Improved mild steel for structural purposes

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THE majority of structures is usually constructed from steels with yield points not more than 17 t.s.i. but the recent developments in this field have led to structural steels with yield points between 20 t.s.i. and 40 t.s.i. Additions of about 2% alloying elements can produce appreciably higher yield strength. These low alloy steels show good impact properties and render them suitable for use at temperatures down to −50°C or even lower. The improvement in yield, tensile and impact properties are not achieved at the expense of weldability because carbon, the principal cause of welding trouble, is maintained at a very low level. The type of steels to be discussed has a low carbon 1·5 Mn base with very small quantity of alloying elements, and it would not be improper to term them as improved mild steel.

By making use of strengthening mechanisms in mild steel containing residual alloying addition, it would be possible to use a cheaper steel in structures needing a more highly alloyed steel, and a significant reduction in cost can be achieved. Also, by replacing mild steel with this improved variety, thinner sections can be used in structures and this would increase the payload of vehicle and a consequent reduction in freight cost.

Possible microstructures and trends in the modern development

The possible microstructures of the low alloy steels are (1) martensitic (2) bainitic and (3) ferritic-pearlitic. Of the possible structures martensitic type is to be eliminated as the steels of this structure type require elaborate heat treatment processes to develop the required strength and toughness. Moreover, the martensitic steels are not favoured because of their poor welding characteristics. The other structural type i.e. the bainitic and ferritic-pearlitic are of great importance for the structural applications and are discussed in detail below.

Bainitic type

Recently extensive work has been carried out by several authors[1,8] on the mode of bainite formation and the characteristics of the bainitic steels.

It has been shown conclusively from the kinetic studies that there are two modes of bainite formation having different values of activation energy of formation[6–7]. Bainite that forms at higher temperature is termed upper bainite and that which forms at lower temperature is the lower bainite. In the intermediate temperature range a mixed structure results. The upper bainite has similarity with very fine pearlite and the lower bainite is more akin to tempered martensite. The changeover from upper bainite to lower bainite has been suggested to take place within the temperature range of 300°C to 400°C[8]. Recently, Sackleton and Kelly[8] distinguished two types of upper bainite, viz. (a) transformed at about 500°C in which cementite precipitates at the boundaries of ferrite laths and (b) transformed at about 400°C where cementite precipitates within the ferrite region as well as boundaries.

Irvine and Pickering[1–3] have shown that the mechanical properties of the bainitic steels are dependent on the transformation temperature and steel composition. At a constant cooling rate, transformation temperatures are dependent upon the composition of the steel. Tensile properties increase progressively with decreasing transformation temperature but the impact properties show optimum values in the temperature ranges at which the high or the low temperature transformation products are obtained. This implies that either the upper or the lower temperature transformation products has the optimum combination of strength and impact toughness. Mixed structures show poor impact transition properties.

Commercially however, bainitic steels have not been very popular in the field of structural steels. Most of the earlier work on the bainitic steels has been carried out on the high carbon compositions having the least attractive properties for structural application. The development of the low carbon bainitic steels on the other hand, was retarded due to the difficulty of producing homogeneous bainitic structures without the formation of polygonal ferrite. This difficulty was however overcome by minute additions of boron (0·003%) in conjunction with 0·5% Mo to low carbon steel base.

The strength of such steels could be further enhanced by suitable alloy additions and weldability properties could be improved by limiting carbon contents to low

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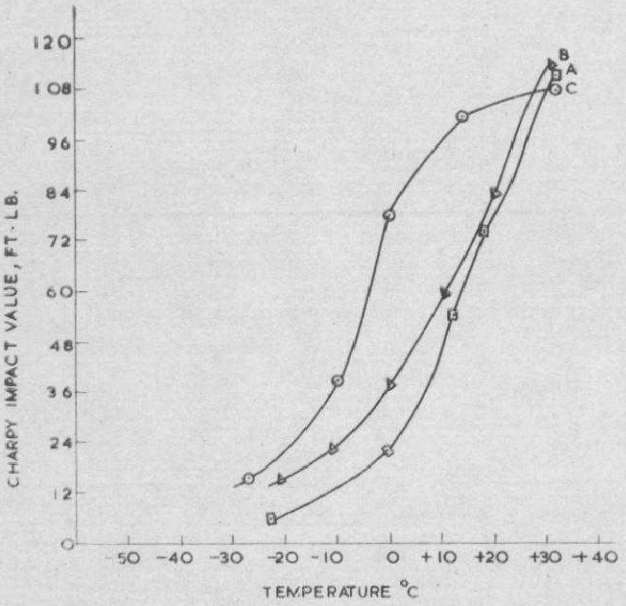


1 *Energy transition curves for niobium treated bainitic steels*

levels. Irvine and Pickering[9] have suggested the following composition of a bainitic steel for structural applications.

Steel No.	Chemical composition weight per cent				Tensile strength t.s.i.
	C	Mn	Mo+B	Cr	
1(Fortiweld)	0·10	0·75	0·50	—	45

The steel has 0·5% Mo+B base composition which gives uniform bainitic structure on normal cooling subsequent to the hot working operations or normalising.

The impact properties of steel 1 can be further improved by reducing the grain size which can be achieved by controlling rolling-finishing temperature or by addition of grain refining elements to the steel.

The authors have carried out an investigation to further improve its impact properties by the addition of niobium as a grain refiner. Compositions of the steels investigated are shown in Table I along with their austenitic grain sizes determined by McQuaid Ehn test. The impact transition curves for different Nb levels are shown in Fig. 1. All the steels were normalised at 950°C and show a gradual improvement in impact properties as the Nb content increases from 0·01% to 0·034%.

Though the bainitic steels possess much higher strength properties than the conventional mild steels and impact toughness of these steels could be improved by controlling grain size, these still suffer from the disadvantage of having relatively poor weldability compared to the ferritic pearlitic steels. Though the carbon content of these steels has been reduced to as low as 0·1%, considerable amount of skill and precaution is needed for the successful welding of these steels.[10]

TABLE I **Chemical compositions of the steels investigated and their austenitic grain sizes**

Steel No.	Chemical composition weight per cent						A.S.T.M. grain size No.
	C	Mn	Mo	B	Si	Nb	
A	0·11	0·65	0·60	0·0018	0·1	0·01	4
B	0·11	0·60	0·60	0·002	0·08	0·018	5
C	0·09	0·65	0·60	0·001	0·07	0·034	6–7

Ferrite-pearlite type

The most widely used structural steel with very low alloy additions is of ferrite-pearlite structure. Improved structural ferrite-pearlite steels are obtained by the additions of Mn, Si, Mo, V, Cr, Ti, Nb, Cu, Al, etc.

The desirable properties of structural steels are a condition of high yield strength, low impact transition temperature and good weldability. Methods that have been employed for obtaining these properties in ferrite-pearlite structural steels are enumerated as below :

(i) Solid solution strengthening : The ferrite can be strengthened by addition of substitutional alloying elements such as manganese and silicon. Interstitial elements such as C and N, though very effective solid solution hardeners have limited solubility in ferrite and adversely affect impact transition properties when present in large quantity and are thus of little importance in this respect.

Manganese is widely used in the structural steels as effective solid solution strengthener. In the ferrite-pearlite steels Mn increases lower yield stress by 2·11 t.s.i. per weight per cent of Mn.[11] Moreover, Mn has a pronounced beneficial effect on toughness and each weight per cent of this element lowers impact transition temperature in ferrite-pearlite steel by approximately 1°F per 0·01% Mn, up to 1·5% Mn[12]. However, it has been found that 1·8% Mn is the maximum limit for a grain refined structure in a normalised ¾″ diameter bar and the higher additions begin to promote bainite structure resulting in a substantial loss in impact properties.

Silicon appears to be a comparatively economic addition because of its more pronounced strengthening effect in comparison to manganese on weight basis, but it adversely affects the impact properties because weight per cent silicon increases impact transition temperature by about 44 to 50°C.[11,12] Therefore, Si addition is restricted to the residual quantities.

(ii) Refinement of grain size : So far as ferrite-pearlitic structural steels are concerned, grain size refinement

seems to be the most important single factor. The refinement of grain size can produce the optimum combination of yield stress and low impact transition temperature. The theoretical aspect of the effect of grain refinement can be clearly understood by the analysis of Petch's equation which can be expressed as follows.[13]

$$\sigma_y = \sigma_0 + K_y d^{-\frac{1}{2}} \qquad \dots \quad (1)$$

where σ_y is the lower yield stress, σ_0 is the friction stress having a temperature independent part σ_0^{I} (arising from the resistance of random solute atoms, fine precipitates, and lattice defects) and a temperature dependent part σ_0^{II} (arising out of Peierls-Nabarro stress). Ky is a measure of dislocation locking of solute atoms, d is the average grain diameter. Equation (1) gives a linear relationship between σ_y and $d^{-\frac{1}{2}}$. This implies a progressive increase in yield stress with decreasing grain diameter. By controlling grain size from the coarsest to the finest obtainable industrially the yield strength can be increased three times.[14]

Petch has also suggested a similar type of relationship for upper yield stress of polycrystalline iron with grain diameter as follows.[15]

$$\sigma_u\gamma = \sigma_0{}^u + \triangle\sigma_0 \log_{10}\left(\frac{1}{Nl^3}\right) + Kl^{-\frac{1}{2}}$$

where $\sigma_0{}^u$ = friction stress if all grains were deforming.

$\triangle\sigma_0$ = change in friction stress produced by a ten-fold change in strain rate.

N = number of grains per unit volume by actually deforming at σ_{uy}.

l = grain diameter.

$Kl^{-\frac{1}{2}}$ = the resistance to the spread of a slip band associated with the presence of the grain boundaries.

This relationship is said to agree well with experimental measurements in which lack of uniformity of the applied stress is minimised.

The temperature dependent part σ_0^{II} can be expressed as follows :

$$\sigma_0^{II} = \text{Constant } e^{-\alpha T} \qquad \dots \quad (2)$$

where α is a constant and T, the temperature.

Provided the temperature independent part σ_0^{I} is not too big, the σ_0 obeys a similar relationship so that

$$\ln\sigma_0 = \ln B - \beta T \qquad \dots \quad (3)$$

where B and β are constants.

It can be shown that ductile fracture will change to cleavage when

$$\sigma_0 = \frac{4\mu y'}{k' d^{\frac{1}{2}}} \qquad \dots \quad (4)$$

where μ is the rigidity modulus, y' is the effective surface energy associated with the growth of the crack and k' is a constant.

Substituting the value of σ_0 in equation (3), where T = Tc (transition temperature)

$$\ln\left(\frac{4\mu\gamma'}{k' d^{\frac{1}{2}}}\right) = \ln B - \beta Tc$$

$$\beta Tc = \ln B - \ln \frac{4\mu\gamma'}{k'} - \ln d^{-\frac{1}{2}}$$

$$Tc = \frac{\ln B}{\beta} - \frac{\ln C}{\beta} - \frac{\ln d^{-\frac{1}{2}}}{\beta} \qquad \dots \quad (5)$$

where $c = \frac{4\mu\gamma'}{k'}$

Equation (5) predicts a linear dependence of Tc on $\ln d^{-\frac{1}{2}}$ or less accurately upon $d^{-\frac{1}{2}}$. The slope of the curve is given by

$$\frac{dTc}{d \ln d^{-\frac{1}{2}}} = -\frac{1}{\beta}$$

This analysis of Petch's equation clearly demonstrates the two-fold effect of grain size refining viz, improvement in yield strength and decrease in impact transition temperature.

As mentioned before the grain size can be controlled either by controlling the rolling-finishing temperature or by the addition of grain growth-inhibiting elements such as Al, V, Nb, Ti, and Zr. In low carbon steels refinement of grain size to ASTM 11–12 would lead to a combination of high yield strength and good impact properties.[16]

Effects of specific grain-refining elements

Aluminium

Aluminium is a strong nitride former and an effective grain refiner in killed steels. The grain coarsening temperature increases with increasing Al content up to 0·08% but thereafter further addition causes slight decrease. Chatterjea and Nijhawan[17] and also Gladman[18] have established that aluminium-nitride is responsible for grain growth inhibition and its agglomeration causes grain coarsening.

Vanadium

Vanadium forms both carbide and nitride and is an effective grain restrainer up to 1000°C and in quantities not exceeding 0·1% in a 0·2%C steel.[19] Higher additions have no further effect on grain size but increases impact transition temperature due to the formation of excessive amounts of vanadium nitride.[20]

Titanium

Titanium is a strong carbide and nitride former and in smaller quantities (0·02%) causes grain refinement. The finer grain size persists at temperatures higher than the Al-killed steels. In plain carbon-manganese steel 0·02% Ti restrains grain growth up to 1100°C.[21] Halley found that when Ti was added in quantities as low as 0·005% to an aluminium treated steel containing 0·03% Al the grain coarsening temperature was raised by about 30°C.[22] The very strong carbide forming tendency of Ti reduces carbon content of the austenite and the amount of pearlite appreciably, when it is present in quantities as much as 0·20%.[23] Thus higher Ti contents decrease the strength and hardness of ferritic-pearlitic steels.

Niobium

Nb forms both carbides and nitride and refines the structure when 0·03–0·04% Nb is present in the steel and the grain coarsening temperature raised to 1050°C.[24] Nb has the advantage over other grain-refining elements that it can be added in semi-killed or balanced steels.

Zirconium

The solubility limit of Zr in ferrite in low carbon steel at room temperature does not exceed 0·14% and when the quantity of Zr added exceeds this limit, it would be present as free Zr C.[25] Zr does not have a pronounced effect as a grain refining element. The maximum grain coarsening temperature of 950°C is reached with a zirconium content of 0·13%, above which there is no further increase in grain coarsening temperature.[21]

TABLE II Grain-refining effects of common grain-restraining elements

Elements	Optimum per cent of elements wt. per cent	Approximate grain coarsening temperature °C
Zirconium	0·1	950
Vanadium	0.1	1000
Niobium	0·03	1050
Aluminium	0·06	1050
Titanium	0·02	1100

(iii) Reduction of carbon content : Plastic deformation of low carbon steel takes place initially in the ferrite grains and hence percentage pearlite has little effect on yield point. Since the carbon content does not affect the yield strength, its percentage can be reduced to 0·10 weight per cent or below. In low carbon structural steel U.T.S. increases by 0·253 t.s.i. per 0·008 wt. per cent increase in carbon (0·253 t.s.i. per 10% pearlite) but if there is about 5 to 6 t.s.i. difference between the U.T.S. and yield stress the U.T.S. is not important and so also the carbon content. On the other hand, the reduction of carbon content has two-fold beneficial effect i.e. (i) decrease in impact transition temperature and (ii) a pronounced improvement in weldability.

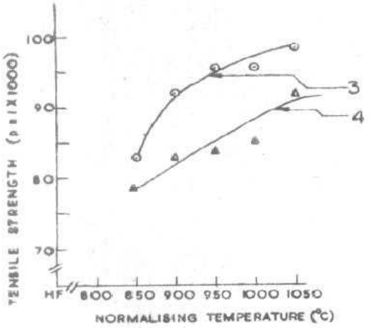
Strengthening by precipitation hardening

It becomes obvious from Petch relationship (eq. 1) that yield strength can be increased by increasing the factors viz. σ_0, K_y and $d^{-\frac{1}{2}}$. The effect of grain size has already been discussed. Usually there are enough interstitials (greater than 10^{-4} weight per cent required to pin all the dislocations in annealed or normalised mild steel) to from Cottrell atmospheres at all available dislocation sites and therefore the K_y value for the commercial low carbon steel is maximum and more or less constant and independent of composition and temperature. However, K_y value can be reduced by the addition of strong carbide or nitride formers such as Nb and V. This decrease in K_y value will lower the yield stress which can however be compensated by processes leading to the increase of σ_0.

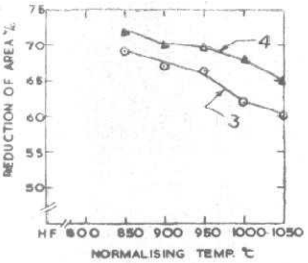
The most important factor affecting strength of low carbon structural steels is σ_0. The temperature dependent part of σ_0 which is related to Peierls-Nabarro force and is thought to be responsible for ductile-brittle transition in b.c.c. metals, is a constant at a constant temperature of testing. The temperature independent factor can be controlled metallurgically and depends primarily on solid solution and precipitation effects. The effect of solid solution has already been discussed.

In recent years extensive work has been carried out on the precipitation hardening effects in low alloy structural steels. The elements used for precipitation hardening should form stable carbides or nitrides having very low solubility in ferrite so that they precipitate out during cooling following normalising, annealing or the hot working processes. For effective strengthening the precipitates should form in the ferrite grains in preference to the ferrite grain boundaries or the ferrite-cementite interfaces.

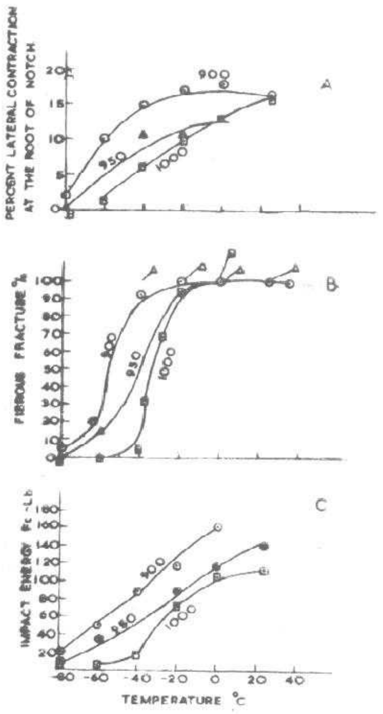
Extensive work has been carried out on the precipitation hardening effects of Nb in low-alloy structural steels.[26–31] The maximum useful addition is shown to be 0·06% and the mechanism of strengthening is attributed to the formation of coherent precipitates of Nb C (N) during cooling after hot working operations. However, on normalising up to a temperature of 1050°C niobium carbonitride acts as a grain restrainer but the strengthening in the hot worked product is lost due to loss of coherency and agglomeration of the niobium carbide precipitates.



2 *The effect of normalising temperature on the ultimate tensile stress of steels 3 and 4*



3 *The effect of normalising temperature on per cent reduction in area of steels 3 and 4*



4 *The effect of normalising temperature on impact transition characteristics of steel 3 ; Figures indicate normalising temperature°C*

Valuable results have been obtained by the addition of vanadium and nitrogen in plain C-Mn steels.[32,35] Vanadium nitride has a lower solution temperature than the niobium carbide (800°–1100°C) which dissolves on normalising and appears as coherent dispersion on cooling.

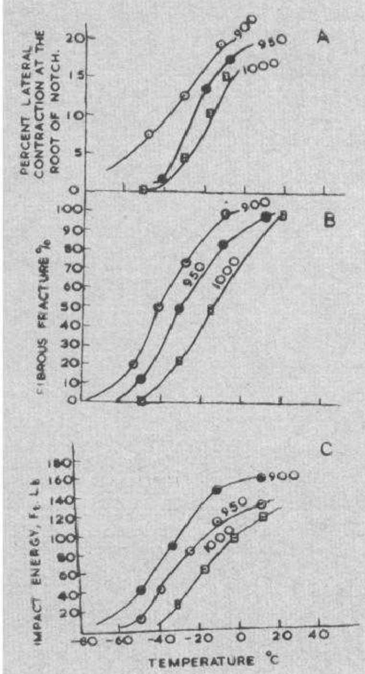
Development of high strength low-alloy weldable steels has followed an alloy design procedure based on the principles as enumerated above. Two compositions were selected based on these principles. The compositions studied are shown in Table III. These steels are of a low carbon base with 1·5% Mn, to which a grain-restraining element and precipitation hardener are added. Steels 3 and 4 were normalised at 850°C, 900°C, 950°C, 1000°C and 1050°C, for one hour. Steel 5 was normalised at 1000°C, 1050°C, 1100°C and 1150°C for one hour.

Steels 3 and 4

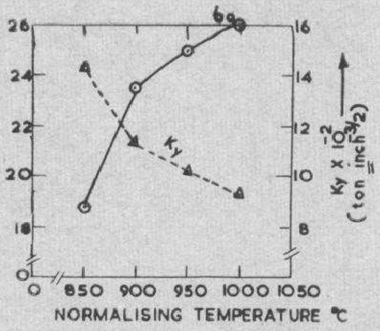
The tensile properties at various normalising temperatures are shown in Figs. 2 and 3. The results of impact tests carried out at room and sub-zero temperatures are shown in Figs. 4 and 5. The pearlite per-

TABLE III Chemical composition of the steels studied

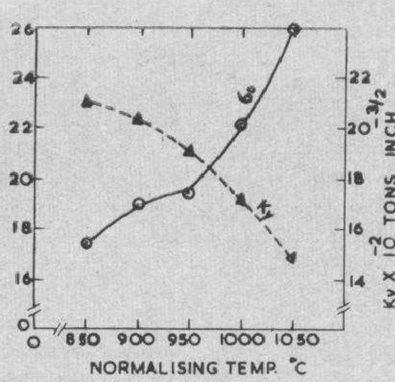
Heat No.	Weight per cent element								
	C	Mn	Si	P	S	Nb	V	Ti	N
3	0·07	1·55	0·187	0·018	0·045	0·06	0·20	—	0·009
4	0·08	1·59	0·081	0·016	0·042	0·06	0·18	—	0·0060
5	0·10	1·64	0·08	0·02	0·03	0·04	—	0·06	—



5 *The effect of normalising temperature on impact transition characteristics of steel 4 ; Figures indicate normalising temperature°C*



6 *Graph of σ_0 and Ky. vs. normalising temperature for steel 3*



7 *Graph of σ_0 and Ky. vs. normalising temperature for steel 4*

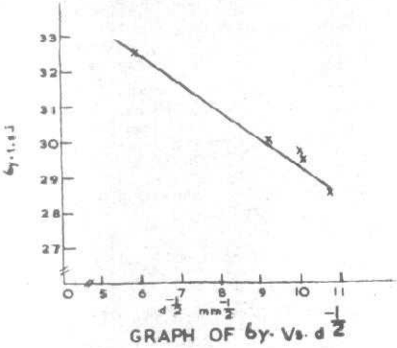
centage and grain diameter were measured by linear intercept method and are shown in Table V. The σ_0 values for different normalising temperatures were determined by extrapolating the stress-strain curves and the Ky values were calculated by utilising Petch's relationship. The values are indicated in Figs. 6 and 7.

Results indicate an increase in U.T.S., L.Y.S. and impact transition temperature with the increasing normalising temperature. The plot of σ_y vs. $d^{-\frac{1}{2}}$ (Fig. 8) shows a negative slope of the curve. With the increase in grain diameter the value of σ_y increases. The explanation of this negative slope lies in the strengthening effect caused by increasing normalising temperature.[25, 26] The increase in pearlite percentage with the increasing normalising temperature is not appreciable to account for the increased strength with higher normalising tem-

TABLE IV Comparative tensile values of Nb, V-N and Nb-V-N steels

Steels	Composition	Treatment	U.T.S. t.s.i.	L.Y.P. t.s.i.
A	C–0·15, Si–0·02, Mn–0·82, Nb–0·045 ...	As rolled	31·0	22·8
	...	Normalised 950°C	28·5	20·5
B	C–0·12, Si–0·03, Mn–0·90, V–0·05, N–0·014. ...	Normalised 950°C	32·1	23·4
C	C–0·21, Si–0·45, Mn–1·65, V–0·13, N–0·018 ...	Normalised 900°C	40·0	16·55
D	C–0·18, Si–0·38, Mn–1·57, V–0·18, N–0·024 ...	Normalised 900°C	39·8	30·5
Steel 3	... ...	Normalised 950°C	42·5	29·5
Steel 4	... ...	Normalised 950°C	37·4	23·7

A Ref : 28, B Ref : 33 D Ref : 32 C Ref : 34



8 *Graph of* σ_y *vs.* $b^{-\frac{1}{2}}$

perature. The increase of σ_0 with the normalising temperature, follows the same pattern as U.T.S., Y.S. and transition-temperature. The increase in temperature independent part of σ_0 may be either due to solid solution hardening or due to precipitation hardening. The solid solution effect of Nb or V in such steels is not large [30–33] and therefore cannot account for the increased strength with higher normalising temperature. Thus, this increase in strength seems to result possibly from the larger amounts of vanadium nitride (or carbide) going into solution with higher normalising temperature giving rise to increased precipitation. The electron micrographs of the extracted replicas show carbides and nitrides of both vanadium and niobium (Figs. 9 and 10).

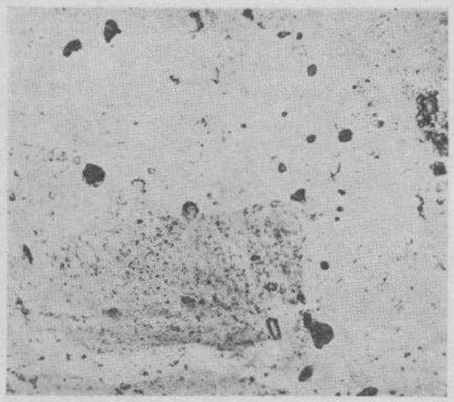
The increase in transition temperature with higher normalising temperatures is caused by the factors enumerated below:

(a) Increase in grain diameter with higher normalising temperature.
(b) Increased amount of vanadium nitride or carbide dissolving in the the steel as the normalising temperature is increased giving rise to progressively greater amounts of precipitates in the ferrite matrix.
(c) Increase of percentage pearlite content with the increase of normalising temperature.

Petch[12] suggested an increase in $\frac{lnB}{\beta}$ (equation 5) of about 5°C with an increase of 1 t.s.i. in σ_0. The values of σ_0 increases from 18·75 t.s.i. (normalised at 850°C) to 26 t.s.i. (normalised at 1000°C) in steel 3 and 17·5 t.s.i. (normalised 850°C) to 28 t.s.i. (normalised 1050°C) in steel 4. Thus the factor (b) seems to be mainly responsi-

TABLE V Properties of the steels studied

Cast No.	Composition	Normalising temparature °C	U.T.S. t.s.i.	L.Y.S. t.s.i.	Yield/tensile ratio	Percentage pearlite	Grain diameter micron	Transition Temp. °C 40 ft. lb. value	Transition Temp. °C 50% Fibrons	V.P.N. 30 kg. load
3	C =0·07 Mn =1·55 V =0·20 Nb =0·06 Si =0·187 S =0·045 P =0·018 N =0·009	850	37·0	27·0	0·73	4·8	7·63	—		161·0
		900	41·0	28·75	0·70	8·3	12·05	−48	42	167·0
		950	42·5	29·5	0·70	10·0	13·05	−40	−30	183·0
		1000	42·5	30·0	0·70	11·5	14·01	−24	−14	188·0
		1050	44·0	—	0·70	11·0	27·0	—	—	192·0
4	C =0·08 Mn =1·59 V =0·18 Nb =0·06 Si =0·081 S =0·042 P =0·016 N =0·006	850	35·0	28·5	0·81	6·4	8·6	—	—	158·0
		900	37·0	29·5	0·80	8·0	9·7	−68	−56	161·0
		950	37·4	29·7	0·80	9·0	9·8	−50	−41	163·0
		1000	38·0	30·0	0·79	10·0	11·9	−35	−35	166·0
		1050	41·0	32·5	0·79	10·0	28·0	—	—	185·0
5	C =0·10 Mn =1·64 Nb =0·04 Ti =0·06 Si =0·08 P =0·02 S =0·03	1000	37·0	28·75	0·78	13·3	16·0	−6	—	164·0
		1050	38·5	29·75	0·79	8·6	19·4	0	—	170·0
		1100	39·5	30·0	0·77	8·0	21·3	+8	—	180·0
		1150	48·25	37·0	0·77	4·6	21·1	—	—	205·0



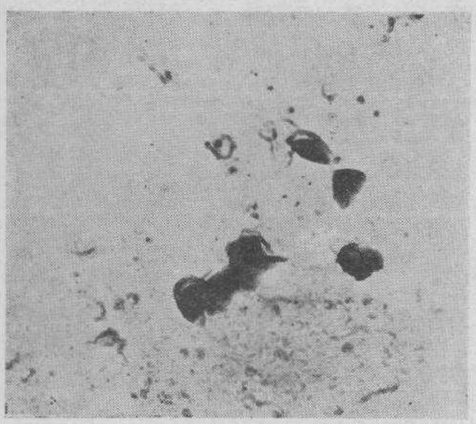
9 *Steel 3 normalised at 950°C carbon extraction replica ×19 000*

ble for the increase in transition temperature with the normalising temperature.

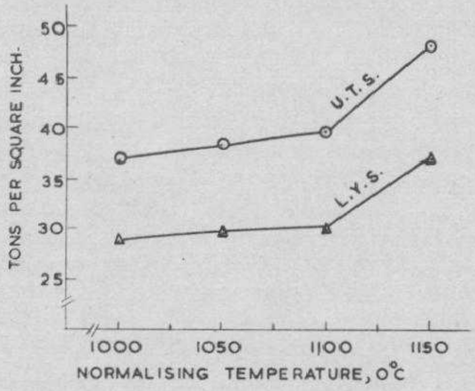
The combined addition of Nb, V and N to a C-Mn steel base resulted in a higher strength level than that obtained by addition of either Nb singly or a combination of V and N. Comparative values of strength properties are given in Table IV which shows a higher order of tensile properties of the experimental steels compared to Nb and V-N steels cited from literature. It should be noted that the steels C and D contain considerably higher carbon contents than the steels under investigation though the strength levels are comparable.

Steel 5

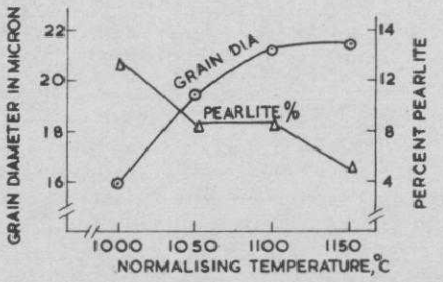
The variation of tensile and yield strength with the normalising temperatures is shown in Fig. 11 and in Table V. The change in grain diameters and pearlite percentage with the increasing normalising temperatures



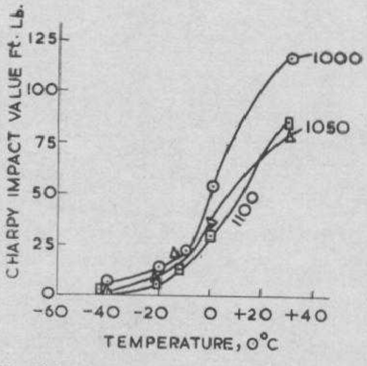
10 *Steel 3 normalised at 950°C carbon extraction replica ×25 000*



11 *Curves for variation of tensile and yield stress with normalising temperature for steel 5*



12 *The variation of grain diameter and per cent pearlite with the normalising temperature of steel 5*



13 *Energy transition curves for steel 5 ; Figures indicate normalising temperatures*

is shown in Fig. 12. The results of impact tests carried out at room and sub-zero temperatures are indicated in Fig. 13.

The yield strength of steel 5 is higher than that of the steels 3 and 4, but the impact-transition temperatures are higher than that of the steels 3 and 4.

There are certain percentages of bainite (below 15%) in the microstructures of these steels. This is possibly due to the comparatively higher normalising temperatures (above 1000°C) used. As the normalising temperature is increased there is a reduction in the proportion of pearlite appearing in the microstructure and a consequent increase in the proportion of bainite.

The precipitation hardening effect of Nb C(N) and the grain refining effect of TiC are possibly responsible for the increased yield strength. The presence of bainite in the structure has adversely affected the impact transition temperatures.

Conclusions

The Nb-V-N steels studied showed high yield strength and lower transition temperature compared to the conventional grades of mild steels. Since these steels also promise good weldability due to low carbon content, they offer a definite advantage as a high strength structural material.

Although the Nb-Ti steel investigated showed a higher yield strength than the Nb-V-N steels, its impact property proved to be inferior. This steel can be recommended for use where high yield strength is needed and some sacrifice can be made of the impact toughness property.

Despite the fact that the impact property of the bainitic 0·5 Mo-B-Nb steel is inferior to the ferritic-pearlitic steels, these would be useful as structural members requiring high order of strength.

Further work is needed to optimise the compositions and working conditions of the steels investigated.

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Discussions

Mr **L. J. Balasundaram** (NML) : What was the criterion used to define transition temperature? Was the same criterion adopted throughout when comparing the results?

Mr **R. Chattopadhyay** (Author) : The criteria for transition temperature used are : (1) energy transition criteria defined by the temperature at which an impact energy value of 40 ft. lb. was obtained. (2) ductility transition criteria defined by the temperature at which the lateral contraction at the root of notch of 5% was obtained and (3) fracture appearance criteria defined by the temperature at which a cleavage area of 50% was obtained on the fracture surface.

When comparing the results, only the energy transition criteria were used.

Dr **A. K. Seal** (B. E. College, Howrah): Did the authors determine the T T T diagram of the steels studied? How did they identify the bainite phase which is stated to be present in these steels in the normalised condition? In the electron micrograph projected it was stated that the Nb-carbides were on the grain boundary and V-carbonitrides inside the grain. How were they identified? My experience is that the usual in situ electron diffraction on the extraction replica is not a very satisfactory method. I would welcome the author's comments on this point.

Mr **R. Chattopadhyay** (Author): The T.T.T. curves of steels studied were not determined. Bainite was identified by optical microscope.

The precipitates of carbonitrides and carbides were identified by selected area electron diffraction technique on carbon extraction replica. This is a widely used method for "in situ" identification of the precipitates.