Theoretical considerations for substitutions in alloy steels

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THE development of materials possessing certain specific properties was followed on empirical basis in the past. This was specially true in alloy steels, where there are plenty of complications and variables, each effecting the properties sought in its own way. As the theoretical background of this behaviour was least understood, development of materials in the nineteenth and early twentieth century was based on very laborious approach which was time consuming on the first hand and costly on the other hand.

Quite lately attempts have been made to rationalise the metal science on more scientific basis, and although the dream of the designing alloys by slide rule, possessing the required properties, is still far from being achieved, the production costs and wastage in time can be avoided to a considerable extent by a proper application of the factors which have been understood lately. It is the object of this paper to describe these factors in detail so that the attempts in future will be based on more realistic and scientific lines than hitherto followed.

Structural steels

These steels are mostly low carbon mild steel having a carbon content 0·25% maximum and 1·5% Mn, giving a yield strength of 18 tons/sq.in. and tensile strength of 30 tons/sq.in. They have been used in the construction of bridges, buildings, autos and railways in the form of beams, girders, channels, joists, sheets, wires and tubes. The important requirements of a steel of this class are cheapness, good weldability and high toughness. Mild steel is an outstanding material on all these counts, and therefore it has been dominating the structural field for quite a long time.

(a) Need of substitution in structural steels

The need of stronger materials has put limitations on the use of conventional mild steel. In the structural field, it has been realised that the use of stronger materials in place of mild steel will reduce the section

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thickness in the design and this will bring considerable savings in weight, which will reduce the cost of the construction.

One obvious way of achieving this aim would be to improve the quality of mild steel, in such a way that it becomes stronger than its present value. Amongst the properties which need improvement are high yield/tensile ratio, low impact transition temperature and good weldability.

(b) Theory of substitution in structural steel

It has been shown[1] that for a majority of plain carbon steels in ferrite-pearlite aggregate, the following relationships hold good.

$$\text{Tensile strength (T.S.) tons/sq. in.} = 19\cdot1 + 1\cdot78\% \text{ Mn} + 5\cdot35\% \text{ Si} + \cdot25\% \text{ pearlite} + 0\cdot10d^{-\frac{1}{2}} \quad (1)$$

$$\text{Lower yield stress (Y.P.) tons/sq. in.} = 6\cdot74 + 2\cdot11\% \text{ Mn} + 5\cdot44\% \text{ Si} + 0\cdot23d^{-\frac{1}{2}} \quad (2)$$

$$\text{Impact transition temperature (I.T.T.C.)} = 63 + 44\cdot1\% \text{ Si} + 2\cdot2\% \text{ pearlite} - 258\% \text{ Al} - 2\cdot3d^{-\frac{1}{2}} \quad (3)$$

It is also known[2] that the weldability prediction from steel composition to avoid heat affected zone cracking is given by the carbon equivalent (C.E.) of the following equation which should not be more than 0·25 in any case.

$$\text{C.E} = \text{C}\% + \frac{\text{Mn}}{6} + \frac{\text{Ni}}{20} + \frac{\text{Cr}}{10} + \frac{\text{Mo}}{50} + \frac{\text{V}}{10} + \frac{\text{Cu}}{40} \quad (4)$$

A close examination of these equations suggest that although the tensile strength can be raised by increasing the amount of pearlite, which increases as the carbon content increases (equation 1), this does not bring simultaneous increase in the yield stress (equation 2). Also, increase in carbon will lead to an increase in the impact transition temperature (equation 3), and also deterioration in the welding properties (equation 4). It therefore seems rather impossible to alter the conventional properties of mild steel by manipulation of carbon or manganese content as an increase of both of

these elements bring an increase in pearlite, which is undesirable.

The basic theory of strengthening suggests that the total strength (T.S.) can be represented by an equation of the type :

$$T.S. = S + D + G + W + O + N + R + V + C,$$

where, S = solid solution hardening

D = dispersion hardening

G = grain size hardening

W = work hardening

O = order hardening

M = martensitic hardening

R = radiation hardening

V = vacancy hardening

C = cluster hardening

The significance of the addition in the above equation has got to be kept in mind while evaluating the effects, because it is the vectorial addition which is important here, and each of the hardening mechanisms influences the contribution from others.

From the point of view of easy control and manipulation, however, solid solution hardening, precipitation hardening and grain size hardening offer the best choice. Pickering[1] et al have calculated the effect of solid solution hardening in ferrite, which is given in Table I.

TABLE I   **Effect of alloy additions per wt. % on the lower yield stress**

C	330 Tons/sq. in.	Cu	2.5 Tons/sq. in.
N	330 Tons/sq. in.	Mn	2.1 ,, ,,
P	25 ,, ,,	Mo	0.7 ,, ,,
Sn	5 ,, ,,	Ni	0 ,, ,,
S	5 ,, ,,	Cr	7.0 ,, ,,
Nb	990 Tons/sq. in.		

This table shows that solid solution hardening does not offer much scope in comparison to precipitation hardening, except in case of carbon and nitrogen ; this choice, however, is limited by the fact that one cannot have more than .15% of these elements in solid solution at room temperature because of limited solid solubility.
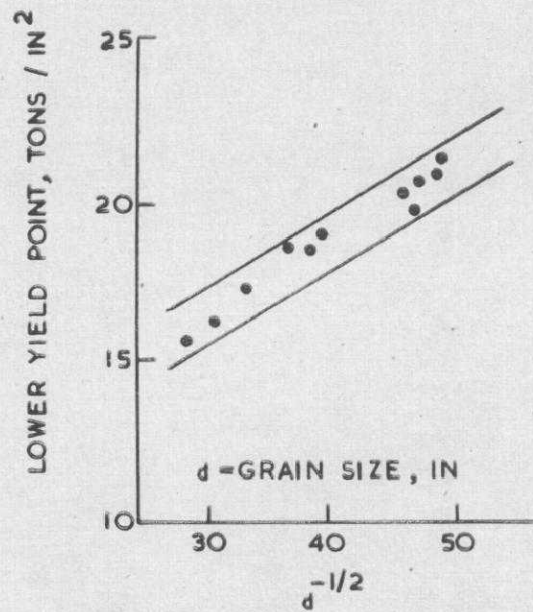
1   *Relationship between grain size and strength for C-Mn steels*

We are, therefore, faced with two possibilities, precipitation hardening and the grain size reduction as a practical solution for improving the properties of mild steels.

(c) Improved and substituted structural steels

That the lower yield stress is dependent on the grain size in a manner as shown by the equation :

$$\sigma_y = \sigma_i + Kyd^{-\frac{1}{2}} \qquad \dots \ (5)$$

was first observed by Petch. Fig. 1 shows the plot of the Petch equation. The reduction of grain size in steel is brought about by a careful control of Al N content, low roll finishing temperature and improved normalising practices. It has been reported[3] that an improvement of 5–10 tons/sq.in. can be brought about by such control. In many cases Nb and V additions have also been tried to reduce the grain size. Killing of steel by vanadium has an advantage that the vanadium-killed ingots do not show a difference in properties in the longitudinal and transverse directions as the Al-killed ingots do.[4] This difference in the directional properties of the Al-killed ingots is a major draw back in big forgings.

The application of precipitation hardening to the structural steels brings about the best overall properties without affecting the weldability. This has been done by substituting the lamellar iron-carbide (Fe_3C) by either NbC or V_3C_4 or some form of nitrides which precipitate very finely in the ferritic matrix. It is felt that the fine dispersion of the precipitate is more effective in impeding the motion of dislocations than iron-carbide in pearlite ; moreover, these precipitates do not
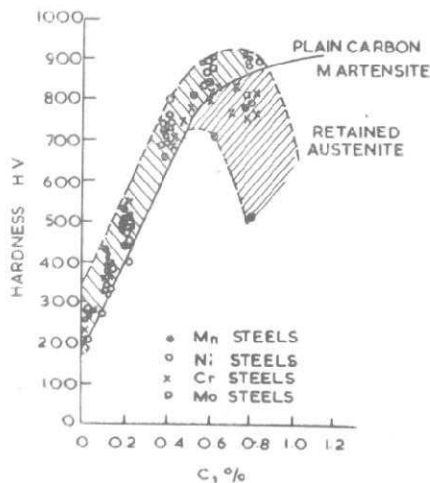
2 *Effect of alloying elements on the hardness of martensite*

introduce those deleterious properties in mild steel which are brought about by pearlite as shown by equations (3) and (4). Improvements on this line have been tried at many places, and pearlite free structural[5] steel and Nb or V treated steels having NbC or V_3C_4 precipitates in the ferritic matrix have been developed. It has been reported that steels having a yield stress of 30-40 tons/sq.in. can be obtained by use of 0.06% of either Nb or V, the use of higher content being not recommended on account of adverse effects. Theoretically, it is possible to dispersion-harden the ferritic matrix during normalising by either vanadium, niobium or titanium carbides or nitrides. The reason that vanadium and niobium carbides are preferred is that these carbides have lower solution temperature than the other carbides, the same applying to their nitrides. Some Russian workers[7] have tried to harden the ferritic matrix in structural steels by means of nitrides of aluminium, manganese and vanadium; however, American workers,[8] on the other hand, appear to favour hardening by carbides of niobium and vanadium, with a little bit of copper addition to increase the atmospheric corrosion resistance in the structural steels. These improved steels are increasingly being used in vehicles, trucks, buses, earth movers, power shovels, tanks and heavy construction equipment. Because of the weight reduction that is possible, these equipments realise benefits in the increased capacity.

Alloy steel used in the hardened and tempered state

These steels are used in the hardened and tempered state in order to develop adequate strength and toughness to meet the service requirements. Typical uses of steels of this class are, pressure vessels, axles, shafts, connecting rods, aero engine cylinders, gears, springs, machine tools and parts requiring resistance to wear and shock.

(a) Need for substitution in steels used in hardened and tempered state

The need of substitution here arises mainly because of scarcity of nickel, molybdenum, vanadium and tungsten. These elements are added in different proportions and combinations with each other in order to develop required characteristics mostly in the quenched and tempered state. It would be great savings of cost indeed, if manganese could substitute nickel in low alloy steels either alone or in combination with chromium and also some substitute could be found out for molybdenum, vanadium and tungsten. In order to know this we must know the reasons for their additions and their behaviours.

Theory of substitution

The important requirements of steels used in quenched and tempered state are : (a) high strength of martensite (b) adequate hardenability (b) high Ms temperature (d) low austenitising temperature (e) resistance to tempering and (f) freedom from temper-brittleness.

High strength of martensite

This is the most essential requirement because the stronger will be the martensite the better will be the inherited properties in the tempered state. Fig. 2 shows a plot of the strength of martensite with various alloy additions, and the strength appears to be completely independent of the extent and nature of the alloy additions except for carbon. As far as the strength of martensite is concerned, there is not very much to choose amongst the alloying elements and one can be easily substituted by the other.

Hardenability

Hardenability is the depth of hardness in a quenched specimen, and is a complex function of specimen shape, size, mass, composition, grain size, quenching media, etc.; the bigger is the hardenability with minimum of alloy additions the better will be the properties.

While it is difficult to evaluate the hardenability in a single parameter, attempts have been made to get a rough idea of it by the formula proposed by Grange and Kieffer as given below :

$$V_c = \frac{A_1 - T}{1.5\, t} \qquad \text{degrees/sec.} \qquad \dots (6)$$

where V_c = the critical cooling velocity

A_1 = the temperature of minimum austenite stability

T = the temperature at the nose of the TTT curve

t = the length of the incubation period at the temperature T, and since both t and T are related a decrease in T will increase t and thereby increase the hardenability.

The important parameter in the above equation is T which will ultimately determine the time t. For most low alloy steels, (where the alloy addition does not exceed 5%), the value[6] of T is between 550 and 600°C, irrespective of the type of alloying addition like Ni, Mn and Cr, except that they modify the shape of the TTT curve in their own way. For heavier elements like W, Mo and V the value of T falls below 500°C and this may be the reason why these elements are mere potent in increasing the hardenability.

High Ms temperature

The Ms temperature should be as high as possible, firstly because a higher Ms will bring less retention of austenite when quenched, and therefore it will avoid the need of sub-zero treatment, or multiple tempering which are needed to overcome the difficulties introduced because of the retained austenite. Secondly, if martensite is formed at higher temperatures, then the matrix will be able to accommodate the deformation stresses introduced due to the transformation, whereas martensite formed at lower temperatures can inherit micro-cracks because of the transformation stresses which will result in the service failure. Thirdly, if martensite is formed at higher temperatures, then it becomes auto-tempered (self-tempered) due to the higher temperature. The properties in the auto tempered state is lot better[14] than the properties of plain martensite which is just a solid solution of carbon in iron. There are various empirical equations connecting the Ms temperature with the composition; the one proposed by Andrews[9], however, seems to be the best.

The equation is:

$$Ms\ (°C) = 539 - 423C - 30.4\ Mn - 17.7\ Ni - 12.1\ Cr - 7.5\ Mo \qquad ...\ (7)$$

A close examination of this equation would suggest that, although manganese is more potent in depressing Ms, however, the difference does not seem to be considerable. Chromium and nickel stand almost on the same level as far as the depression of Ms is concerned.

Low austenitising temperature

A low austenitising temperature is always favoured because this will reduce the grain growth, prevent decarburisation and quench crack and burning. The equations[9] for austenitising is as follows:

$$A_3\ (°C) = 910 - 350C - 2.5\ Mn - 35\ Ni - 11\ Cr + 60\ Si \quad (8)$$

$$A_1\ (°C) = 723 - 10.7\ Mn - 16.9\ Ni + 29.1\ Si + 16.9\ Cr + 290A_5 + 6.38\ W \qquad ...\ (9)$$

It is clear from these equations that most common elements depress the austenitising temperature (A_3) along with the re-austenitising temperature (A_1). Although a low A_3 is always preferred, the simultaneous lowering of A_1 brings undesirable properties, because during tempering at high temperature part of the martensite gets coverted to austenite and therefore gets retained on cooling.

Also in high chromium steels, or steels containing ferrite stabilisers, the range of delta ferrite is also lowered down along with A_3, and therefore, delta ferrite is retained during quenching which is an undesirable constituent. In order to get rid of this trouble high chromium steels require a certain addition of either Mn, Ni or Co which are austenite stabilisers and therefore, raise the range of delta ferrite formation.

Resistance to tempering

This is also a very important property. Apart from producing the necessary hardenability, the other function of the alloying elements is to retard the low temperature tempering reactions, so as to get the best combination of strength and ductility. Table II shows a plot of retardation of tempering due to various alloys additions, according to Irvine.[3]

TABLE II Retardation of tempering (change in Hv per wt. % of alloying elements)

C	—40	Ni	8
Cr	0	W	10
Co	8	Mo	17
Mn	8	Si	20
	V	30	

As the Table indicates that theoretically most elements can bring a retardation of precipitation reactions during low temperature tempering except carbon and chromium which are the worst; however, the best choice would be an element which would cause maximum retardation with the minimum depression of Ms temperature. Silicon and cobalt are such elements.[11] Therefore a combination of Ni-Si-Co or Mn-Si-Co alloy steel would be expected to be most useful in this category. These steels, however, do not retain their high strength at temperatures above 200°C, however for many practical applications such as super heater tubes, pressure vessels, turbine blades, etc., high strength level is required to be maintained up to temperatures of 400–500°C; hence a greater tempering resistance is required.
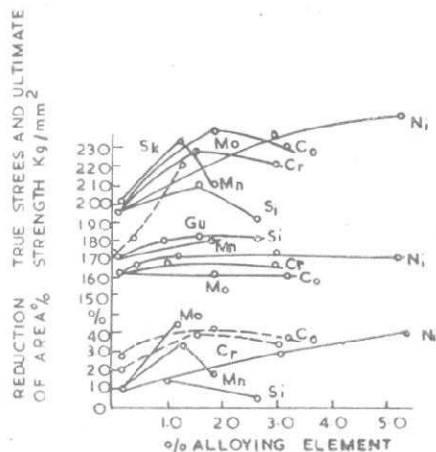
3 *Effect of alloying elements on the mechanical properties of steel with 0·35 per cent C, hardened and then tempered at 200°C (after S. T. Kishkin and S. Z. Bokstein)*

In order to get this strength secondary hardening is introduced in the tempered state. The elements which can impart this property are W, Mo, and V—Mo and V being cheaper they are preferred to W. Vanadium is the most useful element from this point of view because of the slow growth of its carbides; this, however, suffers from the difficulty of high solution temperature. The choice is left to Mo, and the intensity of the secondary hardening can be increased by dissolving elements which would decrease the lattice parameter of ferrite[11] or increase the lattice parameter of carbides by dissolving in it. Silicon reduces the lattice parameter of ferrite, whereas Ti, Nb and V increase the same for molybdenum carbide. Silicon and vanadium are therefore, preferred in comparison to others. Chromium is the worst from this point of view, because it decreases the lattice parameter of carbide and promotes overaging by producing unstable carbides. Ni and Mn also promote overaging. Secondary hardening can also be introduced by the formation of nitrides or the hardness at high temperature can be maintained by introducing inert nitrides which do not coagulate at high temperature. This has formed the basis of development of tool steels.

Most low alloy steels containing Ni, Mn or Cr are susceptible to temper brittleness in the quenched and tempered state. The only remedy seems to be an addition[10] of ½% Mo or P or a combination of both.

Design of alloys in the quenched and tempered state

While designing alloys for use in the above microstructural state, the following considerations have got to be kept in mind : (a) to keep the carbon content as low as possible comensurate with the required strength (b) to keep the amount of alloying elements minimum to give the required hardenability, in greater

amount the alloying elements promote retention of austenite and quench crack, (c) to retard the low temperature tempering by the addition of silicon, when steels are used at low temperature (d) to introduce secondary hardening by molybdenum and vanadium in combination with silicon for steels used at high temperature (e) finally there does not seem to be a lot of difference between nickel, manganese and chromium in low alloy steels from theoretical considerations. This is clear from Fig. 3 where the properties have been compared in the quenched and tempered state for various alloy additions.

It is rather unfortunate that the potentialities of manganese have not been exploited fully in the past for low temperature applications in the quenched and tempered state. Except for the fact that manganese depresses Ms to a greater extent that nickel or chromium, which can be compensated by the addition of silicon, which is further effective in retarding tempering, there does not seem to be any reason from the considerations discussed above as to why it cannot be used extensively in hardened and tempered state. It must be remembered, however, that Mn promotes inclusions in steels, and Mn steels are slightly coarse grained, and, therefore, a certain amount of research is needed before evaluating the effect of all these diverse factors.

Ultra-high strength steels

These are the steels possessing yield strength in excess of 100 tons/sq. in. They have been used for a variety of critical applications including components of aircraft, missiles, naval vessels, pumps, valves, manufacturing machinery, as well as bolts, pins, fittings, shafts, springs, punches and dies.

Theoretical consideration in the development

The various combinations and factors which can be utilised for getting stronger materials have been enumerated earlier. It has already been pointed out that the most potent strengthening effect is produced by the precipitation hardening. The effect depends on (a) the volume fraction of the precipitating phase (b) the number and size of the precipitate (c) the nature of the precipitate itself and (d) the interaction of the precipitate with the matrix.

Meiklejohm and Skoda[12] have proposed a relationship of the type :

$$\sigma_y - \sigma_0 = a\ f^{\frac{1}{3}}\ (k - f^{\frac{1}{3}}) \qquad \ldots \ (10)$$

where $\sigma_y =$ yield stress of the matrix containing the dispersion

$\sigma_0 =$ yield stress of the matrix

f = volume fraction of the precipitate

a = constant

k = constant, depending upon the precipitate shape

This relationship does not hold good for precipitate beyond 8% of the volume fraction and it seems that the precipitate spacing exerts a powerful influence beyond this level. This has been shown by the equation of Ansell and Lend,[13] which shows that

$$\sigma_F = \sqrt{\frac{GbG^*}{2Sc}} \qquad \ldots \text{(11)}$$

where σ_F = flow stress
  G = shear nodulus of the matrix
  G^* = shear nodulus of the dispersed phase
  b = Purger's vector
  S = Particle spacing
 c = constant

It is not possible to get quantitative idea of the extent of hardening by a dispersed phase, but it has been roughly estimated that the yield strength can be increased by a factor of 10 over the matrix phase by proper dispersion of the phase.

Substitution in ultra-high strength steels

This theoretical background has helped to a considerable extent in the development of ultra-high strength steels. The essential requirements are a strong matrix as (equation 10) would suggest and dispersion hardened critically by a strong phase (equation 11). Several combinations are possible and depending on these combinations ultra high strength steels of martensitic, cold rolled austenitic and semi austenitic have been developed. The best known example of the martensitic class is the 18% Ni maraging steels. These steels are martensitic as annealed, but, because of the extra low carbon and high nickel the martensite is quite tough and ductile and hence amenable to welding, forging and machining operations. Several chromium based maraging steels, have also been developed. The Ni in original 18% maraging steel has been partly substituted by Mn, and the maraging steel thus produced seems to be as good as the original ones. The 12–2 maraging as identified by Foote[8] would result in considerable production economics because of lower nickel. Tables III, IV and Fig. 4 give the properties of 12–2 maraging steels.

TABLE III   Mechanical properties of the substitute maraging steel

Elongation	—	12·00—14·00% in 1″
Reduction	—	53—60%
Yield strength	—	90—100 tons/sq. in.
Notch tensile strength	—	154—160 tons/sq. in.
N.T.S./T.S. ratio	—	1·47—1·53
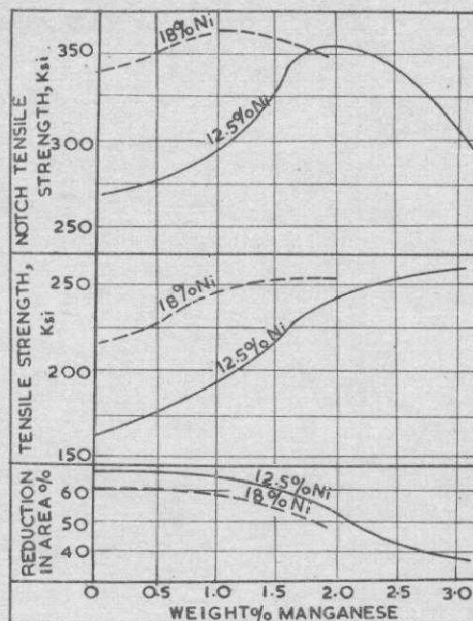
4   *Mechanical properties of the substituted 12–2 maraging steel*

TABLE IV   Nominal composition of the substitute maraging steel

Ni	Co	Mo	Mn	Ti	Al	Fe
12·5%	8%	4%	2%	0·2%	0·1%	Balance

It must be remembered here that the precipitating phase is an intermetallic compound and, therefore, the hardening effect in these steels can be made more or less independent of carbon content. The conceivable intermetallic compounds are B_2A ordered fcc orthorhombic or hexagonal phase, B_2A hexagonal or tetragonal Laves phases, BA bcc phase, B_7A_6 trigonal phase, cubic X phase and tetragonal σ phase, where B is the base transition element and A is the addition element. While any of these compounds can serve the purpose of strengthening, there are certain practical considerations which has got to be met by the hardening phase. These considerations are : (1) the precipitate must be 30% or more by volume fraction (2) interparticle spacing about 500 Å, (3) hardness and the shear strength of the phase should be much greater than the matrix, (4) minimum of mismatching for long time phase stability and (5) ductility.

Judging in these contexts, intermetallic compounds in order of their superiority are B_3A, B_2A and BA type. However, all combination of elements do not form these compounds and a certain electron/atom ratio has got to be satisfied in order that these compounds should form. The most common examples, for these intermetallic compounds are Ni_3Al, Fe_2Mo, Ni-Al, Ni-Ti, Ni-Al-Ti, etc.

In the development of the ultra-high strength steels, one can start with either ferritic, martensitic or austenitic matrix, and then dispersion-harden them by either a BA type of compound as in nimonic alloys or B_2A type compound as in ferritic steels or B_3A type in maraging steels. The requirement of austenitic, martensitic and ferritic matrix besides the requirement of corrosion resistance, seems to imply that Ni-Cr or Mn-Cr appears to be a must addition to a certain extent.

High temperature steels

These are the steels used in the range of 500°–600°C under stress for turbine blades, high temperature tubes, etc. The simple creep equation, in which activation and stress indices as well as the grain size are taken into account is given below.[13]

$$\epsilon = A d^m \alpha^n \exp [-\theta c/RT] \qquad \dots (12)$$

where

ϵ = creep rate (secondary)

A = constant

α = applied stress in p.s.i.

d = mean grain diameter

m = constant

n = constant (stress exponent)

θ_c = activation energy for creep

R = gas constant

T = temperature

If any of the three variables are known, the fourth can be determined. The important parameter here to consider in the designing high temperature materials is A. The smaller the value of A the better will be the properties. The value of A is closely linked with the modulus of elasticity E, and diffusivity D. These parameters in turn are dependent on crystal structure and melting point of the elements. Besides these, there are some comparatively smaller considerations like oxidation to corrosion, evaparation in vacuum, rank of phase stability, etc. which are got to be considered. Tables V and VI show these properties which have been adapted from the work of Decker and Dewitt[16] with slight modification to include Mn for the comparison purposes.

It is interesting to see from these tables that from the point of view of thermal expansion, density, thermal conductivity, Young's modulus of elasticity E, and co-efficient of diffusion, chromium based material seems to be an ideal choice. However, oxidation resistance and phase stability will favour an alloy chromium and Ni. It is rather, unfortunate to see that Mn does not have a bright future as a base material ; however, Mn in combination with chromium will provide a good substitute for Ni-based materials, because from the dislocation point of view, the high temperature materials should be such as to reduce the rate of climb and cross-slip. It is clear that the best parent metals on

which to base high temperature material will be those in which climb and cross slip are difficult, these include fcc and cph metals of low stacking fault energy for which the slip dislocations readily dissociate.

As chromium based materials themselves cannot have an fcc structure, whereas the addition of Mn can provide such structures, this approach seems to be theoretically feasible. Such structures, however, can be precipitation hardened by introducing any of the types of inter-metallic compound hardening which has been discussed earlier.

TABLE V Adapted from Decker and Dewitt[16] creep controlling factors at 700°C[16]

Cr	32·00	10^{-17}
Fe	18·1	10^{-16}
Co	23·4	10^{-14}
Ni	23·2	10^{-15}
Mn	12·0	10^{-13}
Desired	High	Low

TABLE VI Adapted from Decker and Dewitt[16]

Base Cr	Density g/cm² RT 7·1	Coefficient of expansion 10^{-6} per and at RT	Thermal conductivity (cgs at RT) 0·16	Evaporation in vacuum in gm/cm²	Rank of phase stability
Fe (bcc)	7·9	6·5	0·17	3×10^{-2}	4
Fe–18 Cr	7·9	12·1	0·04	8×10^{-3}	5
8 Ni	8·6	14·8	0·16	—	3
(fcc)	8·0	12·5	0·02	$1·4 \times 10^{-3}$	2
Co	8·9	12·5	0·21	—	
Co–28	8·4	13·3	0·04	$2·4 \times 10^{-3}$	1
Cr–22 Fe	7·44	12·5	—	—	
Ni		23·0		—	3
Ni–20 Cr					
Mn					1=most
Desired	Low	Low	High	Low	stable

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Mr **B. N. Das** (NML): I congratulate the authors for a most thought provoking paper. I get confused by the multitude of formulae that have been published for C equivalent. I have come across 180 of them and on trying 5 of the most successful ones for weldability I obtained widely different and puzzling results. Will the authors tell me how to use these formulae?

Secondly in the formula, the authors have presented for the yield point, I do not see any parameter for nitrogen. Aluminium can take care of nitrogen but if the latter is in free stage how would one deal with it? The same also applies to transition temperature ranges. Again I find that none of these formulae takes into account the conditions of the steel, which is of great importance because transition temperature will vary accordingly. I would like to know if the authors have verified these formulae by laboratory work.

Dr **P. Tewari** (Author): These formulae are only an attempt towards a systematic, scientific approach to certain problems. Instead of a random search these formulae can be applied to a limited extent and these do form the basis of future scientific development. It is only the assumption of these points that a formula is postulated and used.

Dr **B. R. Nijhawan** (NML): I might clarify that these formulae very often cover the purest form of metals which we do not get in practice; at times a formula has no real practical importance except as guide lines. I agree with Mr Das that we would not go far if all our designs were to be based exclusively on different formulae. I know of several cases where the theory is not able to explain many of the observed practical phenomena and I have often raised these queries in relation to carbon, nitrogen, grain refinement, impact toughness improvement by stabilisation of steel or through the introduction of aluminium in minute quantities. There are several theories to explain these phenomena but the formulae should not be taken literally. Practical implications in producing an alloy are much more stringent in terms of the properties obtained and that is where lies the value of the technical work. One way of interpreting this paper could perhaps be that it brings out the fallacy of the formulae as much as their usefulness. Trials must be supplemented by practical work to prove or disporve these mathematical expressions but nevertheless if we were to abandon them altogether, we would be eliminating all theoretical thinking and there also we might land ourselves in difficulties.