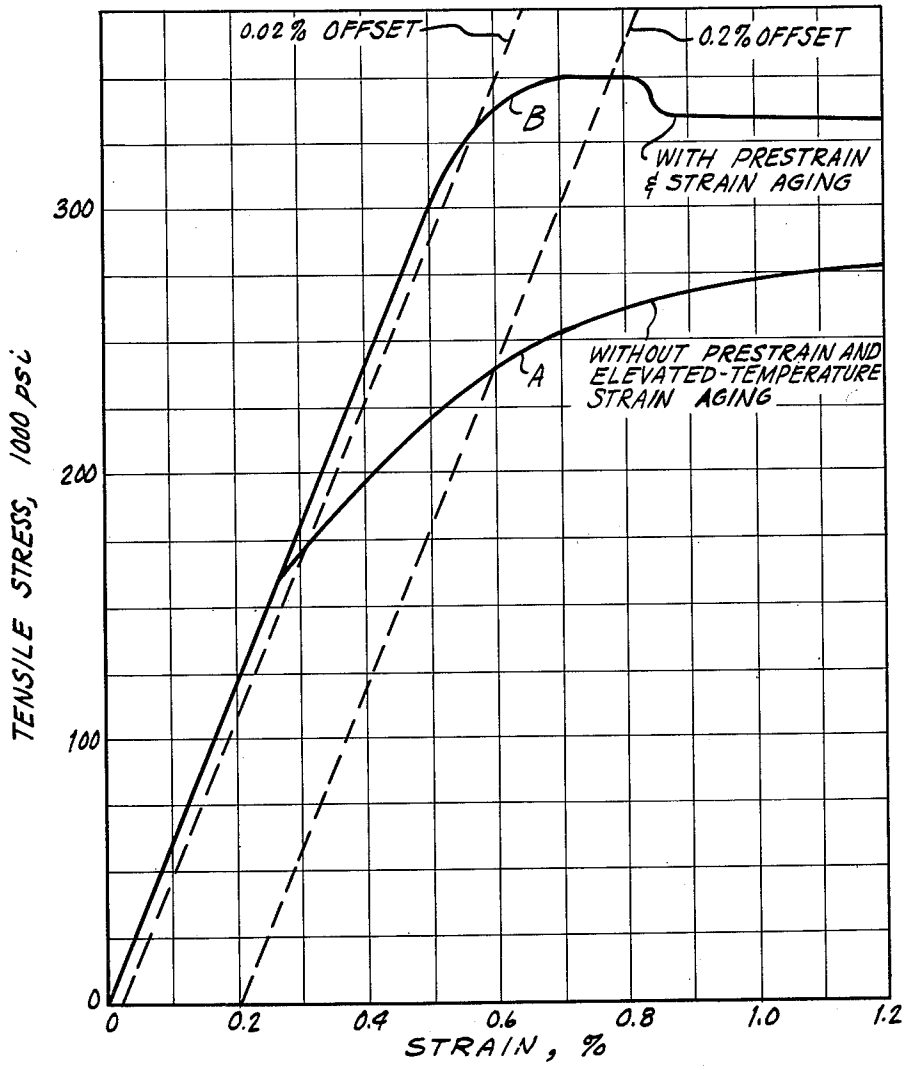
Aug. 3, 1965 J. P. TARWATER 3,198,630
SUPER STRENGTH STEEL ALLOY COMPOSITION AND PRODUCT AND PROCESS OF PREPARING IT
Filed Sept. 21. 1961 3 Sheets-Sheet 1

Fig. 1.



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SUPER STRENGTH STEEL ALLOY COMPOSITION AND PRODUCT AND PROCESS OF PREPARING IT

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Fig. 3.

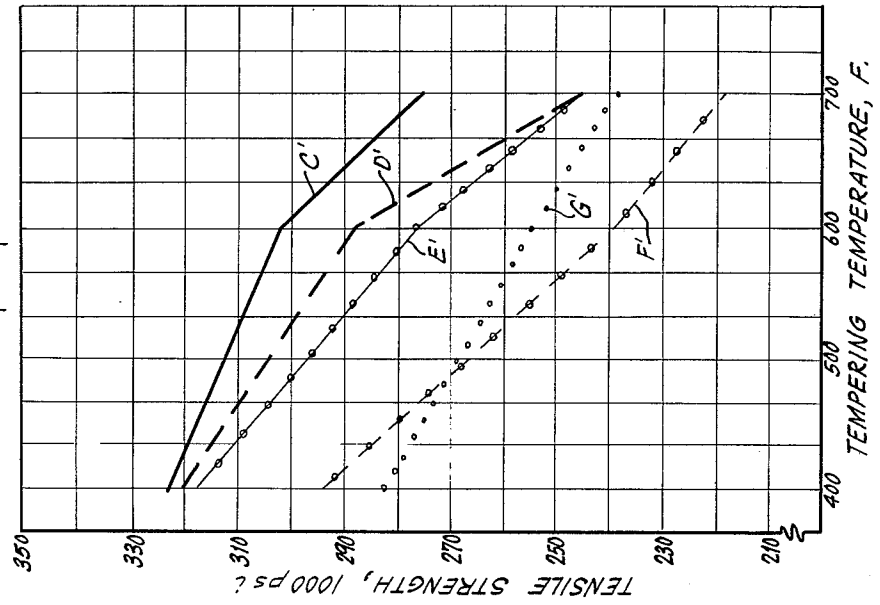
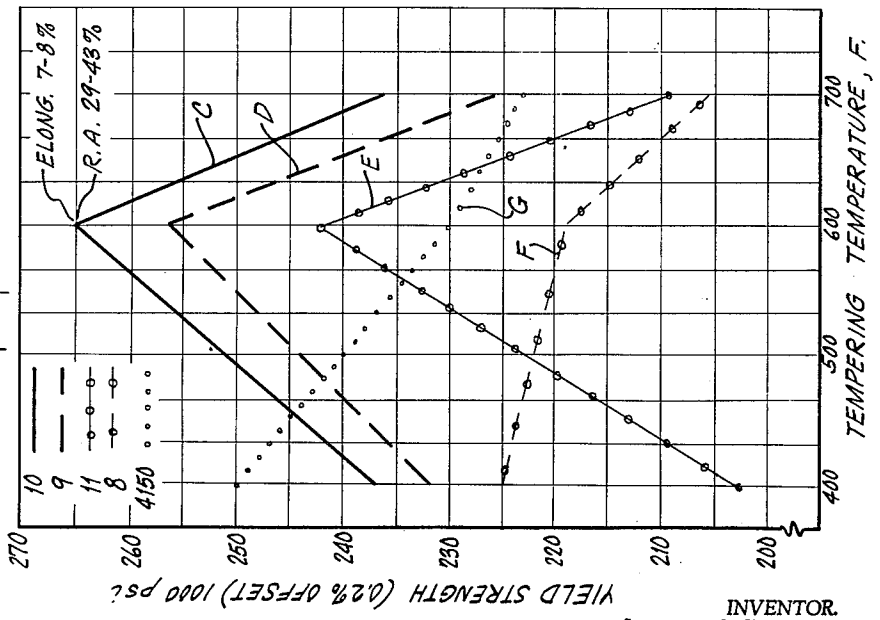


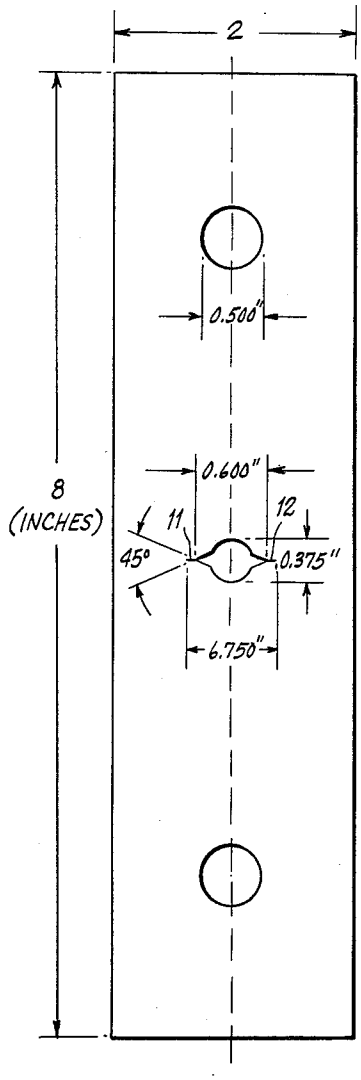
Fig. 2.



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Fig. 4.



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SUPER STRENGTH STEEL ALLOY COMPOSITION AND PRODUCT AND PROCESS OF PREPARING IT

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12 Claims. (Cl. 75—124)

This invention relates to super strength steel alloy composition and product and process of preparing it. More particularly the present invention relates to a steel alloy composition capable of being treated to produce a super strength body as one having a tensile strength substantially above 300,000 p.s.i., while having a substantially definite yield point which is substantially the same as the ultimate tensile strength and further, while having a high degree of ductility.

The invention further relates to the provision of an alloy steel composition of the super strength type, which will have a very high as-tempered strength attained over a substantial range of tempering temperatures, as from about 400° F. to about 900° F., such composition preferably including cobalt and/or aluminum.

The present invention further provides a super strength steel alloy composition which will have high fracture toughness as measured by the resistance of a notched sample of sheet material to crack propagation under stress.

Super strength steels have now become a recognized group of steel alloys, so that various compositions have been disclosed by different manufacturers of such steels along with their claimed tensile strengths, which in practically all instances are substantially less than 300,000 p.s.i. In most instances these steels are in the medium carbon range and include such other alloying elements as manganese, silicon, chromium, molybdenum, vanadium and nickel. In practically all instances, however, the nickel content has been quite low, rarely above 2%; and some other of the elements aforesaid have been present in amounts which are substantially above those contemplated as tolerable in accordance with the present invention, for example, chromium has usually been substantially higher than the upper limit given in the present invention of about 0.5%. The present invention by contrast with these may be termed a medium carbon nickle steel wherein the nickel present is from about 3 to 7% and for certain purposes, further limitation as hereinafter set forth are imposed upon the composition.

The prior art, for example, has also suggested a number of such compositions of what are referred to generally in the art as "Ladish-type" steels, certain of these being set out in U.S. Patent Nos. 2,919,188 and 2,921,849, both owned by Ladish Company, Cudahy, Wisconsin. These steels have a quite low nickel content, it being stated, for example, in the first of these patents:

"The nickel content is always less than the chromium content and is always less than the molybdenum content."

As hereinafter stated in greater detail, the compositions of the present invention attain results of tensile strength, yield strength and ductility which are substantially superior to the Ladish-type steels and to any other known group or type of steels available in accordance with the prior art teachings, both from the point of view of the "as-tempered" characteristics of yield strength and tensile strengths over a substantial range of tempering temperatures and also from the point of view of ductility accompanying the high yield strength and high tensile strength characteristics. Another group of these steels all within the general scope hereinabove outlined has greatly superior fracture toughness as contrasted with known super strength steels, which is a characteristic that is becoming more and more important in some of the uses for which such steels are

required today. One such use of prestrained and strain-aged materials is in the making of high velocity rotors such as are required for high speed precision pumps where close tolerances require a very high 0.02% offset yield strength. Other uses include but are not limited to tensile members of rigging, frames, structures, particularly for aircraft and missiles where high strength-to-weight ratios are required.

The prior art has also suggested a process of treating steel which is in some respects essentially similar to the process herein disclosed; except that this process has heretofore been used only on steels having low nickel content such, for example, as "AISI Type 4340" steel (1.74% nickel) and where the chromium content was substantially higher than that as desired or contemplated in accordance with the present invention (i.e. 0.82% chromium). This AISI-Type 4340 steel and the process of treating it has been described in considerable detail in the "Transactions of the American Society of Metals" Quarterly Edition for March 1961, pages 72–83. The fact that the resulting steel, while having super strength in tension, was fatally deficient in its ductility characteristics is set out on page 82 of this article wherein it is stated—

"the strength is accompanied by an almost complete loss of stable elongation."

In contrast with this, the steels of the present invention may advantageously be treated in accordance with this process, i.e. treating by austenitizing at a sufficiently high temperature, then quenching, then tempering at a relatively low temperature, followed by a plastic prestrain beyond the elastic limit of the metal, and a subsequent treatment which is herein referred to as "strain-aging" in which the steel previously subjected to strain as aforesaid, is held at a desired temperature for a period of time such that the desired characteristics are induced and substantially permanently maintained therein.

When a steel alloy composition within the limits set out in the present invention is treated by the process aforesaid, the resulting steel body will have super strength in the direction in which it was prestrained, accompanied by a yield point which is reasonably definite and with a 0.2% offset yield point as hereinafter defined, which is approximately equal to the tensile strength. Such a steel product is per se a part of the present invention. This product, however, cannot be distinguished from other steels which do not share the same physical characteristics or desirable properties by any of the conventional tests. As such, therefore, this product can be defined, as far as can presently be known, only by the process of making it, coupled with the composition thereof, both of which are necessary in order that the product shall have all the novel and desirable physical characteristics of the present product.

The invention will be better appreciated from a detailed description thereof which follows, and wherein reference is made to the accompanying drawings, in which:

FIG. 1 is a comparison of a steel which has been tempered, but not prestrained or strain-aged as against a steel which is treated completely in accordance with the present invention to include elevated temperature strain-aging, the figure being a chart of tensile stress against strain and also showing the 0.02% offset line and the 0.2% offset line;

FIG. 2 is a plot of yield strength (0.2% offset) in units of 1000×p.s.i. against tempering temperature in degrees Fahrenheit;

FIG. 3 is a chart similar to FIG. 2 and for the same set of test samples of tensile strength against tempering temperature; and

FIG. 4 is a view of a test piece as used for determining the fracture toughness of a sample of steel by determining

the strength level at which a crack propagates rapidly in a sharply notched sample.

From a broad point of view, a group of steels within the general limits as set out will accommodate themselves to three different but allied objects and purposes. All these purposes require a super strength steel, i.e. one with a very high as-tempered or as-heat-treated tensile strength. Many high strength steels do not have a true yield point, so that it has become a custom to get a so-called "yield point" by first plotting stress applied to a test piece, for instance, stress in tension, against the actual elongation or strain on this test piece. During the period of application of force (stress) below and up to the elastic limit of the material, such a plot is a relatively straight line inclined upwardly and to the right. Beyond a point substantially corresponding to the elastic limit of the metal the plot curves off to the right with no perceptible sharp break as a curve which is initially substantially tangent to the previous straight line. For this reason it has become common to draw an arbitrary straight line parallel to the straight portion of the stress-strain plot that lies below the elastic limit, and which is offset therefrom by 0.2% in strain or elongation and to take the point of intersection of this arbitrary line with the principal curve itself as the "yield point" or rather as the "0.2% offset yield strength." The present invention, however, provides steels, which have after treatment substantially a true and quite sharp yield point as shown by the comparison of curves A and B on FIG. 1 of the drawings.

The broad composition of steels within the present invention will first be discussed. This composition is as follows:

About 0.35–0.55% carbon
About 3–7% nickel
About 0.2–0.5% chromium
About 0–2% manganese
About 0–2% silicon
About 0–0.5% molybdenum
About 0–0.2% vanadium
About 0–5% cobalt
About 0–1% aluminum

and not over about 0.01% each of sulfur and phosphorus and the remainder being iron with incidental impurities.

Of the ingredients hereinabove listed, the carbon is selected in a so-called medium range of 0.35–0.60%; as steels having a lower carbon content than that in the range selected present no real advantage over the prior art; while steels having a higher carbon content than the range selected are embrittled, so that the desired characteristics of ductility are not present. The preferred range of carbon from the point of view of providing a steel having a maximum ability to produce a desired prestrained and strain-aged body (to give super strength plus ductility) is somewhat narrower in that for this purpose the carbon content is preferably about 0.45–0.50%. It is further noted that this preferred range carbon content is somewhat higher than that of the AISI-Type 4340 steel (which has 0.40% carbon); and yet the resultant steel alloy has a much higher ductility as treated in accordance with this invention.

The next most important alloying ingredient in the steels of the present invention is nickel. From a broad point of view of this nickel content may be from about 3 to about 7%. From a more limited point of view, it is preferred that nickel shall be from about 3½ to about 5%, excellent results having been obtained at both these limits and no reason being known why the range therebetween should not give equally excellent results for many though not all purposes. The lower limit of nickel content, however, is quite critical in that when the amount of nickel present is substantially below 3%, such as in the AISI-type 4340 steel wherein there is a nickel content of only 1.74%, there is very low ductility for the prestrained and strain-aged bodies. The higher limit for nickel is not as critical, but substantially higher ranges of nickel give essentially different type alloys, which do not follow generally the rules nor have the characteristics applicable to the present group of alloy steels. It is also to be remembered that nickel is much more expensive than are some of the other ingredients, particularly the iron ingredient which is of course present to a very major extent and, therefore, as the percentage of nickel is greatly increased, the cost of the final alloy steel is correspondingly increased.

Manganese is similar in some of the characteristics provided thereby to silicon, in that both provide some degree of hardenability for the steel alloy. Generally, a residual of manganese is maintained to combine with sulfur so as to prevent hot workability difficulties. However, with a judicious selection of raw materials, the manganese additions may be reduced or wholly omitted, so that the lower limit may be said to be zero. The maximum of about 2% is chosen, as there is no apparent improvement in the characteristics of the products with greater amounts of manganese. Thus the upper limit is not a critical limit, but is one dictated to the maximum extent at least by economic factors, rather than by factors having to do with the technical properties of the product.

Silicon is generally found in many steels to some extent and has generally the function of retarding the tempering reaction at tempering temperatures of 600 and less. Generally, silicon is added to combine with oxygen in the melt, however, with special melting techniques, the silicon may be wholly omitted, so that the lower limit may be said to be zero. The maximum value of silicon of about 2% is chosen for the reason that as the amount of silicon is increased, the final product tends to become more and more brittle. Values greater than about 2% thus impart undesired brittleness to the product.

Chromium tends to prevent graphitizing during the heat treating or in service of the steel alloy bodies and is preferably present in the amounts from about 0.2–0.5% in accordance with this invention. The preferred concentration range is from about 0.25% to about 0.35%.

It is desired that sulfur and phosphorus be minimized, as it is well known in all ferrous metallurgy that sulfur tends to render the parts made therefrom brittle when hot, while phosphorus make them brittle when cold. The values for sulfur and phosphorus, therefore, are given as maximum tolerable values throughout this specification; as it is understood that the lowest possible values are desirable, but that it is not practically possible to eliminate these elements altogether under commercial operating conditions.

Another element which is optionally usable in the composition is cobalt, the outside limits of such use in accordance with the present invention being about 0–5%. Thus it is specifically contemplated that compositions having no cobalt at all are to be considered as included in the present invention; while compositions over about 5% are to be considered as excluded. The upper limit in this case is not particularly critical. The function of cobalt, at least in the presence of some silicon, is to improve the temper resistance and provide desirable physical characteristics in the material particularly on an as-tempered basis; in other words, without the steps hereinafter discussed of prestraining and strain-aging. On the other hand, as hereinafter set out, cobalt-containing alloys have been proven not only useful and operative, but highly desirable when prestrained and strain aged.

Aluminum is another optionally usable element, and is desirable for use particularly in steels containing cobalt as hereinafter set out. As far as is known this element acts as a deoxidizer for the steel compositions in the relatively small concentrations contemplated in accordance with the present invention, i.e. from 0–1%. In this instance also the "0" is meaningful in that it is specifically contemplated that many steel alloy compositions in accordance with this invention may not contain any aluminum whatsoever.

Molybdenum and vanadium are also optionally present elements in that they may be absent altogether, which is the reason that the lower limit in each instance as to these elements is given as "0." The maximum may be taken as about 0.5% for molybdenum and about 0.2% for vanadium.

It has been found generally that the elements chromium, molybdenum, vanadium, tungsten and columbium may be collectively termed "carbide formers." They, or some of them, are used in practically all high strength steels to some extent. It is noted, however, that substantial quantities of members of this group of metals tend to render the steel hard and strong, but with very little ductility. In general they are used in steels, which are to be tempered at above 600° F., and therefore, above the range where super strength steels usually exhibit their maximum tensile strength. In the case of most of the steels in accordance with the present invention these elements are kept at or near a minimum, consistent with the desire for strength, and in order that the resulting alloy steel shall have substantial ductility.

The steels of the present invention are characterized more predominantly by the presence of the non-carbide formers such as nickel, silicon, cobalt and aluminum. It has been found that manganese has some of the characteristics of the carbide formers and some of the non-carbide formers, so that it cannot be classed exclusively with either group.

Other elements may be present in relatively small or trace amounts, such as calcium, copper, titanium zirconium, columbium, tantalum and boron. However, the amount of any of these metals, if present in the alloy, or the total of all of them, is so small as not substantially to affect the properties of the alloy as a whole. As such, these alloys can all be classed as incidental impurities in the iron if and to the extend that they are present at all. It is not a part of the present invention intentionally to introduce any of these elements as such.

It is contemplated that most if not all metal parts according to the present invention will be heat treated at least at first by a more or less conventional heat treating procedure which will include first an austenitizing step in which the steel alloy is first heated to and held at a temperature preferably in excess of the range of about 1400–1450° F. for a period of time sufficient to bring the metal to a relatively uniform and stable condition at this temperature. The metal is then quenched in oil or in a fused salt bath as hereinafter particularly noted, this quenching being entirely conventional and hence not being described in any greater detail. Thereafter the quenched body is usually tempered by bringing it to and holding it at a selected tempering temperature which is usually in the range of about 350 to about 600, although in some instances tempering temperatures as high as 800° or more may be used. It is found, however, that for maximum strength, the lower tempering temperatures are usually desirable, with a maximum usually of not over 600° F. and with a preferred tempering temperature for many alloys according to the present invention of about 400°, all temperatures here given being in degrees Fahrenheit.

The usual experience with previously known steels, such as the AISI-Type 4150, is that as the tempering temperature is raised, the yield strength and the ultimate tensile strength is progressively reduced. Data as to 4150-type steel is shown by the dotted lines G and G' in FIGS. 2 and 3 respectively. The steels according to the present invention, however, have increasing yield strengths (0.2% offset points) with increasing tempering temperatures from 400 to 600 and even above 600, the yield strengths of preferred compositions are above those for conventional steels such as the 4150 type. This is illustrated best in FIG. 2 of the drawings showing in graphic form results of testing preferred compositions according to the present invention with respect to type 4150 steel.

The test results are represented by lines C and D on FIG. 2 of the drawings, the composition of the material tested to produce line C containing both cobalt and aluminum and the composition of the material tested to produce line D containing cobalt, but no aluminum, both as hereinafter set out in detail. There is also shown on this same figure a line E representing similar data for a composition similar to that used for the test forming the basis for lines C and D, but in this case containing neither cobalt nor aluminum. These are further to be compared with a composition forming the basis for curve F, which not only contained no cobalt nor aluminum, but did contain both molybdenum and vanadium, thus having an excess of carbide formers in accordance with the preferred compositions of this invention. The composition formed on the basis of line F is relatively undesirable from the point of view of its as-tempered characteristics, which are those shown in FIGS. 2 and 3. Thus this composition, while being within the invention in that it is advantageously usable with the specific process of the present invention including prestraining and strain aging, does not have the desirable characteristics necessary for another phase of the invention having to do with a relatively high as-tempered strength and yield strength over a substantial range of tempering temperatures. All these may further be compared with line G on FIG. 2, which is that for a previously known type of steel, namely, No. 4150 steel. For purposes of record, in accordance with a standard reference book on steel, the composition of No. 4150 steel is as follows:

Carbon—0.48–0.53%
Manganese—0.75–1.00%
Sulfur and phosphorus—0.40% maximum each
Silicon—0.20–0.35%
Nickel—
Chromium—0.80–1.10%
Molybdenum—0.15–0.25%

with the remainder being iron with incidental impuities.

In FIG. 3 the lines C', D', E', F' and G' are drawn with data from stress versus strain tests on the same group of steels respectively as the correspondingly numbered lines on FIG. 2 without the prime marks.

From the data illustrated in FIGS. 2 and 3 it is obvious that while the tensile strengths of the several steels tempered at 400° F. is higher than those tempered at higher temperatures, the steels of the present invention exhibit an unexpected improvement or increase in the 0.2% offset yield strength when tempered up to 600 and higher with respect to prior art steels such as that tested to give the dotted lines G and G'.

Continuing now as to the process of the present invention, applicable to a relatively large group of steels all within the present invention, and following the usual tempering step, there are provided prestraining and strain aging steps. These will now be discussed.

When it is known that a particular part as a steel piece is to have to withstand a particular type of externally applied force such as tension, compression, twisting or torque in a right hand direction or in a left hand direction, but not more than one of these four kinds of force (considering right and left hand torque as two kinds), then it is possible by the process of the present invention to attain super strength characteristics in the desired one of these four directions. Inasmuch as tension is usually considered as a prime method of testing and many steel parts must be made to withstand tensile forces as distinguished from either compression, right hand torsion or left hand torsion, then the part in question is prestrained in tension in the same manner that it is desired to be strong in service. Thus a part which is to withstand tension is prestrained in tension, and the part to with-

stand compression is prestrained compression, etc. This is necessary in order to avoid the so-called "Bauschinger effect." A very general summary of this effect is that while a part which is to withstand tension, for example, may be prestrained in tension and then strain aged; if this part is to withstand compression, prestraining and strain aging in tension is not of significant assistance. Similarly, if a part is to withstand right hand torsion, prestraining and strain aging in left hand torsion is of no assistance, and in fact, may even render the part so treated weaker than a wholly untreated part. With this idea in mind, the prestraining and strain aging in accordance with the present invention must be done in the direction in which it is desired that the body shall have super strength.

The term "in the direction" as so used is intended to distinguish not only between compression and tension on the one hand, and also between right hand and left hand torsion, but also between end-wise force (i.e. compression or tension) on the one hand, and torsion in either direction on the other. This term is used in this manner throughout this application and in the appended claims.

The prestraining in accordance with the present invention is further intended to be restricted to a plastic prestrain, i.e. the application of sufficient force so as to effect a strain beyond the elastic limit of the material, so that there will be a permanent deformation in the body due to and following the prestrain step when the applied force is relieved. This permanent deformation should be of the order of magnitude of about 1 to 6% of the original dimension of the part in question in the direction of the strain and as a permanent strain or deformation to this extent. It can be applied by any suitable apparatus having the necessary strength and gripping means to apply the force in question in the desired direction.

If a body be merely prestrained (without strain aging), for example, in tension in accordance with the teachings herein given and be tested immediately thereafter in tension, the new .2% offset yield strength approximates the stress level at which the prestrain was terminated. If, however, a sufficient time is left following the act of prestraining to give what is known as strain aging, then the desirable effects of the prestraining will be present. This time and the temperatures at which the strain aging is accomplished again are not exactly definite. The strain aging apparently takes place much more rapidly as the temperature is raised and hence is preferably done at an elevated temperature, even though it is theoretically possible to effect stain aging at room temperature if sufficient time is provided. However, due to the desire to secure the results to be attained in a reasonable and limited amount of time, it is ordinarily preferred to use an elevated temperature for strain aging plus a sufficient time. This elevated temperature, however, should not exceed the prior tempering temperature used on the same body without undesired results, in effect eliminating all the desirable results which are sought incident to prestraining and strain aging combined. The elevated temperature for strain aging is preferably about 50° F. less than the tempering temperature. This 50°, however, is not narrowly critical, it being important merely that the strain aging temperature be somewhat and preferably substantially less than the tempering temperature. It has been found that a 50° difference is a preferred differential in this respect. At temperatures of about 50° less than the tempering temperature, strain aging can occur to a satisfactory degree in about two hours, so that increased time beyond two hours does not result in any substantial improvement in the results attained. Again, the two hour period is not narrowly critical, as a greater period may be used with impunity, while somewhat lesser periods of time often attain a large amount of desirable results sought.

The present invention does not rely upon any particular theory as to what takes place during strain aging. It is believed, however, that strain aging is really a diffusion-controlled process in which certain solute atoms such as carbon and nitrogen migrate toward high stress regions created during and as a result of prestrain. It is further believed that the rate of such migration or diffusion is approximately doubled for every 10° C. increase in temperature at which the strain aging is conducted.

The foregoing theory, which is believed to be correct but is not specifically relied upon, tends to explain why strain aging operates better at higher temperature values up to about 50° F. below the tempering temperature and also why it can operate even at room temperature, which, for the purpose of the present invention, may be assumed to be 70° F. The preferred range for the strain aging temperature is about 50° F. to about 100° F. below the tempering temperature and most preferably about 50° F. below the tempering temperature.

One result of the prestraining aging as aforesaid is that steel samples acquire a definite and relatively high yield point as is evidenced from a comparison of curves A and B of FIG. 1 of the drawings wherein the sample tested to give curve A had been tempered in a conventional manner, but had not been prestrained and strained aged; while that to give curve B had been tempered, then prestrained and strain aged.

It is recognized that prestraining and strain aging has been described to some extent in the article hereinabove referred to by Stevenson et al. in the "Transactions of the American Society of Metals." In the tests set out in this article, not only was the type of steel inappropriate for maximum desirable results in accordance with the present invention in that it had a too low nickel content and a too high chromium content; but also the authors of this article did not know or in any case had not investigated and told of the necessity that the strain aging temperature of "retempering," as it was called in that article, should be substantially below the temperature at which the body was first tempered. For this reason, therefore, they did not succeed in obtaining a strengthening of the steel to the ranges which they desired, i.e. over about 300,000 p.s.i., accompanied by reasonable ductility. As aforesaid, they reported a "most complete loss of elongation" which is of course a measure of ductility. As compared with this, by the selection of a proper composition, even in the relatively broad range herein set out as appropriate to the present invention, plus the conduct of the straining and strain aging step at a temperature below that of the original tempering step, the desired characteristics of ductility are retained for the most part, while tremendously improving the strengths of the bodies being produced. While most of the tests hereinafter reported are in tension, and indicate increased strength in tension due to prestraining in tension and subsequent strain aging, similar results will be obtained in compression or in right-hand torque or in left-hand torque. Thus, for example, if a body or article is to be used to withstand left-hand torque, it is prestrained in left-hand torque, then strain aged so as to give a final product which is improved as to its ability to stand left-hand twisting or torque. This body will not, however, be significantly improved in its resistance to right-hand torque. If, on the other hand, a body is to be subjected to right-hand torque during its normal use, then it is prestrained in right-hand torque and strain aged, whereupon its strength to resist or withstand right-hand torque or twisting is greatly improved, but its ability to withstand left-hand torque or twisting is not significantly improved by the prestraining and strain aging.

In a similar way, prestraining and strain aging in tension enhances the strength of a body to withstand tensile forces; whereas prestraining and strain aging in compression makes a body stronger to withstand compressive forces; but prestraining in one direction does not significantly assist in strengthening the body against forces in the or any other direction. This phenomenon is known as the "Bauschinger effect" as aforesaid.

The present invention will be better understood by a consideration of a number of actual examples wherein the various embodiments of the invention will be brought out in detail.

EXAMPLE I

This example is given to illustrate compositions in accordance with the present invention as compared with one prior art composition and to show the results of prestrain and strain aging.

In Table 1 which follows there is given data as to the composition of each of five steels which are in accordance with the present invention and a single steel numbered "PA–1," which is an example of the prior art steel of the Ladish-type, given for purposes of comparison.

Table 1

COMPOSITIONS OF THE NICKEL STEELS, WEIGHT PERCENT

Sample No.	C	Mn	Si	Ni	Cr	Co	Al	Mo	V	P	S
1	0.50	0.19	1.87	3.75	0.26	------	------	------	------	0.003	0.008
2	0.53	0.91	1.71	3.77	0.27	------	------	------	------	0.003	0.008
3	0.47	0.95	1.77	3.73	0.26	------	0.83	------	------	0.004	0.009
4	0.48	1.05	1.92	4.07	0.28	2.29	0.78	------	------	0.004	0.008
5	0.49	0.96	1.71	3.68	0.27	2.11	------	------	------	0.003	0.008
PA–1	0.47	0.78	0.27	0.69	1.13	------	------	1.10	0.08	0.005	0.008

Table 2

EFFECT OF PRESTRAINING AND STRAIN AGING ON THE TENSILE PROPERTIES OF STEELS

[0.357-inch Diameter Tensile Test Specimens]

Sample No.	Austenitizing Temp. (° F.)	Tempering Temp. (° F.)	Plastic Prestrain (Percent)	Strain Aging Temp. (° F.)	Yield Strength (1,000 p.s.i.)		Tensile Strength (1,000 p.s.i.)	Elongation (Percent in 2 in.)	Reduction in area (Percent)
					0.02% offset	0.2% offset			
1	1,500	400	----------	----------	195	250	304	[1] 2	10
2	1,500	400	----------	----------	160	227	319	9	26
3	1,500	400	----------	----------	208	256	294	10	37
4	1,500	400	----------	----------	170	240	313	9	36
5	1,500	400	----------	----------	155	242	311	6	18
1	1,450	400	2	350	343	352	352	6	33
2	1,450	400	3	300	344	359	359	4	23
3	1,450	400	2	350	----------	339	339	5	46
4	1,450	400	2	350	339	353	353	6	39
5	1,450	400	2	300	298	350	350	[1] 3	29
PA–1	1,525	400	3	350	----------	335	335	[1] <1	21
1	1,550	400	2.5	350	338	354	354	3	11
2	1,550	400	3	350	325	369	369	1	5
3	1,550	400	2	350	324	336	336	3	40
4	1,550	400	2	350	320	347	347	5	31
5	1,550	400	3	350	304	352	352	3	16
PA–1	1,700	400	3.5	350	----------	344	344	[1] <1	6
PA–1	1,700	400	5	350	----------	347	351	2	12

[1] Broke at or outside gage marks.

These steels were formulated and test pieces made therefrom. Each test piece of steel of samples 1–5 were tempered at 400° F. as set out hereinafter and further subjected to prestraining in tension and strain aging. The results of these tests are set out in Table II above.

The steels reported on in Table 2 above were, after austenitizing, quenched in molten salt at 500° F. (with the exception of the sample No. PA–1, which was quenched in oil in the conventional manner), held one minute, and then air cooled. The strain aging when carried on was for two hours at the temperature indicated.

The foregoing data is given in full and as the data was taken. It is believed, however, that the data as to percent elongation and reduction in area for the sample 2 as austenitized at 1550 is erroneous, as this data is substantially out of line with the remainder of the samples. Another possible explanation as to this is that this sample had a relatively high carbon content and as such is not within the most preferred range of carbon for this general purpose. This may account in part for the relatively low ductility as contrasted with the other samples tested. It is noted, however, that this sample 2 has the greatest tensile strength and the highest 0.2% offset yield point. These values, however, were attained in this particular sample at a substantial cost in their ductility. For these reasons, therefore, the carbon composition is preferably within the somewhat narrower limits of about 0.45–0.50%. There is attained, however, even in sample 2 when tempered using a relatively high austenitizing temperature, strength characteristics which are beyond those of any other sample tested, so that this sample is considered broadly to be within the scope of the present invention.

EXAMPLE II

This example is given to illustrate the present invention applied to steels having about 5% nickel. Two actual steels were made up and tested in accordance with this example, which are herein numbered as samples 6 and 7. The composition of these steels is set out herein below in Table 3.

Table 3

Sample No.	Percent						
	C	Mn	Si	Ni	Cr	V	Mo
6	0.50	0.41	0.17	4.92	0.38	0.14	0.30
7	0.42	0.87	1.07	5.10	0.39	0.14	0.32

The balance of these compositions consisted essentially of iron with incidental impurities, including a small amount, less than 0.01% each, of sulfur and phosphorus. These samples were tested as before with the results set out in Table 4, which follows.

Table 4

RESPONSE OF 5.0% NICKEL STEELS AND THE EFFECT OF STRAIN AGING TEMPERATURE

Sample No.	Tempering Temp. (° F.)	Plastic Prestrain (Percent)	Strain Aging Temp. (° F.)	Yield Strength (1,000 p.s.i.)			Tensile Strength (1,000 p.s.i.)	Elongation (Percent in 2 in.)	Reduction in area (percent)
				0.02% offset	Yield point	0.2% offset			
6	400			183		238	288	7	37
6	400			163		238	288	8	40
6	400	3	350	325	338	338	338	5	33
6	400	3	350	305	342	342	342	6	33
6	400	2.5	400	245		318	321	4	31
6	400	3	400	265		318	318	5	29
7	400			165		216	285	9	28
7	400			160		210	288	9	28
7	400	3	350	315	332	332	332	6	31
7	400	3	350	317	328	328	328	5	31
7	400	3	400	293	325	325	326	6	29
7	400	2.5	400	268	325	325	325	4	23

The table above shows the strength of the as-tempered bodies made of the respective compositions and also gives the value for a yield point as such, where the yield point is definite and ascertainable. In several instances duplicate samples were run under the same conditions which gives a good approximation of the extent to which the values may be duplicated in repeated tests. It will be noted that in each instance there is a substantial retention of ductility as measured by the elongation and reduction in area, even though the very substantial strengthening incident to the practice of the process of the present invention resulted in some loss of ductility. The remaining ductility is adequate for many and most purposes for which this steel is desired and is much greater than was present in prior art steels treated in a somewhat similar manner. As to each sample, a set of data is also given where the strain aging temperature is substantially equal to the tempering temperature. As will be noted, both the elongation and the reduction in area and hence the ductility under these conditions are less than in the instances where the strain aging temperature is 50° F. less than the former tempering temperature, but such ductility still is substantial and is adequate for most purposes. On the other hand, it is considered undesirable to use a strain aging temperature or retempering temperature higher than the original tempering temperature as this has resulted in substantial decrease in several of the strength characteristics of the product as shown by work reported in the prior art publications of Stephenson et al. above referred to.

EXAMPLE III

This example demonstrates principally the desirable characteristics of a cobalt and/or aluminum containing steel and preferably of a steel alloy composition in which both cobalt and aluminum are present. The steels of this character have the characteristic of having a relatively high 0.2% offset yield strength (230,000 p.s.i. or more) on an as-tempered basis. In the tests carried out in determining the characteristics of a number of steels when tempered at various temperatures in the range of about 400–800° F., a group of steels illustrative of different type compositions were used. The compositions of the several samples tested are given in Table 5, which follows.

Table 5

Sample No.	C	Mn	Si	Ni	Cr	Mo	V	Co	Al
8	0.50	0.41	0.17	4.92	0.38	0.30	0.14		
9	0.51	0.90	1.33	3.59	0.25			1.93	
10	0.53	0.94	1.59	3.65	0.29			2.10	0.78
11	0.50	0.95	1.49	3.69	0.27				

Of the foregoing, sample 8 is not the preferred form in accordance with this phase of the invention as it contains substantial amounts of molybdenum and vanadium; even though this steel is broadly within the general scope of this invention in that it is quite similar to samples 5 and 6 which have quite desirable properties as to products which have been strained and strain aged. Of the four samples hereinabove given, that designated "10" is approximately the preferred composition. The foregoing samples were tested on an as-tempered basis with the results set out in Table 6, which follows.

Table 6

ROOM TEMPERATURE TENSILE PROPERTIES OF EXPERIMENTAL STEELS [1]

[0.357-inch round specimens]

Sample No.	Tempering Temp. (° F.)	Yield Strength (1,000 p.s.i.)		Tensile Strength (1,000 p.s.i.)	Elongation		Reduction in area (percent)
		0.02% offset	0.2% offset		(Percent in 2 in.)	(Percent in 1 in.)	
8	400	183	225	291	8	16	41
8	400	163	225	297	9	15	34
9	400	192	232	321	9	12	21
10	400	170	237	324	9	15	23
11	400	147	203	319	9	15	27
10	500	205	257	316	8	14	38
8	600	210	220	240	9	16	47
8	600	205	217	237	9	16	46
9	600	210	257	283	7	12	32
9	600	190	255	293	8	14	38
10	600	215	265	301	8	16	43
10	600	210	265	304	7	10	29
11	600	198	242	277	9	16	41
11	600	195	243	277	9	15	39
9	700	206	225	244	8	14	36
9	700	208	226	245	7	13	33
10	700	180	236	274	7	12	25
10	700	193	237	276	9	16	38
9	800	205	218	225	9	14	39
8	800	205	218	225	9	15	39
10	800	215	240	260	8	13	28
10	800	218	240	260	7	12	27
10	900	212	223	233	7	10	17
10	900	215	224	233	8	13	30

[1] Austenitized at 1,500° F., quenched into molten salt at 500° F., held 1 minute and air cooled.

From the foregoing it will be seen that the compositions of the present invention provide quite high 0.2% offset yield points which are, however, substantially below the ultimate tensile strengths in most instances, but tend to approach them, particularly as to the composition of sample 10. It is also clear that increases in tempering temperature generally result in a reduction in tensile strength. However, when these data are plotted, it will be noted that the unusual results shown in FIGS. 2 and 3 of the drawings ensue, particularly in that, referring to FIG. 2, there is a peak at 600° F. for 0.2% offset yield point, which is higher than the corresponding yield point for the same samples tempered at 400° F. In the drawings, curves C and C' of FIGS. 2 and 3 are plots of the data given as to sample **10**; curve D of FIG. 2 and D' of FIG. 3 is a plot of the data given for sample **9**, which is next preferred; curve E. of FIG. 2 and E' of FIG. 3 is a plot of the data given for sample **11**; while curve F of FIG. 2 and F' of FIG. 3 is a plot of the data for sample 8. The dotted line curves of FIGS. 2 and 3 are plots of the corresponding data for a prior art composition which is generally known as "Type 4150 steel," the composition of which is given hereinabove.

It will also be noted from Table 6 that as to the preferred compoistion, i.e. that of sample **10**, there is a second relatively low peak (at 240,000 p.s.i.) for the 0.2 offset yield strength versus tempering temperature at a tempering temperature of 800° F. in addition to the peak at 600° F. shown on FIG. 2.

Another and very impressive demonstration of the particular desirability of the composition of sample **10** is the the fact that a 12 inch long button-headed tensile bar of this composition having a diameter of 0.419 inch (less than 7/16 inch) when used in a test demonstration supported the weight of a 45,000 lbs. freight car while serving as its sole support by being interposed between a rigging secured to the car on the one hand and a hook from a heavy crane above the car on the other hand. In this demonstration, the test piece was subjected to a stress of 325,000 p.s.i. In an earlier test on this same piece, to ascertain whether it could stand not only the weight of the car, but also the extra stress incident to lifting the freight car several feet off the rails, it was subjected to a tensile test wherein it sustained a total pull of 50,000 lbs., which was equivalent to 360,000 p.s.i. This test piece had been prepared from the composition of sample **10** by the process herein taught including tempering at 400° F., plastically prestraining the bar about 2% of its original length and strain aging at 350° F. It has been found that the 0.2% offset yield strength of compositions of this character when treated in this way is between about 350,000 and 360,000 p.s.i.

Another steel which is similar in its composition to sample **10** hereinabove referred to and which is in the class of 3.5% nickel steels, is given as another preferred composition of this type steel. This composition is:

Carbon—0.53%
Manganese—0.90%
Silicon—1.60%
Nickel—3.60%
Chromium—0.30%
Cobalt—2.00%
Aluminum—0.75%
Phosphorus—0.006%
Sulfur—0.006%

and the balance iron with incidental impurities.

EXAMPLE IV

This example is to illustrate the fracture toughness characteristics of steels having compositions generally in accordance with the present invention and further to set out preferred compositions within the general scope of the compositions of the present invention which have the best characteristics of fracture toughness.

In order to determine fracture toughness, it is usual at this time to determine the tensile properties of sharply notched steel sheet. In some instances the steel sheets used in tests of this kind are notched on the sides, with the notches facing one another and are imperforate in the center portion to be tested, i.e. between the ends that are gripped in the testing apparatus. It is quite common, however, that the end portions of a sample to be gripped in the testing apparatus may be perforated for a large bar so as to facilitate the gripping of such end portions. A discussion of the fracture testing of notched specimens is contained in the ASTM Bulletins for January and February 1960, pages 29 et seq. in the January issue and page 18 et seq. in the February issue.

In the test work done in connection with the present invention, however, a sample as particularly shown in FIG. 4 of the drawings was used, this figure giving exact dimensions of the test piece which included end portions having perforations of 0.500 inch for gripping purposes and a central hole with the notches disposed on diametrically opposite portions thereof and extending outwardly from the center hole. The central hole and the notches therein were formed by an electrojet apparatus, i.e. a tool employing an electric arc and operating under an oil coolant. The specimens were then fatigue-cracked at the notches prior to testing, the notches serving as fatigue crack starters. This was done by repeatedly alternately applying a high tensile stress equal to about one fifth of the yield strength of the material and a low tensile stress equal to about one-half the high tensile stress for starting and propagating fatigue cracks. These stresses were applied intermittently for about 30,000 cycles. This was carried on until the fatigue cracks as shown at **11** and **12**, FIG. 4, had a total overall length as shown by the dimensioins on that figure of about 0.750 inch from end to end. In order thereafter to determine the length attained by the slowly propagating crack at the inception of crack instability, India ink was applied to the notches near the apex of each notch so as to follow up the crack. This length may in turn be used to compute a value known in the art as "fracture toughness". The ink, while wet and flowable, followed each crack as it was gradually extended up to the point where crack instability set in. This technique was used for estimating the length of the slowly propagating crack and from this, fracture toughness values can be computed. This type of testing is that favored by the ASTM Committee on the Fracturing of High Strength Steel Sheet at its Meeting of November 17, 1960.

The composition (in percent by weight) of various steels which have been tested or known data as to which is herein given for comparison purposes are as follows: (the PA steels being compositions known in the prior art)

Table 7

Sample No.	C	Mn	Si	Ni	Cr	Mo	V	Co
12	0.47	0.95	1.77	3.73	0.26			
13	0.45	0.49	0.18	5.08	0.37	0.30	0.13	
14	0.45	0.48	0.14	5.07	0.31		0.14	2.04
15	0.50	0.51	0.16	5.07	0.30		0.15	4.00
PA-2	0.49	0.74	0.19	0.57	1.07	0.97	0.08	
PA-3	0.49	0.84	0.20	0.57	1.06	1.02	0.09	
PA-4	0.43	0.89	1.78	1.92	0.90	0.43	0.13	
PA-5	0.40	0.75	0.25	1.80	0.80	0.25		
PA-6	0.40	0.75	1.60	1.85	0.85	0.40	0.08	

Samples were prepared from the foregoing materials set out in Table 7, the samples being cut in each instance first with the long axes of the samples, i.e. the vertical direction of the drawing as seen in FIG. 4, extending longitudinally of the roll sheet (i.e. longitudinally of the direction of rolling). In each instance the sample was cut out from the sheet and the holes and notches formed therein prior to conventional austenitizing, quenching and tempering, all of which was then carried on with the formed and punched sheet and then the fatigue cracks formed therein by the alternate application of high and low loads in a longitudinal direction of the sample, the

high load being about twice the low load. It is found that the amount of loading and the number of cycles required to form such fatigue cracks are a function of the frequency of the alternation between high and low loading and hence the data as to the manner as to which these test pieces was fatigue-cracked is not per se characteristic. Suffice it to say that the fatigue cracks plus the center hole diameter and the depths of the notches extended initially a distance of about 0.75 inch as set out in the drawing, FIG. 4. The samples were then tested at room temperature in tension and the maximum tension determined up to the point of crack instability. The length of the crack at the point of crack instability was also determined by the ink technique as above described, as the ink follows the crack propagation up to the point of crack instability. The net notch strength for each sample determined from the above test was calculated as the maximum applied load divided by the uncracked cross section area of the sample at the onset of unstable crack propagation.

As so tested, sample **13**, which is substantially the preferred composition in accordance with the present invention for resistance against crack propagation or fracture toughness as it is sometimes called, showed a net notch strength averaging 246,000 p.s.i. for an average of three specimens tested, the actual values for these specimens being 230,000, 248,000 and 261,000 and a fourth specimen also tested giving the value of 258,000. These specimens were tempered at 400° and had varying thickness between 0.096 inch for the first three in question and 0.116 inch for the sample having the 258,000 p.s.i. test results. In each instance the initial crack length was between 0.750 inch and 0.780 inch and the final crack length also varied from 1.17 inch to 1.27 inch. These figures for notched tensile strength may be compared with the 0.2 offset yield strength of smooth bar test pieces of the same composition of about 210,000–220,000 p.s.i. and a total tensile strength of about 260,000 p.s.i., also for smooth round bar test piece.

These results may further be compared with that of the prior art composition hereinabove given as sample **PA–3** wherein a similar test piece showed a notched tensile strength of only 189,000 p.s.i. Again, sample **PA–5**, which is a steel of the 4340 type, averaged about 206,000 p.s.i. for net notch strength (calculated as aforesaid), with different samples tested giving various values between 202,000 and 209,000 p.s.i. The net notch strength values of the steel according to the present invention may also be compared with the figures ascertainable from prior art literature as to the composition given as **PA–6**, which was the best of the prior art samples and had a net notch tensile strength of 218,000 p.s.i.

The characteristics of the failure of the several samples is also of great interest. Those having compositions according to the present invention, when cut as longitudinally extending samples, were 100% shear failures, which is the desired condition. As contrasted with this, the samples of the several prior art compositions were, in practically all instances, substantially less than 100% in shear, values of 73%, 80% and 95% being common for different samples.

Some samples of the preferred composition were also treated in the following manner, sometimes known as "hot-cold working." This involved austenitizing the material in sheet form at 1450° F., then quenching in molten salt at 600° F. and immediately thereafter and prior to the cooling of the sample, rolling in a single roll pass so as to reduce the thickness of the sheet by 35%, followed by quenching in oil at about 120° F. The samples as thus prepared were then tested to give relatively low net notch tensile strength results of about 87,000 p.s.i. When, however, samples prepared in this way up to the point of testing were further tempered at 400° F., (all these samples in accordance with the present invention having a thickness between 0.054 and 0.060 inch) the test results showed net notch tensile strength values of 239,000 and 262,000 p.s.i. with final cracked lengths of 1.30 inch and 1.32 inch respectively as against final crack lengths for the first group of samples having net notch strengths of 87,000 p.s.i. of only 0.82 and 0.87 inch respectively.

The importance of steel capable of withstanding notch-type tests as aforesaid is believed to be very great in that such cracks have a very small root radius, of 0.001 inch or less. Cracks of this nature could be formed in the manufacture of many articles from sheet steel where it is desired that the article shall have great strength and shall not fail under extreme operating conditions. Such cracks could be generated during heat treating or welding or other fabrication operations in the course of manufacturing articles from sheet steel. Cranks of this kind may not be readily detectable, even if detectable at all, during normal inspection. For these reasons, therefore, resistance to crack propagation is often considered of greater importance than mere tensile strength values of the steel.

Samples **14** and **15** above referred to were tested at both 400° and 600° F. tempering temperatures. The tests made at 400° showed that a sample of the type shown in FIG. 4 and 0.087 inch in thickness had a 100% shear type failure and had a net notch strength as hereinabove defined of 250,000 p.s.i. Sample **15**, also tempered at 400° F., was similarly tested with the sample formed as shown in FIG. 4 and with a thickness of 0.086 inch. The failure in this instance was only 58% of a shear type and the net notch strength only 155,000 p.s.i. This indicates that cobalt at the 2% level is completely tolerable. However, the carbon-cobalt balance in sample No. **15** is somewhat high and consequently samples of this thickness range did not break with full shear failures and high net notch strengths.

When samples **14** and **15**, each of a size and configuration as shown in FIG. 4 and 0.086–0.087 inch in thickness, were tempered at 600° F., the samples broke with full shear failures and high net notch strengths (about 250,000–256,000 p.s.i.). This indicates that at the higher cobalt-carbon balance and at the higher tempering temperatures, full shear failures and high net notch strengths may be achieved, even at the aforementioned thickness. The compositions of samples **14** and **15**, when formed into smooth, round tensile bars 0.357 inch in diameter and tempered at 600° F., when tested at room temperature, had 0.2 offset yield strengths of about 235,000–237,000 p.s.i. and tensile strengths of about 255,000–257,000 p.s.i. It is thus shown that all the net notch strengths for the 600° F. tempered samples Nos. **14** and **15** were substantially in excess of the smooth bar yield strengths.

Another steel alloy composition which has been shown to have good notch properties, in that it is highly resistant to crack propagation of a notched sample as aforesaid, is one having 0.48% carbon, 0.45% manganese, 0.25% silicon, 5.10% nickel, 0.30% chromium, 0.30% molybdenum, 0.12% vanadium and 0.006% (max.) each of sulfur and phosphorus with the balance iron with incidental impurities.

While there has been disclosed herein a relatively broad range of medium carbon alloy steels in accordance with the present invention, certain more limited ranges for particular purposes, and specific compositions also for particular purposes, all as set out herein, and while a process of treating steel to attain super strengths in a particular direction has also been disclosed, other variations and equivalents of the foregoing will become evident to those skilled in the art based upon this disclosure and the appended claims. I do not intend to be limited, therefore, except by the scope of the appended claims, which are to be construed validly as broadly as the state of the art permits.

What is claimed is:

1. A super strength steel alloy composition, consisting essentially of the following:

About 0.35–0.60% carbon
About 3–7% nickel
About 0.2–0.5% chromium
About 0–2% manganese
About 0–2% silicon
About 0–0.5% molybdenum
About 0–0.2% vanadium
About 0–5% cobalt

not over about 0.01% each of sulfur and phosphorus, and the remainder being iron with incidental impurities; the metal being characterized by high as-tempered strength accompanied by good ductility.

2. A super strength steel alloy composition in accordance with claim 1, in which the composition contains about 3½% to 5% nickel.

3. A super strength steel alloy composition in accordance with claim 1, in which the composition contains about 0.45 to 0.50% carbon.

4. A super strength steel alloy composition in accordance with claim 1, in which the composition contains about 3½ to 5% nickel and about 0.45 to 0.50% carbon.

5. A super strength steel alloy composition in accordance with claim 1, in which the composition contains

About 0.45–0.55% carbon,
About 3½–4% nickel,
About 0.2–1.2% manganese,
About 0.2–0.4% chromium,
About 1.7–2% silicon, and

substantially no molybdenum or vanadium, the composition being further characterized by very high strength when austenitized, quenched, tempered, prestrained beyond its elastic limit in the direction in which its super strength is to be utilized and to a permanent strain of about 1 to about 5% beyond its original dimensions and thereafter strain aged at a predetermined elevated temperature which is not substantially more than the tempering temperature thereof.

6. A super strength steel alloy composition in accordance with claim 1, in which the composition contains

About 0.4–0.5% carbon,
About 5% nickel,
About 0.4–0.9– manganese,
About 0.15–1.1% silicon,
About 0.35–0.4% chromium,
About 0.14% vanadium, and
About 0.3–0.35% molybdenum;

the composition being further characterized by very high strength when austenitized, quenched, tempered, prestrained beyond its elastic limit in the direction in which its super strength is to be utilized and to a permanent strain of about 1 to about 5% beyond its original dimensions, and thereafter strain aged at a predetermined temperature which is not substantially above the tempering temperature thereof.

7. A super strength steel alloy composition in accordance with claim 1, in which the composition contains

About 0.35–0.5% carbon,
About 4–6% nickel,
About 0.2–0.5% chromium,
About 0.3–1% manganese,
About 0.2–1% silicon,
About 0.15–0.5% molybdenum, and
About 0.5–0.3% vanadium;

the composition being further characterized by high fracture toughness as measured by the resistance of a notched sample of sheet material to crack propagation under tensile stress.

8. A super strength steel alloy composition in accordance with claim 1, in which the composition contains

Abuot 0.45–0.5% carbon,
About 4.8–5.5% nickel,
About 0.2–0.4% chromium,
About 0.45–0.9% manganese,
About 0.2–0.9% silicon,
About 0.2–0.4% molybdenum, and
About 0.1–0.15% vanadium;

the composition being further characterized by high fracture toughness as measured by the resistance of a notched sample of sheet material to crack propagation under tensile stress.

9. A super strength steel alloy composition characterized by high fracture toughness as measured by the resistance of a notched sample of sheet to crack propagation under tensile stress, said composition consisting essentially of the following:

About 0.48% carbon,
About 5.10% nickel,
About 0.30% chromium,
About 0.45% manganese,
About 0.25% silicon,
About 0.30% molybdenum,
About 0.12% vanadium,

and not over about 0.006% each of sulfur and phosphorus.

10. A super strength steel alloy composition in accordance with clam 1, in which the composition contains:

About 0.4–0.6% carbon
About 3–6% nickel
About 0.2–0.4% chromium
About 0.3–1.3% manganese
About 1–1.8% silicon
About 1–3½ cobalt, and
About 0.2–1% aluminum;

said composition being characterized in that as-tempered articles made thereform have high 0.2 offset yield strengths and high tensile strengths, with excellent ductility over a broad range of tempering temperatures from about 400° F. to about 900° F.

11. A super strength steel alloy composition in accordance with claim 1, in which the composition contains:

About 0.45–0.55% carbon
About 3.5–4.0% nickel
About 0.2–0.4% chromium
About 0.6–1.0% manganese
About 1.4–1.8% silicon
About 1.5–2.5% cobalt, and
About 0.6–0.8% aluminum;

said composition being characterized in that as-tempered articles made therefrom have high 0.2 offset yield strengths and high tensile strengths, with excellent ductility over a range of tempering temperatures from about 400° F.–900° F.

12. A super strength steel alloy composition of the 3½% nickel class having, on an as-tempered basis, a high 0.2 offset yield strength and high tensile strength with excellent ductility over a broad range of tempered temperatures including particularly the range of 400°–900° F., said composition consisting essentially of

About 0.53% carbon
About 3.60% nickel
About 0.30% chromium
About 0.90% manganese
About 1.60% silicon
About 2.00% cobalt
About 0.75% aluminum

and not over about 0.006% each of sulfur and phosphorus and the remainder being iron with incidental impurities.

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