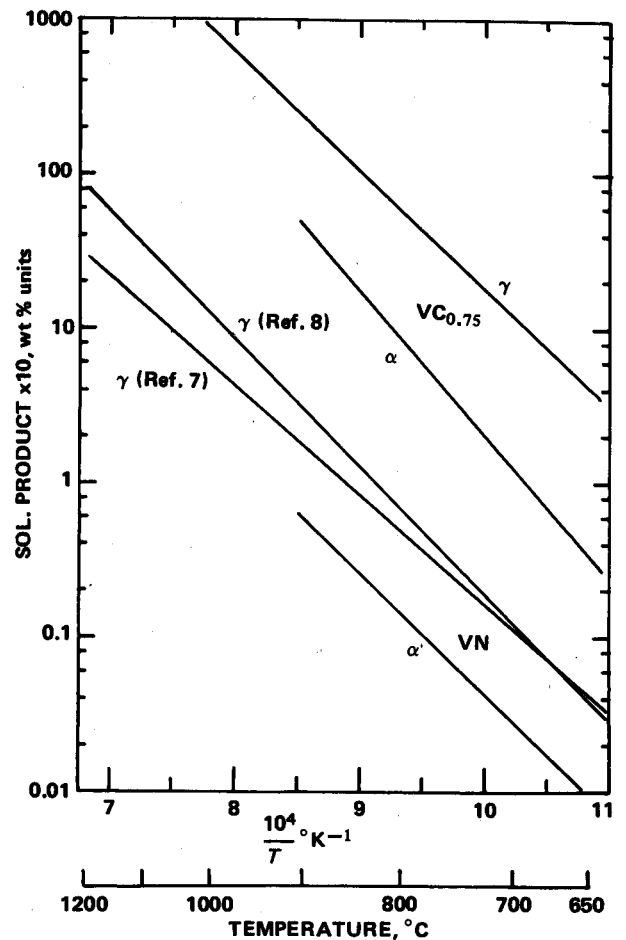


Fig.1 Solubility products for vanadium carbide and vanadium nitride in α and γ iron

with logarithmic scales. The isotherms then become straight lines and in the case of a compound such as VN, only one decade need be plotted as multiplication of one axis by 10 merely implies a division of the other axis by 10; this, of

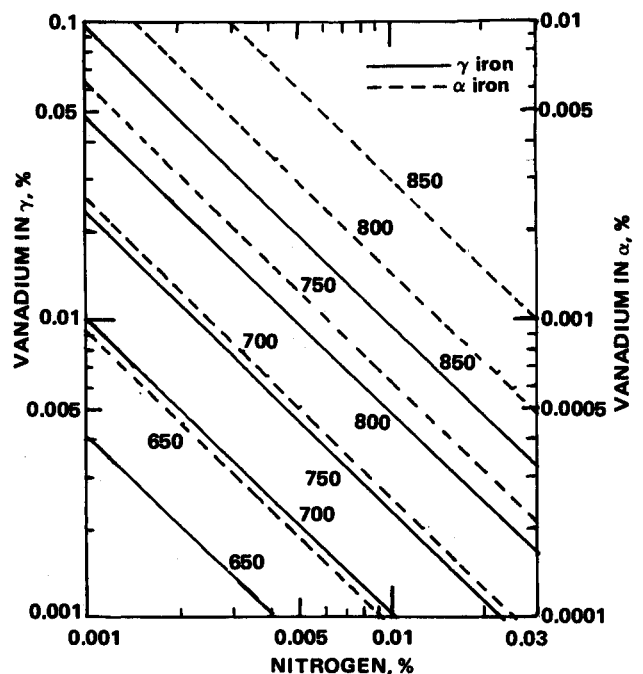


Fig.3 Solubility isotherms for vanadium nitride in α and γ iron

course, does not apply to such stoichiometry as $VC_{0.75}$. Diagrams of this type are shown in Figs. 2 and 3.

Although this seminar is primarily concerned with steels which are strengthened by the direct precipitation of carbonitrides during cooling, it should not be forgotten that vanadium is important in secondary hardening steels, particularly those for use at high temperatures. Secondary hardening may result from the formation of VC but in many cases the influence of vanadium is indirect. In chromium-molybdenum steels secondary hardening is achieved by the precipitation of M_2C but, in service at high temperatures or during longer tempering, this compound gives way to other carbides such as $M_{23}C_6$, and M_6C which are not so effective in strengthening the steel. Vanadium is soluble to some extent in M_2C and has the effect of stabilizing this compound. Similarly in super 13% chromium steels small additions of vanadium stabilise M_2X which is, again, the most effective secondary hardening compound. The stabilities of the various carbides which can form in steels have been reviewed elsewhere.¹¹

HARDENABILITY

The first study of the effect of vanadium on hardenability was made by Zimmerman et al¹² who determined TTT diagrams for eutectoid steels containing 0.27% vanadium.

Their work demonstrated the importance of austenitizing temperature and showed that, provided the austenitizing temperature is high enough, vanadium induces a marked retardation in the formation of pearlite.

The effect of austenitizing temperature is of fundamental importance on interpreting any data relating to the influence of vanadium on hardenability. Only if the temperature is high enough to ensure complete solution of vanadium carbonitrides can one expect to observe the true effect of vanadium. It should be remembered that, at such high austenitizing temperatures and in the absence of precipitate particles, the austenite grain size will become large and this in itself will result in an increase in hardenability. At lower austenitizing temperatures undissolved carbonitride particles will tend to retard austenite grain growth, with a consequent reduction of hardenability. In addition, the undissolved particles can assist the nucleation of high temperature decomposition products of austenite thus giving a further reduction of hardenability. There is also a reduction in dissolved carbon and this also reduces hardenability. Consequently, at low austenitizing temperatures it is quite possible for a vanadium steel to show lower hardenability than a similar plain carbon steel austenitized at the same temperature.

Although Grossman's method of calculating hardenability is of dubious validity, his multiplying factors do provide a means of comparing the effects of solute elements. He

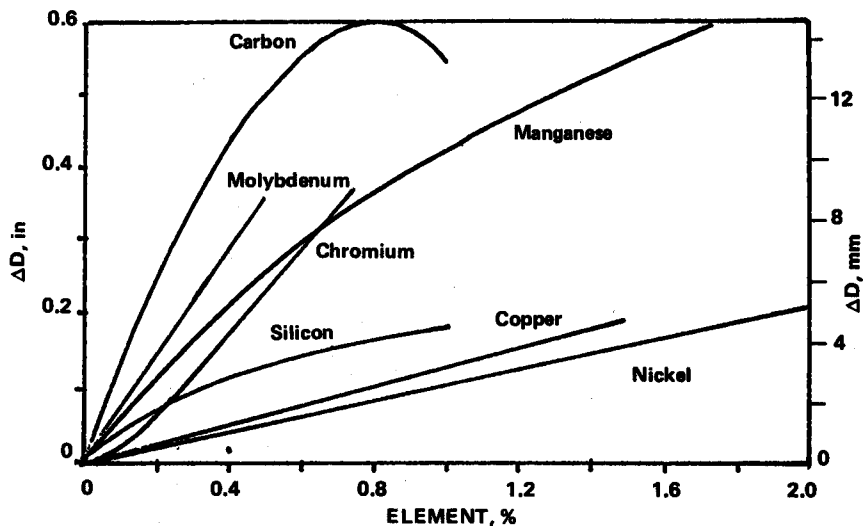


Fig.4(a) Effect of various elements on hardenable diameter (after Grange¹⁴)

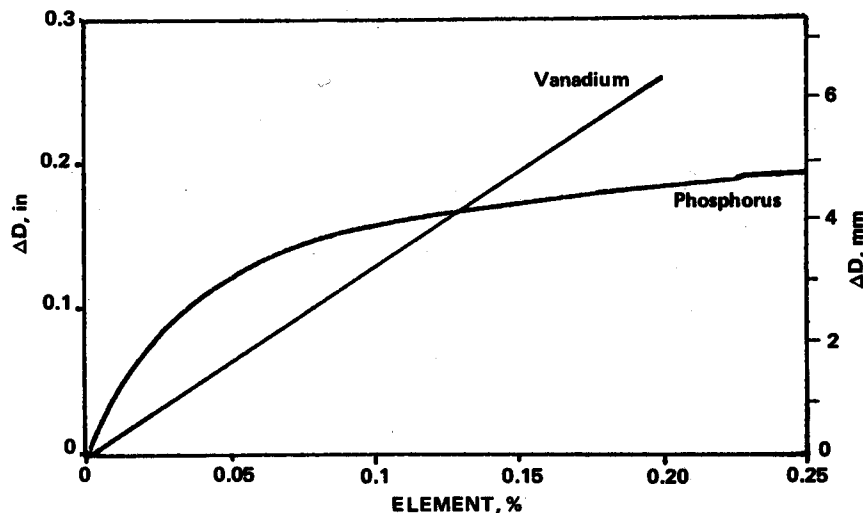


Fig.4(b) Changes in hardenable diameter (after Grange¹⁴)

showed¹³ that small amounts of vanadium enhance hardenability greatly and demonstrated clearly the importance of austenitizing temperature.

Recently Grange¹⁴ has devised the "hot-brine" hardenability test which is particularly suited to lower-carbon steels. His results, which are expressed as the diameters of bars which will give 95% martensite at the centre on water quenching, are of considerable interest. The effect of austenite grain size is shown to be such that the hardenable diameter varies linearly with $d^{1/2}$, where d , is the austenite grain size. The effects of various elements in increasing the hardenable diameter are shown in Fig. 4, for which it can be seen that vanadium is more effective than any of the conventional hardenability additives.

Eldis and Hagel¹⁵ have also shown that vanadium in small amounts is more potent in conferring hardenability than are the more usual elements, although larger additions gave a reduction in hardenability. This was considered to result from a lack of complete solution with consequent loss of both vanadium and carbon from the austenite. By analysis of variance of a factorially designed experiment they were able to reveal significant interactions between elements, thus

quantifying the wellknown synergistic effects of alloy combinations.

Ideally, the effect of vanadium on transformation kinetics could best be shown by means of TTT and CCT diagrams. Many such diagrams are scattered throughout the literature and a number of compilations of diagrams have been published. A brief synopsis of the contents of the major compilations has recently appeared.¹⁶ A systematic assessment of the effect of vanadium is scarcely possible from the available diagrams because of variations in composition, austenitizing temperature and time, sensitivity and reliability of experimental techniques, etc. However, careful examination of the data can leave no doubt that vanadium has a marked effect in increasing hardenability.

VANADIUM IN HSLA STEELS

The factors influencing the properties, transformation characteristics and hot working of HSLA steels have been very well reviewed recently. In particular, the papers by Pickering,¹⁷ Gladman et al¹⁰ and Meyer et al¹⁸ in Micro Alloying 75 give excellent accounts of our present under-

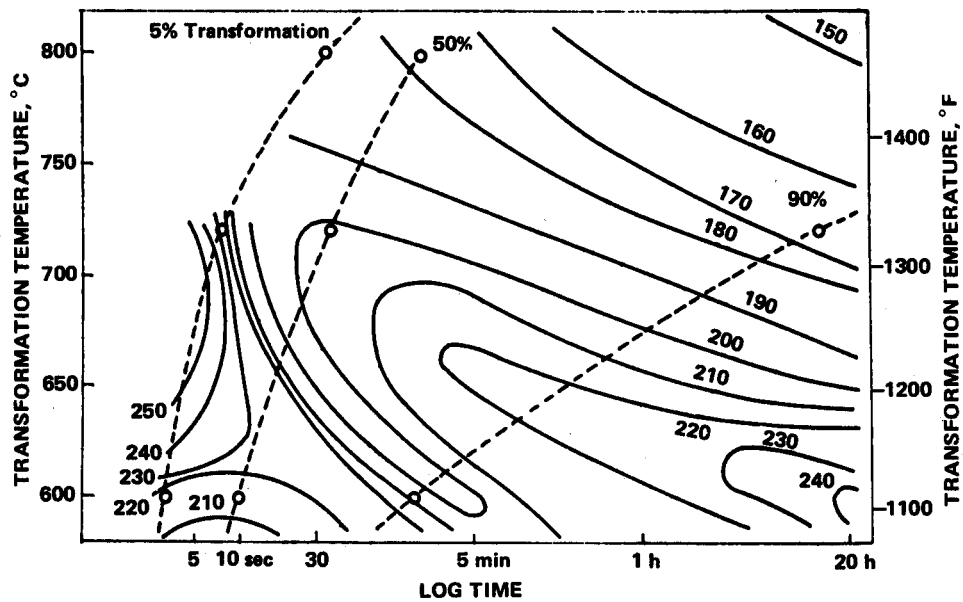


Fig.5 Iso-hardness map for a 0.19% carbon, 0.042% columbium steel²⁴

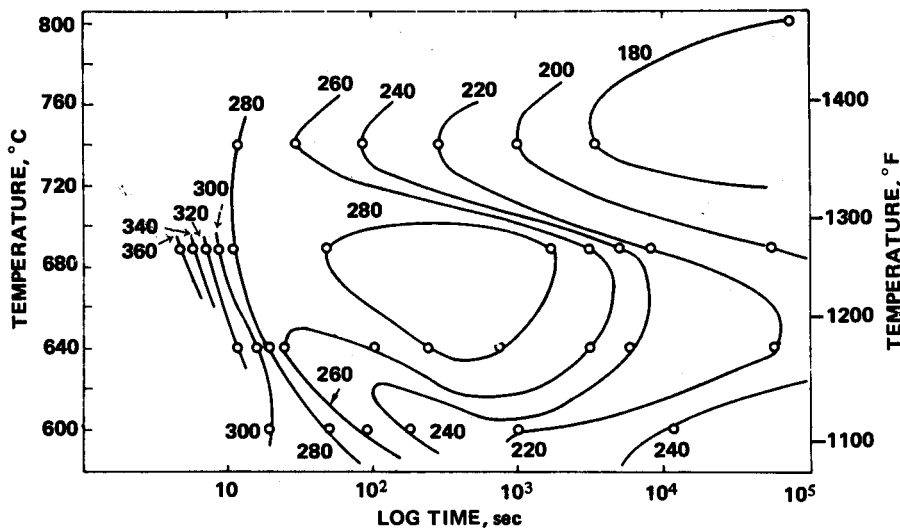


Fig.6 Iso-hardness map for a 0.18% carbon, 0.04% vanadium steel²⁰

standing. Some of the features which are particularly relevant, though by no means exclusively, to the use of vanadium in these steels will be discussed in this section.

As pointed out earlier, the effects of vanadium are, to a considerable extent, dependent on austenitizing temperature. The solubility product relationships show that a steel containing 0.15% and 0.10% vanadium should be free from VC at about 900°C (1650°F). If 0.005% nitrogen is present, the solution temperature is controlled by the solubility characteristics of VN and the solution temperature is about 1000°C (1830°F). Nitrogen contents of 0.015% and 0.020% give solution temperatures of about 1100°C and 1150°C (2010° and 2100°F) respectively. Thus in commercial steels the solution temperature is determined by the nitrogen content; a carbon content of about 0.7% is required to give the same solution temperature for VC in the presence of 0.1% vanadium as is found for VN with only 0.005% nitrogen.

It is reasonable to suppose that, regardless of the composition of precipitates at lower temperatures, their composition steadily approaches VN as temperature rises due to the selective solution of carbon.

The morphology and composition of precipitates and the kinetics of their formation during cooling is still rather obscure. The development of sheets of precipitates parallel to the γ - α grain boundary (interphase precipitates) is well known and was first studied in detail by Gray and Yeo¹⁹ in columbium steels. Webster and Woodhead²⁰ reported this type of precipitation in vanadium steels and subsequently a definitive paper was published by Davenport and Honeycombe.²¹ Honeycombe's group at Cambridge have carried out a great deal of the work on the decomposition of austenite in vanadium and other steels and two very useful reviews have appeared.^{22,23} However, although interphase precipitation is well documented and is frequently observed in both isothermally transformed and continuously cooled HSLA steels, much of the precipitation is general.

It has been shown by Gray et al.,²⁴ using microhardness measurements of ferrite formed during the course of isothermal transformation, that at temperatures below about 700°C (1290°F) there are two precipitation processes in operation. A typical iso-hardness map for a columbium steel is shown in Fig. 5. Similar behaviour is shown in vanadium steels, see Fig. 6.²⁰ The two stage hardening is also observable on continuous cooling, see Fig. 7.

As an aid to understanding the precipitation process it is helpful to consider the supersaturation of the γ and α solid solutions at various temperatures. An attempt has been made to assess the carbon and nitrogen concentrations in the matrix solutions and at the γ - α interface for a 0.15% carbon steel with nitrogen contents of 0.005%, 0.010% and 0.020%. The solubility isotherms shown on Figs. 2 and 3 then give the maximum vanadium which can be in solution at the stated temperatures. The results are given in Tables IV and V.

The nitrogen in solution in austenite at the γ - α interface varies very little from that in the matrix, so the values for soluble vanadium may be regarded as relating both to the matrix and to the interface. Carbon solubility at the interface is little affected by either total carbon or total nitrogen over the range of nitrogen contents considered; the interface solubility of vanadium may thus be considered as practically constant for most HSLA steels.

Carbon in solution in ferrite is not greatly affected by changes in total carbon in the range 0.05% to 0.2% when up to 0.02% nitrogen is present, but it does depend on the nitrogen content. The values given for soluble vanadium in equilibrium with carbon relate to a total nitrogen content of 0.005%. Increasing nitrogen contents cause a reduction in soluble carbon and therefore an increase in the solubility of vanadium with respect to carbon.

It is apparent from Table IV that the supersaturation of vanadium with respect to carbon at the interface is not very different from that relative to nitrogen for a nitrogen content of 0.005% and that even when 0.02% nitrogen is present the difference is not great. In view of the fact that there is much more carbon available at the interface it is reasonable to suppose that interphase precipitates are essentially VC.

In the case of general precipitation in ferrite, Table V shows that, particularly at higher total nitrogen contents the supersaturation of vanadium with respect to nitrogen is much higher than that relative to carbon. In this case the availability of carbon and nitrogen are not very different so one can expect to have a precipitate which approaches VN in composition, but which will, in fact, be a carbonitride with a nitrogen to carbon ratio dependent on the nitrogen content of the steel.

The iso-hardness maps seem to indicate that, at temperatures above about 700°C (1290°F), precipitation is wholly of the interphase type. The precipitate is essentially VC and this averages fairly rapidly to give the steadily falling hardness with increasing time. At temperatures below 700°C, interphase precipitation on a very fine scale gives high initial hardness but, at these temperatures, the interface can move on before precipitation in its vicinity is complete, probably because of the reduced diffusion rate of vanadium at the lower temperatures. The residual vanadium then gives general precipitation of a high nitrogen carbonitride with a typical peaked hardness v time characteristic. During continuous cooling the precipitation will start as the interphase type but, as the start of transformation is likely to be about 750°C (1380°F), general precipitation will soon take over as the temperature falls and a hardness v temperature curve of the type shown in Fig. 7 will be obtained.

The possibility of precipitation in austenite has not, so far, been considered. The solubility product relationships indicate that austenite is supersaturated with respect to both VC and VN at temperatures above the A_3 temperature.

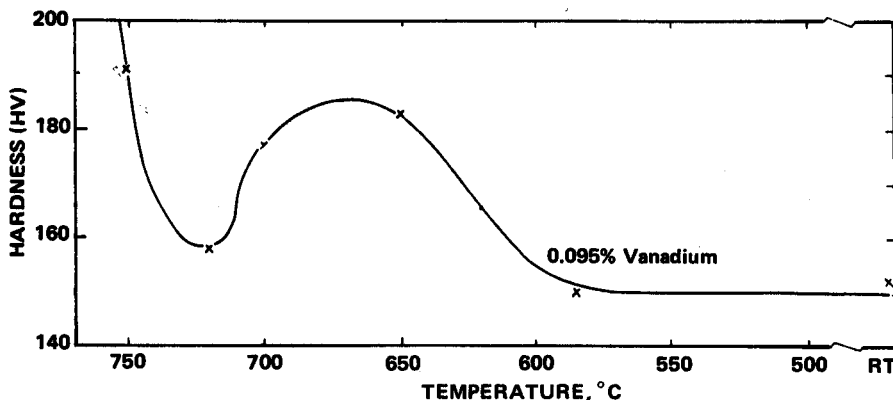


Fig.7 Hardness changes during continuous cooling at 10°C/min for a 0.18% carbon, 0.095% vanadium steel²⁰

However, nucleation of the compounds is difficult and precipitation does not occur other than after very long times. The kinetics of precipitation in austenite has received little attention for vanadium steels although columbium steels have been studied intensively. Recent work by Jonas and Weiss²⁵ has shown that precipitation of CbC in austenite follows a typical C-curve with a "nose" at about 970°C (1780°F) and 100 sec. In view of the lower supersaturations in vanadium steels at any temperature, it is reasonable to suppose that the "nose" for such steels will occur at a lower temperature and after a longer time. Practical experience indicates that precipitation does not occur.

Rates of precipitation are markedly increased by deformation and precipitation in austenite is observed during the rolling of HSLA steels. This deformation induced precipitation is still dependent on the degree of supersaturation in the steel, so one would expect precipitation in vanadium steels to occur at lower temperatures than those found for columbium steels. Precipitation in austenite depletes the matrix in alloying element and therefore, when it occurs, there will be less strengthening resulting from precipitation in ferrite. The austenite precipitates coarsen rapidly and are usually too large to contribute to strengthening.

The effects of hot deformation on HSLA steels are complex. Not only can precipitation be induced but the recrystallisation characteristics of the material are of the utmost importance. The practical aspects of these phenomena have recently been reviewed in depth by Sage.^{26,27} Recrystallisation is retarded by vanadium and columbium. Much confusion has existed concerning the mechanism of the retardation; in particular there have been many arguments about whether the effects are attributable to dissolved elements or to the presence of precipitates. The problem has been clarified by Jonas and Weiss²⁵ who show that at high and low temperatures, where recrystallisation starts before the onset of precipitation, the retardation is due to solution effects but in an intermediate temperature range precipitation starts before recrystallisation and has a very strong retarding effect. Their work related to columbium steels but it seems certain that the same effects are true for vanadium steels. However, since precipitation in austenite is likely to occur at lower temperatures for vanadium steels it is likely that a marked delay in recrystallisation will only be observed on rolling such steels down to temperatures very close to the γ - α transformation temperature. Nevertheless, some retardation through the solute effect should be observable at rather higher temperatures. Pecina²⁸ and Roberts²⁹ have shown that static recrystallisation is very drastically inhibited by vanadium at 800°C (1470°F), but that the effect is small at 900°C (1650°F), whereas in columbium steels the delay is still marked at 1000°C (1830°F). Roberts' work

reveals the interesting observation that vanadium has a stronger effect on dynamic recrystallisation than has columbium.

The major benefit obtained from a retardation of recrystallisation is the possibility of producing either a fine grained recrystallised austenite or a fine unrecrystallised austenite for subsequent transformation to ferrite and the consequent development of a fine grained ferrite with its improved strength and impact transition temperature.

The formation of fine grained ferrite is not solely dependent on the austenite grain size but it also depends on the transformation temperature, and therefore to some extent on cooling rate. It is not so well known that precipitation in ferrite can markedly reduce the ferrite grain size. Some of these effects are illustrated by Fig. 8. An increase in carbon content lowers the curve because of the lower transformation temperature at any cooling rate but it is very apparent that a vanadium-nitrogen steel shows a much finer ferrite grain size as a consequence of the precipitation of vanadium carbonitrides.

In this review little has been said about mechanical properties, rather has the emphasis been on structural and transformational characteristics. However, it must never be forgotten that the major object of our activities should be the development of steels with superior properties and that these properties should be obtained at minimum cost. There is no need for a detailed account of the effects of microstructural and compositional factors on properties. Many such accounts have appeared, that of Gladman et al¹⁰ being outstanding. It will be sufficient here to mention the major factors. Above all, the ferrite grain size must be kept as small as possible as this is our basic tool for minimising the impact transition temperature; it also exerts a considerable influence on yield strength. Further strengthening is, in vanadium steels, most readily achieved by precipitation. The degree of precipitation strengthening depends on the volume fraction of precipitate and on the particle size. The first of these quantities is determined mainly by the amount of vanadium available to combine with carbon and nitrogen. It is generally appreciated that the maximum precipitate volume is obtained when the stoichiometric quantity of vanadium is available for combination with carbon and nitrogen. This quantity is, however, not easy to assess. We have seen that precipitation in ferrite is likely to give a carbonitride which will be relatively high in nitrogen but just how much nitrogen there will be in any particular precipitate must depend strongly on the carbon, nitrogen and vanadium contents of the steel. We might hazard a guess that, on the average, the stoichiometric quantity of vanadium will be about five times the total of available carbon and nitrogen. But what is the available quantity of carbon and nitrogen. Unless a consider-

Table IV. Solubilities of vanadium in γ iron

Temperature,		Solubility of Vanadium in γ -Iron (wt %) with				
°C	°F	0.15% C	C at Interface	0.005% N	0.01% N	0.02% N
800	1470	0.025	0.013	0.010	0.005	0.0025
750	1380	0.011	0.0042	0.0045	0.0023	0.0012
700	1290	0.0045	0.0014	0.0020	0.0010	0.0005

Table V. Solubilities of vanadium in α iron

Temperature,		Solubility of Vanadium in α -Iron (wt %) Saturated with C and N			
°C	°F	C, %	0.005% N _T	0.010% N _T	0.020% N _T
800	1470	0.030	0.0035	0.0028	0.00062
750	1380	0.008	0.0014	0.00078	0.00035
700	1290	0.002	0.00050	0.00025	0.00013

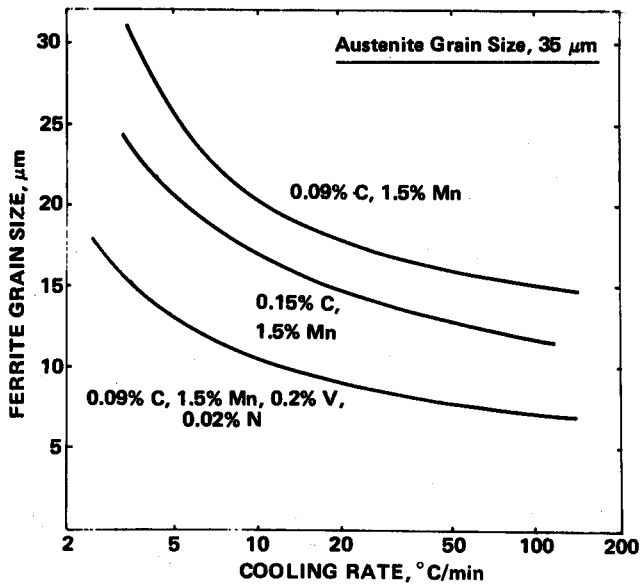


Fig.8 Ferrite grain size v cooling rate for three steels

able amount of interphase precipitation occurs the total carbon content of the steel is irrelevant; precipitation in ferrite must be related to the carbon in solution in ferrite and this is small but difficult to specify with any accuracy. There is some possibility that the effective carbon for precipitation may be rather greater than this simple concept suggests since carbon removed from ferrite by precipitation may be to some extent replenished by diffusion from the austenite. In the times available this is not thought likely to have a great effect.

The nitrogen content of the ferrite should be very close to that of the austenite from which it forms at relatively low precipitation temperatures. It seems likely that the available carbon and nitrogen will be in the range 0.02% to 0.04% depending largely on the nitrogen level. Thus it is unlikely that any increase in precipitate fraction will result from increasing the vanadium content beyond, say, 0.2%.

Precipitate size will depend to a large extent on the temperature at which precipitation occurs and therefore, in part, on cooling rate and transformation temperature. However, if this temperature becomes too low, the rate of precipitation will be adversely affected and a reduction in precipitate fraction will follow. Precipitation kinetics may be influenced by composition and it is interesting to note that Niltawach³⁰ has found that the addition of aluminum to a vanadium-nitrogen steel results in finer precipitates and therefore greater strengthening over a wide range of cooling rates.

Precipitation strengthening always carries a penalty in the form of an increased impact transition temperature. It is often stated that this penalty is directly proportional to the strengthening. However, this is not strictly true. Niltawach³⁰ has shown that precipitates with high nitrogen contents have less effect on transition temperature than is found for the same degree of strengthening produced by low nitrogen precipitates.

Although, in a short Paper it is not possible to cover in detail all aspects of the physical metallurgy of vanadium in steels, it is hoped that sufficient material has been included to show that many things can be understood and that, although some gaps remain to be filled, our knowledge of the subject is in a sufficiently healthy state for practical development to continue.

REFERENCES

- G. Brauer and W. D. Schnell: *J. Less Common Metals*, Vol. 6, 1964, p. 326.
- C. W. Haworth and C. D. Morton: to be published.
- C. D. Morton: Ph.D. thesis, University of Sheffield, 1975.
- G. Brauer and W. D. Schnell: *J. Less Common Metals*, Vol. 7, 1964, p. 23.
- K. Bungardt, K. Kind and W. Olsen: *Archiv. Eisenhüttenwesen*, Vol. 27, 1956, p. 61.
- Sekine, Inoue and Ogasawara: *Trans. I.S.I. Japan*, Vol. 8, 1968, p. 101.
- M. G. Froberg and H. Graf: *Stahl u. Eisen*, Vol. 80, 1960, p. 539.
- K. J. Irvine, F. B. Pickering and T. Gladman: *J.I.S.I.*, Vol. 205, 1967, p. 161.
- J. Wadsworth, J. H. Woodhead and S. R. Keown: *Metals Science*, Vol. 10, 1976, p. 362.
- T. Gladman, D. Dulieu and I. D. McIvor: *MicroAlloying 75 Proceedings*, 1977, p. 32.
- J. H. Woodhead and A. G. Quarrell: *J.I.S.I.*, Vol. 203, 1965, p. 605.
- J. G. Zimmerman, R. H. Aborn and E. C. Bain: *Trans. ASM*, Vol. 25, 1937, p. 775.
- M. A. Grossman: *Trans. AIME*, Vol. 150, 1942, p. 227.
- R. A. Grange: *Met. Trans.*, Vol. 4, 1973, p. 2231.
- G. T. Eldis and W. C. Hagel: "Hardenability concepts with applications to steel," *ASM Symposium*, Chicago, 1977.
- G. T. Eldis: *Hardenability concepts with applications to steel*, *ASM Symposium*, Chicago 1977.
- F. B. Pickering: *MicroAlloying 75 Proceedings*, 1977, p. 9.
- L. Meyer, F. Heisterkamp and W. Mueschenborn: *MicroAlloying 75 Proceedings*, 1977, p. 153.
- J. M. Gray and R. B. S. Yeo: *Trans. ASM*, Vol. 61, 1968, p. 255.

(Concluded on page 10)

20. J. H. Woodhead and D. Webster: J.I.S.I., Vol. 207, 1969, p. 51.
21. A. J. Davenport and R. W. K. Honeycombe: Proc. Roy. Soc., Vol. 322, 1971, p. 191.
22. R. W. K. Honeycombe: Met. Trans., Vol. 78, 1976, p. 915.
23. R. W. K. Honeycombe: Scand. J. of Metallurgy, Vol. 8, 1979, p. 21.
24. J. M. Gray, D. Webster and J. H. Woodhead: J.I.S.I., Vol. 203, 1965, p. 812.
25. J. J. Jonas and I. Weiss: Metal Science, Vol. 13, 1979, p. 238.
26. A. M. Sage: "X70 Vanadium Steel, Report on Development of Steel for Linepipe," Vanitec, 1975.
27. A. M. Sage: Sheffield International Conference on Hot-working and Forming processes, 1979.
28. A. M. Pecina: Ph.D. thesis, University of Sheffield, 1975.
29. W. Roberts: Swedish Institute for Metals Research, Report No. 1211, 1973.
30. S. Niltawach, S. R. Keown and J. H. Woodhead: to be published.

DISCUSSION ON

The physical metallurgy of vanadium steels

J. H. Woodhead

A. T. DAVENPORT (Republic Steel Corp.) On the basis of our extensive studies on a whole variety of air cooled HSLA steels containing vanadium and columbium, we have found that the only form of precipitation in the ferrite is interphase precipitation. The same thing has just been shown independently for vanadium-bearing high carbon steels. Did you not find interphase precipitation in the steels you examined?

J. H. WOODHEAD It's true that you will often have interphase precipitation when the apparently random dispersion of particles will lead you to believe that you haven't. My experience with laboratory materials indicates that much of the precipitation is not interphase, and no matter how hard we tried to reveal regularity or precipitate rows by tilting the sample in the electron microscope, we failed. However, the issue is still in considerable doubt and one would be rash to claim one thing or the other with absolute certainty.

M. KORCHYNSKY (Union Carbide Corp.) If you assume that there can be two modes of precipitation, interphase and

general, the parameter which controls the preference for one or the other is the cooling rate: the faster the cooling rate, the more prevalent will be the general precipitation. Is there any reason to believe that the effect of either form of precipitation on mechanical properties is different and, if so, is there any reason for us to control the precipitation mode in order to maximize mechanical properties?

J. H. WOODHEAD My guess is that the mode of precipitation would not make a great deal of difference to the properties. They are going to be controlled by the size and the dispersion of the particles, and whether these occur with some degree of regularity or not should have only a minor influence on that effect.

A. M. SAGE (Highveld Steel & Vanadium Corp.) Under accelerated cooling conditions, such as one would obtain in strip or thin plate and where one may have supersaturation of nitrogen, would the interphase precipitate be a vanadium carbonitride rather than a vanadium carbide?

J. H. WOODHEAD So long as you are getting an interphase precipitation, I don't think the cooling rate and transformation temperature will have much effect. If the transformation start temperature is lower, the carbon content of the austenite at the interface is going to be even higher, and with this higher carbon content at the interface I would expect that if an interphase precipitate forms, it would certainly be a high carbon compound.