

# DEVELOPMENT OF CREEP-RESISTANT ALLOY STEELS

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**T**HE major part of the industrial uses of high creep-resistant alloys at elevated temperatures and fairly high stresses has been due to the development of the gas turbine and the jet-propulsion engines. Although Howe<sup>1</sup> as far back as 1885 made investigation on the extension of metals as function of time, and creep data were reported by Andrade<sup>2</sup> in 1911, there was very little technical interest in the phenomena of creep or in creep-resistant metals and alloys until shortly after World War I, when better materials were required for aero-engine parts, for the petroleum and chemical industries at fairly high temperatures (400°-600°C.) and sometimes under severely corrosive conditions, for parts of industrial furnaces at high temperatures and for steam turbines. The demand of materials resistant to creep steadily increased during World War II. But the greatest resistance to creep at high temperatures and fairly high stresses have been due to the development of gas turbine and jet-propulsion engines. An attempt has been made in this paper to indicate briefly some of the metallurgical factors involved in the development of high creep-resistant alloy steels with particular reference to turbine materials.

## A Brief Outline of the Problem

(a) The gas turbine parts, particularly the blades, need a material resistant to fatigue and corrosion. The designer would be happy if, in addition, he is given a material with low thermal coefficient of expansion and fairly low density to minimize the centrifugal stresses at high speeds. The manufacturer would be happy to shape a material if it has good working properties and high

machinability. The alloy to be developed should thus possess:

- (i) resistance to creep at high temperatures and at fairly high stresses
- (ii) resistance to corrosion
- (iii) resistance to fatigue
- (iv) good forgeability
- (v) good machinability
- (vi) low density
- (vii) low thermal coefficient of expansion.

In order to get the desired alloy it has to be borne in mind that some elements when added to the base metal may be beneficial for a certain group of properties but quite harmful for some others. Some additions, for example, are helpful for resistances against creep and oxidation but detrimental to forging characteristics and machinability. Tungsten alloys, for example, may have very good creep resistance but will have poor oxidation resistance. Moreover, due to the high specific gravity of tungsten, the centrifugal stresses may be several times than those for iron base alloys. Cobalt helps resistance against creep, but seriously affects forgeability. Ferritic steels are better for some properties while austenitic steels are better for some other properties. A variety of metallurgical features are thus involved in the development of high creep-resistant alloys.

(b) When some desired specification of an alloy is selected, extra care need be given to each of the different stages, viz. for melting, casting, forging and heat treatment of the alloys. A good alloy may be spoilt due to faulty forging or heat treatment.

(c) Creep is a structure-sensitive property to an extent far greater than most of other mechanical properties. Hence no matter how well an alloy is melted, cast, forged and/or heat-treated, it is very essential that it is

tested at or near the operating temperature. In fact, 'the reputation for reliability', to quote Mr. Oliver, the Director of Research, the B.S.A. group, and Mr. Harris, the Research Manager, Jessop & Sons Ltd., Sheffield, 'achieved by British gas turbine builders is to some extent a result of the British practice of rejecting ruthlessly any products which appear to be only slightly sub-standard'. There is no reliable method yet known to predict the creep properties of an alloy. To obtain reliable design data it is essential to carry out actual creep tests.

(d) Availability and cost of the raw materials have to be seriously considered in developing an alloy. There are sufficient theoretical and practical reasons to expect that one or more of the high melting point elements in the periodic table as indicated in Table 1 will be helpful to impart increased creep resistance. But for reasons of high cost and low availability many of these are

not ordinarily used and only the following elements need be considered:

- (i) Cr, Mn, Fe, Co, Ni
- (ii) Ti, V, Mo, Cb, W, Zr, Ta.

Elements in class (i) form the large bulk of most of the creep-resisting alloys of both the ferritic and austenitic groups. Their effects are too well known. Elements in class (ii) are added primarily to form complex carbides and thus to enhance creep properties by increasing the so-called 'complexity effect'. Of these, the first four, viz. Ti, V, Mo and Cb, are more often used.

**Some Basic Considerations**

Creep occurs essentially due to flow of a material under a given set of temperature and stress conditions. The creep strength of a material is defined as the maximum stress that can be applied at a given temperature to cause flow of the material either at a given rate or not exceeding a certain amount

**TABLE 1 — ELEMENTS USEFUL FOR THE DEVELOPMENT OF HIGH CREEP-RESISTANT ALLOYS**

mp °C →	1800	1735	1950				
	Ti	V	Cr	Mn	Fe	Co	Ni
gm/cm <sup>3</sup> →	4.5	5.7	7.1				
mp °C →	1750	2000	2625				
	Zr	Nb	Mo				
gm/cm <sup>3</sup> →	6.4	8.6	10.2				
					2500	1970	1555
					Ru	Rh	Pd
mp °C →	1700	3000	3010				
	Hf	Ta	W				
gm/cm <sup>3</sup> →	11.4	16.6	19.4				
					2700	2454	1774
					Os	Ir	Pt
					22.5	22.4	21.5

N.B. ALL HAVE HIGH MELTING POINTS, LOW ATOMIC VOLUME  $(\frac{\text{ATOM WT}}{\text{DENSITY}})^{1/3}$  AND LOW ATOMIC HEAT.

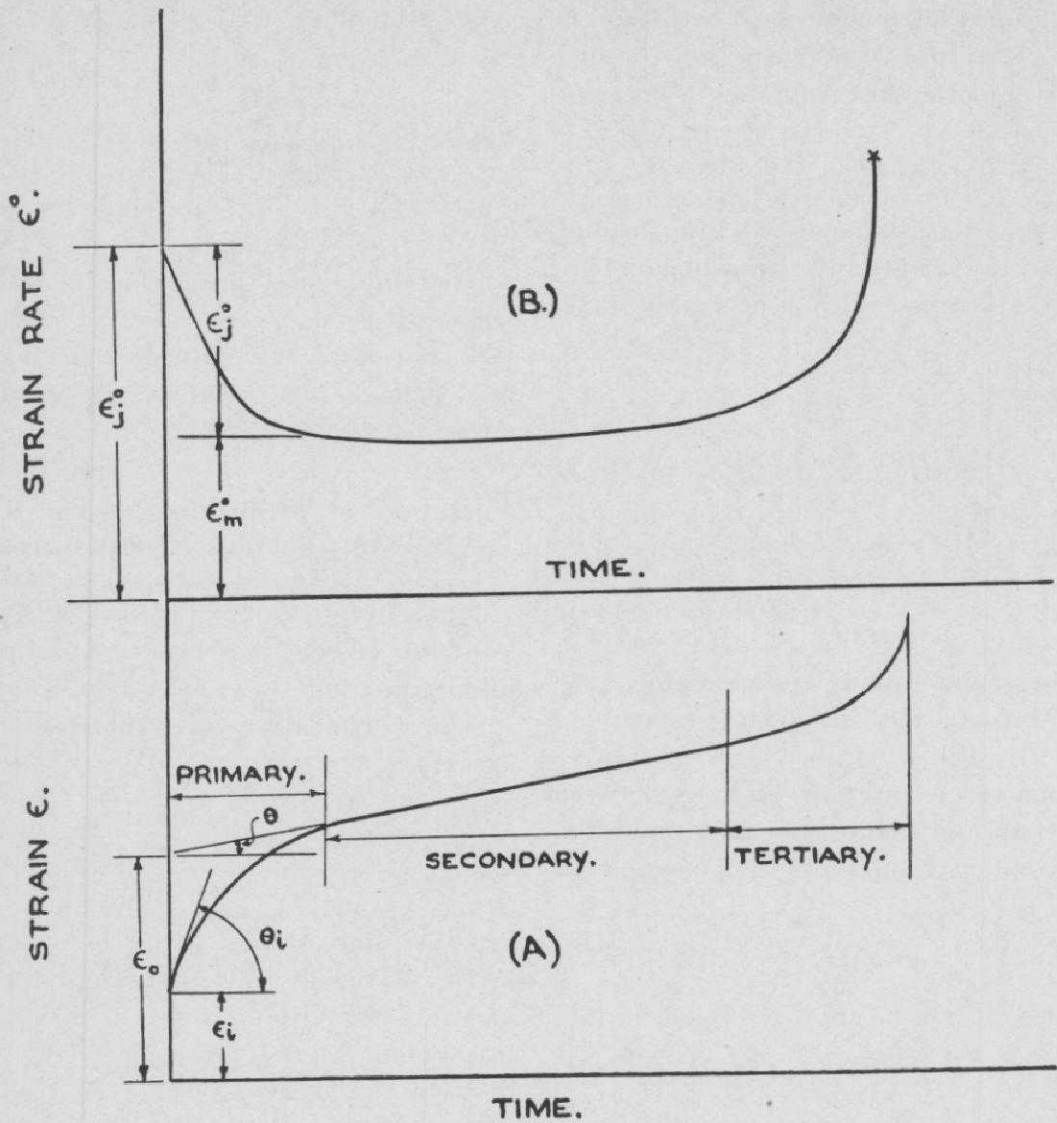


FIG. 1—(A) CREEP STRAIN  $\epsilon$  AS A FUNCTION OF TIME 't'. (B) STRAIN RATE  $\dot{\epsilon}$  AS A FUNCTION OF 't'

of extension (say, 0.1 per cent) in a specified number of hours (say, 1000 hr.). A typical creep curve at constant load and temperature is shown in Fig. 1(A) where the extension  $\epsilon$  is plotted as function of time  $t$ .

The corresponding strain rates  $\frac{d\epsilon}{dt}$  or  $\dot{\epsilon}$  are plotted as function of  $t$  in Fig. 1(B). It may be noted that

- (i) there is an immediate initial extension  $\epsilon_i$  consisting both of elastic and plastic deformations (the initial extension rate being given by  $\tan \theta_i$  and represented as  $\dot{\epsilon}_i$  in Fig. 1(B);

- (ii) in the 'Primary Stage', the strain rate  $\dot{\epsilon}$  progressively decreases and tends to approach a constant minimum value  $\dot{\epsilon}_m$  asymptotically [ Fig. 1(B) ];
- (iii) in the 'Secondary Stage', the strain rate  $\dot{\epsilon}_m$  remains constant and thus the linear portion of the plot in Fig. 1(A) may be represented as

$$\epsilon = \epsilon_o + \dot{\epsilon}_m t$$

where  $\dot{\epsilon}_m$  is the constant minimum strain rate represented by  $\tan \theta$ ;

- (iv) in the 'Tertiary Stage' the rate  $\dot{\epsilon}$  progressively increases leading ultimately to fracture.

The empirical relationship between the length  $l$  and time  $t$  in tensile creep which was first reported by Andrade<sup>2</sup> is expressed as

$$l = l_0(1 + \beta t^3)e^{Kt} \quad (\text{Andrade}^2) \dots\dots (1)$$

where  $l_0$  = length immediately after loading and  $\beta$  and  $K$  are constants depending on temperature and stress. If  $K$  is negligibly small, the strain rate is given by

$$\frac{dl}{dt} = \frac{1}{3}l_0\beta t^{-2} \dots\dots\dots (1a)$$

Thus strain rate progressively decreases corresponding to the primary stage of creep. If  $\beta$  becomes small, the strain rate is given by

$$\frac{1}{l} \frac{dl}{dt} = K \dots\dots\dots (1b)$$

Thus the strain rate is approximately constant during the secondary stage of creep.

Various other empirical relations between creep strain and time  $t$  have been indicated by different investigators. For example:

$$\epsilon = \epsilon_0 + At - \frac{B}{a}e^{-at} \quad (\text{McVetty}^3) \dots\dots (2a)$$

$$\epsilon = a \log(1 + bt) \quad (\text{Tapsell \& Prosser}^4) \quad (2b)$$

$$\epsilon = a \log(1 + bt) + \dot{\epsilon}_m t \quad (\text{Chevenard}^5) \quad (2c)$$

$$\epsilon = a \log \frac{t}{t_0} + \dot{\epsilon}_m t \quad (\text{Weaver}^6) \dots\dots\dots (2d)$$

$$\epsilon = \epsilon_1 t^K \quad (\text{Sturm, Dumont \& Howell}^7) \quad (2e)$$

where  $A, B, a, b$  and  $K$  are material constants and  $\dot{\epsilon}_m$  is the minimum creep rate. From the above it is evident that all creep-time plots cannot be represented by one mathematical relation. It is sometimes convenient to classify creep-time plots<sup>8</sup> (prior to accelerating creep in the tertiary stage) into three groups:

(a) Parabolic:  
 $\epsilon = at^m + b \quad (1 > m > 0, a > 0) \dots (3a)$

(b) Exponential:  
 $\epsilon = a \log t + b \quad (a > 0) \dots (3b)$

(c) Hyperbolic:  
 $\epsilon = at^m + b \quad (m < 0, a < 0) \dots (3c)$

During the secondary stage of creep the total deformation at some time beyond that of the test period may be obtained from

$$\epsilon = \epsilon_0 + \dot{\epsilon}_m t \dots\dots\dots (4)$$

where  $t$  implies the time within the secondary stage. It is necessary, however, to know in this connection the nature of the variation of  $\dot{\epsilon}_m$  and  $\epsilon_0$  under different conditions of stress and temperature.

**Variation of Minimum Strain Rate  $\dot{\epsilon}_m$  and Strain Constant  $\epsilon_0$  with Stress and Temperature**

At low stresses it is believed that creep is essentially due to viscous flow and hence  $\dot{\epsilon}_m$  is proportional to  $\sigma$  and, therefore, may be expressed as

$$\dot{\epsilon}_m = A\sigma \quad (\text{linear}) \dots\dots\dots (5a_1)$$

where  $A$  is sometimes called the 'flow-ability' constant or the minimum strain rate per unit stress. At very low stress-levels,  $A$  depends on temperature  $T$  but not on stress  $\sigma$ . With increasing stress, however, it is well known that creep is not truly viscous in nature (quasi-viscous) and  $A$  then depends both on  $\sigma$  and  $T$  and is sometimes expressed as

$$A = \frac{\dot{\epsilon}_m}{\sigma} = A_0 e^{\sigma^{\frac{1}{2}}} \dots\dots\dots (5a_2)$$

and  $A = \frac{\dot{\epsilon}_m}{\sigma} = A'_0 e^{-Q/RT} \dots\dots (5a_3)$

It is evident, therefore, that plots of  $\log(\dot{\epsilon}_m/\sigma)$  as function of  $\sigma^{\frac{1}{2}}$  or  $\frac{1}{T}$  should yield straight lines. This has been found to be true by Kanter<sup>9</sup> for a 0.7 per cent C steel. The energy of activation  $Q$  is found to be  $90 \times 10^3$  cal./gm.-atom which, it is interesting to note, is of the same order as the latent heat of evaporation and self-diffusion in iron ( $78 \times 10^3$  cal./gm.-atom).

For many alloys including steel the following two relations hold good within wide ranges of stresses and strain rates ( $10^{-5}$  to  $10^{-3}$  per cent per hour):

$$\dot{\epsilon}_m = C_1(1 - e^{\sigma/\sigma_1}) \dots\dots (5b)$$

(exponential or semilogarithmic relation<sup>10</sup>)

and  $\dot{\epsilon}_m = a\sigma^n$  (power relation) ... (5c)

For a large variety of steels the power relation (5c) is found to be superior and is, therefore, widely used. At low stress and strain rates (less than  $10^{-5}$  per cent/hour), however, the available test data deviate appreciably<sup>11</sup> from (5c). Moreover, all these relations are rare or less empirical. Nadai<sup>11</sup> from theoretical consideration has suggested that

$$\dot{\epsilon}_m = \dot{\epsilon}_0 \sinh \frac{\sigma}{\sigma_0} \dots\dots\dots (5d)$$

(hyperbolic sine relation<sup>11</sup>)

It may be noted that (i) equation (5d) reduces to the forms of (5a) and (5b) according as  $\sigma$  is small or large and (ii) a power function defines a tangent to a hyperbolic line curve and thus equation (5c) may be a close approximation to (5d) under certain conditions.

It is of interest to note that the strain constant  $\epsilon_0$  as function of  $\sigma$  and  $T$  follows more or less the same relations as those of  $\dot{\epsilon}_m$  and thus may be expressed as

$$\epsilon_0 = \epsilon'_0 \sinh \frac{\sigma}{\sigma'_0} \dots\dots\dots(6a)$$

and  $\epsilon_0 = \epsilon''_0 e^{-Q/RT} \dots\dots\dots(6b)$

It is obvious that equation (6a) reduces to linear or exponential function according as  $\sigma$  is small or large. This confirms McVetty's suggestion that at moderate and high stress levels  $\epsilon_0$  varies exponentially with stress.

Combining equations (1), (5d) and (6a) one gets

$$\epsilon = \epsilon_0 \sinh \frac{\sigma}{\sigma'_0} + (\dot{\epsilon}_0 \sinh \frac{\sigma}{\sigma'_0})t \dots(7)$$

A more generalized expression based on the reaction range theory<sup>12</sup> has recently been de-

veloped by Condon<sup>13</sup> and others<sup>14-16</sup> in which creep is visualized to occur due to an average jump or flow  $x$  of unit domains (groups of atoms or ions in the lattice) having energy greater than that of the average potential barrier. This leads to the expression

$$\gamma = \gamma_0 \sinh \frac{\omega}{kT} \dots\dots\dots(8)$$

where  $\omega = c\alpha x\tau$

and  $\gamma_0 = \left( \frac{2}{3} \frac{x}{y} \frac{kT}{h} e^{-\Delta F/kT} \right) \dots\dots\dots(8x)$

$\gamma$  is the constant shear creep rate under tensile stress  $\tau$  at the absolute temperature  $T$ ;  $x$  is the average jump of the unit domains separated by a distance  $y$ ;  $x$  is the projected area of the domain along the stress plane;  $k$ ,  $h$  and  $R$  are the Boltzmann, Planck and gas constants respectively, and  $c$  is a stress-concentration factor. It may be easily shown that if the work done in carrying a unit domain from the normal to the activated position be small, that is, if  $\omega$  be small compared to  $kT$ , equation (8) becomes

$$\gamma = \gamma_0 \frac{\omega}{kT} = \gamma_0 \frac{c\alpha x}{kT} \tau \dots\dots\dots(8a)$$

Hence at a constant temperature the strain rate  $\gamma$  becomes proportional to the stress  $\tau$  and thus the flow is of a viscous or Newtonian type [cf. equation (5a<sub>1</sub>)]. For larger values of  $\tau$  similarly it may be shown that equation (8) leads to

$$\gamma = \gamma_0 e^{\omega/kT} \dots\dots\dots(8b)$$

If at a given temperature  $T$ ,  $\log \gamma$  be plotted as function of tensile stress  $\tau$ , the slope of the lines gives ( $c\alpha x$ ) and the intercept the value of  $\gamma_0$ . If now  $\log \frac{\gamma_0}{T}$  be plotted as function of  $\frac{1}{T}$ , equation (8x) indicates that the slope of the line gives  $\Delta F$  and the intercept helps to compute the value of  $\frac{x}{y}$ . The general feature of metallic flow seems to suggest that flow occurs by means of so-called

'dislocations' forming at relatively few points in the lattice. The lateral dimensions of the dislocations are strongly temperature dependent. It is also found that the free energy of activation  $\Delta F$  in equation (8x) does not vary widely from metal to metal. Metals with large value of  $\gamma$  (creep rates) have a large lateral extension of dislocations while metals which do not flow easily have comparatively smaller extension, that is smaller values of  $(c\alpha x)$ .

Such theoretical discussion is, therefore, helpful in the development of creep-resistant alloys. For a given stress  $\tau$  and temperature  $T$ , the value of flow rate  $r$  will be small if  $\Delta F$  be large and  $(c\alpha x)$  be small.

$\Delta F$  may be made larger by using metals of high elastic moduli and  $(c\alpha x)$  may be made smaller by the introduction of so-called 'dislocation' barriers and 'carbide' formers.

In spite of the good attempts to develop mathematical relations explaining the primary and secondary stages of creep, there is no quantitative physical theory which successfully explains the creep phenomena as a whole. Also there is no safe quantitative relation between creep and the mechanical properties of steel or other metals and alloys. Metallurgists and engineers have, therefore, to rely on actual creep rupture and extension time data and the factors affecting creep.

### Factors Affecting the Creep Properties

Creep properties are affected by (1) temperature, (2) stress, (3) structure, (4) treatment, (5) environment, and (6) composition.

(1) *Temperature* — In connection with the effects of temperature it may be noted that:

- (i) at a lower temperature there is a prolonged stage of constant rate secondary creep and fracture may not occur if load be low;
- (ii) with increasing temperatures the stage of secondary creep rapidly decreases;

- (iii) at high temperatures the primary stage of decelerating creep rate is soon followed with a point of inflection by the rapidly accelerating rate of tertiary creep. The secondary creep may be altogether absent.

In general, creep increases almost exponentially with temperature [ cf. equations (5a) and (6b) ].

(2) *Stress* — The effects of increasing stresses on creep curves are more or less similar to those of increasing temperatures. Different theoretical and empirical relations have been developed by various investigators as indicated before. It should be noted here that most of the alloys undergo structural changes when subjected to stress and high temperature for long periods of time and hence do not obey any of the empirical relations. It is extremely unsafe, therefore, to rely on extrapolated data from short-time creep tests. To obtain the correct design data at high temperatures and stresses it is essential that creep tests are carried out for periods at least equal to those for which the material will be stressed in service.

(3) *Structure* — Creep is highly structure-sensitive property and is affected by minor variations to a greater extent than any other mechanical property. Some of the common structural factors affecting the creep properties are grain size, soluble elements due to soluble impurities and/or alloy additions, insoluble phases due to insoluble impurities, phase changes and/or precipitation and allied phenomena.

It may be shown from theory that yield stress  $Y$  is nearly a linear function of the concentration of the solute atom. It may also be shown that  $Y$  varies as the square of the differences of the atomic radii. Similarly for small concentration of soluble elements, resistance to creep may be a simple function of the concentration or differences of the atomic radii. At large concentrations, however, the relation may be rather complicated due to the diffusion coefficient increasing the

creep and thus counteracting the effects of lattice distortion. It is not unlikely, therefore, to expect a minimum in the plots of creep as function of atomic concentration. It has been found in some cases ( for example in Ni-Cr alloys precipitation-hardened by Ti) that precipitating element should be present in slight excess of the solubility limit at the service temperature to get the optimum creep strength.

(4) *Treatment* — From what has been discussed before it will be evident that for services at low temperatures when atomic migrations due to recovery or recrystallization are small, the heat treatment which corresponds to maximum strength and hardness also corresponds to maximum resistance against creep ( e.g. quenching or quenching followed by ageing to maximum hardness ). For higher temperatures, however, when atomic diffusion is large resulting in the so-called 'place-change-plasticity' and causing recovery, recrystallization, phase changes and/or precipitation, the heat treatment should be directed towards attaining strain-free and metallurgically stable phases. For heat-treatable alloy steels, normalizing at comparatively high temperatures and for longer periods of time, and quenching followed by tempering is usually recommended. For age-hardening type alloys, solution heat treatment at high temperatures ensuring complete homogenization and large grain growth followed by stabilization of the precipitate at about the operating temperatures is recommended.

(5) *Environment* — The atmosphere to which a metal or alloy is exposed in service seriously affects its creep properties. Selective oxidation or corrosion of the grain boundaries and/or elements in the alloy may cause accelerated rate of creep. Since the surface properties of the alloys are primarily responsible for such reactions, the alloys in question are often given a protective treatment to inhibit surface reactions.

(6) *Composition* — Detailed quantitative knowledge about the effects of different elements (either individually or in combination) on the creep strength of steel is rather meagre. Elements in general affect the creep properties in two ways which may be looked upon as:

- (i) solid-solution effect, and
- (ii) precipitation effect.

Creep strength is increased initially more or less by all elements in solid solution in iron. But the order of solid solution strengthening of elements at ordinary temperature is not of much help to get creep resistance at high temperatures. Silicon<sup>17</sup>, for example, which is one of the most effective hardeners of ferrite is one of the most ineffective at higher temperatures. The effects of different elements<sup>18</sup> on the creep strength ( stress to produce creep rate of  $10^{-4}$  per cent/hour ) of commercially pure iron at 800°F. are shown in Fig. 2. The very large effect of Mo, the very

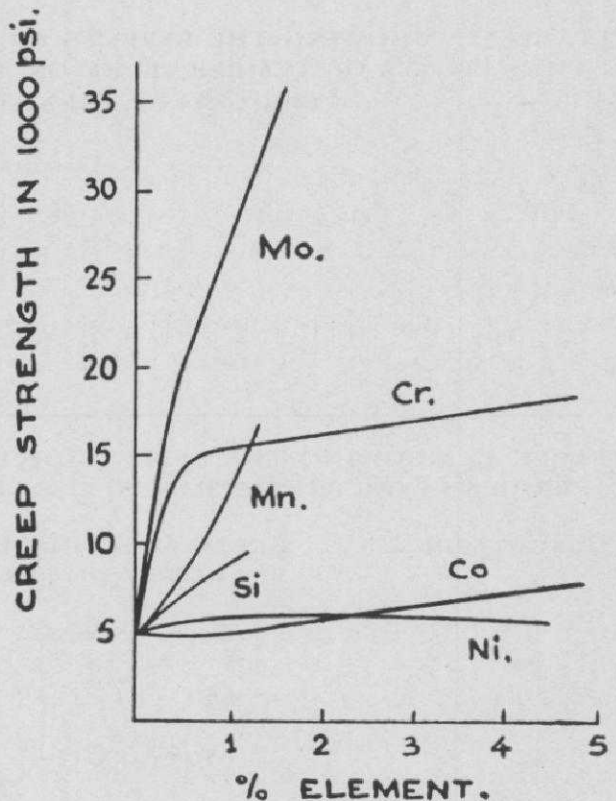


FIG. 2—CREEP STRENGTH IN 1000 P.S.I. TO PRODUCE CREEP RATE OF  $10^{-4}$  PER CENT PER HOUR AS A FUNCTION OF PER CENT ELEMENT ADDED TO COMMERCIAL PURE IRON AT 800°F. ( 425°C. ) ( AUSTIN<sup>18</sup> )

little effect of Ni and the intermediate effect of Mn may be noted. There do not seem to be any relation between gamma-stabilizing and ferrite-forming elements. Carbide-formers like Mo, Cr and Mn appear to be more effective than Si, Co or Ni. The positions of the elements Cr and Mn on the one hand and Co and Ni on the other with respect to iron in the periodic table are also noteworthy.

A comparison of Tables 2 and 3 shows that elements which are effective in raising the crystallization temperature of iron are also effective in increasing creep strength.

**Experimental Results and Discussion**

On the basis of what has been indicated before several alloy steels have been melted in a 40 kW. induction furnace at the Bengal Engineering College Metallurgical Labora-

tories. Compositions of the steels are shown in Table 4. The properties of several steels have been tested with reference to their tensile properties at room temperature, resistance against oxidation at 700°-900°C. and short-time creep behaviour at 650°-700°C. Typical results are shown in Figs. 3 and 4 and Tables 4 and 5.

It may be noted from Fig. 3 that so far as oxidation resistance in air is concerned steel I is the best obviously due to its higher chromium content. And for the same reason steel IV has better oxidation resistance compared to those of steels II and III. It is obvious from Fig. 4 that steel I has the best creep resistance (except nimonic 80 which is indicated in the figure for comparative study). All the steels I-IV were tested for creep after ageing for 8 hr. at 650°C. A typical plot indicating the effects of

**TABLE 2 — SHOWING THE EFFECTS OF ELEMENTS WITH OR WITHOUT C AND THE EFFECTS OF TEMPERATURE ON THE ORDER OF ELEMENTS AFFECTING THE CREEP STRENGTH OF IRON (Austin<sup>18</sup>)**

TEMP. °F	%C	%ELEMENT.	ELEMENTS IN DECREASING ORDER OF THEIR EFFECTS ON CREEP STRENGTH.									
800	NIL	1%	Mo	-	-	-	Cr	Mn	-	Si	Ni	Co
800	"	2%	Mo	-	-	-	Mn	Cr	-	Si	Co	Ni
750	0.1	0.5%	V	Mo	W	Cu	Cr	Si	Ti	Mn	Co	Ni
930	0.1	0.5%	Mo	V	Ti	Cu	Cr	Mn	Co	W	Si	Ni

**TABLE 3 — SHOWING THE EFFECTS OF ELEMENTS (AT DIFFERENT REDUCTION PER CENT) ON THE RECRYSTALLIZATION TEMPERATURE OF IRON (Austin<sup>19</sup>)**

%REDUCTION	ELEMENTS IN DECREASING ORDER OF THEIR EFFECTS ON RECRYSTALLISATION TEMPERATURE						
5%	Cr	Mo	Mn	Si	Ni	Co	
20%	Mo	Cr	Mn	Si	Ni	Co	
40%	Mo	Cr	Mn	Si	Ni	Co	
75%	Mo	Mn	Cr	Si	Ni	Co	
90%	Mo	Mn	Cr	Si	Co	Ni	



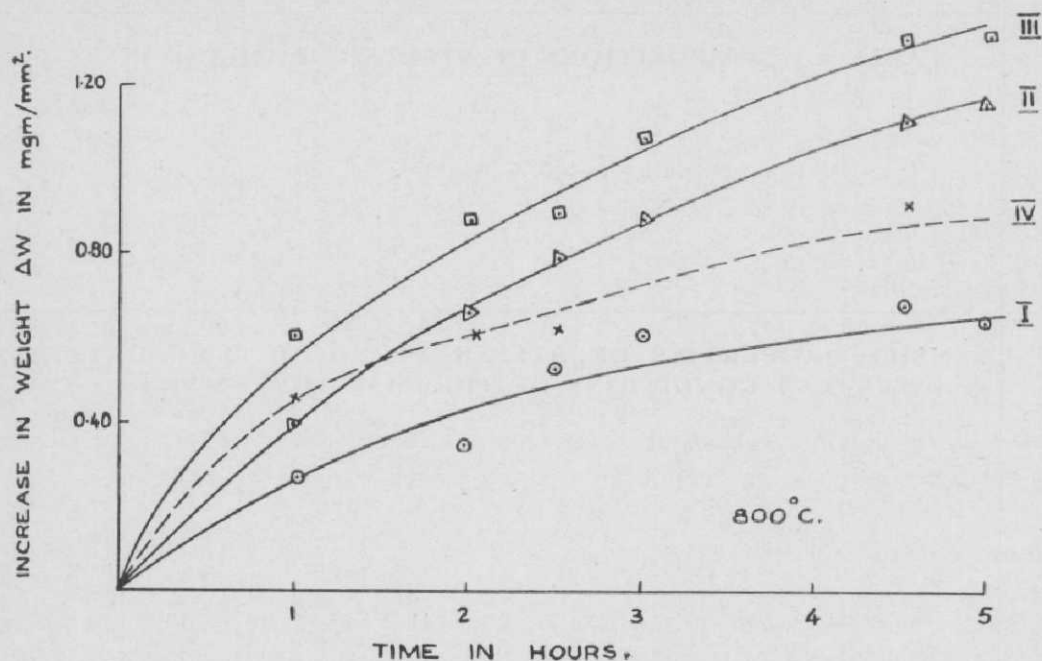


FIG. 3 — INCREASE IN WEIGHT  $\Delta W$  IN  $\text{MGM./MM.}^2$  AS A FUNCTION OF TIME ON HEATING IN AIR AT  $800^\circ\text{C}$ .

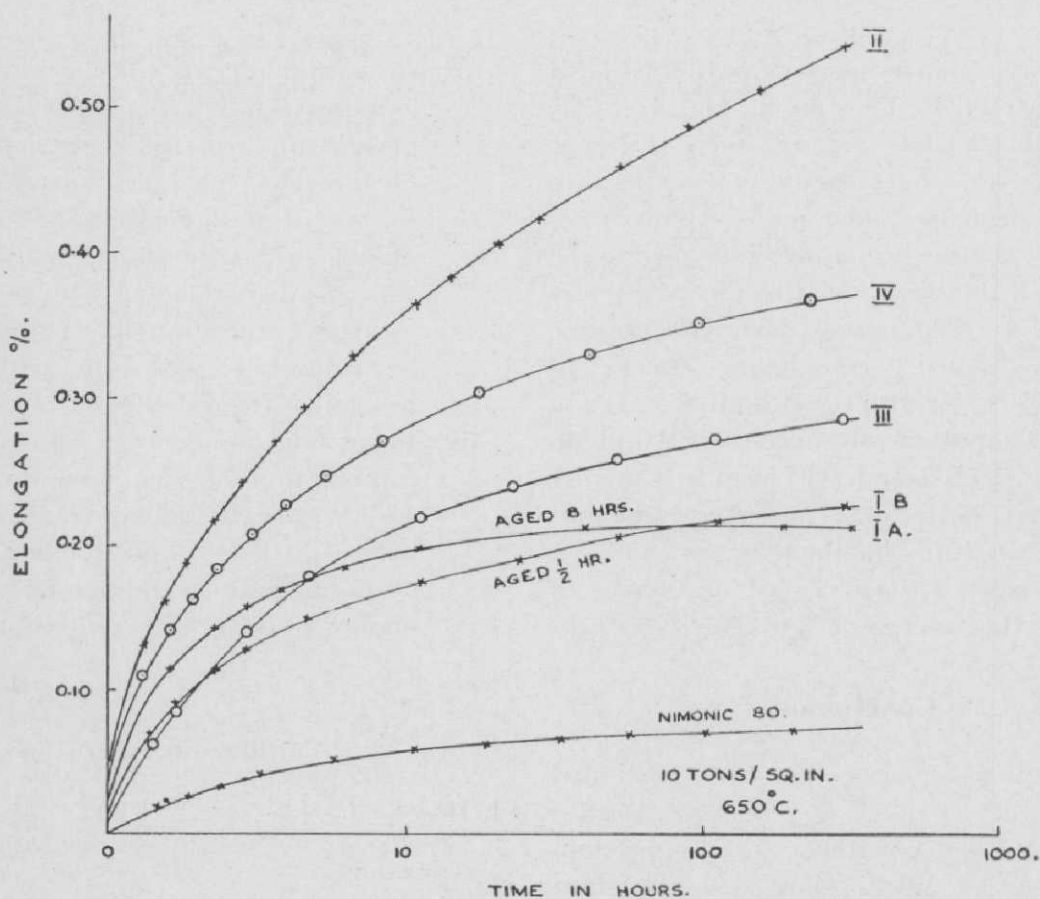


FIG. 4 — ELONGATION PER CENT AS FUNCTION OF TIME FOR STEELS I, II, III AND IV (O.O.  $1100^\circ\text{C}$ , 1 HR.; AGED  $650^\circ\text{C}$ , 8 HR.) UNDER A CONSTANT LOAD OF 8 TONS/SQ. IN. AT  $650^\circ\text{C}$ .

TABLE 4—COMPOSITIONS OF STEELS I, II, III AND IV

STEEL	C	Si	Mn	Cr	Ni	Ti	V	C	Mo	W
I	0.3	0.8	17.7	11.8	0.7	—	0.3	3.5	0.25	0.6
II	0.2	0.7	18.2	8.9	1.0	—	0.8	—	—	—
III	0.3	0.8	18.6	8.6	—	1.2	0.1	1.0	0.10	0.5
IV	0.3	0.7	19.2	9.8	—	—	1.5	1.0	—	—

TABLE 5—TENSILE PROPERTIES OF STEELS AT ROOM TEMPERATURE UNDER DIFFERENT CONDITIONS OF THERMAL TREATMENTS

STEEL	HOT FORGED			ANNEALED 900°C.			O.Q. 1100°C. 1 HR.			AGED 650°C.			AGED 650°C. 8 HR.		
	M.S. tons/ sq. in.	E., %	R.A., %	M.S., tons/ sq. in.	E., %	R.A., %	M.S., tons/ sq. in.	E., %	R.A., %	M.S., tons/ sq. in.	E., %	R.A., %	M.S., tons/ sq. in.	E., %	R.A., %
I	46	21	32	43	26	37	53	37	41	55	30	39	54	30	38
II	50	25	33	49	27	39	55	38	46	58	34	45	57	34	44
III	56	46	43	56	48	48	65	52	53	59	58	55	59	57	54
IV	54	48	38	53	52	43	58	54	48	56	58	45	55	56	46

ageing for a shorter period of time is also shown in Fig. 4 (Plot IB in Fig. 4). Although the initial rate of creep is larger (Plot IA) for longer ageing period, the rate tends to decrease much more rapidly than in IB. It is also clear from a reference to the analyses of the steels that the comparatively better creep resistances of I and III are due to the so-called 'complexity effect' of elements V to W with or without Ti. It appears that simultaneous increase of V and Mo (compare steels I and III) even to a smaller extent has a better effect on creep resistance than the addition of about 1 per cent Ti. Further work, however, is necessary to establish this fact.

### Conclusion

From the work so far done it is concluded that

- (i) there is a possibility for the development of Mn-Cr alloy steels resistant to oxidation and to creep with Mn content around 17-20 per cent;

- (ii) chromium appears to be most effective in increasing oxidation resistance when present in excess of 9 per cent;
- (iii) for keeping the steel austenitic and for reasons of forgeability and machinability and at the same time to retain the 'complexity effect' by the ferrite formers—the limit of Cr appears to be around 11 per cent with 17-20 per cent Mn and 0.2-0.3 per cent C;
- (iv) to get fair amount of creep resistance, combination of elements Mo, V, Ti and W has to be taken recourse to with as little W as possible for reasons of specific gravity and to guard against poorer resistance to oxidation.

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## PAPERS DISCUSSED

1. The Effect of Variation of Ruling Section on the Mechanical Properties of Carburizing Steels, by J. E. RUSSELL & B. LUDGATE.
2. The Molybdenum-boron Structural Steel, by J. F. SEWELL & W. E. BARDGETT.
3. Micro-metallurgy of Alloy Steels, by B. R. NIJHAWAN.
4. Boron Steels, by THOMAS G. DIGGES.
5. The Properties of High-speed Steel Derived from Magnetic Iron Sand, by ENG. SADAO KOSHIBA.
6. The Cast Structure of High-speed Steel, by E. INESON & G. HOYLE
7. Sintered Materials on the Base of Iron with Special Regard to Alloyed Sintered Steels, by R. KIEFFER & F. BENESOVSKY.

**DR. B. R. NIJHAWAN** (Dy. Director, National Metallurgical Laboratory)

The work done at the Mond Nickel Co. Ltd. in the U.K. on the hardenability of boron-free and boron-treated triple alloy steels containing 0.15-0.53 per cent carbon was investigated by means of Jominy tests on specimens from 4 in. billets and showed

marked variations in the hardenability in boron-treated steels. But on the average there is considerable improvement in the hardenability whatever the boron content was. This subject has been recently worked upon in the U.K. by some who have noted extremely marked variations in the hardenability tests. This subject should be examined most thoroughly in order to dispell the doubts about the boron-treated steels which very often go a long way to prevent their acceptance in many countries, including the U.K., and perhaps it will be a more deterrent factor in this country where any doubts expressed very often preclude its use entirely.

**MR. E. H. BUCKNALL**

I would like to ask Mr. Sewell regarding the corrosion resistance of  $\frac{1}{2}$  per cent Mo-B steels.

**MR. J. F. SEWELL** (Samuel Fox & Co., Stocksbridge)

The corrosion resistance of these steels is superior to the ordinary structural steels.

**DR. V. G. PARANJPE (Tata Iron & Steel Co. Ltd.)**

I would like to ask Dr. Nijhawan the advantages obtained by treating steels with mischmetal and its relative cost.

**DR. B. R. NIJHAWAN (Dy. Director, National Metallurgical Laboratory)**

With the additions of rare earths to high alloy steels the yield increases by 6-15 per cent. In case of 25 Ni - 20 per cent Cr steels the figure goes as high as 10-15 per cent. Thus the cost of rare earth additions is counterbalanced by the increase in yield. I would like to refer to a paper by Evans in the journal *Iron and Steel* in which he has shown the improvement in the yield by minute additions of cerium and mischmetal.

**MR. E. H. BUCKNALL (Director)**

In the manufacture of a steel, such as Fortiweld, quite clearly 'know-how' is an important factor. The authors of the paper Molybdenum-boron Steels can hardly be held responsible for difficulties encountered while making experimental heats. These may be due to unfixated nitrogen.

The boron-effect has been attributed to soluble boron which has a tendency to segregate at the grain boundaries. But Mayer emphasized how effective boron was on nucleation of acicular ferrite near the nose of the S-curve, which is formed in the body of the grains.

The morning's papers have shown that the tonnage of heat-treatable boron steels being made is falling. It seems possible that the high tensile type will soon reverse the trend.

Against the Mo-B high-tensile steels being applied in India is the belief that the Indian atmosphere is generally usually corrosive despite B.I.S.R.A. observations on the conditions applying to the Delhi pillar. It does not seem to be accepted that

Mo is beneficial, and some prejudice has been caused by an early remark of Bardgett that Cu in this class of steel is not useful.

In the case of the materials like high-speed steel one can possibly take a leaf out of the book of the cast iron metallurgists and cause the carbide formation to adopt the spherulitic form. Something of that sort has been done in connection with silicon in certain aluminium alloys. If this could be achieved, I think it could represent the very important result, not only for high-speed steel but for all the high carbide classes of material. If one has the patience and manual dexterity to smith forge, white cast iron can become a very nice wrought material, and a good tool steel. But because of the usual distribution of the carbide in the as-cast material, the operation is very laborious and requires the greatest skill. If the distribution could be modified so that one could get rid of the sharp angularities, the operation would become much easier. This, of course, is a very pious wish, but something which workers in that field should not regard as a completely unattainable goal. An alternative line of thought, with which Mr. Sewell may be familiar, is to use the high-speed steels in as-cast condition. During the war, quite a lot of milling cutters were made by the lost wax process.

**DR. B. R. NIJHAWAN (Dy. Director, National Metallurgical Laboratory)**

In certain high-speed steels I observed diametrical segregation of carbides irrespective of the size of the ingot. This segregation persisted however much the section was reduced.

**MR. S. VISVANATHAN (Tata Iron & Steel Co. Ltd.)**

If the forging temperature is low, the carbide segregation is maximum. Once this occurs, this cannot be removed.

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