Influence of vanadium on hot ductility of steel

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The hot ductility of V + N steels has been examined and compared with that obtained from a Nb containing steel. When tensile samples from hot rolled plates were solution treated and cooled to test temperatures in the range 700–1000°C, the V containing steels exhibited higher hot ductility than the Nb (0·03 wt-%) steel. Increasing the V and N contents caused both the depth and the width of the low ductility trough to increase, due to increasing precipitation of VN. In accordance with this, there was a good relationship between the product of the total V and N concentrations and the ductility. This product had to be as high as $1{\cdot}2 \times 10^{-3}$ (corresponding to 0·1%V, 0·012%N) for ductility levels to approach the low values of reduction of area exhibited by the Nb containing steel. Furthermore, with precipitation being the factor that controls the hot ductility of the steels, marked fine dynamic precipitation was found to be present in the Nb and the highest V and N containing steel examined (0·1%V, 0·011%N), whereas little precipitation was observed in samples taken from a 0·05%V, 0·005%N steel. Recovery of ductility at the high temperature end of the trough corresponded to when dynamic recrystallisation occurred, and the greater the degree of precipitation, the higher this temperature. In view of the close relationship observed between the hot ductility behaviour in the trough and the likelihood of transverse cracking, it is recommended that, where transverse cracking is a problem, V should be considered as a replacement for Nb. IS/1068

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INTRODUCTION

Transverse cracking is a serious problem in the continuous casting of steel and Nb grades have been found to be most susceptible to this defect.[1] In those cases where the strand is cast in a curve, the cracks are thought to propagate when the strand is straightened. The cracks are intergranular and meander along the prior γ grain boundaries.[1] The straightening operation is performed in the temperature range 700–1000°C, which coincides with the interval in which steel exhibits a ductility minimum in laboratory hot tensile tests. Therefore, the simple hot tensile test has proved very useful in assessing the likelihood of cracking in steels.[1] In the tensile test, conditions are maintained as close as possible to those of the commercial operation. In general, the tensile samples are heated to a high temperature (1300–1350°C) to dissolve all the microalloying additions and to produce a coarse grain size reminiscent of the as cast grain size. The rate of cooling from the solution to the test temperature and the strain rate are selected to approximate those undergone by the strand during cooling and straightening (typically 60 K min^{-1} and 10^{-3}–10^{-4} s^{-1}, respectively). Under these conditions, it has been possible to use hot ductility testing as an aid to investigating and suggesting ways of minimising cracking problems.

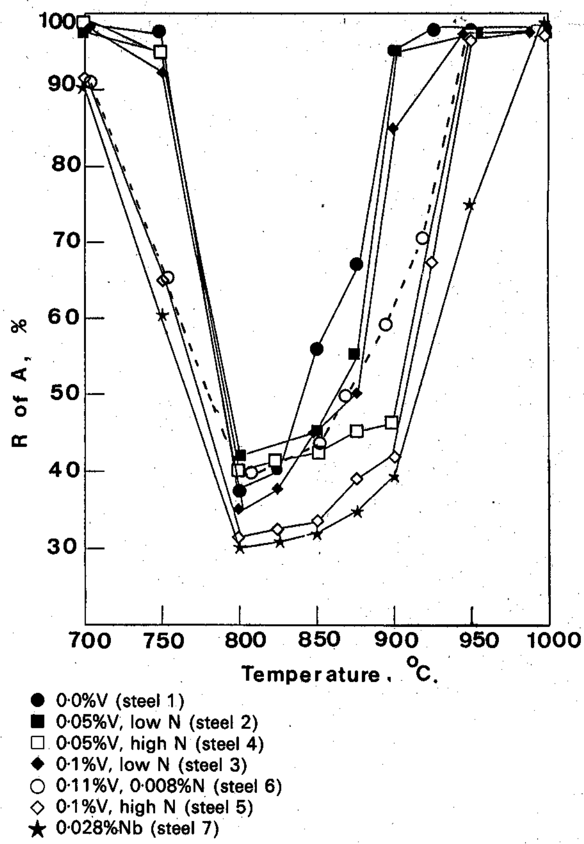
The curves of tensile hot ductility *v.* temperature have been found to be very sensitive to composition. Increasing the Al content or introducing Nb into the steels both deepens and widens the trough to higher temperatures.[1–3] In accordance with this behaviour, transverse cracking also increases because of these elements.[1,4] Most research work has been carried out on Nb containing steels, because these steels have the highest incidence of cracking. Vanadium offers itself as an alternative to Nb, but there is little information available concerning the severity of relevant transverse cracking. A statistical analysis of works data from Oxelund by Hannerz failed to show any significant influence of V on transverse cracking.[4] His hot ductility work suggested that, with a high N steel (0·016%N), the V content had to be in excess of 0·07% to produce a significant deterioration of ductility. This is in agreement with other data which indicate that high V, high N steels are susceptible to transverse cracking.[5]

From these limited data, V appears to be less detrimental to hot ductility than Nb.[6] This may be related to the higher solubility of VN in austenite. Both Nb(C,N) and VN can precipitate rapidly during testing at strain rates $<10^{-1}$ s^{-1} and, hence, can have a very important influence on hot ductility and transverse cracking.[7,8] For steels that are solution treated and cooled to the test temperature, Nb is more effective in extending the trough to higher temperatures than V. This is probably because, for typical microalloy compositions, the nose of the temperature–time curve for Nb(C,N) precipitation without concomitant deformation is at 950°C, whereas it is at about 885°C for VN. Vanadium may also reduce the amount of precipitation of AlN at grain boundaries. Crowther and Mintz[9] have also suggested that V precipitates in both a coarser and more random manner than Nb, which would again favour higher ductility.

In the present work the influence of V and N on the hot ductility of C–Mn–Al steels is systematically examined and the ductility compared with that exhibited by a Nb containing steel.

EXPERIMENTAL

Much consideration was given to the testing route most likely to be relevant to the transverse cracking problem. It has been shown[10] that, for microalloyed steels, solution treatment is the preferred route rather than, as might be expected, direct casting of the tensile samples and cooling to the test temperature. Direct casting is only necessary when solution treatment is difficult, as is the case for Ti treated steels, and when examining the influence of S on hot ductility. For Nb and V treated steel, solution treatment allows all the microalloying additions to be available for precipitating in fine form at the test temperature, enabling their influence on hot ductility to be precisely defined. Therefore, the latter method was adopted.



1 **Hot ductility curves for all steels examined**

The composition of the steels is given in Table 1. The steels were supplied by British Steel as 50 kg vacuum melts, hot rolled to 13 mm thick plate. The casts all have the same base composition: 0·1%C, 1·4%Mn, 0·3%Si, 0·03%Al. Two contents of V, 0·05 and 0·1%, and three contents of N, 0·005, 0·008, and 0·01%, were examined. For comparison purposes, a plain C–Mn–Al steel without any V addition (steel 1) and a Nb containing steel (steel 7) were included.

Hot tensile samples of 4·4 mm dia. and 25 mm gauge length were machined from the plates in the rolling direction. The samples were nickel plated and heated to 1330 ± 5°C in an argon atmosphere, held for 5 min, and cooled at 50 K min^{-1} to test temperatures in the range 700–1050°C ± 5°C. The samples were held for 5 min at the test temperature and strained at a strain rate of 3×10^{-3} s^{-1} on an Instron tensile machine. The samples were ice–brine quenched immediately after fracture to enable the prior γ grain size to be determined as well as to provide evidence for dynamic recrystallisation. The fracture surfaces of all the samples were examined using a Jeol T100 scanning electron microscope (SEM). Carbon extraction replicas were taken from selected steels from transverse sections approximately 1 mm behind the fracture surface and examined using a Jeol 100 kV transmission electron microscope (TEM).

RESULTS

Hot ductility curves

The hot ductility curves of reduction of area (RA) against test temperature are shown in Fig. 1. A ductility trough was found in all the steels. Values of RA at 1000°C were almost 100% and at 700°C were ≥90%. Ductility was a minimum at 800°C. The width and depth of the troughs varied with alloying additions. As expected, the low N C–Mn–Al steel had the narrowest trough and one of the higher RA values at 800°C, while the Nb containing steel exhibited the lowest ductility. Adding 0·05%V at the lowest N content caused a slight widening of the trough and a further increase to 0·1%V produced only minor changes (a slight deepening and widening of the trough).

More marked changes were apparent when the N content was increased from 0·005 to 0·008 through to 0·01%. At the lower V content this was manifested by a widening of the trough, but no change of depth. At 0·1%V, both widening and deepening occurred.

The 0·1%V, 0·011%N steel gave a curve which was now similar to the Nb steel in the temperature range 700–850°C and only showed improvement in the higher temperature range.

The stress *v.* total elongation curves for the steels are shown in Figs. 2 and 3. The occurrence of dynamic recrystallisation can be detected from these curves, by either an abrupt decrease or oscillations of the flow stress. It would appear from these curves that the temperature for the onset of dynamic recrystallisation increases with the concentration of V and N and is highest for the Nb containing steel. Estimated temperatures were 900, 900, 900, 950, 950, 950, and 1000°C for steels 1–7, respectively.

Metallography

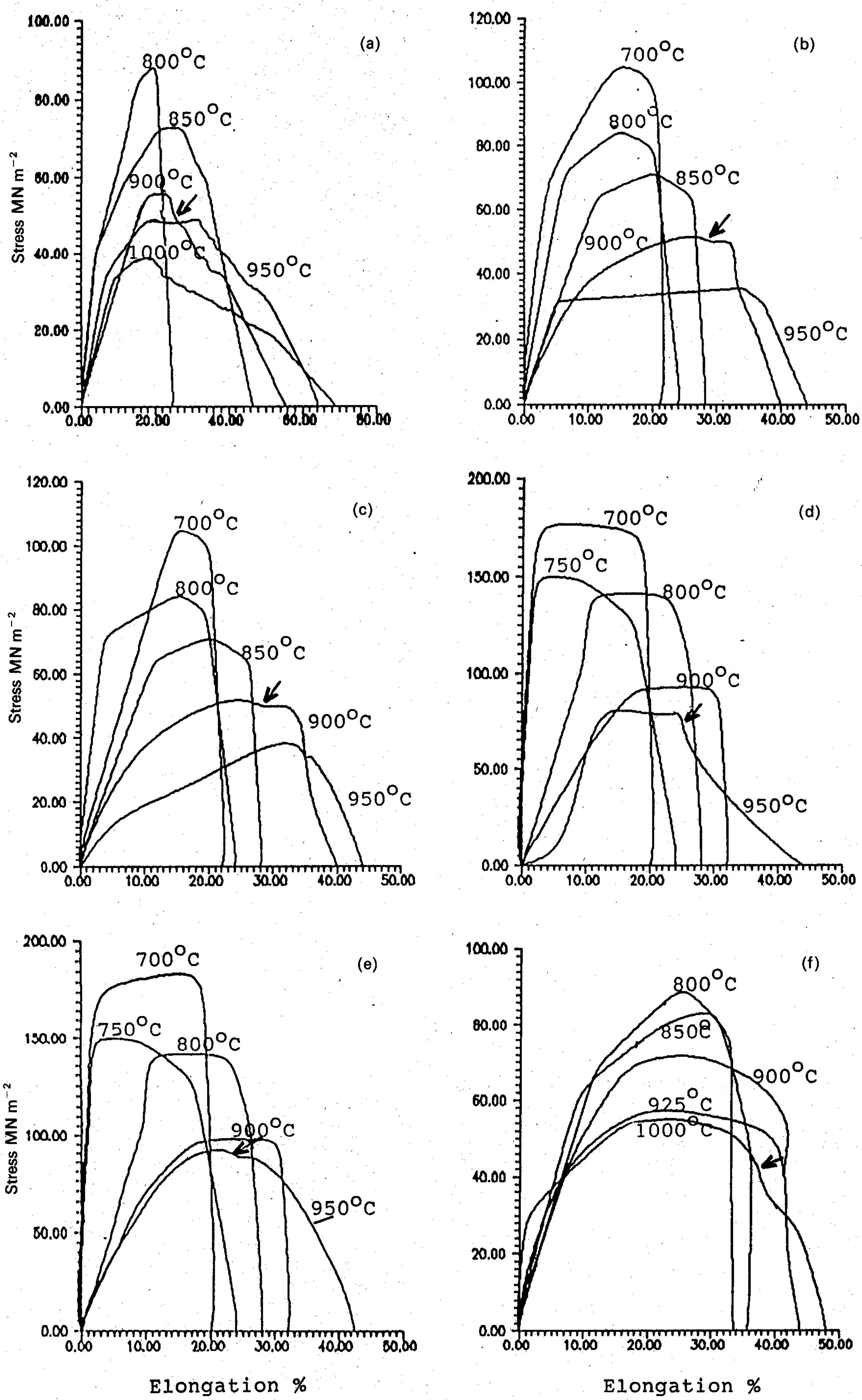
The γ grain size after solution treatment was found to be 400 μm for all the steels examined. This is close to the bottom end of the commercial as cast grain size range (500–5000 μm)[6] for continuously cast steel. At the temperature for minimum ductility, cracks were observed to form along the γ grain boundaries (Fig. 4), these often being wedgelike in appearance, typical of cracks formed by grain boundary sliding. Thin films of deformation induced ferrite (Fig. 5) were also found to be present at this temperature.

Fracture examinations

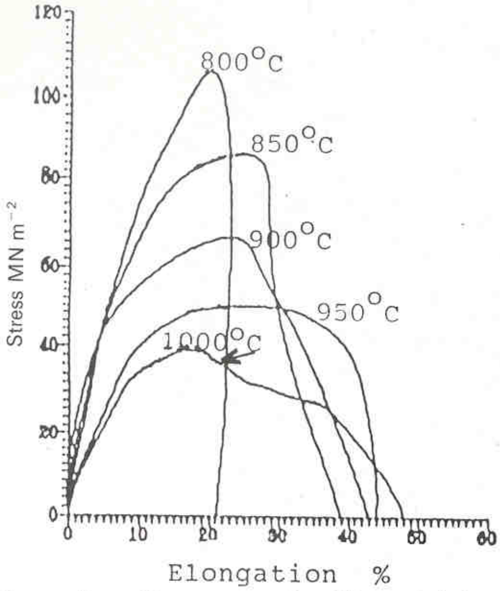
Fracture behaviour was similar for all the steels examined. At temperatures >950°C, high temperature ductile rupture failure occurs (Fig. 6). Large cavities are present on the fracture surface not associated with the second phase particles. It is thought that these are formed as cracks at the boundaries at an early stage of deformation by grain boundary sliding.[11] However, because the temperature is sufficiently high to allow dynamic recrystallisation to occur (Fig. 7), these cracks become isolated from the prior boundaries as a result of grain boundary migration. The original cracks are then distorted into elongated voids until final failure occurs by necking between these voids. When the temperature is decreased to the minimum ductility temperature, the fracture appearance changes to a mixture

Table 1 Composition of steels used, wt-%

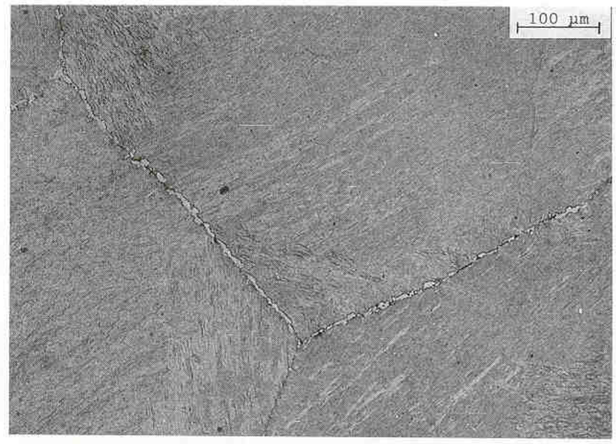
Steel	C	Si	Mn	P	S	Al	N	Nb	V
1	0·09	0·31	1·43	0·016	0·003	0·030	0·0056	...	...
2	0·10	0·33	1·39	0·015	0·003	0·031	0·0051	...	0·050
3	0·10	0·33	1·39	0·011	0·003	0·031	0·0053	...	0·100
4	0·10	0·33	1·37	0·018	0·002	0·030	0·0110	...	0·052
5	0·10	0·33	1·35	0·018	0·002	0·028	0·0100	...	0·110
6	0·11	0·32	1·29	0·016	0·002	0·027	0·008	...	0·110
7	0·09	0·30	1·41	0·016	0·003	0·032	0·0052	0·028	...



2 Stress–elongation curves for V steels; arrows indicate temperature for start of dynamic recrystallisation



3 **Stress–elongation curves for Nb containing steel (steel 7); arrow indicates temperature for start of dynamic recrystallisation**



5 **Evidence of deformation induced ferrite in Nb steel tested at 850°C (Ae_3 temperature = 860°C)**

of flat smooth facets and faces covered with microvoids (Fig. 8*a*), the latter mode becoming more pronounced as the temperature decreases (Fig. 8*b*). Intergranular failures are now obtained. The flat facets are characteristic of failure by grain boundary sliding in the γ and microvoid coalescence failures are due to precipitation and/or MnS inclusions causing cavitation at the boundaries within the thin films of deformation induced ferrite (Fig. 5).[12] These films have been shown to form between the Ae_3 and Ar_3 temperatures.[12] Ferrite is softer than austenite and, hence, when the steels are deformed in tension, strain concentration occurs in these films, causing cavitation around the particles, and these cavities link up to give intergranular failure. The proportion of flat facets was observed to increase with temperature.

Below 800°C, the fracture appearance again changes and, at 750°C, ordinary ductile failure (i.e. transgranular, with the joining up of the voids surrounding the second phase particles) was apparent and ductility was high.

Replicas

The precipitation quenched in from the austenite in the Nb containing steel at the minimum ductility temperature was found to be intense, with a fine matrix precipitation of Nb(C,N) as well as a coarser γ grain and subboundary precipitation, as shown in Figs. 9*a* and 9*b*, respectively. In the high V, high N steel, precipitation was generally coarser and less intense (Fig. 10). The size distribution for the VN and Nb(C,N) precipitates in steels 5 and 7 are shown in Fig. 11, the average particle size being 16 and 6 nm, respectively. In the low V, low N steel, only isolated precipitates were observed in accord with its good ductility.

DISCUSSION

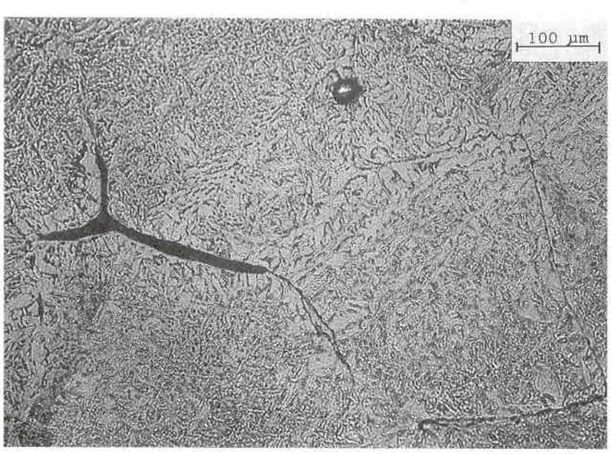
Influence of Nb

For Nb containing steels, intergranular failure in the higher temperature range (≥900°C) invariably occurs by grain boundary sliding in the austenite. Nb(C,N) precipitation is particularly effective in reducing the hot ductility and widening the trough for the following reasons.[6]

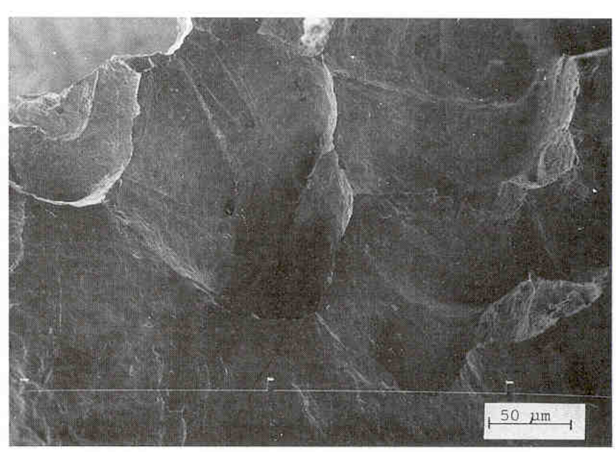
1. The very fine matrix precipitation of Nb(C,N) increases the stress required for deformation and, therefore, increases the stress in the grain boundary regions.

2. Often, precipitate free zones are present at the boundaries, concentrating the strain into the boundary regions. The situation is then very similar to that prevailing when thin films of ferrite are present.

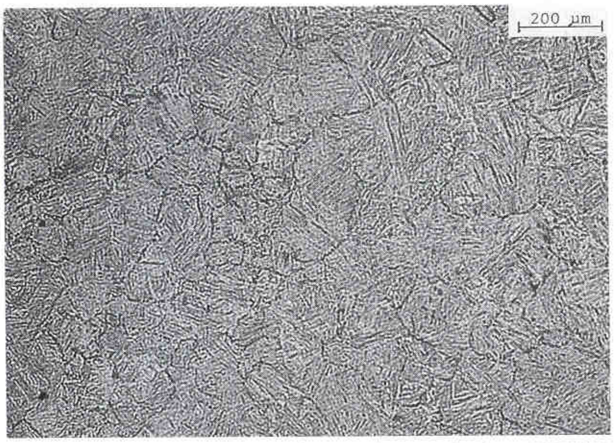
3. There is always marked precipitation at the austenite grain boundaries and this will encourage voiding and extension of cracks formed by grain boundary sliding. If the precipitation is fine with close spacing of particles, as is found in Nb containing steels, cracks can readily link up. Previous work[13] on hot rolled samples heated up to 1330°C has tended to indicate that all the Nb(C,N) will



4 **Wedge cracks observed in 0·11%V, 0·01%N steel (steel 5) tested at 850°C**



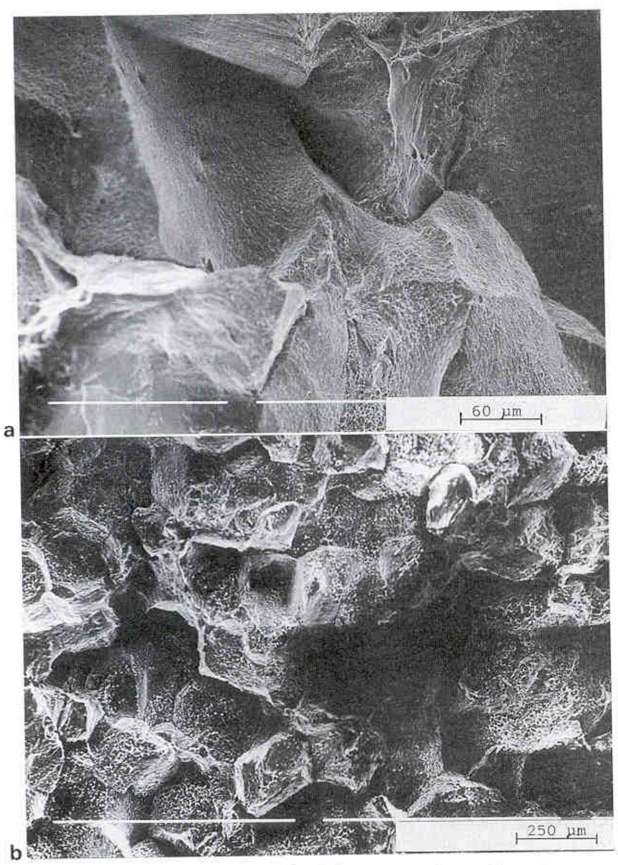
6 **Example of ductile rupture observed in high V, high N steel (steel 5) tested at 950°C**



7 Dynamic recrystallisation observed in 0·05%V, 0·005%N steel (steel 2) tested at 950°C

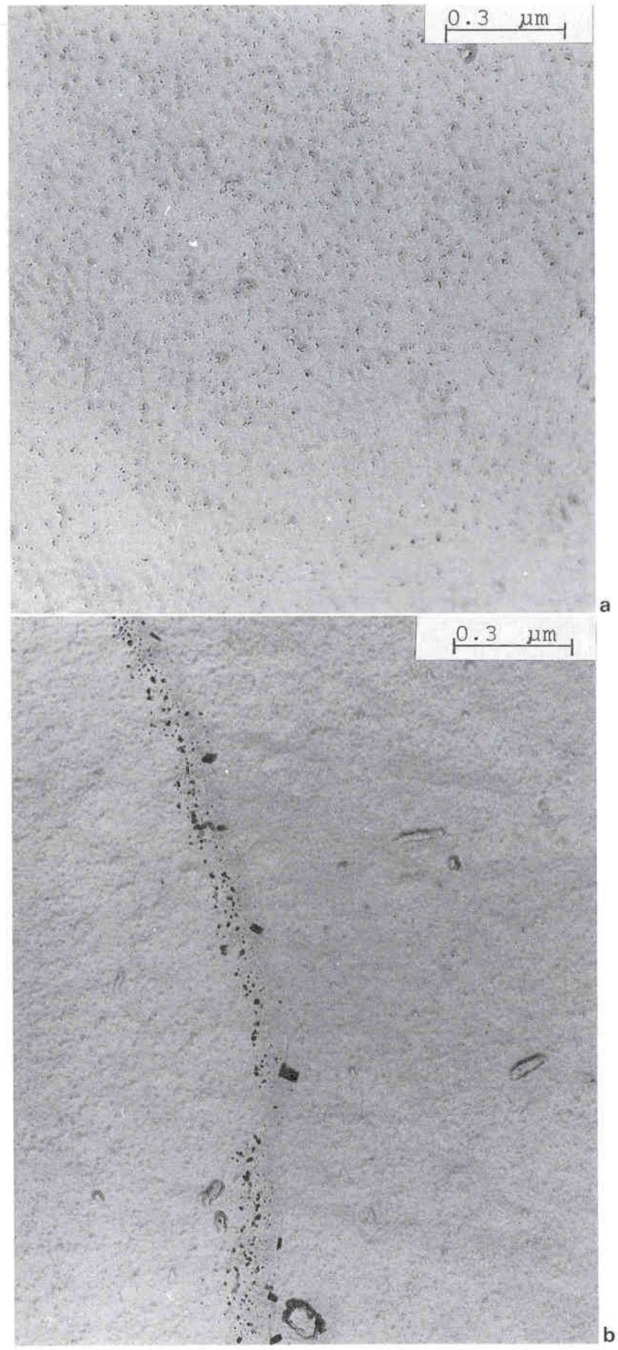
have gone back into solution and may all precipitate out in a very fine form during deformation at the test temperature (Fig. 9). It has been shown that dynamic precipitation occurs very rapidly at temperatures close to the nose of the precipitation–temperature–transformation (PTT) diagram, i.e. at 900°C for C–Mn–Nb–Al steels, and it can often be assumed that the rate of reaction is sufficiently rapid to maintain equilibrium.[5]

At lower temperatures, deformation induced ferrite is also formed (800°C) and this, in conjunction with the precipitation, leads to continued poor ductility. However, fracture now is by microvoid coalescence at the MnS inclusions



a at 850°C, showing mixture of intergranular microvoid coalescence and intergranular decohesion modes of failure; *b* at lower temperature of 800°C, showing mainly intergranular microvoid coalescence as mode of failure

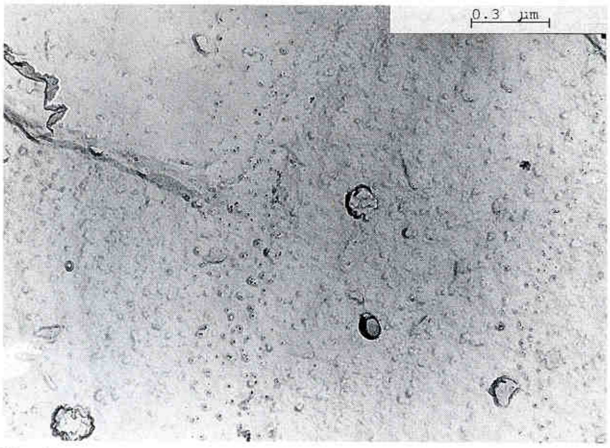
8 Fracture surface of C–Mn–Al–Nb steel



9 *a* Fine matrix Nb(C,N) precipitation, as well as *b* coarser grain/subboundary precipitation, in Nb containing steel tested at 800°C

present in these thin films, and this is exacerbated by the Nb(C,N) precipitation. Recent work[14] has shown that full recovery of ductility at the high temperature side of the trough is associated with dynamic recrystallisation. Improvement in ductility can occur on increasing the temperature above the Ae_3 due to deformation induced ferrite no longer being able to form, but, unfortunately, this improvement is small, since grain boundary sliding in the austenite then occurs and this is often almost as detrimental to ductility. Some improvement in ductility can also occur at higher temperatures, by either reducing the degree of precipitation or coarsening existing precipitates. However, the main improvement in ductility is associated with dynamic recrystallisation.

For Nb containing steels, the fine precipitation of Nb(C,N) produced on deformation delays the onset of



10 Sparser and generally coarser precipitation of VN in high V, high N steel (steel 5) tested at 800°C

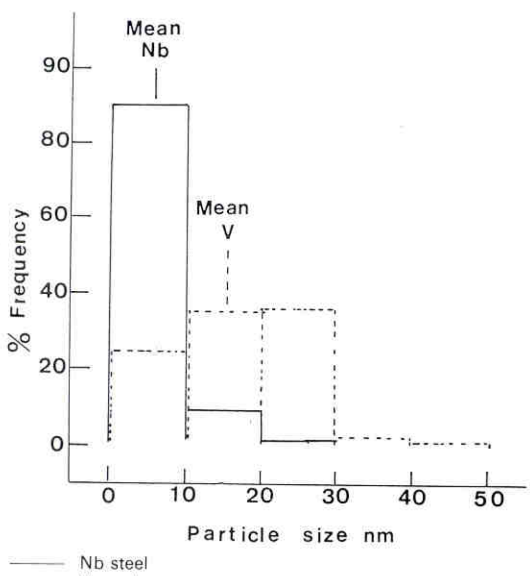
dynamic recrystallisation to higher temperatures, thus extending the trough to higher temperatures.

Ductility recovery at the low temperature end of the trough generally corresponds to the presence of a significant volume fraction of ferrite, so that strain concentration does not occur.[6] Also, as the temperature is decreased, the strength differential between the γ and ferrite is reduced.[11] Ferrite has higher ductility than γ, because it recovers more readily, tends to produce scalloped boundaries, which makes grain boundary sliding difficult, and has a finer grain size.[6]

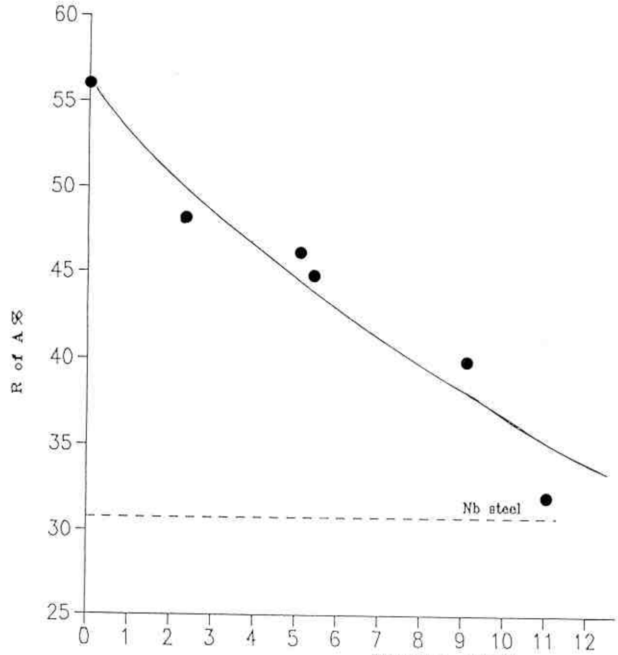
Influence of V

It can be seen from Fig. 1 that increasing the V or N content reduces the hot ductility. This arises because increasing the concentrations of V and N will favour more precipitation of VN in the temperature range of the trough.

Therefore, it is not surprising in the solution treated condition that a strong relationship exists between the product of the total V and N concentrations, $[V_t][N_t]$, and both the depth and the width of the trough (Figs. 12 and 13, respectively). Only when this product increases to about $1{\cdot}2 \times 10^{-3}$ (0·1%V, 0·012%N) does the ductility become close to that of the Nb containing steel (0·028%Nb).



11 Particle size distribution for Nb (7) and high V, high N (5) steels; a total of 500 particles was measured for each steel in 10 different areas



12 Influence of product of total V and N contents, $[V_t][N_t]$, on RA values; steels were tested at 850°C in trough

Examination of the carbon extraction replicas shows that, at these V and N contents, precipitation is less extensive but coarser than that shown by the Nb containing steel (*see* Figs. 9 and 10).

Recovery of ductility at the high temperature end of the trough in the V containing steels is again related to the temperature at which dynamic recrystallisation can occur and this is increased as the precipitation becomes more intense.

It has been shown[8] that, for Nb containing steels deformed at temperatures close to the nose of the PTT diagram (900°C), precipitation can be so rapid that equilibrium precipitation can be attained during the tensile test. The volume fraction of NbC precipitated can be calculated from the equilibrium data of Irvine *et al.*[15]

$$\log[Nb_s][C_s] = \frac{-6770}{T} + 2{\cdot}26 \qquad (1)$$

(where subscript s refers to the elements being in solution) and these calculated values for temperatures of 800, 850, and 900°C are given in Table 2.

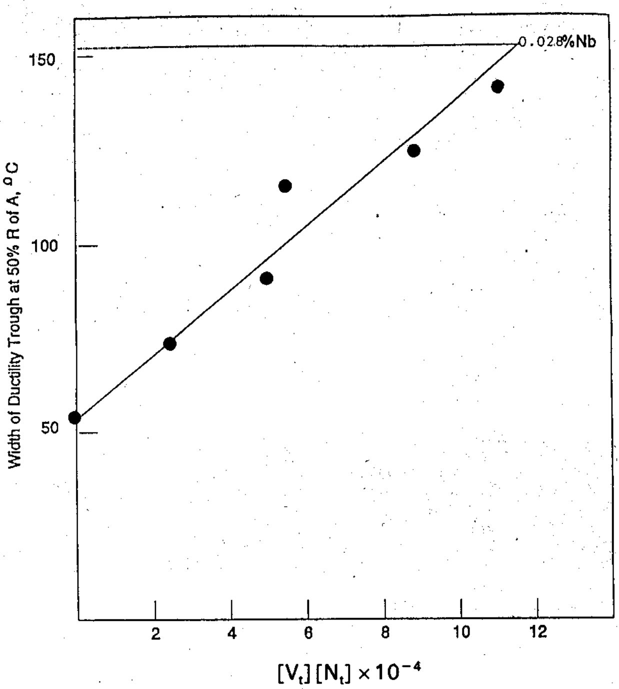
Similar calculations can be carried out for VN, assuming equilibrium conditions, using the following equation[15]

$$\log[V_s][N_s] = \frac{-8830}{T} + 3{\cdot}46 \qquad (2)$$

Previous work[6] has shown that AlN precipitation is very sluggish and AlN is not generally precipitated under the conditions of the test, provided the $[Al_t][N_t]$ does not

Table 2 Calculated volume fractions of NbC and VN, assuming equilibrium conditions

		Temperature, °C		
Steel	$[V_t][N_t] \times 10^4$	800	850	900
2	2·55	0·027	0·024	0·018
3	5·30	0·031	0·029	0·026
4	5·70	0·057	0·054	0·044
5	11·0	0·058	0·056	0·054
6	8·8	0·046(5)	0·045	0·042
7	...	0·032	0·030	0·029



13 Effect of product of total V and N contents on width of ductility trough at 50%RA

exceed 3×10^{-4}. In the present instance it is possible that a small amount of N may have been removed from solution as AlN, leaving less N for precipitation as VN, but only at the higher N contents. For the purpose of calculation, this has been ignored.

It can be seen from Table 2 that increasing the product of the V_t and N_t contents generally leads to an increase of the volume fraction of VN precipitated. However, this is not always so, as can be seen on comparing the volume fractions for steels 4 and 6. This is because increasing the N content is more effective in inducing precipitation than increasing the V content.

Table 2 shows that, under equilibrium conditions, the volume fraction of VN for the high N, high V steel (steel 5) is about twice that of the Nb containing steel (steel 7), yet its ductility is similar. Such behaviour would be consistent with the observation that precipitation in the V containing steel is coarser than in the Nb. However, work by Akben *et al.*[16] suggests that equilibrium conditions will not be obtained in the present exercise.

These authors have shown that the kinetics of dynamic precipitation of Nb and V are similar and very fast. They examined[16] a 0·05%C, 1·2%Mn steel at 0·035%Nb and 0·115%V contents with 0·006%N and found, using a strain rate of $5 \cdot 6 \times 10^{-3}$ s^{-1}, that the nose temperature was approximately 900°C for the Nb containing steel and slightly lower (880°C) for the V steel. Above about 880°C Nb precipitated more rapidly than V, while below 880°C the reverse was true. At 900°C about 5 min were required to complete precipitation for the Nb containing steel and 8 min for the V steel. The time to complete a test in the present exercise was 2–4 min, so it is not likely that precipitation has gone to completion and that the data in Table 2 can be used. However, it is likely that, even if equilibrium conditions are not maintained, the high V, high N steel (steel 5) would in a given time have a greater volume fraction of precipitation present than the Nb containing steel. Certainly, in the present work the value of the product of the V and N concentrations has been shown to be a reasonable criterion to use, as might be expected where non-equilibrium conditions for precipitation prevail. A higher volume fraction would be consistent with the observation that the VN precipitation is coarser than the Nb(C,N) precipitation. From the Lifshitz–Wagner theory of particle coarsening,[17] precipitates would also be expected to coarsen more rapidly in the V containing steel due to the higher solubility of V in austenite compared with Nb. Thus, it is probable that V is less effective than Nb in reducing ductility, because, under the conditions experienced during solidification of continuously cast strand, it forms a coarser precipitate and, hence, more of it is required before ductility deteriorates. It should be noted that, although these V rich precipitates are coarser than the Nb(C,N) precipitates, they are still sufficiently fine to be taken rapidly into solution during reheating for rolling, thus making the V available for subsequent precipitation strengthening.

COMMERCIAL IMPLICATIONS

Since dynamic recrystallisation is not possible during the straightening operation because of the low applied strain (~2%) and the coarse grain size present (500–1000 µm), it has been found that the depth of the trough before the onset of dynamic recrystallisation has the most relevance with regard to the continuous casting operation. In this respect, the curve shown in Fig. 12 can be used to compare the tendency of a steel to exhibit transverse cracking, since the temperature selected for comparison purposes is below the temperature for the onset of dynamic recrystallisation.

It is apparent from these curves that the ductility decreases with an increase of V and N and that, provided the V, N steels have a product $[V_t][N_t] < 1 \cdot 2 \times 10^{-3}$, they will show a reduced propensity to cracking compared with the Nb containing steel. Thus, it is reasonable to suggest that, where transverse cracking is a problem, V may usefully be considered as a replacement for Nb.

CONCLUSIONS

1. Vanadium has been shown to be less detrimental to hot ductility than Nb, provided the product $[V_t][N_t] < 1 \cdot 2 \times 10^{-3}$. Increasing the V or N content both deepens and widens the hot ductility trough.
2. The ductility has been shown to be related to the degree of precipitation and this increases with increasing V or N content.
3. Under the conditions which are likely to exist during the continuous casting of strand, V is probably less detrimental to ductility than Nb, because it precipitates in a coarser form and a greater volume fraction of precipitate is required to cause the ductility to deteriorate.
4. Because of the generally good relationship between the hot ductility behaviour as measured by the reduction of area values in the trough and the likelihood of transverse cracking occurring, it is recommended that, where cracking is a problem, V should be considered as a replacement for Nb.

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