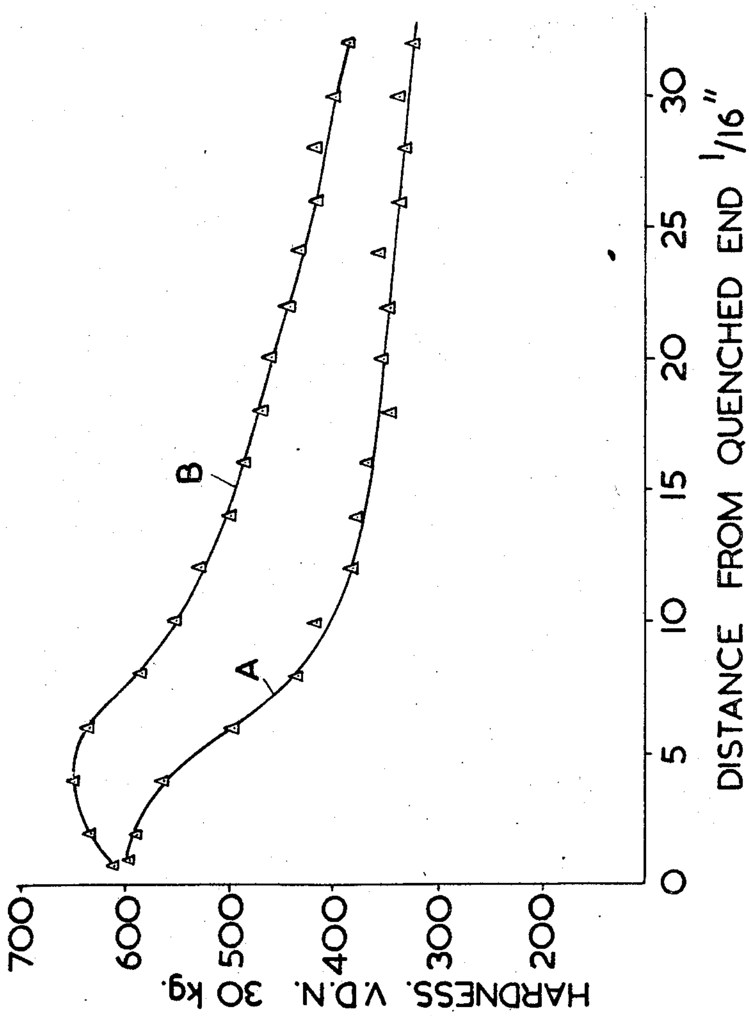

HARDNESS. V.D.N. 30 kg.
700
600
500
400
300
200
DISTANCE FROM QUENCHED END 1/16"
0
5
10
15
20
25
30
A
B

FIG. 1.

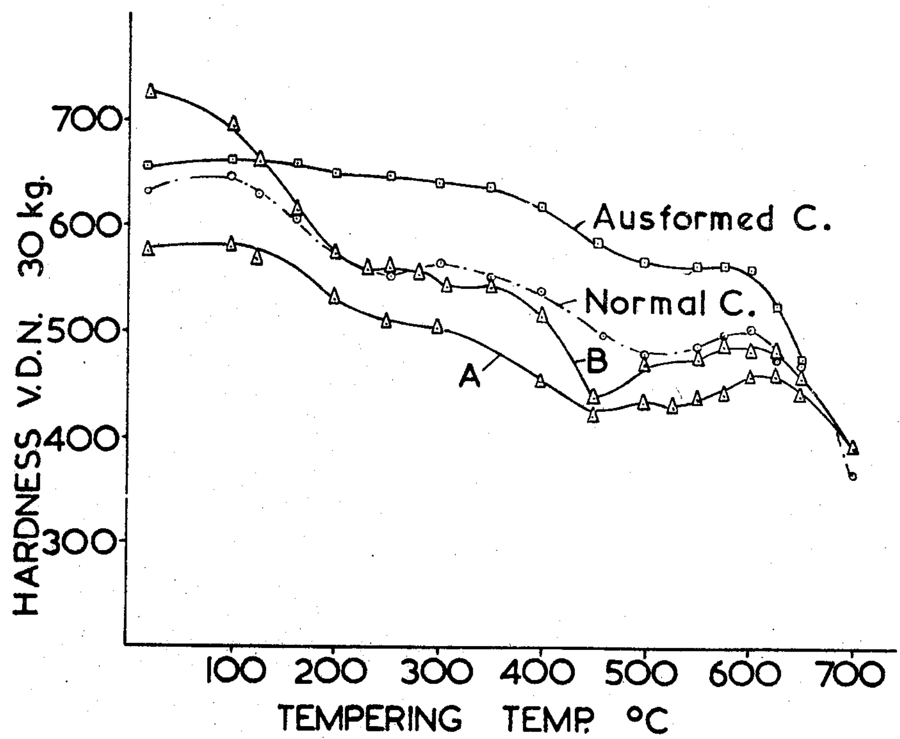

HARDNESS V.D.N. 30 kg.
700
600
500
400
300
Ausformed C.
Normal C.
B
A
100
200
300
400
500
600
700
TEMPERING TEMP. °C

FIG. 2.

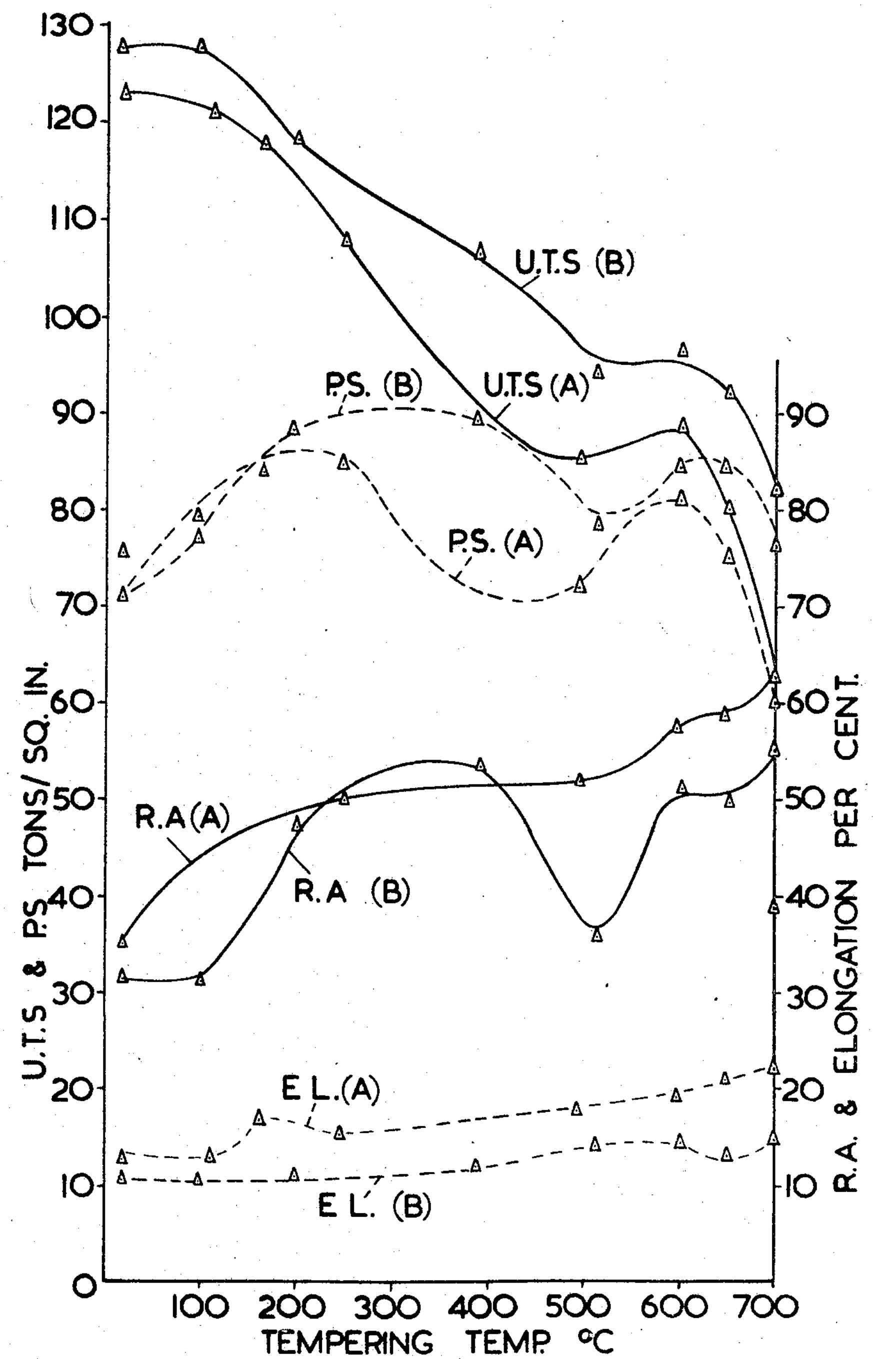

U.T.S (B)
U.T.S (A)
P.S. (B)
P.S. (A)
R.A (A)
R.A (B)
E L.(A)
E L. (B)
U.T.S & P.S. TONS/SQ. IN.
R.A. & ELONGATION PER CENT.
0
10
20
30
40
50
60
70
80
90
100
110
120
130
100 200 300 400 500 600 700
TEMPERING TEMP. °C

FIG. 3.

HEAT TREATABLE ALLOY STEELS

Filed Dec. 19, 1966

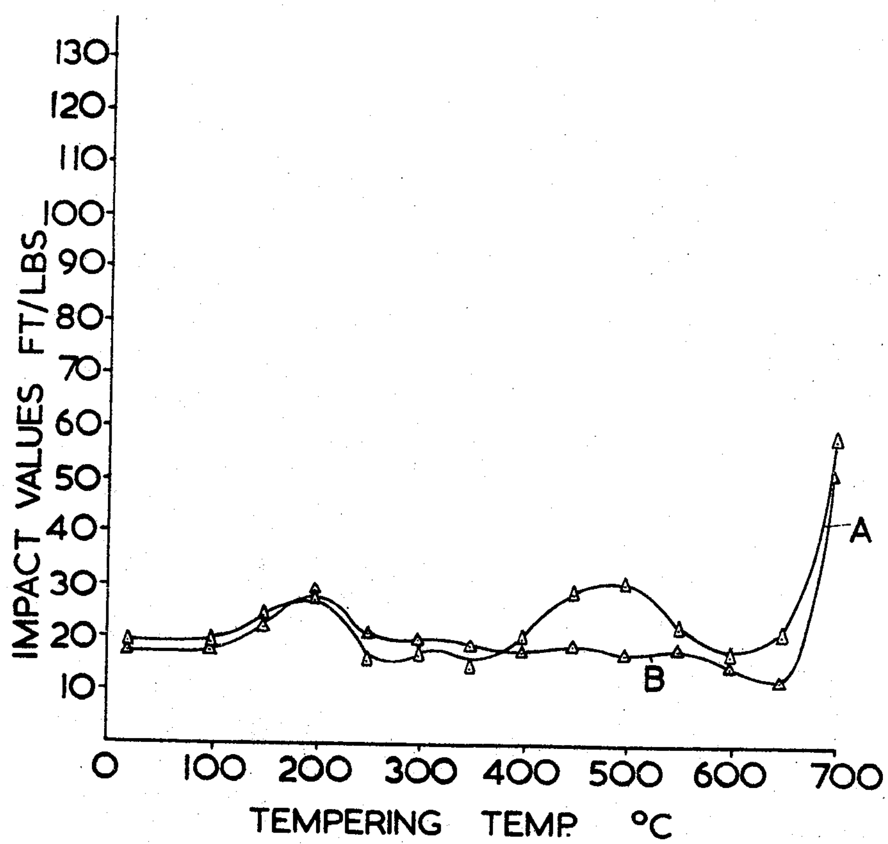
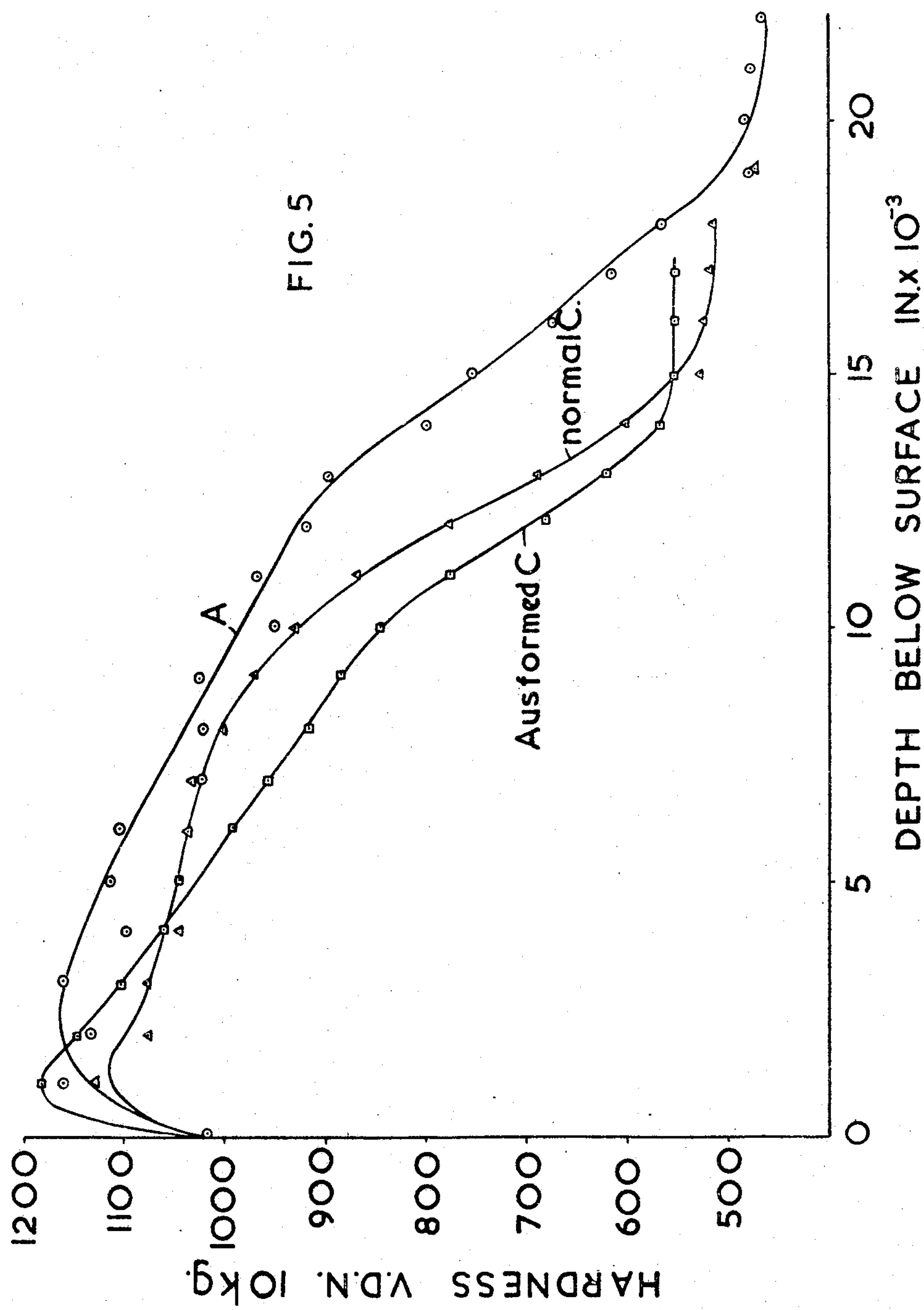


FIG. 4.

HEAT TREATABLE ALLOY STEELS

Filed Dec. 19, 1966 7 Sheets-Sheet 5

FIG. 5



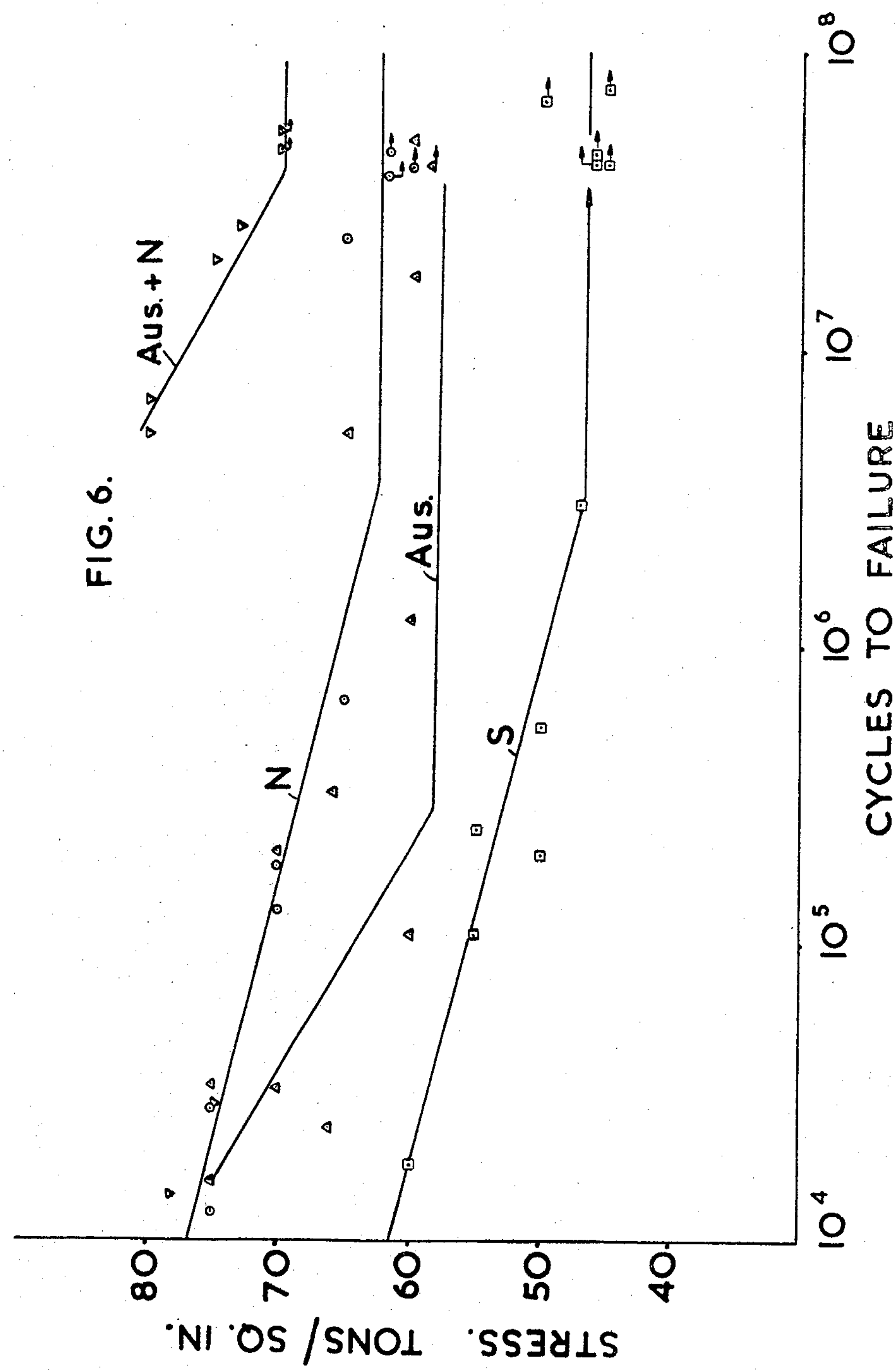

Aus.+N
Aus.
N
S
80
70
60
50
40
STRESS. TONS/ SQ. IN.
10⁴
10⁵
10⁶
10⁷
10⁸
CYCLES TO FAILURE

FIG. 6.

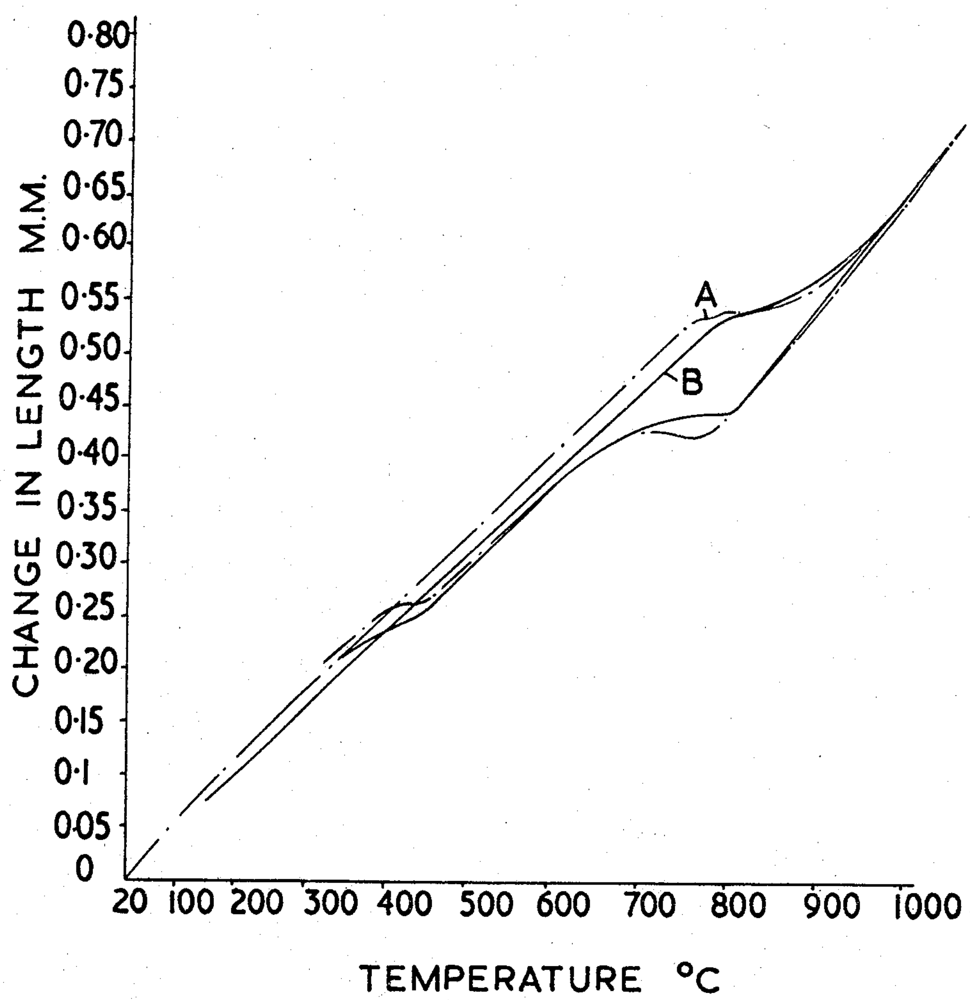

0·80
0·75
0·70
0·65
0·60
0·55
0·50
0·45
0·40
0·35
0·30
0·25
0·20
0·15
0·1
0·05
0
CHANGE IN LENGTH M.M.
A
B
20 100 200 300 400 500 600 700 800 900 1000
TEMPERATURE °C

FIG. 7.

United States Patent Office 3,492,116
Patented Jan. 27, 1970

3,492,116

HEAT TREATABLE ALLOY STEELS

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Claims priority, application Great Britain, Dec. 22, 1965, 54,329/65
Int. Cl. C22c 39/04, 41/02
U.S. Cl. 75—124 3 Claims

ABSTRACT OF THE DISCLOSURE

High strength medium-carbon alloy steels are provided consisting of about 0.1–0.6% carbon, about 0.25–5.0% manganese, about 0.5–2.0% aluminum, about 0.5–3.0% molybdenum, about 0.01–2.0% silicon and about 0.2–1.0% vanadium, the balance being essentially iron. These alloys which may be heat treated by soaking above the AC_2 transition temperature, quenching and tempering, retain the properties of great resilience, hardness, strength, and good ductility to a remarkable degree when the steel is tempered and subsequently used at temperatures up to 600° C. or even higher temperatures.

The present invention relates to high-tensile alloy steels which are heat-treated by heat-soaking above the AC_3 transition temperature, quenching in oil or water, for example, to avoid transformation into bainite, and then tempering. In the course of the heat-treament the alloy steels may be ausworked before quenching.

The invention provides a new heat-treated manganese alloy-steel which has great resilience, hardness and strength and good ductility and which is very resistant to tempering, so that it retains these properties to a remarkable degree when the steel is tempered and subsequently used at temperatures up to 600° C. or even high temperature.

According to the invention, an improved medium-carbon, manganese alloy-steel contains about 0.1–0.6% carbon, about 0.25–5% manganese, about 0.5–2% aluminium, about 0.01–2% silicon, about 0.5–3% molybdenum and about 0.2–1% vanadium, the remainder being essentially iron with any of its common impurities, e.g. nickel and chromium and the non-metallic impurities sulphur and phosphorus in small amounts usual in commercial steels.

According to a particular feature of the invention, the alloy-steel contains about 0.5–2% silicon and as a result the hardenability and tensile strength throughout the whole tempering temperature range is extremely high. On the other hand if the silicon content is in the range 0.01–0.5% the alloy steel can have greater impact strength at ambient temperatures whilst still having very high tensile strength and hardenability.

Preferably the carbon content of the alloy is 0.15–0.45% and the manganese content between about 0.5 and 3% in particular between about 0.75 and 1.5% provided the steel is not subsequently ausformed. The aluminium content of the steel is preferably 0.75–1.5% while the molybdenum content is preferably 0.5–1.5% and the vanadium content is preferably 0.2–0.5%. The preferred silicon content of the alloy steel is between 0.5 and 1.5%.

Alloy steels in accordance with the invention have properties in many respects equivalent or superior to those of more expensive steels such as those containing significant amounts of chromium or of other expensive alloys such as titanium alloys. When the heat treatment of the alloy steel has involved tempering at temperatures of at least 500° C.

and up to about 650° C. its properties are particularly good. For example, an alloy steel in accordance with the invention, having a carbon content of 0.35%, when heat-treated with tempering at 600° C. can have an ultimate tensile strength and 0.1% proof stress of 80–90 tons/sq. in. and 75–85 tons/sq. in. respectively, and a hardness of about 450 V.D.H. (Vickers Diamond Hardness) and a Charpy impact value of over 16 ft. lb. Alloy steels in accordance with the invention have very good fatigue properties and workability e.g. hot-working such as forging or rolling.

Alloy steels in accordance with the invention may be ausformed, i.e. soaked above the AC_3 temperature, cooled to a temperature within the austenite range, at a rate sufficient to prevent transformation of the steel into ferrite and pearlite, ausworked, quenched and tempered. Preferably steels to be ausformed contained at least about 2% manganese.

Ausforming has the effect of raising the already outstanding tensile strength of those steels by about 20 tons/sq. in. for very little loss in ductility. The resistance to tempering, fatigue limit and charpy impact results are also effected benefiicially by ausforming.

Very high values of surface hardness, e.g. around 1,000 V.D.H., can be imparted to alloy steels by nitriding after heat soaking, quenching, and tempering. However, all the nitriding alloys steels commercially used at present are expensive in that they contain significant proportions of chromium. Furthermore, these steels have the disadvantage that they do not retain good mechanical properties at temperatures around 500° C. at which the nitriding process must take place. Thus high surface hardness is gained at the expense of core strength, and their use in highly stressed components, or components operating in high temperature conditions, tends to be limited.

An outstanding property of the alloy steels in accordance with the invention is that they can be nitrided to produce a very deep rate of high surface hardness, without impairing the advantageous properties of the core as above described. On the contrary, notwithstanding the retention of high strength at high temperatures, the fatigue strength of the alloy steels is still considerably increased by nitriding; (the fatigue limit can be increased from a stress of about ±45 tons/sq. in. to over ±60 tons/sq. in.) so that alloy steels in accordance with the invention can have, when nitrided, most exceptional properties, combining high tensile strength with very high fatigue strength, high surface hardness and excellent workability, which properties are highly desirable for highly stressed components such as gears which are subject to wear and sliding contact in service. The high fatigue strength also makes the steel valuable for use in torsion bars and springs. Furthermore, these exceptionally good properties are possessed by an alloy steel which is not so inherently expensive as special purpose steels and other alloys and so can find wide application in the field of more common commercial steels particularly, for example, where a high strength-weight ratio is desirable.

An alloy steel in accordance with the invention as above defined may, after a heat-treatment which involves tempering at a temperature of more than 500° C., be nitrided at a temperature less than the previous tempering temperature. The nitriding is advantageously carried out at the normal temperature of about 500° C. and the tempering is then carried out at a temperature above 500° C. selected to give the alloy steel properties as appropriate as possible for the designed use. In any case, as the nitriding temperature is less than the preceding tempering temperature, nitriding will not cause deterioration of the mechanical properties of the alloy steel.

Preferably an alloy steel as above defined is heat-treated which involves tempering at about 550° C. to about 650° C. followed by nitriding at about 500° C.

The depth of the surface region having an increased hardness increases with the duration of the nitriding treatment. An increased hardness can be produced in the surface layer to a depth of about 0.018 in. by nitriding the alloy steel for about 72 hours after quenching and tempering to 600° C.

Steels in accordance with the invention may be ausformed and nitrided, that is to say heat-soaked at above the AC_3 transition temperature, cooled to a temperature within the austenite range, worked preferably deforming the steel by at least 50%, quenched, tempered at between 550° C. and 650° C., and nitrided. By this treatment it is possible to improve the excellent tensile, impact, and hardness properties of the steel as nitrided but without ausforming and as a result to impart to the steel a fatigue limit hitherto thought unattainable and which is in fact greater than the tensile strengths of some mild steels, i.e. over 70 tons/sq. in. The ausformed steel may be used in high performance gears and tools operating at high temperatures; for example they may be used in tools used for ausforming.

The outstanding properties of the alloy steel of the invention are illustrated by the results of tests carried out on three typical alloy steels (called A, B, and C) in accordance with the invention. The percentage content of the alloying elements in alloy steel A was as follows: 0.35 carbon, 1.0 manganese, 1.0 aluminium, 0.75 molybdenum, 0.03 silicon and 0.25 vanadium.

Alloy steel B had the same composition as did steel A but with 1.04% silicon. The steel C contained 0.35% carbon, 2.05% manganese, 0.85% molybdenum, 0.26% vanadium, 1.13% aluminum, and 0.56% silicon.

The results of the tests on these alloy steels are shown graphically in the accompanying FIGURES 1 to 7, in which:

FIGURE 1 is a graph showing the end quench hardenability of steels A and B,

FIGURE 2 is a graph showing resistance to tempering of the steels A and B, and C when soaked above the AC_3 temperature, quenched and tempered, and of the steel C when ausformed and tempered,

FIGURE 3 is a graph showing the effect of tempering on the tensile properties of the steels A and B,

FIGURE 4 is a graph showing the effect of tempering on the Charpy impact (V notch) values of the steels A and B,

FIGURE 5 is a graph showing the case hardnesses of the steel A when quenched and tempered at 600° C. and nitrided for 72 hours at 500° C., and of the steel C when ausformed (Aus) and when ausformed, tempered at 600° C. and nitrided for 72 hours at 500° C. (Aus+N),

FIGURE 6 is a graph showing the effect of nitriding on the fatigue properties of the steel A and the ausformed steel C, and

FIGURE 7 is a dilatometer curve for steels A and B.

The results of dilatometric analysis of the three steels are given in the Table 1 below for heat-soaked, quenched and tempered steels A, B and C and ausformed steel C.

TABLE 1

Steel	Critical Range, ° C.	Martensite breakdown	
		Stage I, ° C.	Stage III, ° C.
A	762–959	114	247–495
B	788–997	104–202	391–515
C	718–930	89–206	393–500
Ausformed C	727–905	----------	420–553

A steel of comparable composition but without aluminium had a critical range of 768–895° C., a stage I of 77–195° C., and a stage III of 383–540° C. The presence of aluminium has therefore raised the martensite breakdown temperatures considerably, with relatively little effect on the critical range. Silicon is known to raise the critical temperatures of steel, but it has been hitherto unknown how it would act when the steel also contains aluminium, whose effect on the martensite breakdown temperatures is considerable. The results for steel A and steel B show that the effects of silicon and aluminium in this respect are additive. The results for normal and ausformed specimens of steel C show that while ausforming has little effect on the critical range, the third stage of martensite breakdown is raised. The first stage of martensite breakdown was eliminated by ausforming.

FIGURE 1 shows the hardenability of specimens of steels A and B under standard end quench conditions (S.A.E. Handbook 1947), at various depths from an end which has been water quenched after heat soaking at 1075° C. for one hour. Both steels are good but the silicon-containing steel B shows definitely higher hardenability values.

FIGURE 2 shows the effect on the hardness of the quenched alloy steel of tempering carried out at various temperatures. The values of Vickers Diamond hardness V.D.H. at a 30 kg. load given on the graph clearly shows the high resistance to softening of the steel on tempering up to temperatures of about 600° C. Steel B is shown to be generally harder and its resistance to tempering greater than the basic alloy steel A. The steel C in its heat-soaked, quenched, and tempered (normal) condition shows results which are commensurate with those for steels A and B. By ausforming, however, the steel is rendered even more resistant to tempering and remained harder even when tempered to 700° C. A further point of interest is that ausforming eliminates the secondary hardening peaks normally found when tempering steels containing molybdenum and vanadium despite a carbon content of 0.35%.

FIGURE 3 shows the effect on the tensile properties of steels A and B of the standard heat-treatment involving tempering for one hour at various temperatures. The specimens used in the tests had a gauge length of 1.4″ (3.556 cm.). Good tensile properties of the new alloy steels are maintained up to temperatures of about 650° C. They are particularly good after tempering at 600° C., the higher silicon-containing steel not just being the stronger, but being relatively stronger to an increasing extent with increase in tempering temperature.

The following Table II compares the mechanical properties of steels A, B and C (normal) all after the standard heat-treatment and ausrolled steel C, after tempering at 600° C. in each case.

TABLE II

Steel	0.1% P.S., tons/sq. in.	U.T.S. tons/sq. in.	Elongation	R.A.
A	81.2	89	15.2	58.5
B	84.25	96.1	14.6	51.0
Normal C	85	96	14.3	49.7
Ausrolled C	105.6	110.7	13.7	45

The results for the steels A and B show that silicon raises the tensile strength of the steel by about 6% for little loss in ductility. The ausformed steel C, however, shows a 20% increase in tensile strength, again for little loss in ductility.

FIGURE 4 shows the Charpy impact values obtained from tests on standard specimens (10 x 10 x 56 mm. with a 45° notch 2 mm. deep) of the steels A and B subjected to the standard heat-treatment involving tempering for one hour at various temperatures. The graph shows that adequate values of over 14 ft. lb. are obtained for tempering steel A at temperatures up to 700° C., while a particularly good value of 30 ft. lbs. is obtained at 500° C. The higher silicon-containing steel B, however, exhibits a value of about 13 ft. lb. if tempered at 650° C., and in the preferred tempering range steel B is more brittle than steel A. Steel C, when quenched and tempered at 600° C. had an impact value similar to that of steel B (13.0–13.7 ft. lb.). However, ausforming and tempering at 600° C. had the effect of raising the impact value of steel C to 16.6 ft. lb.

Table III below shows the effect of temperature variation on the Charpy impact value of the alloy steel A, when it has been tempered at 600° C.

TABLE III

Testing temp., ° C.	Energy ft. lb.
100	32.9
Zero	22.5
−20	19.1
−40	18.1
−60	16.6

If plotted these results show that the steel has an impact transition temperature of about +20° C. The steel has, of course, been tempered at a temperature above the stage III of martensite breakdown (495° C.). Had it been tempered below 495° C. the results shown in Table III would have had a linear relationship, the impact valves below the transition temperature being then higher than, and those above the transition temperature lower than, the values shown at present. The presence of silicon in the steel would raise the transition temperature, and as stated earlier it raises the stage III temperatures. If the steel is to be used at temperatures lower than the impact transition temperature and also at temperatures up to 500° C., and an impact value of over 20 ft. lb., say, were required at −40° C. these conditions could be fulfilled by an alloy steel containing at least 0.5% silicon, tempered at 500° C. without passing through the stage III. If on the other hand the steel is required to have an impact value of about 30 ft. lb. at ambient room temperatures (20°–30° C.) with impact values up to or ever higher than 530° C. then a steel with less than 0.5% silicon would be preferable.

FIGURE 5 shows the hardness of the surface layer obtained by nitriding specimens of the alloy steels A and C after the standard heat-treatment with tempering at 600° C. and by nitriding specimens of steel C after ausforming and tempering at 600° C. The specimens were nitrided for 72 hours at 600° C. in an atmosphere of dried ammonia. The exceptionally great depth of the hardened case for all three steels is clearly shown.

FIGURE 6 shows a fatigue curve S for a set of specimens of alloy steel A given the standard heat-treatment involving tempering at 550° C. only and a fatigue curve N for another set of specimens of the same material similarly treated and then nitrided as above described for 72 hours. The graph clearly shows the considerable improvement in fatigue limit produced by nitriding, i.e., an increase in stress from about ±45 tons/sq. in. to over 60 tons/sq. in. FIGURE 6 also show a fatigue curve (Aus) for specimens of alloy steel C when ausformed and tempered at 600° C. and a fatigue curve (Aus+N) for specimens of alloy steel C ausformed, tempered at 600° C. and nitrided for 72 hours at 500° C. While the fatigue limit for the ausformed steel (58 tons/sq. in.) is good when compared with many known steels, that for the steel C ausformed and nitrided (70 tons/sq. in.) is outstanding. The fatigue limit for steel C given the standard heat treatment with tempering to 600° C. is 55 tons/sq. in. and that for steel C treated similarly and nitrided as above is 65 tons/sq. in. Thus it is consistently shown that nitriding imparts to steels of the invention fatigue endurance limits of exceptional magnitudes.

FIGURE 7 shows the dilatometer curves for steel A and steel B. Both of these steels exhibit a change in rate of volume change which is considerably less than that which accompanies the heating of most steels through the critical range. This means that the risk of quench cracking when cooling the steels of the invention is considerably reduced.

By tempering at 600° C., the steels are assured of good hardness and tensile properties as shown in FIGURES 2 and 3 and satisfactory Charpy impact values as shown in FIGURE 4 and Table III. Nitriding can be carried out at a conventional temperature, i.e. about 500° C., and excellent surface hardness as shown in FIGURE 5 and fatigue strength as shown in FIGURE 6 are obtained.

Commercial steels may contain up to about 0.5% silicon if no effort is made to restrict the amount. The control of silicon content, i.e. to less than 0.1%, can readily be achieved however, and as shown above the fulfilment of differing requirements for the steels can depend upon absence or presence of a significant proportion of silicon in the steel.

We claim:

1. A high strength medium-carbon alloy steel consisting of about 0.1–0.6% carbon, about 0.25–5% manganese, about 0.5–2% aluminum, about 0.5–3% molybdenum, about 0.01–2% silicon and about 0.2–1% vanadium, the balance being essentially iron.

2. An alloy steel according to claim **1** containing about 0.5–2% silicon.

3. An alloy steel according to claim **1** containing about 0.01–0.5% silicon.

References Cited

UNITED STATES PATENTS

3,326,675 6/1967 Kenneford ---------- 75—124

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W. W. STALLARD, Assistant Examiner

U.S. Cl. X.R.

148—12.1, 12.4