Materials for Winding Plant Components

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SYNOPSIS

Materials for critical applications such as winding plant components are reviewed in terms of the basic requirements which have to be complied with. It is shown that tensile testing alone is not sufficient and that impact testing also is necessary if the incidence of brittle fracture is to be avoided. As a criterion it is suggested that the brittle transition temperature (B.T.T.) determined by impact testing should in all cases be well below the lowest anticipated service temperature.

A number of steels were evaluated by impact testing and it is shown that a small ferrite grain size, obtained either by chemical means such as by the use of appropriate alloying elements in the steel or by means of a quenching heat treatment, is effective in attaining a low B.T.T. The speed of strain aging as well as the influence of strain aging on the B.T.T. was also evaluated. On the basis of the results obtained it is recommended that steels for critical applications must conform to an appropriate grade and class of steel according to BS 4360 specification. It is further recommended that such steels be exempted from the customary six-monthly heat treatment prescribed by the Mines and Work Act, Regulation 16,18.

OPSOMMING

Materiale vir kritiese toepassings soos hyser onderdele word beoordeel volgens die basiese vereistes waaraan dit moet voldoen. Daar word aangetoon dat trektoetse alleen nie voldoende is nie en dat slagtoetse ook nodig is om te verseker dat die voorkoms van brosbreuke vermy moet word. As 'n keterium word daar voorgestel dat die brosoorgangstemperatuur soos bepaal met behulp van 'n slagtoets in alle gevalle laer as die laagste verwagte werkstemperatuur moet wees.

'n Aantal stale is deur middel van slagtoetse beoordeel en daar word aangetoon dat 'n klein korrelgrootte wat verkry kan word deur die gebruik van geskikte legeringselemente in die staal of deur middel van 'n afblus hittebehandeling effektief is om 'n lae brosoorgangstemperatuur te verkry. Die spoed van rekveroudering en die invloed van rekveroudering op die brosoorgangstemperatuur is ook ondersoek. Op grond van die resultate van die toetse word dit aanbeveel dat stale vir kritiese toepassings aan die toepaslike graad en klas van staal volgens die BS 4360 spesifikasie moet voldoen. Daar word verder aanbeveel dat sulke stale ook vrygestel word van gebruiklike sesmaandelikse hittebehandeling soos vereis deur die Wet of Myne en Bedrywe-regulasie 16,18.

INTRODUCTION

A high degree of integrity is required in materials which are used for critical applications such as various parts of a winding plant especially where the safety of personnel is involved. Many engineers consider that a high degree of safety can be realised simply by using a high safety factor. This approach, however, is full of pitfalls and if the safety factor is the only criterion it can be a very dangerous practice. The overall safety is also dependant on a variety of other properties of the materials which determine the integrity of the structure.

In a winding plant where the loading conditions are fixed and where with the aid of experimental stress analysis and computers the applied stresses can be calculated to a high degree of accuracy the use of a safety factor is still necessary. The safety factor which should be used is determined by the uncertainties in relating the properties of the materials determined by tensile testing to the behaviour of the completed structure. The material used in practice may have undetected structural defects such as hairline cracks, segregation or pipe, or otherwise be subjected to fatigue and impact loading as well as to conditions of corrosion, abrasion, fatigue or strain aging to which the tensile test specimen was not subjected. In actual fact therefore the integrity of a material can only be fully assessed by also subjecting it to each of the abovementioned factors or to various combinations depending on the operating conditions. Even so, the final result must necessarily depend on judgment and a reasonable margin of safety would still be necessary.

TENSION TEST DATA

Apart from the deficiencies of a tension test a great deal of insight into the reaction of a steel to all sorts of different operating environments can be deduced from the result of a tension test.

The yield strength for example limits the maximum static stress which can be applied without appreciable permanent deformation. Since the yield stress is usually lower than the tensile strength and since the factor of safety is traditionally and, to my mind, erroneously based on the tensile strength it stands to reason that the safety factor must be greater than one. The smallest safety factor which could be used is the ratio tensile strength to yield stress. Consequently the tensile strength: yield stress ratio is of significance in interpreting tensile data.

The total percentage elongation or the percentage reduction in area obtained in a tensile test is indicative of the ductility of the steel and a certain minimum percentage elongation is usually considered necessary and also included in most steel specifications. Apart from applications where cold forming of the steel is to be performed there does not seem to be much unanimity amongst engineers on the minimum ductility which may be required. Very few structures for example will still be serviceable if the material due to overloading has yielded more than 5 per cent. Most structural steels have ductilities of 20 per cent-30 per cent, far in excess of that which can be usefully employed.

If 5 per cent elongation is the maximum ductility which can be used in service, the ductility requirements

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'n most specifications do not relate to the engineering requirements of service but rather to the level of ductility that accompanies the required yield strength in normally processed material. From a quality point of view low percentage elongation value — although from an engineering point of view adequate for the purpose for which it is intended — is always viewed with suspicion. Such low values of elongation in comparison with material in the same structural condition may be indicative of a lack of soundness or cleanliness of the steel. Thus material specifications for various grades of steel have fairly strict limits on the minimum elongation in spite of the fact that it is far in excess of that required for the particular end use. The useful limit of plasticity depends on the design of the article and the type of service which it must withstand. A higher degree of plasticity would therefore not increase the merit of the steel as far as that particular application is concerned.

Another point about which designers do not always agree is the allowance which must be made for stress raisers. In the case of fatigue loading, for example, due allowance should be made. In the case of static loading, however, the local stress at such stress raisers cannot exceed the level of the yield stress by any large measure due to localized plastic yielding. In a steel with a high degree of ductility the stress concentration will therefore be relieved to a large extent.

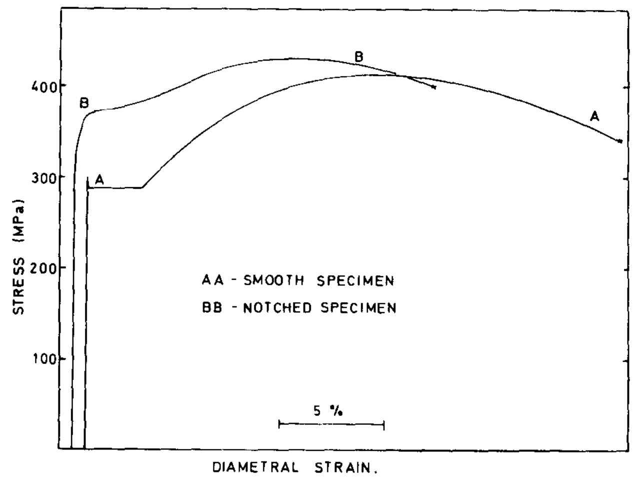

Fig. 1—Stress-Strain curve of a notched and an un-notched specimen in the normalized condition

Fig. 1 shows a comparison of the stress-diametral reduction curves for a conventional tensile test specimen and a specimen with a sharp 60°V notch around its circumference for mild steel in the normalized condition. In spite of the severe elastic stress concentration neither the tensile strength or the ductility has been affected seriously. In fact the plastic constraint exerted by the presence of the notch has actually increased the yield strength by about 20 per cent.

The results shown in Fig. 1 must, however, not be generalised. Fig. 2 for example shows the same type of graph for a medium carbon steel. In this case there is a striking difference in that the notched specimen suffered premature brittle-failure. This test very dramatically illustrates the shortcoming of the conventional tensile

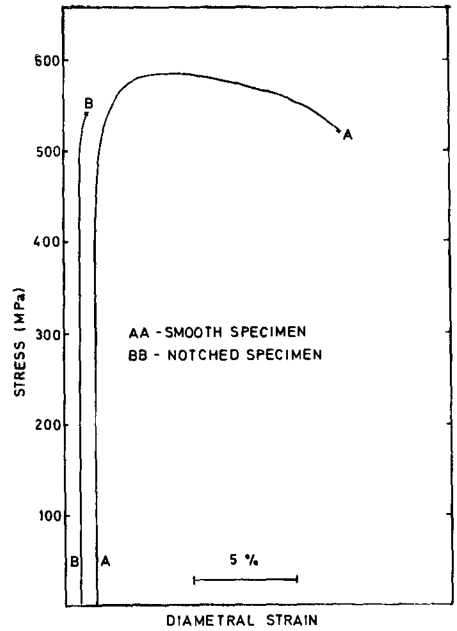

Fig. 2—Stress-Strain curve of a notched and an un-notched specimen for a medium carbon steel

test and exposes the fallacy of basing the design stress solely on the tensile strength by using appropriate safety factors. Even if the stress concentration factors are allowed for, and the design stress is limited to values far below the yield stress, the actual stresses due to the presence of residual stresses may locally still exceed the yield stress and some measure of resistance against brittle-failure is a prerequisite. Even if an appropriate stress relieving heat treatment is used to relieve the residual stresses due to fabrication, all commercial steels contain to a greater or lesser extent defects such as seams, laps, laminations, hairline cracks and non-metallic inclusions whence brittle cracks may nucleate unless there is a sufficient measure of resistance against the propagation of brittle cracks.

RESISTANCE AGAINST BRITTLE FRACTURE

It is now fairly universally recognised that resistance against brittle fracture of a pearlitic steel is best evaluated by measuring the energy necessary to fracture a notched specimen in a test such as the Charpé impact test. The results of such impact tests on mild steel are

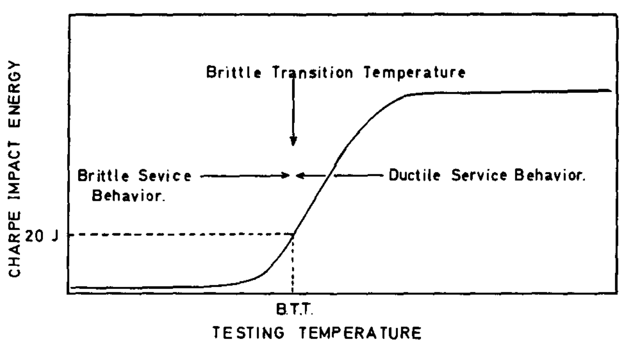

Fig. 3—Charpé impact strength as a function of the testing temperature

shown schematically in Fig. 3. The fracture energy depends very sensitively on the testing temperature, the fracturing mode changing from a shear (ductile) fracture to a brittle (cleavage) fracture as the testing temperature is lowered. The temperature at which this transition in fracture mode occurs is known as the brittle transition temperature (B.T.T.). The B.T.T. is usually defined as the temperature where the energy absorbed equals 20 Joules (15 ft. lbs). The actual impact values themselves are of no direct value in design. A material with an impact energy of 100 J is not five times as good as a material with an impact energy of 20 J. The two may indeed be equally good or equally unsuitable depending on the transition temperature in relation with the lowest service temperature.

To be of value for design purposes an impact toughness test needs to be made on actual structural components having the exact size and shape of those in service and previously subjected to the same treatment during use, such as fatigue and possibly strain ageing, as the actual components are subjected to during their lives. This, however, is not very practical. From the large number of investigation of actual brittle fractures which have occurred during the past decades it has in most cases been found that ductile service behaviour can be expected when the service temperature is above the 20 J-B.T.T. An example is the brittle fracture of the U.S.S. Ponagansett[1] which split in two while tied at dockside. The air temperature at the time of failure was 2 °C at which temperature the 25 mm thick plate absorbed only 10 J energy in a Charpé V-notch test producing a fracture surface of which only 5 per cent was fibrous. The 20 J-B.T.T. for the plate material was 10 °C. In another instance[1] the 31 mm thick plate of a pressure vessel failed during hydraulic testing at a temperature of 7 °C whereas the 20 J-B.T.T. of the plate material was 15 °C. Ten other similar incidences of brittle failure of pressure vessels revealed that the Charpé impact strengths at the failure temperatures were between 5 and 24 J. Similar investigations in the brittle fracture of dipper sticks on power shovels and turbine generator rotors, to mention only a few, have been recorded with very much the same result. Although the actual service conditions are usually not as severe as that in an impact test, it makes engineering sense to ensure that the B.T.T. is at all times well below the lowest service temperature.

EXPERIMENTAL

For the purpose of the investigation, five types of steel were used with analysis given in Table 1.

Except where indicated, the following heat treatments were used.

Full anneal — heat to 900 °C followed by furnace cooling.

Normalizing — heat to 900 °C followed by air cooling.

Quenching — heat to 900 °C followed by quenching in oil or water.

Stress relieving — heat to 650 °C followed by air cooling.

Strain aging — Plastic deformation of 6 per cent followed by aging 15 min. at 150 °C.

TABLE I

Element	Mild Steel	Cor-ten A	25/30 Carbon steel*	Vulcan	Mangear*
% C	0,15	0,09	0,3	0,24	0,12
% Mn	0,52	0,36	0,875	0,95	1,53
% P	0,017	0,011	0,022	0,015	0,030
% S	0,025	0,012	0,019	0,015	0,025
% Si	0,22	0,59	0,06	0,35	0,21
% Ni	—	0,31	0,036	3,46	—
% Cu	—	0,31	0,028	—	—
% Cr	—	0,72	0,02	1,68	—
% Mo	—	—	< 0,005	0,32	—
% V	—	—	0,009	—	—
% Al	—	—	< 0,005	—	—

* Mangear and '20/30' carbon steel is covered by the B.S. 2772: Part 2: 1956 specification which applies to iron and steel for colliery cage suspension gear, tub and mine car drawgear and couplings and rope sockets. Mangear steel is specially produced for skip and cage attachments and is exempt from Government regulations regarding periodic annealing.

Impact testing was carried out on standard Charpé V-notch specimens by prior cooling or heating in a liquid bath followed by testing a few seconds after removal from the liquid bath. Mangear and '20/30' carbon steel specimens from actual Humble hooks were machined transverse to the rolling direction from a plate 57 mm thick. Cor-ten specimens were also taken transverse to the rolling direction from 12 mm thick plate whereas the mild steel samples were tested in the longitudinal direction. Tensile testing was carried out on an Instrom testing machine by automatic plotting of the load and the deformation on an X-Y plotter.

RESULTS

Figs. 4 and 5 shows the energy absorbed as well as the percentage shear fracture as a function of the testing

TABLE II

INFLUENCE OF HEAT TREATMENT ON THE B.T.T.

Heat treatment	20 J — B.T.T. (°C)				
	Mild steel	Cor-ten A	'20/30' carbon steel	Vulcan	Mangear
(a) as received	—	—	+27 °C	< −80 °C	−40 °C
(b) normalized	−25 °C	−41 °C	− 7 °C	< −80 °C	—
(c) normalizing followed by strain ageing	+25 °C	− 5 °C	+37 °C	—	−10 °C
(d) Oil Quenched and stress relieved	< −80 °C	—	—	—	—
(e) annealed	− 5 °C	—	—	—	—

temperature of Mangear and Cor-ten steels in the normalized and strain aged condition respectively. Fig. 6 shows reproductions of stress-strain curves of mild teel after different heat treatments. Table II summarises the results shown in Figs. 4 and 5.

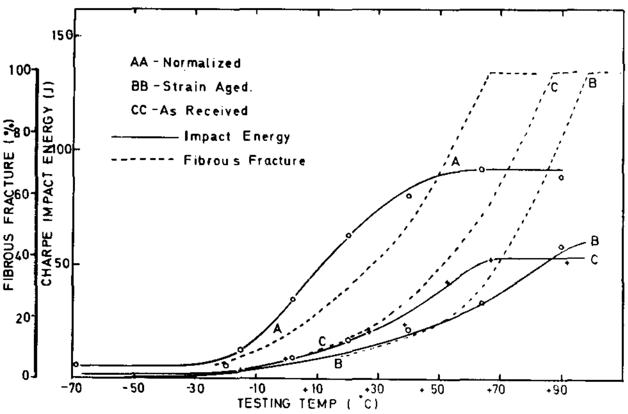

Fig. 4—Influence of the testing temperature on the fracture mode and impact energy of '20/30' carbon steel Charpé specimens

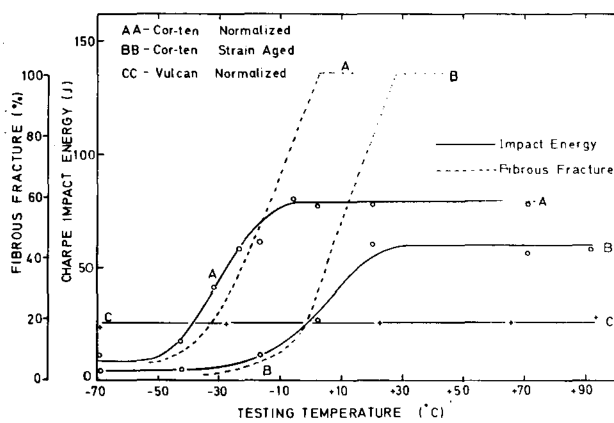

Fig. 5—Influence of the testing temperature on the fracture mode and impact energy of Cor-ten Charpé specimens

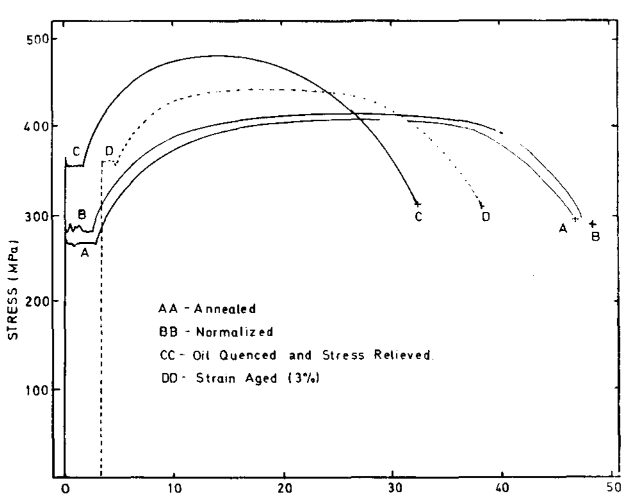

Fig. 6—Stress-Strain curves of mild steel after different heat treatments

DISCUSSION

From the results presented in Figs. 4 and 5 as well that in Table II it is clear that not only does the chemical composition as reflected by the variety of steels tested influence the B.T.T., but that the metallographic structure as a result of the prior treatment influences the B.T.T. profoundly. The metallographic structure of '20/30' carbon steel as received (presumably normalized as 60 mm thick plate) is shown in Fig. 7 (a). As a result of the manganese segregation during solidification of the original casting coupled with the relatively slow cooling during normalizing of the thick plate, gross banding as well as a very large ferrite grain size resulted. Subsequent normalizing of small Charpé specimens, however, resulted in a much more uniform structure as well as in a smaller grain size Fig. 7 (b). As a result of the

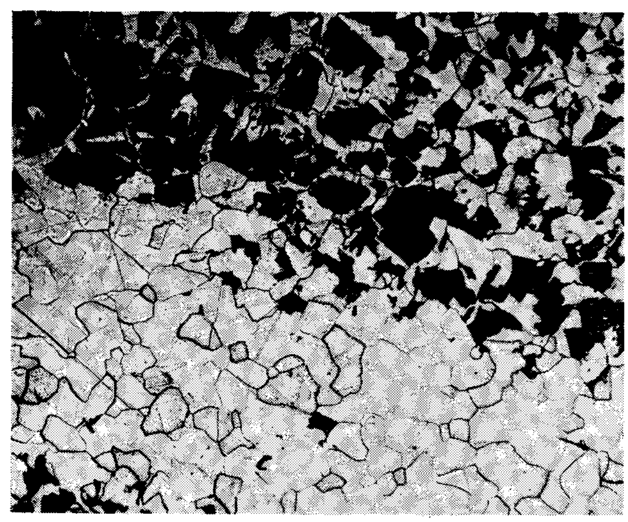

Fig. 7 (a)—Metallographic structure of '20/30' carbon steel in as received condition x82

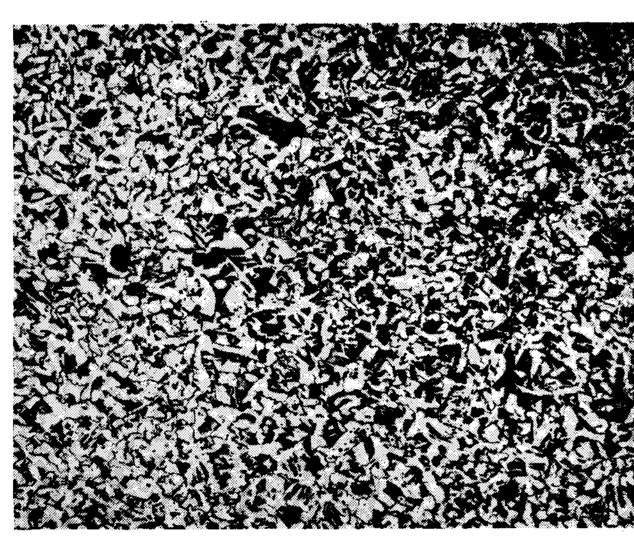

Fig. 7 (b)—Metallographic structure of '20/30' carbon steel after normalizing of small Charpé specimens x82

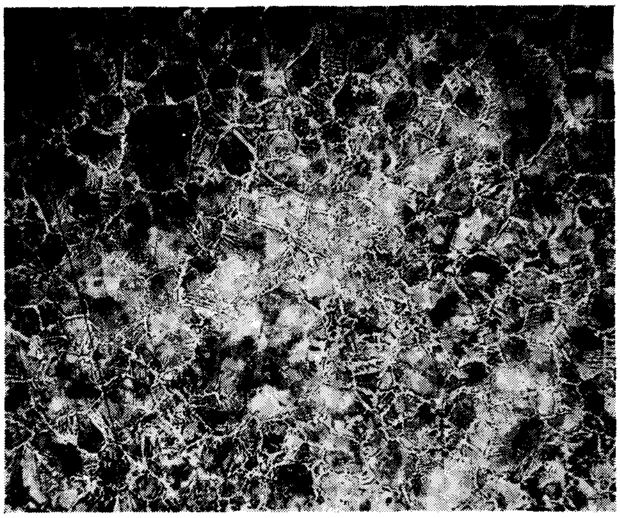

Fig. 7 (c)—Metallographic structure of '20/30' carbon steel after oil quenching x82

smaller grain size the B.T.T. was lowered from +27 °C to − 7 °C in comparison with the "as received" condition. Fig. 7 (c) by comparison shows the metallographic structure obtained in the centre of a 150 mm × 150 mm by 57 mm thick section which was oil quenched. The structure consists as in Figs 7 (a) and 7 (b) of ferrite and pearlite. Pre-eutectoid ferrite has however been suppressed to a larger extent and the B.T.T. can be expected to be lower than any value given in Table II for '20/30' carbon steel. By comparison Fig. 7 (d) shows the very fine structure which was obtained in the case of normalized cor-ten steel.

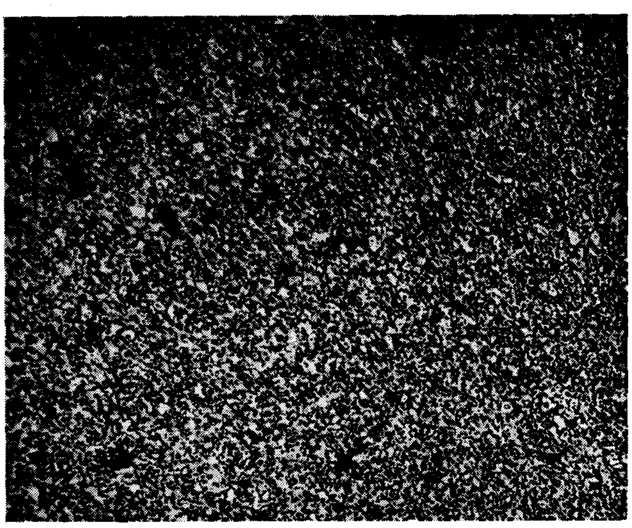

Fig. 7 (d)—Metallographic structure of Cor-ten after normalizing x82

Fig. 8 shows a hardness profile taken on a transverse section of a '20/30' carbon steel 57 mm thick plate which was sectioned through the centre after it had been quenched in water and oil respectively. Quenching of such a massive section (57 mm thick plate) in oil results in a relatively slow cooling rate of 30 °C/sec. on the surface and 5 °C/sec. in the centre of the section at 700 °C. These cooling rates are far below the critical cooling rate to form martensite in this type of steel.

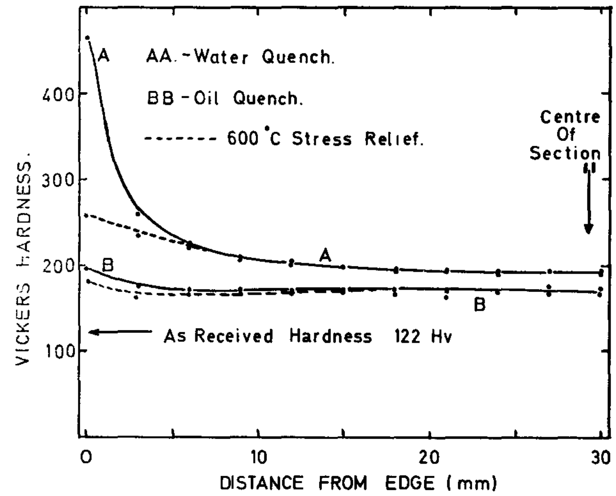

Fig. 8—Transverse hardness profile of '20/30' carbon steel plate which water and oil quenched respectively

Insofar as the metallographic structure is concerned, oil quenching has merely resulted in a fast normalizing heat treatment. In the case of water quenching the cooling rate at the surface was fast enough to give some hardening as can be seen in Fig. 8. Towards the centre of the section the cooling rate was again too low and a pearlitic structure was obtained. A stress relieving heat treatment at 600 °C subsequent to the oil and water quenching (Fig. 8) hardly affected the hardness except on the surface of the water quenched specimen where tempering of the martensite took place. The specimen after quenching was perfectly sound with no cracks at all. Even if quenching results in the formation of martensite the danger of cracking is negligible on account of the low carbon content and the consequent ductility of any martensite which may form. Quenching of the low carbon steel, apart from resulting in a beneficial small ferrite grain size and a higher yield strength as well as a low B.T.T., may, as a result of the high thermal, stresses induced by quenching, result in a component with high residual stresses. A stress relieving heat treatment of 600-650 °C is therefore recommended in all instances after quenching. In the case of long or thin flat components quenching may result in excessive distortion. Any straightening of such components should then be undertaken prior to the stress relieving heat treatment. On account of the warping which may take place it is also recommended that heat treatment be done before machining is undertaken. Quenching results in a structure which is highly resistant to crack propagation and it is consequently an ideal parent metal structure if any welding is to be undertaken.

The beneficial effect of a fast cooling rate through the transformation temperature is clearly evident in Table II which shows that mild steel in the fully annealed condition had a B.T.T. of − 5 °C. By normalizing this was improved to $-25\,°C$ whereas quenching in oil followed by stress relieving resulted in a B.T.T. which was below $-80\,°C$.

The stress-strain curves in Fig. 6 show that quenching and stress-relieving has also resulted in a modest increase in the yield and tensile strength and consequently also in the fatigue strength. The apparent decrease in ductility caused by quenching as shown in Fig. 6 is misleading. Quenching and stress relieving results in an increase in the yield to tensile strength ratio as well as in a decrease in the work hardening coefficient. Consequently the amount of uniform elongation before necking is reduced and consequently also the total percentage elongation. The true ductility as measured by the reduction in area in the annealed condition was 65 per cent in comparison to 71 per cent in the quenched and stress relieved condition.

In comparison with the pearlitic structures which are obtained when low carbon steels are quenched, the B.T.T.'s of tempered martensitic structures, which are obtained when higher carbon steels with higher hardenabilities are quenched and tempered, are generally also very low.

In the case of Vulcan steel (Table I), which on account of its high alloy content is air hardening, the B.T.T. in the normalized (undertempered martensitic) condition was also below $-80\,°C$.

STEEL SPECIFICATIONS

Fig. 9[2] shows the yield stress, tensile strength, and the percentage elongation of 25 mm diameter carbon steels in the hot rolled condition as influenced by the carbon content. If the tensile strength and percentage elongation were the only requirements it is clear that a high carbon steel would be the most economical to use since its ductility in terms of service requirements is still adequate. Fig. 9 however also shows the Charpé impact energy at 20 °C plotted from the work of Rinebolt[3]. The inferior impact values for steels with a carbon content in excess of about 0,3 per cent is not so much due to the inherent brittleness but rather due to the fact that steels with more than about 0,3 per cent C have a B.T.T. above room temperature. Consequently 0,3 per cent C is about the maximum carbon content which is allowable in structural steels in the hot rolled condition if the danger of brittle fracture is to be avoided This limitation of course is not applicable in the case of quenched and tempered condition.

Instead of using carbon as an alloying element to obtain high strengths it is also possible to use a variety of other elements in small quantities to obtain higher strengths than that of mild steel. In such low alloy structural steels the compositions have to be carefully balanced in order to obtain high strengths with acceptable values of B.T.T. An example of such a low alloy structural steels in which the percentage carbon is kept below 0,2 per cent is cor-ten B. It provides for a minimum yield strength of 350 MPa* and a tensile strength of 480 MPa in sections up to 100 mm thick.

Reference to the literature[4] shows that, with the exception of manganese and nickel, most alloying

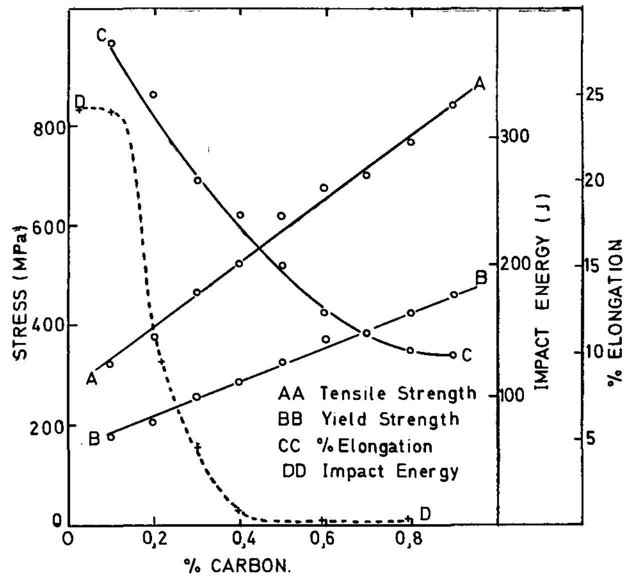

Fig. 9—Influence of carbon content on the mechanical properties of carbon steels in the hot rolled condition

elements increase the B.T.T. Table III shows very approximately the effect of a variety of elements.

TABLE III

INFLUENCE OF ALLOYING ELEMENTS ON THE B.T.T.

Alloying elements	Composition range (%)	Change in B.T.T. per 0,1% alloying element
Nitrogen	0,003-0,02	+184 °C
Phosforous	0,01 -0,06	+ 56 °C
Carbon	0,1 -0,3	+16,7 °C
Chromium	0,005-1,00	+ 2,8 °C
Copper	0,03 -2,00	+ 2,2 °C
Sulphur	0,02 -0,05	0
Nickel	0,04 -2,5	− 2,5 °C
Manganese	0,01 -1,5	− 5,6 °C

The alloying elements silicon and aluminium have been omitted since in small quantities, these elements have a beneficial effect whereas in larger percentages they result in an increase in the B.T.T. This is due to the high affinity of silicon and aluminium for oxygen.

In small percentages these elements are added for the purpose of deoxidation and consequently decrease the B.T.T. by removing the deleterious effect of oxygen. Aluminium has the further advantage that it also combines with nitrogen, and in addition promotes the development of a fine grain steel. Any silicon and aluminium in excess of that required to combine with oxygen and nitrogen will, however, result in an increase in the B.T.T.

Apart from the influence of the above alloying elements, the B.T.T. is also lowered by 14 °C for every increase in one ASTM grain size number (decrease in grain size). Elements such as niobium and vanadium are sometimes also added to structural steel to facilitate the attainment of a small grain size.

From a consideration of the above it is quite clear that it is pointless to specify structural steel according to its chemical composition. Its strength and, more important, the B.T.T. will also be influenced by the steel

* 1 MPa is approximately 145 lbs. per sq. in.

making process, the rolling temperatures and sequences and the section thickness.

This particular fact has in fact been recognised by the latest British specification for weldable structural steels (BS. 4360 Part 2, 1969). The specification provides for four grades of steels (40, 43, 50 and 55) with minimum tensile strengths of 400, 430, 500 and 550 MPa respectively. Each grade is subdivided into classes A B C D & E providing (except for class A) for steels with guaranteed impact values at various sub zero temperatures. Grade 40, class B for example requires a minimum Charpé value of 27 J at room temperature in plate material up to 50 mm thick whereas classes C, D & E require the same impact value at 0 °C, −20 C and −50 °C respectively. Reference to Table II shows that Cor-ten A, as far as its mechanical properties are concerned, complies with grade 50 D. After strain-ageing it still complies with the requirements of grade 50 C. '20/30' Carbon steel in the "as received" condition could not even meet the requirements of a grade 40 B steel.

HEAT TREATED STEELS

By quenching and tempering at appropriate temperatures it is possible to produce superior properties in terms of strength and B.T.T. A large number of steel manufacturers today market pre-quenched and tempered constructional alloy steels (HSLA) in the form of plates, bars and a limited number of sections. The carbon content of these steels is usually kept below 0,2 per cent C with the result that with appropriate preheating welding can be effected in the heat treated condition. In order to obtain the necessary hardenability in thick sections liberal use has been made of a large number of alloying elements such as 1 per cent nickel, 0,50 per cent chromium, 0,50 per cent molybdenum, 0,50 per cent copper and 0,005 per cent boron. In this way minimum yield strengths of 700 MPa and tensile strengths of 900 MPa coupled with a percentage reduction in area of 40-50 per cent and a B.T.T. below −30 °C can be realized in sections up to 60 mm thick. With this strength level the steel is still reasonably machinable with modern cutting tools. By using this type of prehardened and tempered steel instead of mild steel a weight saving of about 50 per cent can be effected which makes it very attractive for use on skips where the payload can be increased proportionately. Iscor is at present installing equipment to heat treat large plates for this purpose.

STRAIN AGING

Strain aging in mild steel occurs as a result of aging after prior plastic deformation. During aging, which can occur naturally at room temperature or be induced artificially by heating, diffusion of carbon and nitrogen atoms takes place to dislocations with the result that they are anchored in position. Consequently the yield and tensile strengths are increased after aging whereas the ductility is decreased slightly. Apart from the beneficial increase in strength, strain aging is unfortunately also very detrimental due to the fact that it increases the B.T.T. The influence of 6 per cent prior strain followed by aging for 15 minutes at 150 °C in the case of '20/30' carbon steel and cor-ten steel, is shown in Figs. 4 and 5. Based on impact values of 20 J, strain aging in the case of '20/30' carbon steel in the normalized condition increased the B.T.T. from −7 °C to 37 °C representing an increase of 44 °C. This compares with an increase of 50 °C and 36 °C in the case of mild steel and cor-ten A respectively.

During service plastic straining can occur at localised positions of high stress especially where high working stresses are used. Subsequent aging during use can by virtue of the fact that the B.T.T. is then possibly increased to above the service temperature, result in the nucleation of a brittle fracture at some small structural defect.

This problem of strain aging can in theory be solved in three ways. In cases where the maximum applied loads are well known and where the exact magnitude of the resultant stress at localised positions of stress concentration can be calculated, the component can be so dimensioned that localised plastic straining cannot occur at all under the worst possible combinations of operating conditions. This in many cases involves the employment of unrealistically high safety factors since the stress concentration factors presented by undetected defects are never known accurately. Alternatively the component can be heat treated appropriately at specified intervals before any appreciable aging has occurred at the service temperature. This approach is the basis of Regulation 16,18 of the present Mines and Works Act which requires that vital connections between the rope and the skip be annealed or given proper heat treatment or shall be discarded at intervals of not more than six months. From the results which have been presented previously it is clear that a periodic full anneal would be disastrous since this would lead to a very high B.T.T. Even a subcritical anneal cannot be expected to relieve the dislocation density of slightly strained material and cannot be expected to be really effective in restoring the original condition of the material. The only really effective method to restore the original properties[5] would be normalizing. This requires a fairly high temperature which if not properly controlled can lead to surface decarburization and a consequent reduction in the fatigue strength. In the case of irregular massive components high residual stresses may be induced and warping of the component may occur as a result of normalizing. If normalizing in such cases is not immediately followed by a subsequent stress relieving heat treatment, plastic straining, followed by strain aging, is bound to occur during the first reapplication of the load. For this reason it has become standard practice simply to discard such components after six months of service.

The last alternative would be to use a steel with such a low B.T.T. that even after strain aging the B.T.T. would still be below the lowest service temperature and therefore still safe even though strain aging has occurred. Steels which are highly resistant to aging (non-aging) would be of value in this context.

Reference to Table I shows that the '20/30' carbon steel which was tested contained the maximum amount of carbon which the specification allows, whereas the percentage silicon was 0,4 per cent below specification. In the "as received" condition the B.T.T. was 27 °C.

Strain aging of the "as received" material would most probably have increased the B.T.T. to a value in the neighbourhood of 60-70 °C. If this is compared with the lowest operating temperature of possibly −10 °C it can at best be termed dangerous. These results certainly do not serve to instill confidence in '20/30' carbon steel as a constructional material for vital components such as Humble hooks and at the same time illustrate the fallacy of specifying a chemical analysis instead of a grade and a class of steel according to the recent BS. 4360 spec. However by an appropriate heat treatment of this material such as quenching and stress relieving no objection to its use can be raised.

Table II shows that in the case of Mangear the B.T.T. after strain aging is still only −10 °C and as such acceptable. Mangear, a 1,5 per cent Mn 0,15 per cent C mild steel, has in terms of the provisions of the Mines and Works Act been exempted from the requirements of periodic heat treatment. Similarly reference to BS. 4360 for example shows that 1,5 per cent manganese steel fully deoxidised and with 0,1 per cent Nb added can be expected to meet the specification for a grade 50 D steel with a guaranteed B.T.T. below −30 °C. Such a steel could indeed be exempted from the customary six-monthly heat treatments.

SIX-MONTHLY HEAT TREATING INTERVALS

The speed of aging after straining is dependent on a large number of factors such as temperature, the percentage carbon, and nitrogen in solid solution as well as the type of straining. For any particular steel the time (t) required for artificial aging at a temperature (T °K) can be related to the time (t_r) required for aging at room temperature (T_r °K) by the following relationships[6] in the case of carbon and nitrogen aging respectively

$$\log \frac{t_r}{t} = 4\,400 \left(\frac{1}{T_r} - \frac{1}{T}\right) - \log \frac{T}{T_r} \text{ (carbon aging)}$$

and

$$\log \frac{t_r}{t} = 4\,000 \left(\frac{1}{T_r} - \frac{1}{T}\right) - \log \frac{T}{T_r} \text{ (nitrogen aging)}$$

From these relationships the influence of the equivalent carbon aging time at various temperatures is shown in Table IV.

TABLE IV
EQUIVALENT AGING TIMES AT VARIOUS TEMPERATURES

15 °C	21 °C	33 °C	100 °C	150 °C
1 year 4 weeks	6 months 2 weeks	1½ months 1 week	4 hours 20 min.	10 min. —

Table IV shows that an increase in room temperature of 6 °C doubles the rate of aging. If two hypothetical mines are compared where the average temperatures are respectively 21 °C and 33 °C (a coal mine in Scotland and a deep gold mine in South Africa for example) Table IV shows that heat treatment of detaching hooks should occur every 1½ months in the case of a hot mine if 6 months is considered the maximum allowable time in the case of a cool mine. The legal requirement of six-monthly intervals regardless of the service temperature seems rather arbitrary.

Fig. 10 shows the increase in hardness of '20/30' carbon steel and mild steels during strain aging in boiling water. In both cases the steels in the "as received" condition were first strained 6 per cent immediately before aging. From the graph it is clear that aging was virtually complete after 20 minutes at 95 °C. At a service temperature of 27 °C full aging can be expected in about 2 weeks.

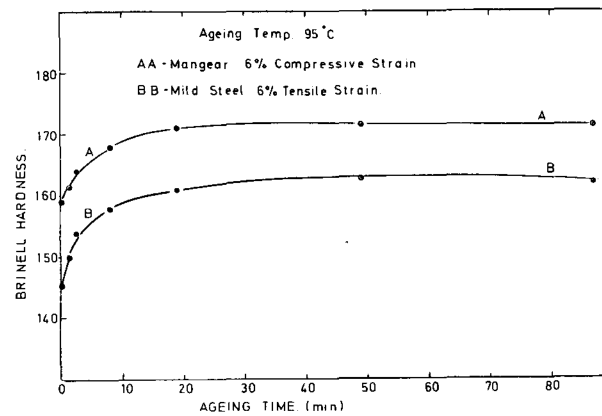

Fig. 10—Increase in hardness during strain aging in boiling water

In practice aging would occur with the components under stress. Available experimental evidence suggests that strain aging under stress (stress aging) takes place even faster necessitating even shorter heat treating intervals. This shows that the legal requirement of six months for such heat treatment is much too long to prevent strain aging.

Apart from the aging temperature the speed of aging is also determined by the percentage carbon and nitrogen in solid solution in the steel. It has been shown[7] that under equilibrium conditions the solubility of nitrogen is about 100-1 000 times greater than that of carbon atoms. Consequently the rate of strain aging can be reduced considerably by the addition of aluminium to steel which will then by chemical combination reduce the percentage nitrogen left in solid solution. In practice, however, the cooling rate during normalizing heat treatment is too high to reduce the percentage C to its equilibrium value and a large degree of super saturation is available for carbon strain aging.

Even the type of straining apparently plays a role in the speed of aging. Hundy[6] reported that aging occurred in a fully killed extra deep drawing quality mild steel sheet within 3 days at room temperature in the case of samples strained 5 per cent in a tensile testing machine whereas 3 months were required if the straining was performed by 4 per cent temper rolling.

The safety factors which are at present used in the design of rope attachments are very conservative and if any plastic straining occurs under the worst combination of service condition in material which is free of residual stress it will be of a very localised nature at positions of high stress concentration. Such positions will subsequent to straining contain a favourable residual stress pattern in opposition to the applied stress. In addition, local aging will then increase the yield strength above which it was prior to plastic straining with the result that the chance of subsequent straining at the same position is rather remote. Fig. 6 shows that 3 per cent strain aging can increase the yield stress by about 28 per cent. Such strain aging can be very beneficial in assisting the material to support the applied stress. If on the other hand the applied stresses are so high that repeated plastic yielding takes place at such positions in spite of strain aging, it is quite clear that the material will be subjected to fatigue as well and can be expected to fail.

Considering all the factors which influence strain aging and the speed with which it takes place at room temperature it appears as if the legal requirement of "annealing" every six months is absurd. A much more sensible clause would rather be a stipulation of the quality of the steel in terms of the maximum B.T.T. (say $-30\,°C$) which is allowed for critical components of winding plant. It is also clear that suitably low B.T.T. can be obtained in Mangear steel of the correct composition but also in ordinary mild steel and '20/30' carbon steel if a water quenching followed by a stress relieving heat treatment is used. Even in the case of Mangear an improvement in quality is possible by a quench and stress relieving heat treatment[8].

FATIGUE FAILURE

Most of the experimental work shows that within limits a high B.T.T. does not influence the fatigue strength materially. In cases where cyclic loading is present careful attention is still necessary in the design stage to ensure that the danger of metal fatigue is reduced as far as possible. For this purpose it is not only necessary to eliminate stress raisers in the design but also to follow a rigid quality control inspection procedure to ensure that vital components are, as far as possible, free from metallurgical defects such as laps, seams, cracks, segregation, laminations and surface decarburization which can reduce the fatigue strength. The service conditions should also be appraised critically. Components such as rope attachments which are subjected to corrosion should as far as practical be protected by hot dip galvanizing, or if that is impossible by zinc spraying. Protection against corrosion is especially important if prehardened and tempered alloy steel components with high fatigue strengths are contemplated to effect a weight saving.

Although a high B.T.T. does not have an adverse effect on the fatigue strength, fatigue cracks can be expected to propagate much further in a material with a low B.T.T. In the case of a high B.T.T. material rather small cracks can be expected to propagate in a brittle fashion. Thus, a low B.T.T. lends further advantage because it gives the inspection personnel a better chance of detecting such cracks during the regular inspections.

CONCLUSIONS AND RECOMMENDATIONS

Material specification

Mangear steel of the correct composition should be a perfectly satisfactory material for critical applications. Due to the fact that the final properties, and especially the B.T.T., are dependent on the steel making practice it is suggested that in addition to a specification on the usual mechanical properties a specification on the B.T.T. is advisable. It is recommended that this is best done by the use of the BS. 4360 specification and that a class C material as a minimum requirement should in all cases be specified. By specifying grade 55C it would be possible to reduce the weight of such components about 25 per cent in comparison with manganese mild steel used at present.

Since the higher grades 50 and 55 of BS. 4360 are sometimes difficult to manufacture due to the fact that controlled rolling may be required, the production in relative small volumes in South Africa may not be economical. As an alternative it is recommended that either Cor-ten B or C be considered. The higher alloy content and the lower carbon content of the Cor-ten variety of steel make it much easier to obtain the required properties. Due to the high demand for Cor-ten for other structural purposes it would be much easier to obtain the required sections in small quantities from local production by Iscor, eliminating the import necessity. As far as mechanical properties are concerned Cor-ten B is approximately equivalent to a grade 50 and a Cor-ten C to a grade 55.

Heat treatment

Both Mangear to BS. 4360 or Cor-ten are normally supplied with a pearlitic structure and no additional heat treatment should be necessary. By water quenching and stress relieving some improvement in properties can be achieved, depending on the ruling section. In small sections quite a considerable improvement is possible whereas in thicker sections only marginal improvement in tensile strength can be expected.

In order to benefit fully from a heat treatment such as quenching and tempering, a steel with a higher alloy content and consequently a higher hardenability should be used. By the use of USS—T1 type A or B for example a weight saving of almost 50 per cent is possible in comparison with manganese mild steel. This type of steel is normally supplied in the quenched and tempered condition and no further heat treatment would be necessary. It is, however, not yet manufactured in South Africa and supplies would have to be imported. Iscor is at present installing the necessary equipment to heat treat plates from 5-150 mm thick, 3,2 m wide, and up to 13,7 m long. Fully heat treated plates can be expected to be available early in 1972. Sections and rounds, however, will still have to be imported.

Mines and Work Regulation **16,18**

From the arguments presented it is clear that any material with an impact energy in excess of 20 J below -30 C would satisfy general safety requirements and should qualify as Mangear for relief from regulation 16,18. On the other hand it has been shown that material with a marginal B.T.T. may be embrittled by strain aging to such an extent that the statutory requirement of "annealing" every six months is not sufficient to eliminate the danger of a brittle fracture, and it is suggested that mild steel without any specification with respect to B.T.T. should be prohibited altogether for critical application.

When Regulation 16,18 was initially incorporated in the Mines and Works Act no really effective non-destructive testing means was practical. Today the situation has changed radically and this fact should be recognized.

As an alternate to the statutory requirement of "annealing" every six months it is suggested that the Act be revised so as to require dismantling and testing by non-destructive methods of all vital components of a winding plant every year.

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