FERRITE GRAIN REFINEMENT AND PRECIPITATION STRENGTHENING IN V-MICROALLOYED STEELS

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INTRODUCTION

Vanadium is best known as an eminent element for strong and easy controllable precipitation strengthening. The principal reason for this is the relatively large solubility product of its carbo-nitrides resulting in a lower solution temperature and a larger capacity to dissolve them at elevated temperatures. A special feature of V as compared to Nb is that its nitride is much less soluble than its carbide and this gives N a very important role in V-steels, especially in their precipitation strengthening. In order to maximise the precipitation strengthening it is necessary to understand the roles of nitrogen and carbon in formation of high volume fractions of finely dispersed carbonitrides. Previous data strongly indicate that the role of nitrogen is clear[1]. It has been demonstrated that the yield strength increases almost linearly with increasing nitrogen content for given vanadium and carbon levels making N an eminent choice for strong and easily controllable precipitation strengthening[2]. Carbon content has usually been considered not relevant to precipitation strengthening when the precipitation occurs in ferrite. This was deduced from the fact of very restricted solubility of carbon in ferrite (which is normally supposed to be independent of the total carbon content in the steel). Most published literature does not suggest that differences in carbon contents in the range for structural steels (0.04-0.3%) should affect significantly the response of vanadium in these steels[3]. However, this viewpoint must be revised in the light of the later work[2].

Recent studies at SIMR[4] and literature data[5-7] strongly suggest that vanadium can also by effectively used for ferrite grain refinement. There is also evidence that vanadium promotes the formation of acicular ferrite microstructure steels[8]. It was suggested that the VN particles which precipitated inside austenite grains during/or after hot rolling show strong potential for nucleation of intragranular ferrite. Although, vanadium does not readily precipitate in austenite, the precipitation process can be enhanced with increasing nitrogen in the steel or by plastic deformation (strain-induced precipitation). In the case of precipitation in ferrite, the Baker-Nutting (B-N) orientation relationship is observed between the V(C,N) particles and matrix[9]. This fact may be very important for the nucleation of ferrite on cooling as the interfacial energy between ferrite and vanadium nitride is very low for the B-N relationship. In fact, intragranular ferrite idiomorphs were observed to nucleate at vanadium nitrides with the B-N orientation[7]. These observations suggest that intragranular ferrite can nucleate on VN and maintain coherent, low energy interfaces with respect to vanadium nitride. Thus, it was concluded that the main factor governing the formation of intragranular ferrite is the presence of VN precipitates in austenite which can develop coherent, low energy, interphase boundaries with ferrite[10]. The precise mechanism by which vanadium additions may enhance the nucleation rate of ferrite is not known[7,10,11], but this new role of vanadium can be extremely important for thermo-mechanical processing as grain refinement leads to a significant increase in strength which is accompanied by a marked improvement in toughness.

The aim of the present work was to investigate the ferrite grain refinement by intragranular ferrite nucleation on VN and to explore the role of N and C in precipitation strengthening of V-microalloyed steels.

EXPERIMENTAL PROCEDURE

The investigation was carried out on seventeen V-microalloyed steels with 0.04 - 0.22% C, 0.06 - 0.13% V and various levels of nitrogen. The chemical compositions are given in Table I. All the steels were prepared as 2 kg laboratory ingots (15 x 120 mm in cross section) with a cooling rate during solidification designed to simulate commercial continuous casting.

Table 1 Chemical composition of the experimental steels.

Steel	**C**	Si	Mn	P	S	Al	**V**	**N**	$T_{aust.}$(°C)
A5	**.10**	.36	1.38	.006	.008	.005	**.12**	**.005**	1075
A8	**.10**	.40	1.38	.008	.010	.003	**.12**	**.008**	1100
A14	**.10**	.40	1.38	.008	.010	.003	**.12**	**.014**	1100
A25	**.10**	.37	1.36	.007	.009	.004	**.13**	**.025**	1150
B25	**.10**	.38	1.40	.007	.010	.005	**.06**	**.025**	1100
C5	**.04**	.38	1.39	.008	.009	.003	**.12**	**.005**	1050
C9	**.04**	.39	1.39	.008	.010	.004	**.12**	**.009**	1075
C15	**.04**	.39	1.39	.007	.010	.004	**.12**	**.015**	1100
D1	**.22**	.40	1.37	.008	.014	.006	**.12**	**.001**	1125
D13	**.22**	.39	1.39	.006	.016	.006	**.12**	**.013**	1160
D19	**.22**	.40	1.40	.008	.015	.005	**.12**	**.019**	1140
D22	**.22**	.40	1.40	.008	.013	.006	**.12**	**.022**	1125
E6	**.23**	.40	1.38	.006	.013	.004	**.06**	**.006**	1105
E12	**.21**	.40	1.39	.008	.016	.006	**.06**	**.012**	1110
E23	**.21**	.39	1.38	.006	.014	.004	**.09**	**.023**	1125
G13	**.15**	.40	1.43	.009	.014	.005	**.12**	**.013**	1040
G23	**.15**	.40	1.39	.008	.016	.005	**.13**	**.023**	1170

The ingots were homogenised at 1250°C for 4 hours and water quenched before heat treatment. Then, careful examination of the austenite grain size and dissolution of V(C,N) was performed in order to determine the appropriate heat treatment temperature for each steel. The criteria for the determination of the heat treatment temperature were that the steels had the same austenite grain size (100 µm) after austenitisation and that the temperature was such that all vanadium had gone into solid solution for subsequent precipitation. The actual heat treatment temperatures that were adopted are also given in Table 1.

The effect of V on the intragranular ferrite formation was examined by using a quenching dilatometer. The specimens were austenitised for 10 minutes and cooled at a rate of 2°C/s to isothermal holding temperatures of 850-875°C. After holding for 1 hour the specimens were quickly transferred to salt baths for isothermal transformation at temperatures 750-450°C. After holding for a short time the specimens were gas quenched to room temperature.

Examinations of the precipitation kinetics of VN in austenite were performed on stabilised specimens which were first solution-treated at 1300°C for 1 hour followed by water quenching and then aged at different isothermal temperatures for 24 or 48 hours. The equilibrium precipitation of VN was analysed in the temperature range of 850-1000°C. The amount of V in VN was measured by the inductively coupled plasma (ICP) emission method on electrolytically extracted compounds.

The ferrite microstructure was examined by optical microscopy after etching in nital. Precipitate morphology and general structural effects were studied on electrolyticaly polished specimens using SEM and on thin foils using TEM. Chemical compositions of inclusions and small precipitates were determined by the EDS technique. The ferrite grain size, the volume fraction of pearlite, precipitate morphologies and size distribution of V(C,N) particles as well as yield strength of the steels were evaluated.

RESULTS

Precipitation Kinetics of VN in Austenite

Thermo-Calc calculations of the equilibrium precipitation of V(C,N) in steels microalloyed with vanadium are shown in Fig. 1. Also shown in this figure is the mole fraction of nitrogen in carbo-nitrides at various temperatures and for various nitrogen contents from hyperstoichiometric levels to zero. These results show that vanadium starts to precipitate in austenite as almost pure nitride and the precipitation start temperature depends strongly on the nitrogen level. For 0.10%V and 0.03%N the precipitation starts at ~1110°C and the precipitation start temperature decreases to below 950°C at 0.003%N. When the nitrogen is about to be exhausted there is a gradual transition to form mixed carbo-nitrides, Fig. 1(b). Enhanced precipitation during the transformation is a result of the solubility drop of the vanadium carbo-nitrides associated with the transformation from austenite to ferrite at a given temperature. A significant feature of the solubility of VC in austenite is that this is considerably higher than the solubility of VN, suggesting that vanadium carbide will not precipitate in austenite.

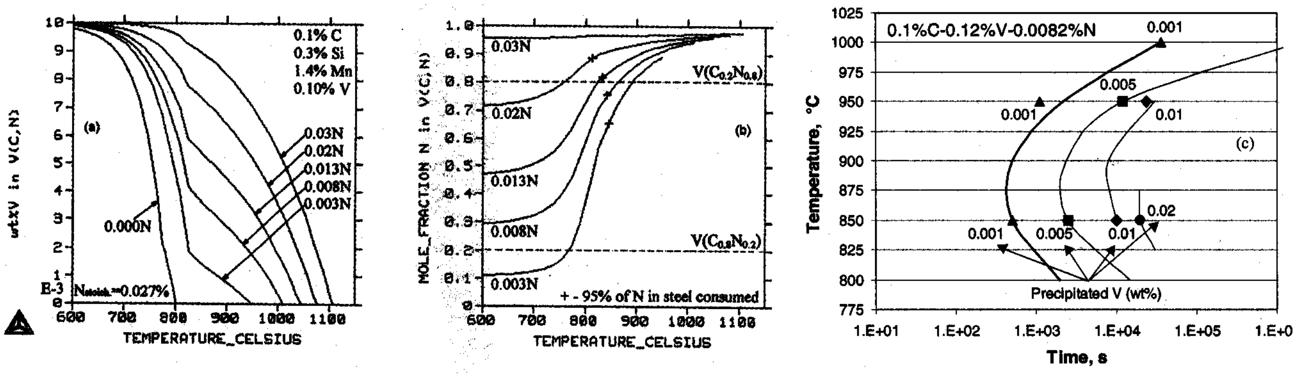


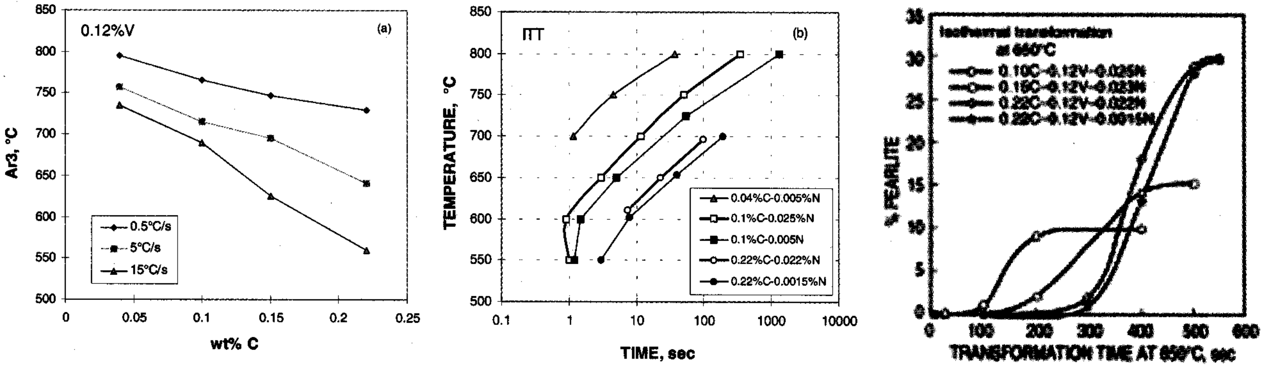
Fig. 1 Theoretical Thermo-Calc calculations showing the precipitation of nitrides, nitrogen rich carbonitrides and carbides in 0.10% V steels at various nitrogen contents (a), (b) and the experimental precipiation-time-temperature diagram for vanadium carbo-nitrides in undeformed austenite (c).

The precipitation-temperature-time diagram for VN in undeformed austenite is shown in Fig. 1(c) for the steel A8. The experimental results yield the well-known "C-curve" associated with the kinetics of VN precipitation. The curves indicate precipitated vanadium at the level of 0.001%, 0.005%, 0.01% and 0.02%. It is clearly seen from Fig. 1(c) that the precipitation process of VN in undeformed austenite is very sluggish. After holding for 1 hour at 850°C less than 10% of the equilibrium amount of VN was, in fact precipitated. The maximum rate of precipitation of VN in austenite was evaluated to be in the range of 850-875°C.

Transformation Behaviour

Transformation temperatures were determined by dilatometry at three different cooling rates, 0.5°C/sec, 5°C/sec and 15°C/sec and in all cases showed the expected reduction in Ar_3 with increasing cooling rate. Results which are summarised in Fig. 2(a) show that a reduction of carbon content from 0.22 % to 0.04 % has a significant influence in raising the transformation temperature as would be expected from the phase diagram. In previous investigations[12] some tendency for the Ar_3 temperature to increase at higher nitrogen levels was observed, especially for the faster cooling rates.

The isothermal austenite to ferrite transformation was more rapid in low C steels than in high C steels, Fig. 2(b). A small influence of nitrogen content was seen during isothermal phase transformation in the range 550°C to 700°C. Raising the nitrogen level from 0.005 % to 0.025 % accelerated the transformation, reducing the transformation start time by about 25 %. Raising the carbon content delays the pearlite formation from ~100 seconds at 0.10%C to ~300 seconds at 0.22%C, Fig. 2(c).

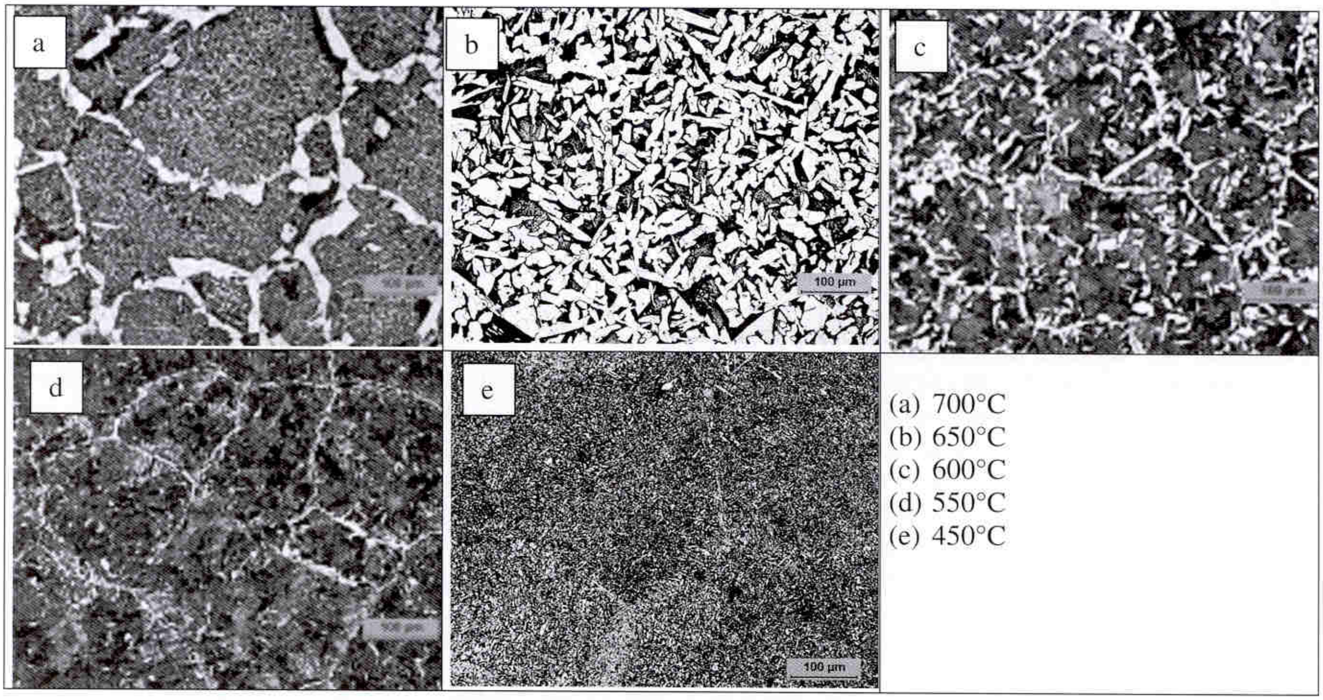


(a) (b) (c)

Fig. 2 Transformation characteristics of the investigated steels during continuous cooling (a) and isothermal transformation (b) and pearlite start time during isothermal transformation at 650°C (c).

Formation of Intragranular Ferrite in V-Steels

The microstructures of the specimens isothermally treated at 850°C for 1 hour and partially transformed at the temperatures 700-450°C followed by gas quenching to room temperature are shown in Fig. 3. High transformation temperatures resulted in largely reconstructive transformation at the prior austenite grain boundaries. The microstructure produced on a partial transformation at 700°C and gas quenching is shown in Fig. 3(a). It consisted of coarse grain boundary ferrite. Within the prior austenite grains a mixture of acicular ferrite with a second phase was formed as a result of the secondary transformation at a lower temperature during gas quenching. The microstructure obtained in his steel after a partial transformation at 650°C followed by gas quenching is shown in Fig. 3(b). Comparing to the specimen transformed at 700°C there was less grain boundary ferrite at the prior austenite grain boundaries, but a high density of intragranulary nucleated polygonal ferrite inside the prior austenite grains. Some of the intergranulary nucleated ferrite crystals were elongated with sharp edges which is probably characteristic of the faster cooling after the interrupted transformation. Quenching from 600°C produced a thinner band of grain boundary ferrite and higher number of intragranulary nucleated ferrite crystals, Fig. 3(c). There is also a change in the morphology of the intragranular ferrite which become less polygonal and more sideplate. However, it is not clear whether this plate morphology of ferrite was associated with reconstructive growth mechanism during isothermal transformation at this temperature or whether it formed at lower temperatures after the gas quench. There were also some ragged ferrite plates growing from the grain boundary ferrite allotriomorphs. Fig. 3(d) shows the microstructure resulting from a partial isothermal transformation at 550°C followed by gas quenching. There was very little grain boundary nucleated allotriomorphs which a very thin skeleton of ferrite at the prior austenite boundaries. Within the prior austenite grains a high density of small ferrite crystals was nucleated. A large proportion of this intragranular ferrite was in the form of thin elongated plates, some of which were observed to have grown from particles. The microstructure produced on isothermal transformation at 450°C is shown in Fig. 3(e). Almost no grain boundary nucleated allotriomorphic ferrite is seen. The general matrix microstructure formed at this temperature was very fine acicular ferrite.



(a) 700°C
(b) 650°C
(c) 600°C
(d) 550°C
(e) 450°C

Fig. 3 Microstructures of steel D13 austenitised at 1150°C, cooled at a rate of 2°C/s to 850°C, held at this temperature for 1 hour and partially isothermally transformed between 700-450°C prior to gas quenching to room temperature.

Number Density of Grain Boundary and Intragranular Ferrite

Metallographic examination of these specimens was carried out to determine the amounts of grain boundary ferrite, intragranular polygonal ferrite and intragranular acicular ferrite as a function of the transformation temperature. Measurements were made of the number density of ferrite grains to quantitatively describe the refining effect of intragranular ferrite formation. Fig. 4 summarises the effect of isothermal transformation temperature on the number density of grain boundary ferrite and intragranular ferrite. It can be seen that the grain boundary ferrite started to form directly below the transformation start temperature and the number density of grain boundary ferrite grains increased slightly with decreasing the transformation temperature. More complex behaviour was observed for intragranular ferrite. There was no intragranulary nucleated ferrite at higher transformation temperatures, above about 650°C. The temperature range where the large density of intragranular polygonal ferrite was formed was narrow, between 650-600°C. On further lowering the isothermal transformation temperature, below 600°C, there was a tendency for the intragranular ferrite to change morphology and grow as ferrite sideplates. The number density of intragranular ferrite grains at 550°C was higher than that for 600°C.

The number density and morphology of the ferrite grains changed considerably after isothermal transformation at 450°C. Cooling to 450°C led to the formation of exclusively acicular ferrite structure with a more than 4 times higher density of acicular grains. The tendency for grain boundary ferrite formation was reduced and it was difficult to estimate the number density of grain boundary ferrite grains as they have a similar acicular morphology.

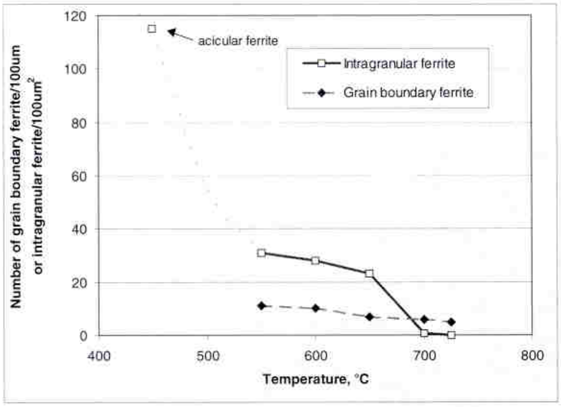
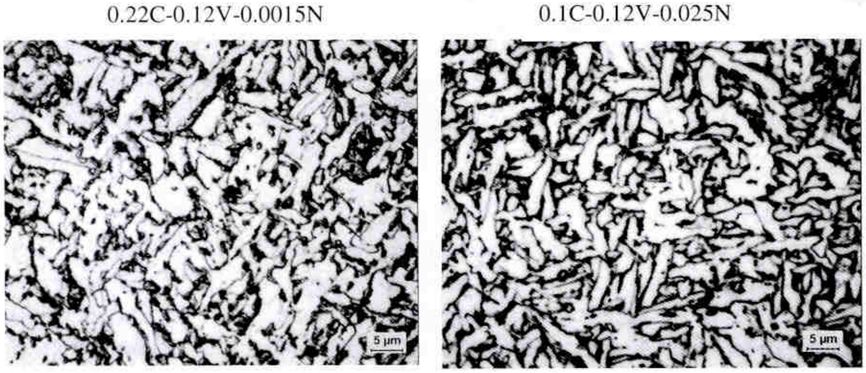


Fig. 4 Number density of grain boundary ferrite and intragranular polygonal and acicular ferrite grains as a function of transformation temperature in steel D13.

Effect of Vanadium on the Formation of Acicular Ferrite.

The experimental results described above indicate that intragranular acicular ferrite was the dominant microstructure after isothermal transformation at 450°C. This treatment was repeated for steels with low and high nitrogen contents in order to verify the role of N as well as VN in the formation of acicular ferrite in V-steels. The resulting microstructures are shown in Fig. 5. It is clearly seen that a similar acicular ferrite structure was obtained in the V-microalloyed steel with a very low N content of 0.0015%N and high N content of 0.025%N. In the low N steel no VN precipitates were expected to grow in austenite. This result confirms the suggestions of He and Edmonds[8] that there exists an effect of vanadium on the formation of acicular ferrite microstructure even in low nitrogen steels.


0.22C-0.12V-0.0015N

0.1C-0.12V-0.025N

Fig. 5 Microstructure of low N steel, D1 and high N steel A25 isothermally transformed at 450°C.

Analysis of Nucleation Sites

Nucletion of intragranular ferrite and the morphology of nucleating precipitates/inclusions was examined in several of the steels using both scanning and transmission electron microscopy. Fig. 6 shows a micrograph of the sample A25, which was water quenched from 630°C after ~5% transformation. In the intragranular regions, a number of inclusions exhibited ferrite growing with different morphologies. There were loops, blocks and caps of ferrite coating part of the inclusion surfaces as well as wedges and ferrite growing with more of a plate type of morphology. The formation of very long plates from individual inclusions was also observed.

A number of spherical non-metallic inclusions were present in the investigated steels. The majority of these inclusions were identified by EDS as manganese sulphides, Fig. 6. Other inclusions which were identified were oxides, mainly silicon-manganese oxides. Vanadium-rich particles, presumably nitrides were always visible in connection with inclusions. The fact that VN particles had formed in the austenite prior to transformation and nucleated ferrite crystals is clearly evident from SEM micrographs.

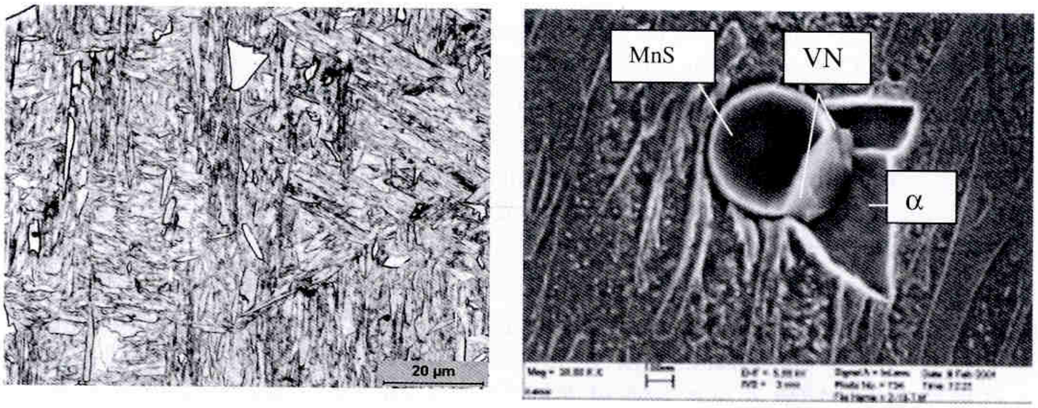


Fig. 6 Intragranular ferrite nucleus in steel A25, water quenched from 630°C after ~5% transformation.

Precipitation of V(C,N) in Ferrite

Precipitation of V-carbonitrides can occur randomly in ferrite behind the migrating austenite-ferrite (γ-α) boundary – general precipitation – or by interphase precipitation characterised by the development of sheets of particles parallel to the γ/α-interface formed repeatedly with regular spacing.

Interphase precipitation - Fig. 7 shows the typical morphology of interphase precipitation of V(C,N) in the investigated 0.10%C-0.13%V steels. Already from its appearance one can conclude that such a microstructure is formed in sheets parallel to the γ/α-interface by repeated nucleation of particles as the transformation front moves through the austenite.

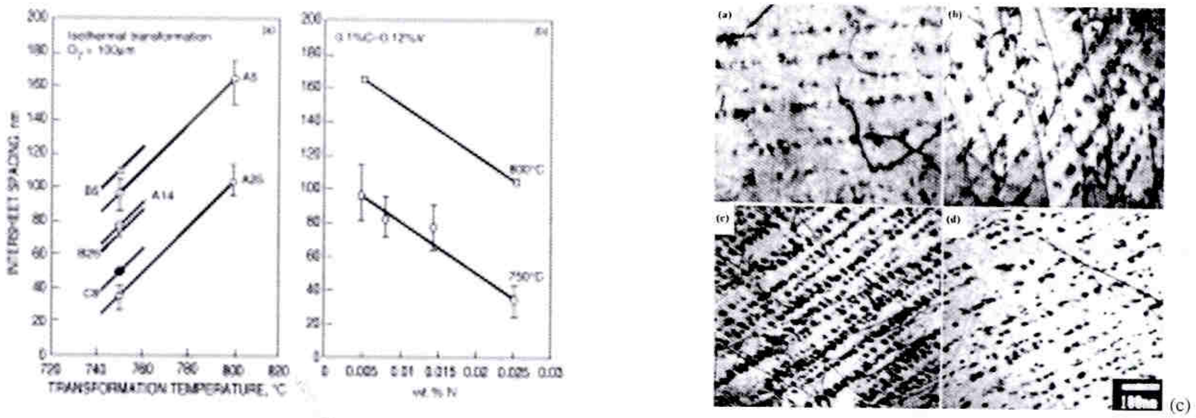


Fig. 7 The effect of transformation temperature (a) and N-content (b) on intersheet spacing of V(C,N) interphase precipitation for the investigated steels and electron micrographs of the selected steels (c) isothermaly transformed at 750°C for 500 s, (a) 0,0051%N (b) 0,0082%N (c) 0,0257%N (d) 0,0095%N, 0,04%C.

At high transformation temperatures, ~800°C, the interphase precipitation consists of irregularly spaced, and often curved sheets of V(C,N)-particles. With decreasing temperatures the incidence of curved rows of precipitates diminishes and the dominant mode is regularly spaced, planar sheets of particles. From about 700°C the interphase precipitation is commonly found to be incomplete, and random precipitation from supersaturated ferrite after the γ/α-transformation takes over progressively with decreasing temperature. A characteristic feature of interphase precipitation is that it becomes more refined at lower temperatures. Fig. 7 shows that the intersheet spacing is affected considerably by the nitrogen content of the steel. As shown there, it is diminished to almost one third at 750°C on increasing the nitrogen content from 0.005 to 0.026%.

Raising the carbon content above 0.10% C, will similarly decrease the A_3 temperature of the steel and suppress interphase precipitation. According to the present investigation, interphase precipitation was visible within proeutectoid ferrite or pearlitic ferrite only at transformation temperatures very close (within 40-50°C) to the eutectoid point. At lower transformation temperatures, formation of solute supersaturated proeutectoid ferrite and pearlitic ferrite occurs instead with random distributions of V(C,N).

A model for interphase precipitation based on the present results was developed recently[13]. This model is based on an analysis of the diffusional drainage of V due to of a moving α/γ-boundary to a sheet of interface precipitation, Fig. 8. The V-concentration in the interface increases gradually as it advances and reaches eventually a critical value when a new row of precipitation can be nucleated. It was shown that volume diffusion of V cannot explain the observed intersheet spacings and that a faster diffusion process is required. Hence, a new mechanism was put forward where the transport of V occurs by boundary diffusion in the moving α/γ-interfaces. The model exhibits good agreement with observed values of intersheet spacing, as well as the transition from interphase to general precipitation, and their dependences on C-, V-, N-contents and temperature.

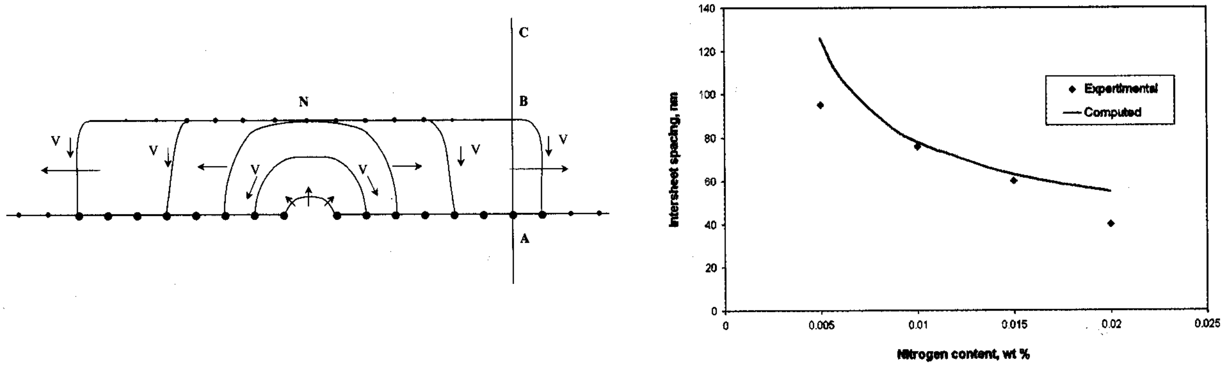


Fig. 8 Left figure shows schematically how the γ/α-interface bows out, expands sideways and reaches eventually material with sufficient V for renewed precipitate nucleation to occur. The transfer of V by boundary diffusion to the lower precipitate row is indicated. Right figure compares computed and observed intersheet spacings in 0,10C-0,12V steels as a function of N content.

General precipitation - Randomly distributed V(C,N) particles are most common in samples transformed at 550-650°C at all carbon contents. At these temperatures the precipitates were randomly distributed, either nucleated homogeneously in the matrix or on dislocations in the acicular ferrite structures. The fact that these particles had formed in the ferrite after transformation and not on the inter-phase boundary during transformation was evident both from their random dispersion and from the existence locally of several different variants of the B-N orientation relationship[9].

The results of detailed measurements of particle sizes on steels having different carbon and nitrogen contents are summarised in Fig. 9 after isothermal transformation at 650°C. As indicated in this figure both nitrogen and carbon cause a decrease in the particle size and at the same time an increase in volume fraction, so giving the lowest planar spacing at a given transformation temperature. For steel having 0.005% nitrogen and 0.1% carbon the particles were relatively sparse with an average diameter of ~ 11 nm. They became smaller and more dense with increasing nitrogen content, reaching ~ 6 nm at 0.022% nitrogen. An increase in carbon content from 0.10% to 0.22% causes significant refinement (about a factor of two) of the vanadium-rich particles at all nitrogen levels. A higher density of smaller V(C,N) suggests that carbon play an important role in precipitation of V(C,N). This agrees with the suggestion that metastable carbon content increases the supersaturation levels achieved in ferrite and the intensity of V(C,N) precipitation is higher.

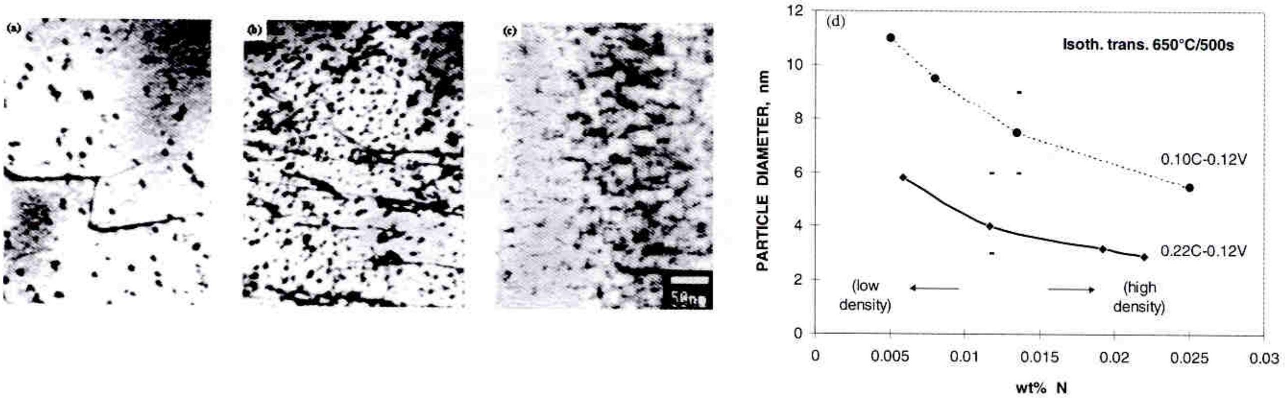


Fig. 9 Electron micrographs showing V(C,N) precipitation after isothermal transformation at 650°C in 0,12%V-0,013%N steels with different C-contents; (a) 0.04% C, (b) 0.10% C and (c) 0.22% C , and the effect of N and C on the size and density of V(C,N)-precipitates during isothermal transformation at 650°C for 500s (d).

Precipitation Strengthening

The precipitation hardening component, ΔRp, is the strength increment after the effects of the solid solution elements, ferrite grain size as well as pearlite content and pearlite interlamellar spacing have been excluded from the strength of the steel. The results presented below were obtained by subtracting the yield stress of the base steel, at the measured ferrite grain size as well as the contribution of pearlite from the observed yield stress at each transformation temperature and cooling rate.

Plots of precipitation strengthening against the nitrogen and carbon contents of the steels after isothermal transformation at 650°C are given in Fig. 10 (a) and (b) respectively. Each of the six classes of steel chemistry (Table 1) shows a similar dependence of precipitation strengthening on carbon and nitrogen contents. Therefore, the analysis was limited to isothermal temperatures 650°C where an equilibrium ferrite-pearlite structure was obtained. The following remarks can be made concerning the effects of carbon and nitrogen on precipitation strengthening results:

- the amount of ΔRp increases approximately linearly with nitrogen over the ranges investigated (~ 6 MPa for every 0.001% N). This agrees with the well established effect of nitrogen on ΔRp of V-steels,
- the effect of carbon on ΔRp is unexpectedly strong. For the present steels the strengthening contribution of carbon is approximately 5.5 MPa per 0.01% C.

These results demonstrate very clearly that the precipitation strengthening of V-steels increases significantly with the C-content of the steels. The role of carbon in precipitation strengthening of V-steels is discussed below.

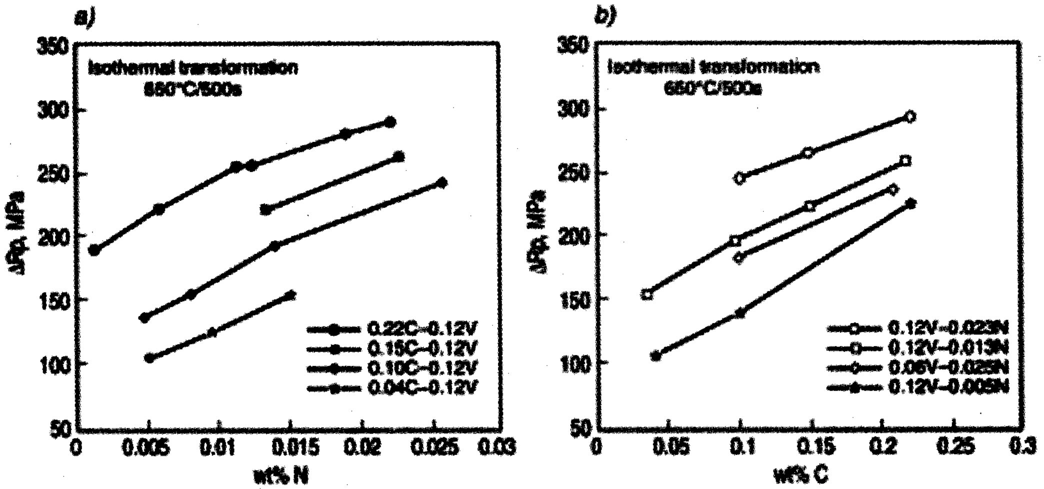
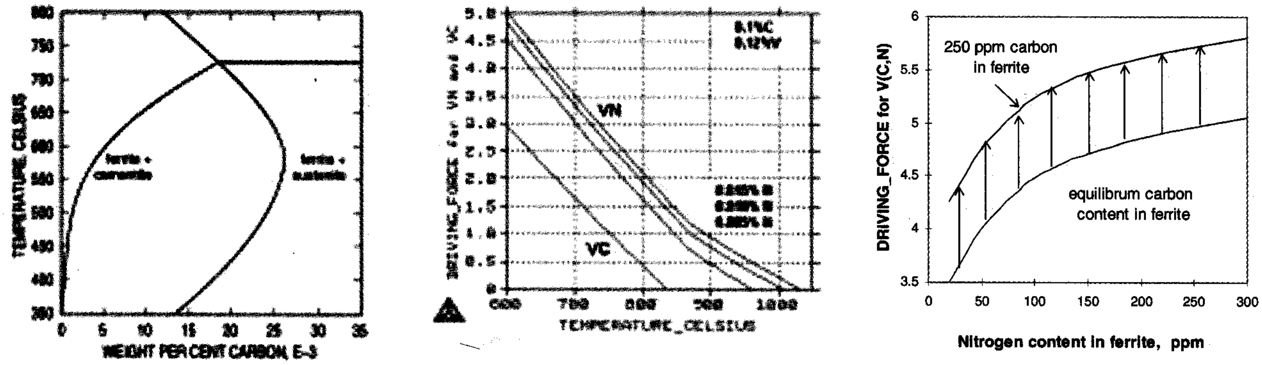


Fig. 10 Deduced values of precipitation strengthening for isothermally transformed V-steels (650°C/500s). (a) as a function of N-content (b) as a function of C-content.

The Role of Carbon in Precipitation Strengthening

Carbon content has usually been considered not relevant to precipitation strengthening when the precipitation occurs in ferrite. This was deduced from the fact of very restricted solubility of carbon in ferrite (which is normally supposed to be independent of the total carbon content in the steel). Most published literature does not suggest that differences in carbon contents in the range for structural steels (0.04-0.3%) should affect significantly the response of vanadium in these steels[4]. However, it was suggested recently that the effective carbon for precipitation in ferrite may be much greater in the times available during phase transformation[12]. Fig 11(a) shows the considerable difference in solubility of C in ferrite when governed by the γ/α and cementite/α equilibria below A_1. The solubility at 600°C is 5 times larger in γ/α than cementite/α. This metastable equilibrium between ferrite and undercooled austenite can greatly increase the solubility of carbon in ferrite thereby contributing to profuse nucleation of V(C,N) particles. The driving forces for nucleation of V(C,N) calculated for two different levels of dissolved carbon content equivalent to the equilibration ferrite+pearlite or metastable ferrite+austenite are shown in Fig. 11(b), (c). In the extreme case of 250 ppm carbon in ferrite the driving force for nucleation increases by about 25-30% in comparison with the equilibrium content at 650°C. This demonstrates that the nucleation of V(C,N) is indeed a function of the carbon content dissolved in ferrite. In reality the carbon content in ferrite should lie somewhere between these values depending on transformation kinetics, mainly phase boundary mobility.

The total C-content affects the kinetics of the transformation of γ to α and an increase in total carbon content displaces the pearlite formation to longer times as was shown in Fig. 1(c). The implication of this is that a larger chemical driving force for V(C,N)-precipitation will remain longer and hence promote profuse nucleation. Consequently, denser V(C,N) precipitation nucleates in ferrite, as was shown in Fig. 9. This role of carbon in precipitation strengthening is particularly significant in steels with higher carbon contents where the metastable condition is more extreme and prolonged.



(a) (b) (c)

Fig. 11 (a) Solvus lines for C in ferrite in equilibrium with cementite and austenite. (b) Chemical driving force, $\Delta G_m/RT$, for precipitation of VC and VN in 0.12% V steel. (c) Effect of equilibrium and metastable carbon content in ferrite on driving force for precipitation of V(C,N) at 650°C.

SUMMARY

The present work has concentrated on two strengthening mechanisms in V-microalloyed steels: (i) grain refinement by promoting the formation of intragranular ferrite, and (ii) the role of nitrogen and carbon in precipitation strengthening by interphase and random precipitation of V(C,N) in ferrite.

The experimental results strongly indicate that vanadium can by effectively used not only for precipitation strengthening but also for ferrite grain refinement. It was shown that vanadium contributes to the formation of two types of intragranulary nucleated ferrite; polygonal (idiomorph) ferrite and acicular (sideplate) ferrite. Intragranular polygonal ferrite nucleates on VN particles which grow in austenite during isothermal holding or slow cooling throughout the austenite range. The intragranular polygonal ferrite forms in the narrow temperature range, between 650-600°C for the investigated compositions. Acicular ferrite microstructure forms in V-microalloyed steels during isothermal transformation at lower temperatures (~450°). The acicular ferrite microstructure was obtained in V-microalloyed steels containing high, medium or very low nitrogen levels. This suggests that vanadium on its own can promote the formation of the acicular ferrite microstructure.

Vanadium is an effective and easy controllable precipitation strengthening element. The degree of precipitation strengthening of ferrite for a given vanadium content depends on the available quantities of carbon and nitrogen. It was confirmed that nitrogen is a very reliable alloying element, increasing the yield strength of V-microalloyed steels by some 6 MPa for every 0.001% N, essentially independent of processing conditions. Experimentally it is demonstrated that the V(C,N)-precipitation becomes denser and finer with increasing N-content. Carbon content, on the other hand, has usually been considered not relevant to precipitation strengthening when the precipitation occurs in ferrite because of the very small carbon content in solution in ferrite at equilibrium. The present results have shown that the precipitation strengthening of V-microalloyed steels increases significantly with the total C-content, ~5.5MPa/0.01% C. The explanation is that the C-content of the steel delays the pearlite (ferrite+cementite) formation and thereby maintains the higher, metastable C-content in ferrite given by the austenite/ferrite equilibrium for a longer time. This effect of carbon is particularly significant for medium carbon steels typically used for hot rolled bars and sections.

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