United States Patent [19]

Hultgren et al.

[11] **3,723,194**

[45] **Mar. 27, 1973**

[54] **METHOD OF PROVIDING SUPERPLASTIC STEEL AND OF PRODUCING ARTICLES BY DEFORMATION THEREOF**

[75] Inventors: **Frank A. Hultgren,** Burton; **Richard A. Kot,** Parma, both of Ohio

[73] Assignee: **Republic Steel Corporation,** Cleveland, Ohio

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Related U.S. Application Data

[63] Continuation-in-part of Ser. No. 98,674, Dec. 16, 1970, abandoned.

[52] **U.S. Cl.** **148/12**

[51] **Int. Cl.** **C21d 7/14**

[58] **Field of Search** 148/12, 12.3, 12.4, 37

[56] **References Cited**

UNITED STATES PATENTS

3,574,002 4/1971 Hayden, Jr. et al. 148/37

Primary Examiner—W. W. Stallard

Attorney—Robert P. Wright

[57] **ABSTRACT**

Procedure for converting steel, notably ordinary and alloy grades of low carbon, ferritic character, to a superplastic state, e.g. affording very high ductility, and for deforming such superplastic steel in a desired manner, embraces: rapidly heating a body of steel to a temperature, advantageously in the alpha-plus-gamma phase field, where the steel is then found, over a brief interval, to experience a transitional state of severe microstructural instability and to be characterized by superplasticity; and applying stress to the body in such interval to effect the desired deformation. Enhanced effectiveness of the process can be attained in suitable situations where the steel is extensively cold worked, e.g. by reduction upwards of 70 percent, without annealing, prior to the heating stage, the occurrence or progress of recrystallization at the elevated temperature being then found to coact in the transitional condition of microstructure that affords superplasticity. Large and complex deformations can be achieved by the process, at conveniently high strain rates.

18 Claims, 15 Drawing Figures

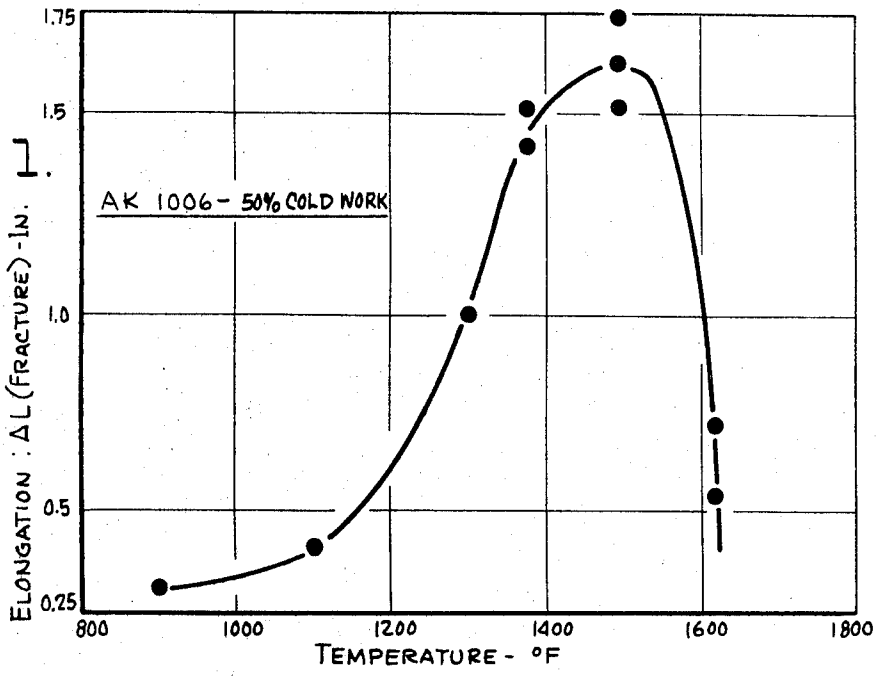

1.75
1.5
1.0
0.5
0.25
AK 1006 - 50% COLD WORK
ELONGATION : ΔL (FRACTURE) - IN.
800
1000
1200
1400
1600
1800
TEMPERATURE - °F

Fig. 1.

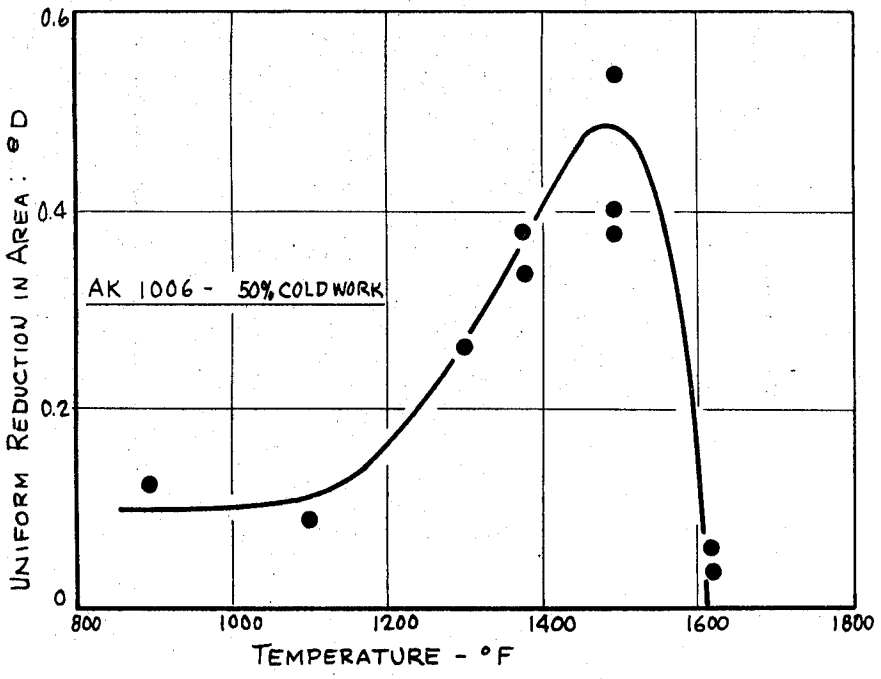
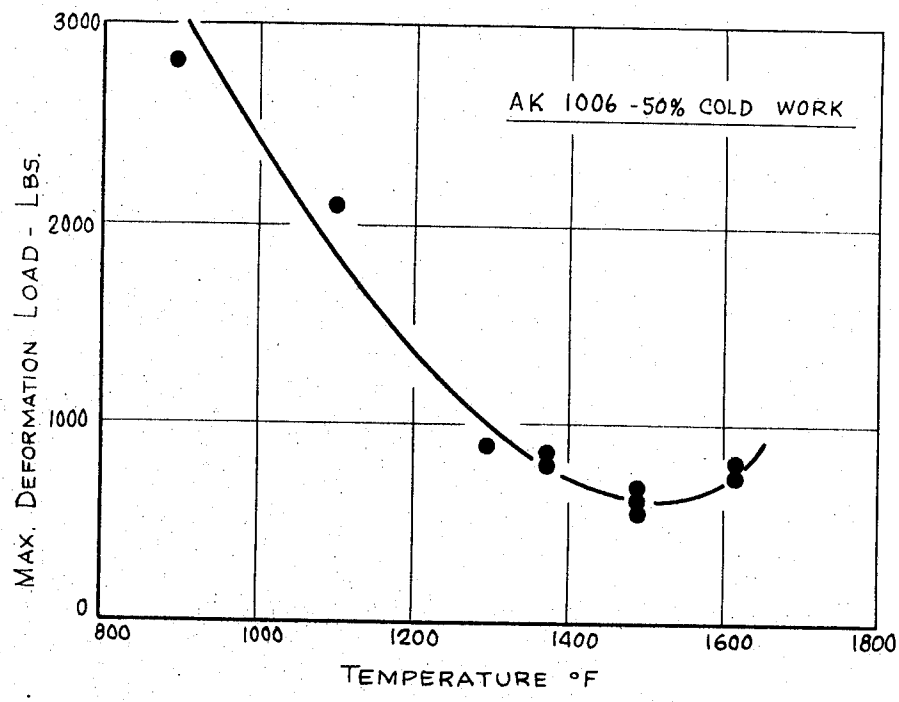

0.6
0.4
0.2
0
AK 1006 - 50% COLD WORK
UNIFORM REDUCTION IN AREA : ϵD
800
1000
1200
1400
1600
1800
TEMPERATURE - °F

Fig. 2.

Fig. 3.


3000
2000
1000
0
MAX. DEFORMATION LOAD - LBS.
AK 1006 -50% COLD WORK
800
1000
1200
1400
1600
1800
TEMPERATURE °F

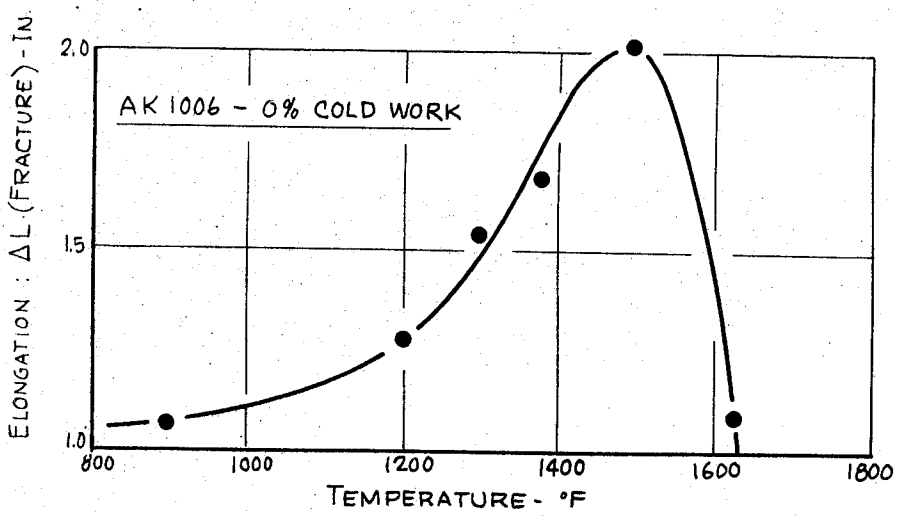

2.0
1.5
1.0
ELONGATION : ΔL (FRACTURE) - IN.
AK 1006 - 0% COLD WORK
800
1000
1200
1400
1600
1800
TEMPERATURE - °F

Fig. 4.

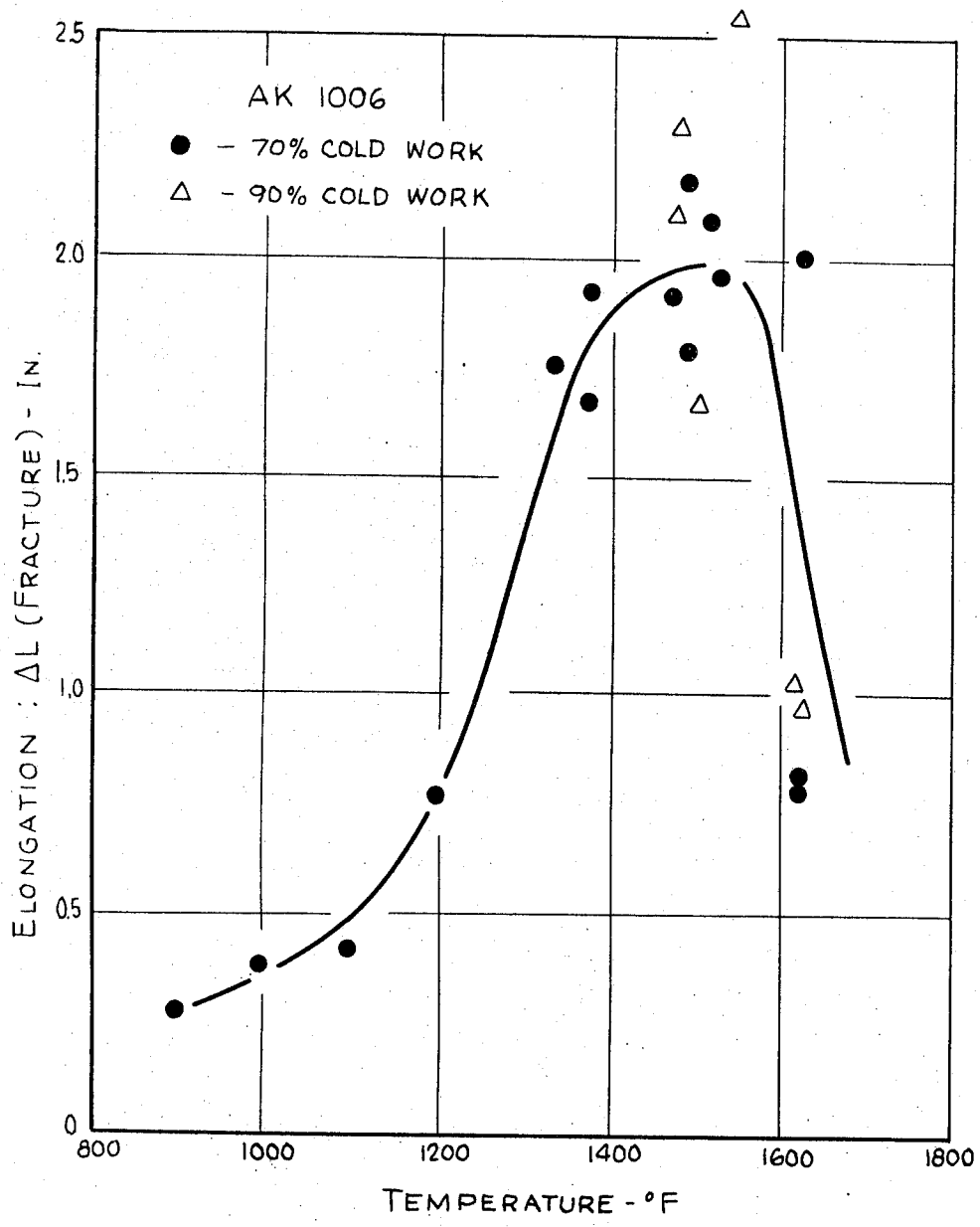

AK 1006
● - 70% COLD WORK
△ - 90% COLD WORK
ELONGATION : ΔL (FRACTURE) - IN.
2.5
2.0
1.5
1.0
0.5
0
800
1000
1200
1400
1600
1800
TEMPERATURE - °F

Fig. 5.

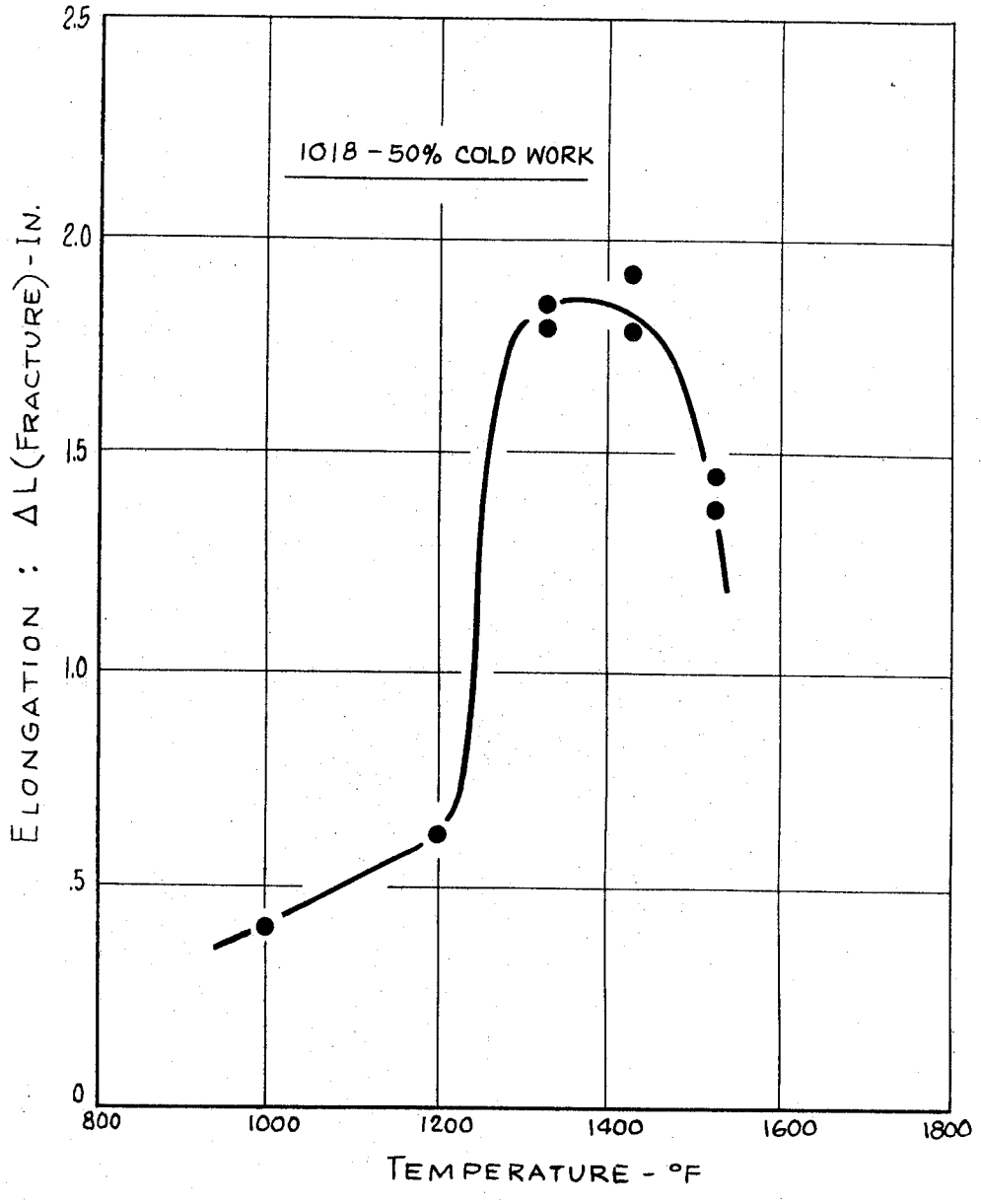

1018 – 50% COLD WORK
ELONGATION : ΔL (FRACTURE) - IN.
2.5
2.0
1.5
1.0
.5
0
800
1000
1200
1400
1600
1800
TEMPERATURE - °F

Fig. 6.

Fig. 7.

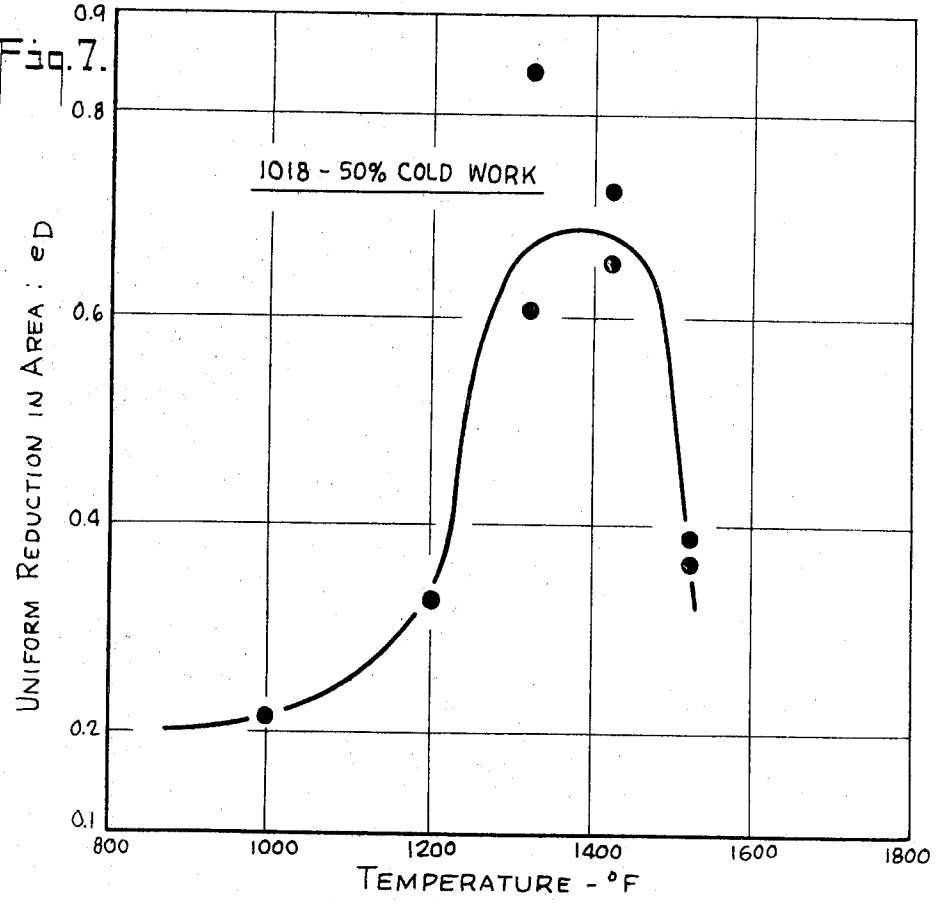

0.9
0.8
0.6
0.4
0.2
0.1
1018 - 50% COLD WORK
UNIFORM REDUCTION IN AREA : eD
800
1000
1200
1400
1600
1800
TEMPERATURE - °F

Fig. 8.

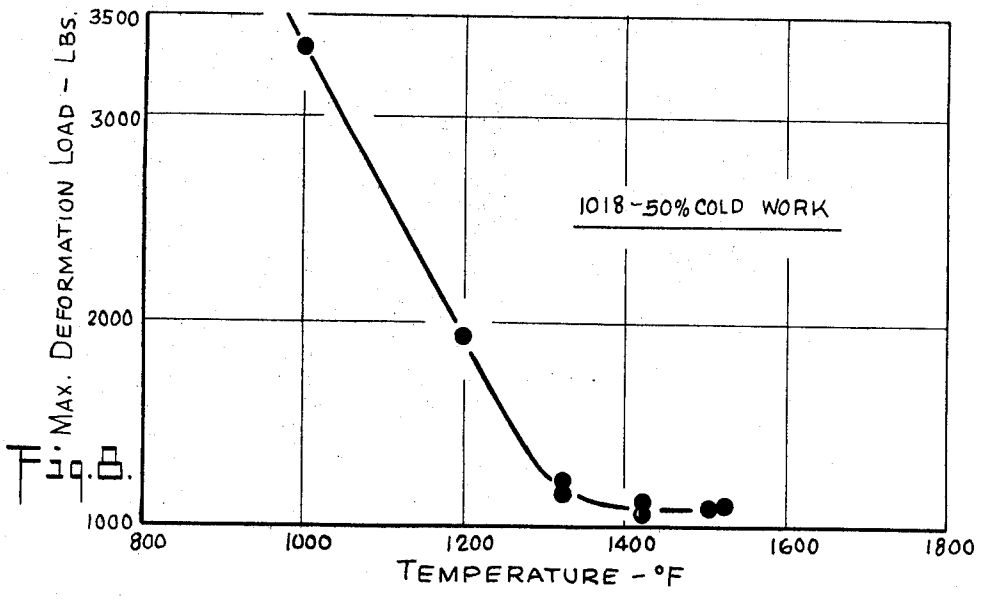
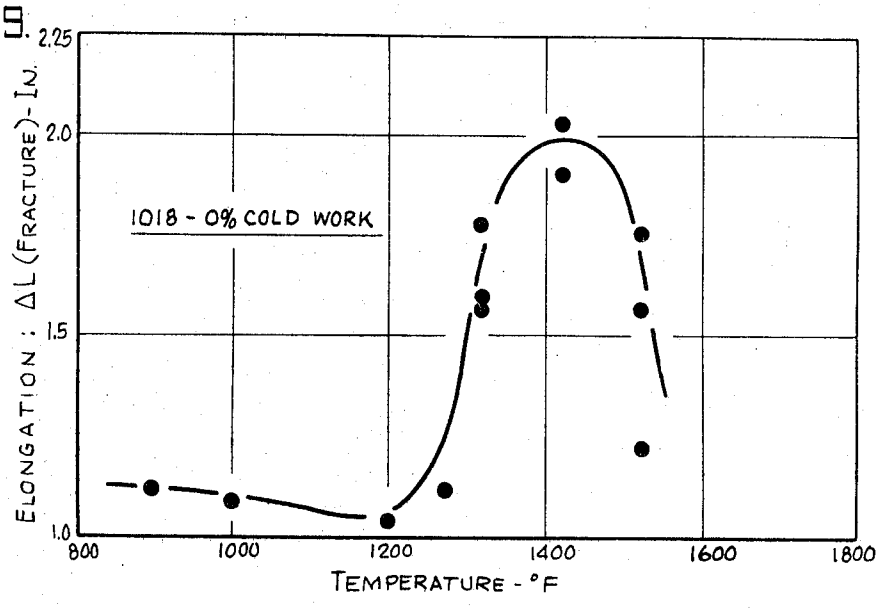

3500
3000
2000
1000
1018 - 50% COLD WORK
MAX. DEFORMATION LOAD - LBS.
800
1000
1200
1400
1600
1800
TEMPERATURE - °F

Fig. 9.


ELONGATION : ΔL (FRACTURE) - IN.
2.25
2.0
1.5
1.0
1018 - 0% COLD WORK
800
1000
1200
1400
1600
1800
TEMPERATURE - °F

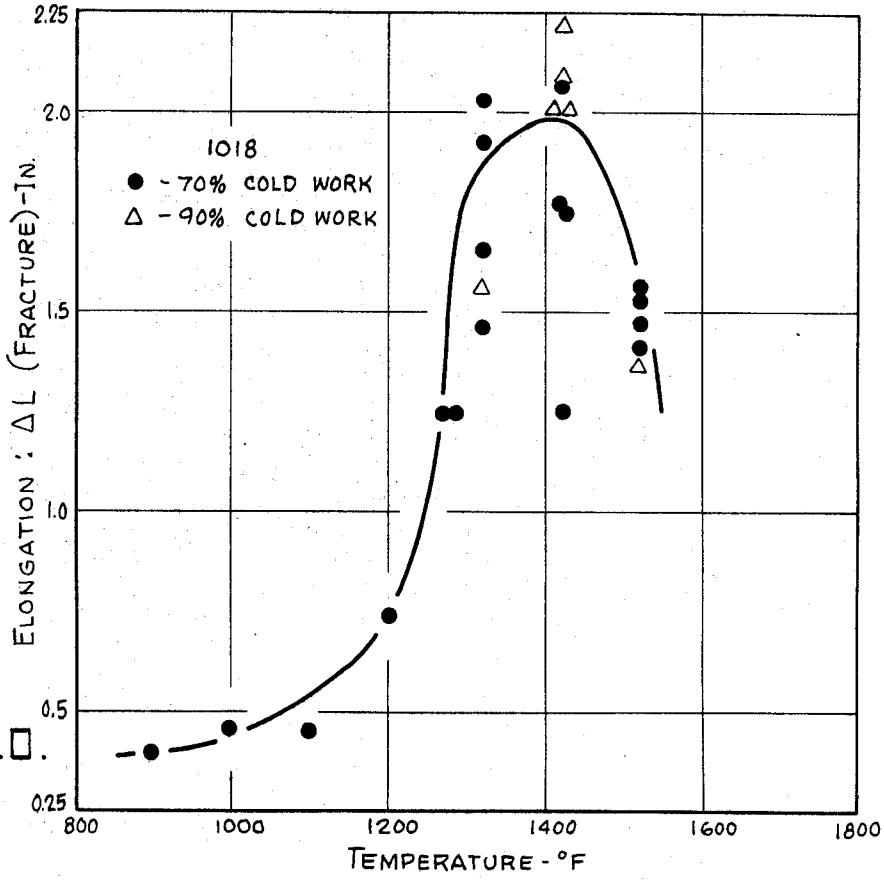
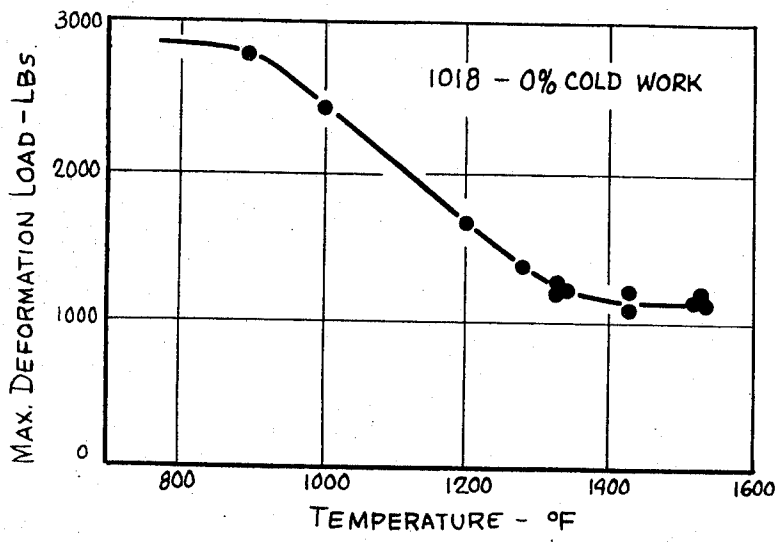

ELONGATION : ΔL (FRACTURE) - IN.
2.25
2.0
1.5
1.0
0.5
0.25
1018
● - 70% COLD WORK
△ - 90% COLD WORK
800
1000
1200
1400
1600
1800
TEMPERATURE - °F

Fig. 10.

Fig. 11.


3000
2000
1000
0
MAX. DEFORMATION LOAD - LBS
1018 - 0% COLD WORK
800
1000
1200
1400
1600
TEMPERATURE - °F

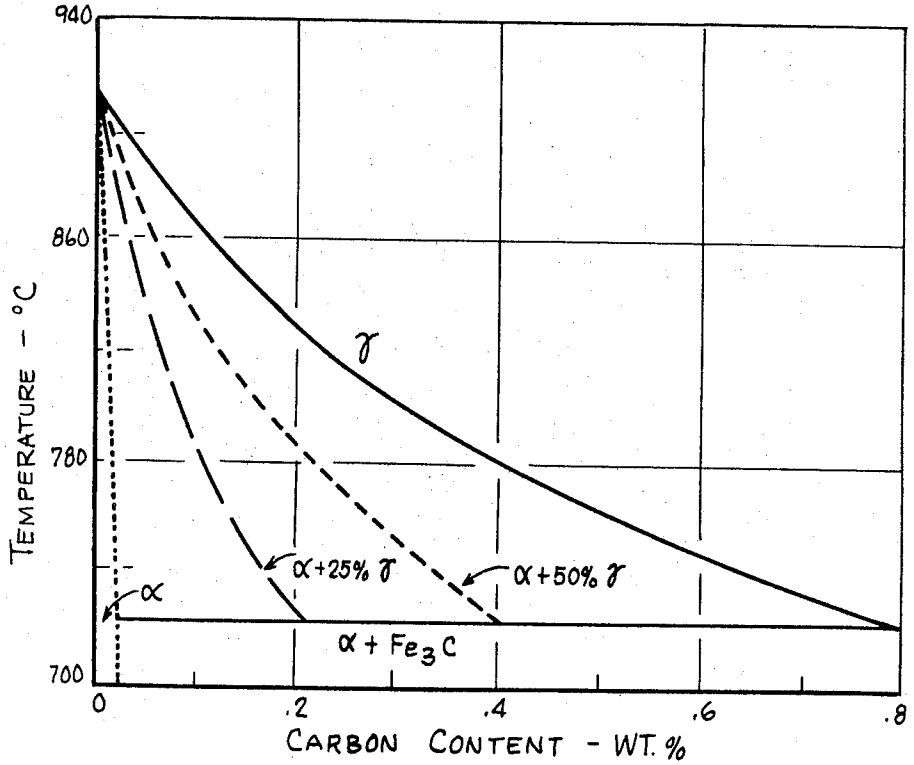

940
860
780
700
TEMPERATURE - °C
γ
α+25% γ
α+50% γ
α
α + Fe3C
0
.2
.4
.6
.8
CARBON CONTENT - WT. %

Fig. 12.

Fig. 13.

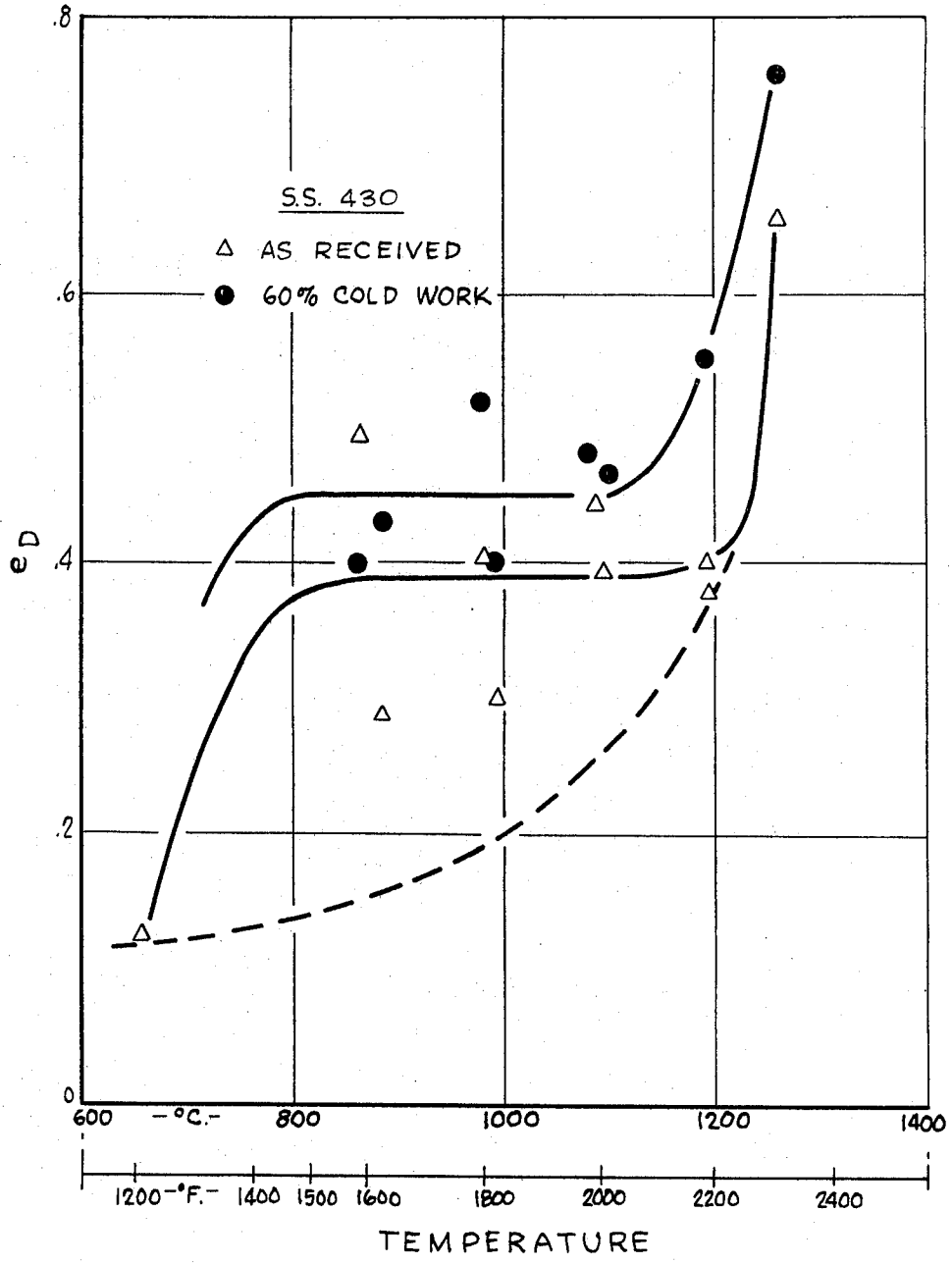

.8
.6
.4
.2
0
eD
S.S. 430
AS RECEIVED
60% COLD WORK
600
800
1000
1200
1400
-°C.-
1200
1400
1500
1600
1800
2000
2200
2400
-°F.-
TEMPERATURE

Fig. 14.

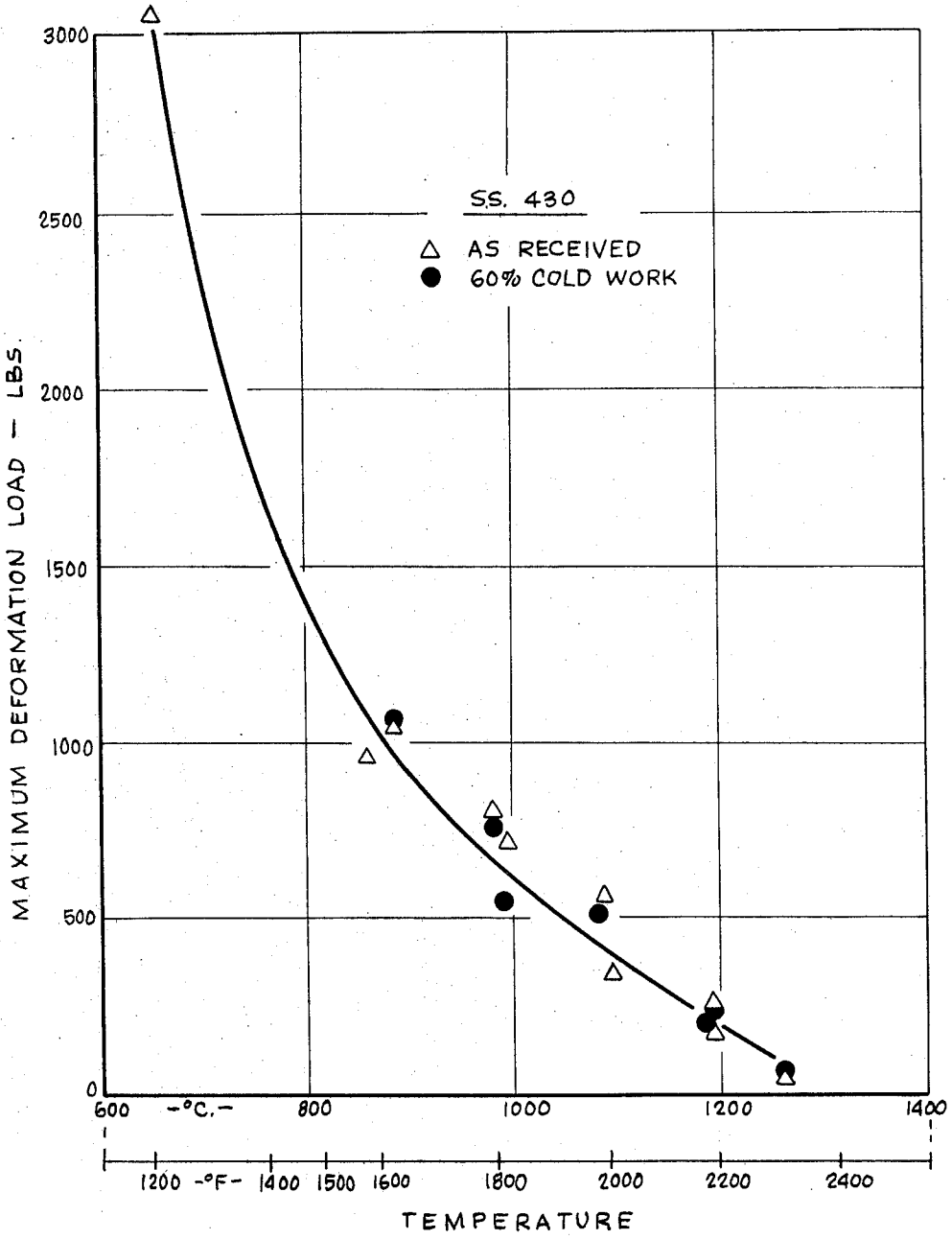
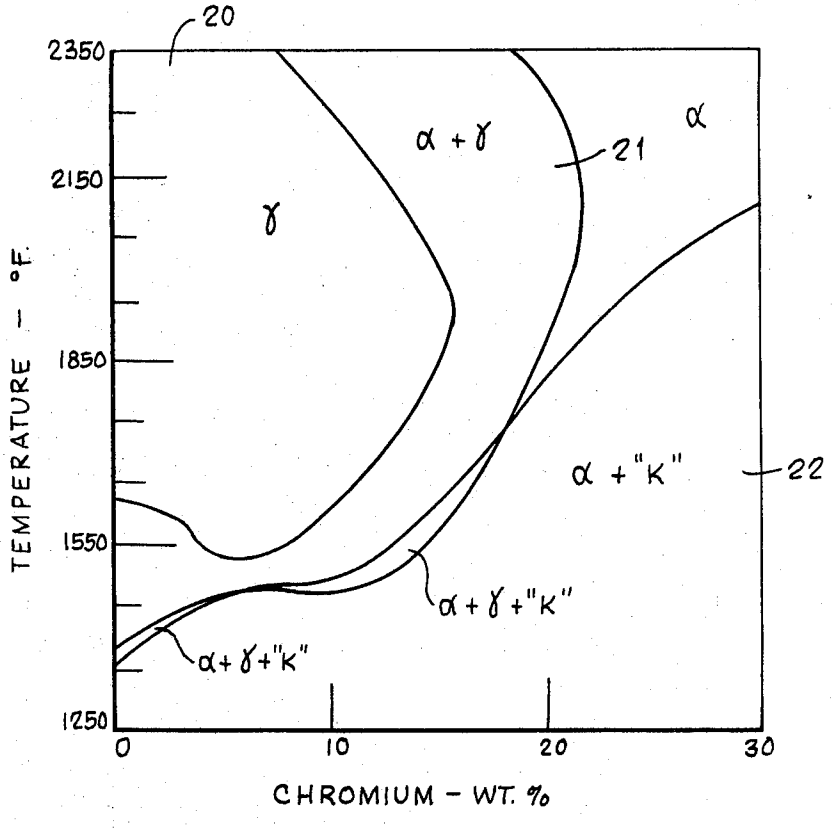

SS. 430
AS RECEIVED
60% COLD WORK
MAXIMUM DEFORMATION LOAD - LBS.
3000
2500
2000
1500
1000
500
0
600
-°C.-
800
1000
1200
1400
1200
-°F-
1400
1500
1600
1800
2000
2200
2400
TEMPERATURE

Fig. 15.


20
2350
2150
1850
1550
1250
TEMPERATURE - °F.
α + γ
γ
21
α
α + "K"
22
α + γ + "K"
α + γ + "K"
0
10
20
30
CHROMIUM - WT. %

METHOD OF PROVIDING SUPERPLASTIC STEEL AND OF PRODUCING ARTICLES BY DEFORMATION THEREOF

CROSS REFERENCE TO RELATED APPLICATION

This patent application is a continuation-in-part of our copending application of the same title, Ser. No. 98,674, filed Dec. 16, 1970, and now abandoned.

BACKGROUND OF THE INVENTION

This invention relates to procedure whereby or wherein steel is converted to a superplastic state, e.g. a state of unusually high ductility, and in a more specific and particularly important sense, the invention is concerned with procedure for making steel articles by establishing superplasticity in steel and deforming the metal by application of stress to achieve desired characteristics of dimension, shape or the like which may involve a large extent of such deformation. Thus for example elongation by stretching is contemplated to an extent of the order of 100 percent or more and similarly large or difficult changes of shape or dimensions by operations such as pressing, drawing, coining or the like. The improvements are related to a wide variety of low carbon steels including ordinary or so-called standard steels and alloy steels of many conventional compositions, and other normally non-austenitic steels such as a number of stainless steel grades.

Superplasticity is a phenomenon known to be attainable in some metals, that has been used or proposed mostly for certain non-ferrous alloys, such as zinc-aluminum compositions, and aluminum-copper, lead-tin, and tin-bismuth alloys. In general, the techniques used to achieve this state of extraordinarily high ductility, in such alloys, have been directed to establishing an extremely fine grain or microstructure, as by recrystallization with or without precipitation effects; a stable condition of superplasticity is then attained for the desired large extent of deformation. Another principle sought to be utilized is to take some advantage of change of phase in metals characterized by different phase conditions, one expedient being to apply heat and cooling so that the metal object is cycled slowly back and forth through a phase change region while stress is applied over a relatively long time, to effectuate deformation.

Although it is readily seen that the attainment of superplasticity or extreme ductility in steel would permit an expedited and simplified mode of fabricating a variety of articles where large changes of dimension or shape are required from an initial piece or blank, or where an article could be made in one piece to have a complex configuration of relatively thin section, not heretofore so produced, there has been no practical success, as presently known, in achieving superplasticity with steels of ordinary types, particularly those mentioned above.

Whether attempted by utilization of the techniques described for non-ferrous metals or by some similar or other approach, superplasticity of practical and commercially useful nature does not appear to have been reached in any ordinary steels, especially those that at room temperatures or other normal temperatures of use are essentially non-austenitic. A principal difficulty is that when superplastic steel bodies have been established, capable in theory of very large elongation

under stress, the strain rate attainable for such elongation (i.e. without rupture) has been extremely low, e.g. of the order of 0.0001 to 0.1 per minute. It will be understood that strain rate is a measure of the rate or speed of elongation, determined as the ratio (e.g. in inches per inch) of the increment of extension to the original unextended length, per unit of time. Strain rates as low as 0.1 per minute, and notably the still lower rates mentioned above are such as to have essentially no practical or commercial value. Formation of a single article to have anything like a deformation of the order of 100 percent would require a time interval of at least many minutes, and most likely of the order of an hour or hours, which is manifestly not feasible.

It will be understood that references to deformation and elongation as related to desired superplasticity or ductility, are herein concerned with substantially uniform changes of length or section over the entire extent of the article subject to deformation. This is in contrast to the ordinary situation of steel where, for example, an application of tension, as in standard tensile testing of a bar or rod, results first in moderate degree of uniform stretch, and then produces a phenomenon known as necking, whereby a localized reduction of sectional area occurs, producing a very thin region at a single place, further stress then leading to rupture at such locality. The present procedure is related to superplasticity in the sense that deformation or reduction of cross-sectional area occurs substantially uniformly throughout the length of a piece being stretched, without necking or other appreciable local consequences — i.e. short of ultimate rupture at the ultimate limit of superplastic elongation.

The invention is therefore designed to afford what are believed to be novel characteristics in steels, particularly to achieve a significant condition of very high ductility, and further and most particularly, to provide novel forming or shaping operations, producing large dimensional changes or other changes of configuration, in a single step, not heretofore generally possible. As will be understood, a primary feature of the invention is to attain the stated superplasticity with the characteristic of a practical strain rate, e.g. upwards of one per minute and indeed desirably in the range of 100 to 1,000 per minute.

SUMMARY OF THE INVENTION

To the above and other ends the invention embraces the discovery that superplasticity can be established in commercial steels with commercially feasible strain rates, by a relatively simple treatment, or sequence of steps, which develops a high order of microstructural instability in the metal, and in a special sense, which permits rapid application of deforming stress in substantial coincidence with the development of a maximum of such instability.

An essential feature of the process involves heating the steel, as from room or other ordinary low temperature, extremely rapidly to an elevated temperature, specifically a temperature where the alpha-to-gamma transformation can then proceed while the steel is thereafter held at such temperature for a short but convenient interval. In particular, the temperature is advantageously selected as one which would afford, at equilibrium, a combination of the alpha and gamma

phases of iron, e.g. such as characterizes the well-known alpha plus gamma phase in the conventional temperature-versus-carbon equilibrium phase diagram for ordinary steel. That is to say, it has been found that instead of attempting to achieve a stable or temporarily stable condition of very fine grain size or the like, or instead of endeavoring to cycle the metal somehow through variations in equilibrium condition between one phase and another, it has been found that the attainment of a severely unstable microstructure — specifically by the procedure of rapid heating upward to a selected elevated value and then holding at said value for a short interval while relatively rapid deformation is effected — an unexpectedly very high ductility is attained. When formed, moreover, the metal of the article promptly proceeds to an equilibrium or stable condition, and the resulting piece, e.g. upon cooling in any desired manner, reattains substantially the characteristics of the selected steel composition with its normal mechanical and other desired properties, or even in some instances superior properties. There is thus no need for any special step or operation to terminate or destroy the superplastic condition, after the forming step.

In many instances, the process of the invention can be circumstanced or modified to take advantage of two cooperating mechanisms or phenomena that coact to establish a special degree or extent of superplasticity. Thus by first subjecting the steel to drastic cold reduction, as for example cold rolling to an extent of 70 percent or preferably higher, e.g. 90 percent reduction, the process is capable, indeed inherently by virtue of the range of temperatures selectable for the above-described unstable state, of providing a correlated characteristic of instability, with respect to recrystallization and with phase-change instability.

This special procedure may be explained by reference to examples, as in the use of ordinary low carbon steel, e.g. AISI 1006 or AISI 1018, respectively 0.06 percent and 0.20 percent carbon. The steel, after the usual first stages of production and hot rolling, is cold rolled to 90 percent reduction, and then the desired piece, without annealing, is rapidly heated, as at a rate of 100° to 200° F. per second, to a temperature within the alpha plus gamma phase area, advantageously midway in such area, this being the area where the ferrite phase begins substantial transformation to the austenite phase and indeed where at equilibrium, e.g. over a period of time, the structure would consist of both ferrite and austenite. In accordance with the procedure of the invention, the article is held at the selected temperature for a short interval, say 5 seconds, and then subjected to rapid deformation, as by drawing, stretching or pressing, to the desired large extent. Conveniently the total time of continuing deformation may be relatively short, of the order of a few seconds or less.

In this way a high degree of microstructural instability occurs in the piece as it is held at the selected temperature, notably after the stated holding interval, and the instability is characterized, according to present understanding, by the initiation and continuance of extensive recrystallization simultaneously with the rapid, vigorous, phase transformation from ferrite toward and to austenite. During and as a heretofore unappreciated result of these conditions of severe instability, a substantial state of superplasticity, i.e. a condition of very high ductility, is found to occur and thus to permit the desired deformation.

As stated, the invention is nevertheless basically advantageous, according to present understanding, without the cold reduction of the metal and thus can be applied to steel produced without cold reduction or to steel which has been cold reduced and annealed, even though in such cases reliance for superplasticity is then placed primarily on the high degree of transformation instability after rapid heating to the selected temperature. It is presently believed, moreover, that in at least some instances the superplastic property may be enhanced by what can be called a continuing recrystallization brought about by the fact of elongation or other deformation — i.e. a secondary phenomenon or mechanism which may well be advantageously occurring in all cases.

As used with any of a large variety of steels, most advantageously low carbon steels but also alloy steels and likewise non-austenitic stainless steels, e.g. the compositions classed as ferritic and martensitic in the 400 series, the procedure can be effective for the production of many highly or intricately deformed shapes, with great economy and efficiency.

BRIEF DESCRIPTION OF THE DRAWINGS

FIGS. **1** to **5** inclusive are graphs where various significant values of deformation properties are plotted against testing temperature, being temperatures to which specimens of a selected steel, AISI 1006, were very rapidly heated and then held, in accordance with the invention, these graphs showing the unusual results attained at the temperature of severe microstructural instability, as follows:

FIG. **1** being a plot of elongation to fracture, against testing temperature, for the steel which had been cold worked to reduction of 50 percent;

FIG. **2** being a plot of uniform reduction in area (i.e. non-necking reduction) against testing temperature, again after 50 percent cold work;

FIG. **3** being a plot of maximum deformation load against testing temperature (50 percent cold work);

FIG. **4** being a plot of elongation to fracture, like FIG. **1**, but for specimens which had not been cold worked; and

FIG. **5** being a plot of elongation to fracture, like FIG. **1**, for groups of specimens that had respectively been cold worked 70 percent and 90 percent;

FIGS. **6** to **11** inclusive are graphs of deformation properties against testing temperature, similar to FIGS. **1** to **5**, for specimens of another steel, AISI 1018, as follows:

FIG. **6** being a plot of elongation to fracture, for specimens having 50 percent cold work;

FIG. **7** being a plot of uniform reduction in area, 50 percent cold work;

FIG. **8** being a plot of maximum deformation load, **50%** cold work;

FIG. **9** being a plot of elongation to fracture, for specimens with no cold work;

FIG. **10** being a plot of elongation to fracture, for groups of specimens that had respectively been cold worked 70 percent and 90 percent; and

FIG. 11 being a plot of maximum deformation load, for specimens with no cold work;

FIG. 12 is an iron-carbon equilibrium phase diagram, plotting temperature against carbon content for pure iron-carbon systems, showing selection of temperatures especially suited for practice of the invention, e.g. for the hypoeutectoid compositions there are two intermediate boundaries shown which predict phase mixtures of 75 percent alpha, 25 percent gamma and 50 percent alpha with 50 percent gamma, each so labeled in the diagram;

FIG. 13, illustrating deformation properties in accordance with the invention, is a plot of uniform reduction in area against testing temperature for specimens of stainless steel grade 430, both in hot rolled condition and after 60 percent cold work;

FIG. 14 is a plot of maximum deformation load against testing temperature for specimens of stainless steel grade 430, as in FIG. 13;

FIG. 15 is an equilibrium phase diagram for stainless steels of the 400 series, plotting temperature against weight percent chromium, exemplified at a carbon level of 0.05 percent.

DETAILED DESCRIPTION

More particularly explained, the procedure is notably applicable to so-called hypoeutectoid compositions, meaning compositions where the carbon content is less than the eutectoid value, e.g. 0.8 percent in ordinary low carbon steel. Preference for this low carbon range is related to the fact that in the ordinary equilibrium diagram plotting carbon percentage and temperature for such steels, FIG. 12, there is an area (commonly designated by the letters alpha plus gamma, and here bounded by the gamma, alpha and alpha-cementite regions) in a temperature range from generally about 1,350° F. to varying higher points up to about 1,650° F. (about 725° to 900° C.), where a stable structure can be reached having a composition consisting of both alpha and gamma phases of iron. It will be understood, however, that the procedure of the invention in no sense involves the attainment of such equilibrium for providing superplasticity, but on the contrary resides in the discovery that a severe microstructural instability, which in turn is found to exhibit the desired superplasticity, occurs when the steel has been heated rapidly from low temperature (meaning any temperature sufficiently below the selected value as not to involve transformation and as not to involve recrystallization where there has been substantial cold reduction) up to the selected value and then holding at said value for development of such instability and for employment of it, so to speak, in deformation under stress. A particularly important consideration is that in this alpha-plus-gamma zone, the phase-change is severe, and though brief, lasts long enough to accommodate the desired, rapid operation of deformation.

Thus the invention is specially applicable to steels, such as the above mentioned hypoeutectoid low carbon standard steels, where the equilibrium diagram exhibits an area of coexistence of two phases, e.g. both ferrite and austenite, or as more ganerally represented on such phase diagram, an area representing, at equilibrium, a combination of alpha iron and gamma iron. Whereas in ordinary steels the eutectoid is a composition containing about 0.8 percent carbon, corresponding characteristics are shifted or different in alloy steels and notably in the non-austenitic stainless steels, e.g. so-called ferritic or martensitic steels of the 400 series. Nevertheless, it is apparent that an austenite-ferrite area exists in a wide variety of steels, notably of low carbon content, and can be availed of for optimum practice of the invention, by corresponding selection of a terminal and holding temperature after the stated fast heat-up, for achieving a high order of temporary instability.

To a degree, however, it is also conceived that the procedure can be utilized with other steels, advantageously of low carbon content (e.g. not more than 1 percent), where a value of elevated temperature can be selected and maintained such as to yield a state of transformation instability which, even though very short, may be sufficient to permit deformation under stress. As will be understood, however, selection and maintenance of such temperature may require special care, notably to avoid so rapid a phase transformation as not to permit practical utilization of the instability of microstructure in accordance with the principles explained above. Moreover, if advantage is also to be taken of the action of recrystallization while it is going on — such action being conceived as the conversion of large, elongated crystal structure resulting from cold reduction, to a fine, equi-axed crystal form — care must be taken to time the occurrence of such action with the occurrence of the transformation instability.

The procedure can afford an unusual forming operation effective for production of a variety of articles such as greatly elongated parts, cuplike elements, or similar tubular casing or pieces closed at one end having cylindrical or other shapes, various other hollow articles such as refrigerator shells, hub caps, automobile body elements, and the like. In general, these are parts which may have heretofore required a series of forming steps, sometimes a long series, as by successive stages of cold forming, e.g. pressing, drawing or other deformation, interspersed with annealing steps. Inasmuch as the strain rate afforded by the invention generally resides in the range of 100 to 1,000 per minute and inasmuch as the property of superplasticity is found to be accompanied by a marked decrease of the required deformation load, the process results in an unusually rapid and highly economical operation for fabricating finished or semi-finished structures in one or no more than a few forming steps, from a wide variety of steels of recognized commercial type, including steels of essentially inexpensive character such as the ordinary low carbon compositions.

Referring further to specific examples of the invention, an extensive series of tests was carried out with specimens of ordinary carbon steels as noted above, being AISI 1006 (0.06 percent carbon) and AISI 1018 (0.20 percent carbon), both well below the eutectoid. The 1006 steel was aluminum killed, but presence or absence of such condition is not understood to bear significantly upon the invention. The steels had the usual manganese contents (below 0.39 percent and 0.67 percent respectively) and were entirely conventional non-alloyed compositions, with usual limits of phosphorus and sulphur. Ordinary processing with hot rolling had been followed in each case.

For test purposes, cylindrical specimens were prepared, ¼ inch diameter by 8 inches long, by machining from suitable pieces. Where cold working was employed prior to test, the pieces were of material which had been reduced by cold rolling in conventional manner, without intermediate or final anneal. Specifically, specimens were employed for which the metal had zero cold working and cold working to reductions of 50 percent, 70 percent and 90 percent respectively. These percentages refer to reduction in thickness, the cold rolling being effected with the necessary series of passes, preferably in the same direction, for production ease. In order to take full advantage of the occurrence of recrystallization in practice of the invention, the selected degree of cold working should be performed without intermediate or subsequent anneal; any previous treatment (not employed in these tests) which may have involved a cold work and anneal, is disregarded. For the invention, cold working can be achieved in any suitable manner, as by rolling, forging or like operations.

All of the tests in these series involved heating the specimen rapidly, i.e. at a rate of 120° F. per second or thereabout, to a test temperature, where the specimen was held for a predetermined short time, which was 5 seconds for the 1,006 specimens and 10 seconds for the 1,018 specimens, and then each specimen (while remaining at such temperature) was subjected to rapid deformation, e.g. at a cross head speed of 90 inches per minute. The equipment used, for heating, holding and deformation, was an apparatus commercially available for test purposes (under the name of Gleeble), designed for other tests of high temperature properties as in welded structures.

In general, the tests involved determination of deformation characteristics at various test temperatures, being the selected temperature to which the specimen was heated, and at which it was held, and the results demonstrated the occurrence and utilization of superplasticity in accordance with the invention, being the effect of severe microstructural instability when the temperature was in the austenite-ferrite range for these steels. For best representation of superplasticity, all deformation tests in these examples were of tensile characteristics, and all deformation was effected by application of tensile stress alone, exerted between elements gripping the end portions of the specimens. It was fully apparent, indeed confirmed by other investigation, that the extraordinarily high ductility revealed by these tests was accompanied by other, unusually superior properties of deformability, as in pressing to deep or complex shapes, full conformity with intricate die configuration, and the like.

One index of superplasticity was taken to be the extent of elongation to fracture, under tensile stress. For simplicity of comparison among specimens of identical shape and size, identically held by the tension-applying elements of the test apparatus, this elongation was simply measured as the total increment of length, in inches, exhibited by the stretched specimen, to the point of fracture. The fact that in each case the specimen necked down, just prior to fracture, at the fracture locality did not affect the comparative significance of the results, it being fully apparent that the large values of increment in length obtained at the test temperatures of the invention were fairly representative of a large comparative increase in neck-free elongation of the free central portion of the test specimen which had an effective gauge length of about 1 inch.

FIG. 1 shows the result of such tests for various specimens of the 1006 steel that had been cold worked 50 percent, the specimens which were prepared from the cold worked metal being rapidly heated to various test temperatures as shown in the figure, held and subjected to stress deformation as described above. As will be seen, a very pronounced maximum of elongation to fracture was achieved for temperatures in the range of about 1,375° to about 1,550° F., with optimum results at approximately 1,500° F. These values coincide with the alpha-plus-gamma region of the equilibrium diagram (see FIG. 12), the value of 1,500° F. (about 815° C.) being substantially that for which at equilibrium, e.g. after prolonged holding and stabilization the metal (0.06 percent carbon) would consist of approximately 75 percent ferrite and 25 percent austenite from the figure for pure iron-carbon. The actual phase boundaries for commercial alloys will be shifted due to the presence of Mn, P, S, Al, Si, etc.

Another and somewhat more significant value was measured as indicated in FIG. 2, being the extent of uniform reduction in area, obtainable under tensile elongation, at the several test temperatures. Specifically this is in effect a measure of the maximum cross-section reduction which is reached in each case before necking occurs. This is believed to be an effective measure of superplasticity, or of high ductility, inasmuch as the purpose of the invention is to achieve a high order of deformation, without necking or equivalent localized effect in the workpiece subjected to the desired deformation. The value actually determined in these tests is a function of the ratio of the original cross-sectional area of the test rod to the final area (just prior to necking), conveniently a logarithmic value e_D, defined as

$$e_D = 2\ 1n\ (D_0/D_f)$$

where D_o is the initial diameter and D_f is the final diameter, with factor 2 providing the conversion to a ratio of areas. The values of e_D obtained at various test temperatures are shown in FIG. 2, where the test specimens (of 1006 steel) were the same as in FIG. 1. Percent elongation is definable in reference to e_D data; thus e_D values of 0.4 to 0.8 represent uniform elongation of about 50 percent to 125 percent. Normal uniform elongation for these steels is about 25–30 percent or less.

It will be noted that in FIG. 2 a sharp maximum of useful result again appeared at about 1,500° F., with useful high values also apparent at adjacent points within the range of the austenite-ferrite region. Specifically the measured quantity e_D peaked to values varying from 0.4 to above 0.5 (i.e. above 65 percent elongation) at the effective temperature range from values of 0.1 (10 percent elongation) and less both below and above it.

A further and extremely desirable characteristic of superplasticity is a large reduction in required deformation load, i.e. stress required to deform the workpiece. With the specimens of the 1006 steel which had received 50 percent cold work, measurements were

made of the maximum load needed for tensile deformation at various test temperatures, i.e. the maximum reading of applied load reached during the brief interval of deformation. Results were as plotted in FIG. 3, showing a low minimum of such deformation load at about 1,500° F., with a useful range of superplasticity generally similar to that exhibited in the preceding figures. Specifically, the required tensile load dropped to well below 1,000 pounds, indeed close to 500 pounds, from values in the range of 2,000 to 3,000 pounds at lower temperatures. Other investigation has indicated that the maximum tensile load likewise rises at higher temperatures than as shown in this graph; certain limitations of the test equipment prevented extension of this and other curves above 1,600° F. in these sets of tests.

Tests of elongation to fracture, similarly made and measured as those reported in FIG. 1 and following identical procedure of treatment of the specimens were also made of specimens of the 1006 steel where the prior treatment of the metal was different from that in FIG. 1. Specifically, tests on specimens of steel that had received no cold working are reported in FIG. 4, showing a significant maximum of ductility at essentially the same value, i.e. about 1,500° F. Like tests on specimens of which the steel had received relatively large amounts of cold working (without anneal) are plotted in FIG. 5, the solid circles representing specimens of steel that had been cold reduced 70 percent, and the triangles being specimens of metal that had received cold reduction to 90 percent. The superiority in elongation to fracture, at and adjacent to the optimum temperature of 1,500° (reach increments well above 2 inches) is very great, being especially so for the 90 percent cold worked metal.

The second example of the invention is illustrated by the tests similarly made with specimens of the 1018 steel as explained above. For metal that had been cold worked to 50 percent reduction, FIG. 6 shows the values of elongation to fracture (maximum nearly 2 inches), while FIG. 7 correspondingly shows the extent of uniform reduction in area (maximum e_D, 0.6 to 0.8+), and FIG. 8 indicates the maximum deformation load (lowest 1,000 pounds), all for various test temperatures. The curves and results agree qualitatively with those in FIGS. 1, 2 and 3, indicating a very large increase in ductility and a large decrease in required tensile stress, at the optimum temperatures for this alloy, specifically a temperature in the range of about 1,350° to 1,450° F., notably temperatures aroung 1,400° F.

Again, tests of the 1018 steel which had not been cold worked at all and which had received higher degrees of cold working were made, under conditions otherwise the same as those for FIG. 6. Specifically, in FIG. 9, the steel without cold working showed a fairly sharp maximum of elongation to fracture at around 1,400° F. (peak value about 2 inches), while very high values were achieved in the steels tested in FIG. 10, the solid circles being 70 percent cold worked material and the triangles 90 percent cold worked material (peak values well over 2 inches), the optimum temperatures, around 1,400° F., being clearly similar to those revealed in FIGS. 6 to 9.

Corresponding to FIG. 8, tests of maximum deformation load (for tensile stress) were made as reported in FIG. 11, relative to the 1,018 specimens (see FIG. 9) of which the metal had received zero cold working. A minimum value (about 1,000 pounds, as against other values up to 3,000 pounds), being an extremely desirable attribute of superplasticity, appears in the vicinity of the optimum temperature of 1,400° F.

From all of these tests, which in effect represent practice of the invention at the optimum and near-optimum temperatures, and from other investigations by test, it has been demonstrated that the present procedure affords unusually high ductility, representative of the desired superplastic state, and thus permits feasible and rapid deformation to large elongations, such as 50 percent and well above, and usually of the order of 100 percent and greater, e.g. 300 percent in optimum conditions. Effective results are achieved with material which has received no cold working at all, but significantly superior advantage is demonstrated for metal cold worked to a reduction of about 70 percent or more. It will be seen that the process is eminently practical, requiring only a rapid heating of the piece up to the selected transformation temperature, then holding the piece at such temperature for a brief time to develop severe microstructural instability, and rapidly performing the desired deformation, indeed under relatively low stress.

Summarizing the above examples and taking into account other experimental work of similar sort relative to the invention, a presently preferred practice, for obtaining excellent deformability at high strain rates in the case of ordinary carbon steels having a carbon content below 0.8 percent is as follows: the piece is heated rapidly at a rate of 100° to 200° F. per second to a selected temperature in the two-phase field, preferably a temperature which on holding to the point of stabilization would produce a microstructure of about 50 percent ferrite and 50 percent austenite. Thereupon after a short holding time, e.g. 5 to 10 seconds, the piece is deformed under stress, conveniently so at a strain rate upwards of 10 per minute and advantageously in the range of 100 to 1,000 per minute. Maximum superplasticity appears to be obtained if the steel has previously received a significant cold reduction, e.g. 70 percent and upwards, and preferably about 90 percent. For the AISI 1006 steel, the selected temperature (apparent optimum 1,500°) can be in the range of 1,475° to 1,550° F., while a preferred value for AISI 1018 is about 1,400° to 1,425° F.

It is essential that the treatment temperature be reached by heating the metal up from a much lower value, conveniently from room temperature or usually at least from a value several hundred degrees F. below the transformation point. Indeed where recrystallization effects are to be significant, the heat-up must progress from a temperature well below any annealing value. The rate of heating is preferably quite rapid, as in the range mentioned above, although it is understood that substantially lower rates can be useful, especially where the onset of severe instability does not occur before reaching the desired temperature. In general heat-up rates should be at least 10° F. per second, more suitably 20° F. per second and upwards; a good range appears to be from 50° to 300° F. per

second. While there is no theoretical upper limit, convenience is served by rates below 500° per second; and indeed in some cases a lower degree of superplasticity has inexplicably been noted with extremely fast heat-up times, e.g. above 1,000° F. per second. The time taken, from room temperature, has normally been of the order of 10 to 12 seconds, with little change in result at times as high as 60 seconds or as short as 1 or 2 seconds. Correlation of the heat-up rate with occurrence of recrystallization (if desired) from cold worked condition is relatively simple to determine, usually in the light of known or ascertainable characteristics for annealing the selected steel.

In most cases, some holding time will be required, suitable values for the low carbon steels having been indicated above. In other situations the optimum time is at most a matter of simple test. For example, after selecting a conveniently rapid heat-up time and selecting an appropriate terminal temperature from the phase diagram for the metal (where the minimum is usually about 1,330° F. for ordinary steel) a short series of tests, e.g. utilizing several different holding times and then applying deformation at the desired strain rate, will readily reveal the holding interval for best results. Indeed selection of the treatment temperature itself may be correlated with the strain rate needed for a given type of deformation or other needs, in that within an available range of temperature, the useable range of strain rates appears to rise with temperature.

Once the parameters have been selected, as to treatment temperature, heat-up rate and holding time, practice of the invention is a convenient and expedited operation. Suitable provision is made for heating the workpiece rapidly, as by induction heating, and then by appropriate thermal control, maintaining the piece at selected temperature for the desired time, including the holding interval and the very brief period required for actual deformation. The latter task may be performed in any suitable fashion, with equipment adapted to hold the piece at the desired temperature, e.g. as part of the heating equipment, or to receive the piece from such equipment into whatever dies (e.g. preheated) or other elements may be needed for engaging and for mechanically exerting stress in the desired forming operation.

Thereafter, treatment of the shaped article may be of any sort desired, essentially without regard to the fact that a condition of superplasticity has temporarily existed. Indeed such condition, including the stated instability, will usually disappear at once, more or less simultaneously with the completion of forming or very briefly thereafter. The piece can be allowed to cool slowly or rapidly and can be subjected to heat treatments or other operations as may be conventionally desirable for the ultimate service of the article and for agreement with the metallurgical properties of the selected alloy. Indeed it is found that in a number of cases the resulting article has properties superior to those of articles formed by conventional procedure (such as repeated cold working and annealing), in being characterized by a finer grain size, better toughness, smoother or otherwise improved surface quality, and indeed in the case of stainless steel, a lessening or avoidance of the surface shape characterized as roping.

If the nature of the required fabrication is such that a single deformation is insufficient, it will be understood that the piece, after cooling, can be further worked in the same manner, as by again rapidly heating and holding at the temperature of microstructural instability, while a second deforming operation is performed, which may, of course, be similar to or different from the first. While in such repetition or repetitions it may not be possible to take advantage of recrystallization effects from a cold worked state, substantial superplasticity is nevertheless attainable.

Although the invention is not dependent on theoretical considerations, it is believed that the superplasticity is at least in substantial part occasioned by the change in structure (while it occurs) of some of the material from the body-centered cubic structure of alpha iron, which is the low temperature phase, to the face-centered cubic structure of gamma iron, which is the high temperature phase. Thus there is a driving force, so to speak, which produces such transformation and which may be deemed to contribute to the severity of instability. At the same time where the material has been heavily cold worked, the concurrent recrystallization affords a like driving force in converting the elongated crystals to very small equiaxed ones. Indeed tests have been made by quenching the deformed product immediately after such deformation, and examination of the microstructure has indicated a very fine grain, and has further indicated that where recrystallization has occurred, the crystals are small and equiaxed.

It is also believed that even where there has been no previous cold working, the deformation, during transformation instability, may be accompanied by a so-called continuing recrystallization, which enhances the unstable condition. The preference for operation in the two-phase field is that the instability is there considerably prolonged, so as to afford full opportunity for deformation time and for avoidance of difficultly close control of time and temperature. Indeed it may be noted that exact treatment temperature values do not usually have to be maintained, e.g. in the sense that if the theoretical optimum is 1,500° F., the selected temperature can actually be a range, as from 1,475° to 1,550° F.

The invention is applicable, as stated above, to a variety of steels, which are normally non-austenitic, i.e. which are not austenitic at room temperature as commonly processed, and especially in such steels having a two-phase region of austenite plus ferrite at specific elevated temperatures. These include, in addition to the ordinary carbon steels (most advantageously up to about 0.5 percent C), any of a wide variety of alloy steels, of which a few examples of compositions that are appropriate for making articles of the sort for which the invention is especially suitable, are grades 4340, 8620 and 10B20, the latter being a boron steel corresponding to grade 1020. It is likewise conceived that a number of high alloy grades can be treated. Particular utility is contemplated for stainless grades of the 400 series as mentioned above, these being generally straight chromium compositions, with chromium content ranging from 11 percent to 27 percent, and generally lacking nickel as an intended alloying element, except for one or two grades with nickel up to 2.5 percent. Specific examples of stainless steel are

AISI types 409, 410, 430, 434 and 436, which have carbon contents up to about 0.2 percent. All of the above steels, whether of ordinary, alloy or stainless types, and including those which are conventionally classed as martensitic, may be generically deemed to be of normally ferritic compositional character.

The forming operations conceived as useful include drawing and stretching, pressing, coining and similar die procedures, and indeed a variety of operations often attempted in the cold state, such as those of the nature of heading, upsetting, and like impact methods. Not only is there obtainable a very large extent of deformation, preferably upwards of 100 percent in elongation, but a better conformity with intricate die configurations, thus permitting single step fabrications in many cases not heretofore attained. Indeed parts or shapes which could not at all be produced in one piece may now be so made. As also stated, suitably high strain rates are realized, well above 1 or preferably 10 per minute, and indeed up to 1,000 per minute or higher if feasible.

A further example of the invention is represented by application of the process to stainless steel of the 400 series, i.e. so-called straight chromium compositions, particularly stainless steel type 430 having a nominal chromium content of 14 percent to 18 percent. For these tests a commercially produced 430 grade stainless steel was used, containing, by weight, chromium 16.25 percent and carbon 0.073 percent, with the usual maximum limits for incidental elements (e.g. Mn and Si each under 1 percent).

The tests were performed in essentially the same way as in the other examples above. The steel had been produced by ordinary processing, including hot rolling to appropriate plate thickness. A part of this plate material was subjected to cold rolling to a reduction of 60 percent (without subsequent annealing), and another portion was left in the as-received condition, i.e. with no further work after hot rolling. Cylindrical specimens of both types of material were prepared, ¼ inch in diameter and 7½ inches long, providing a grip separation of 4½ inches. The same test equipment was employed as in preceding examples, having provision for heating the specimen rapidly to a selected temperature, and then allowing it to be held at such temperature while deformation was effected, i.e. by stretching. The desired determination was with respect to superplastic conditions, as represented by stretch deformation to a maximum reduction of cross-section before occurrence of necking. In particular, the tests regarded as significant for this steel, within the competence of the equipment described above, were measurements of uniform reduction in area at various test temperatures, and likewise the deformation loads required, i.e. stresses needed to deform the work piece.

The results of the tests, illustrating the effectiveness of the invention in this type of stainless steel, are shown in FIGS. 13 and 14. Each plotted point represents a separate test, being the same test for the two figures, wherein a specimen was subjected to rapid heating, was held at the indicated temperature and was promptly deformed while at such temperature and specifically, while it remained in such state of microstructural instability, if any, as occurred. In each case the specimen was heated at a rate of about 125° F. per second, requiring a time of about 12.5 seconds to reach the test temperatures in the optimum localities of the range. The specimen was held for 10 seconds at the selected temperature and then immediately subjected to deformation by axial stretching, while remaining at temperature, at a cross head speed of 90 inches per minute. In each of the figures, the plotted triangles represent the as-received specimens, with no cold working, and the circles represent specimens which were characterized by 60 percent cold reduction.

As illustrated in FIG. 13, the tests showed that when the specimens were heated to a temperature in the range of about 1,470° F. to about 2,100° F., they exhibited highly desirable values of e_D as defined above, particularly values of the order of 0.4 or better, indicating deformation with uniform reduction in area, and no necking, to elongation of about 50 percent or higher, in contrast to normal uniform elongation for 430 stainless steel, of not more than about 20 percent. It was evident that at these temperatures the metal was temporarily in a condition of microstructural instability while alpha to gamma transition was occurring.

The cold worked specimens showed significantly higher values of e_D, understood to result from the contribution of the occurrence of recrystallization to the total microstructural instability that afforded the superplastic state. FIG. 13 expressly shows that the increase in ductility was roughly three-fold from the situation of specimens at 1,200° F., which is understood to be approximately the same as at room temperature. It is also understood that if the phase transformation and recrystallization phenomena were not present, i.e. as by holding the metal for a long time at a given temperature so that the microstructure was stabilized before deformation, the values of e_D would be represented approximately by the dotted line curve in FIG. 13.

FIG. 14 shows the values of maximum load required to deform the specimens at the selected temperatures. As in the case of other steels mentioned above, a drastic decrease in deformation load requirement was exhibited over the temperature range of superplasticity, including a three-fold decrease between 1,200° F. and 1,580° F., thereby further substantiating the applicability of the invention to the steel, in attainment of this additional advantageous property. Strain rates, as will be understood, were well within the preferred ranges indicated above, the test operations being at the same cross head speed as in the other examples.

It will be noted in FIG. 13 that the improved elongation characteristics were found over a plateau of temperatures from about 1,470° F. to 2,100° F., and rose substantially at much higher temperatures. Similarly the deformation load requirements decreased, as shown in FIG. 14, over the plateau temperature range indicated in FIG. 13 and beyond. It will be understood, however, that the major significance of these tests is the improvement experienced in the lower temperature values of the plateau range, particularly in that higher temperatures are less commercially desirable, and indeed at exceptionally high values would become very difficult to utilize. At the same time, increases of temperature above about 1,580° F. do not significantly improve the ductility until values of about 2,100° F. are reached. In consequence optimum temperatures for the selected steel are represented by the range of about

1,500° to about 1,700° F., preferably 1,550° to 1,600° F.

The above tests, as illustrated in FIGS. **13** and **14**, afford abundant confirmation that the invention is applicable to alloy steels such as the stainless grades of the **400** series and that the described, large increase in ductility or superplasticity is attained for a brief but sufficient interval when the steel has been heated rapidly to a convenient temperature in the alpha-gamma phase transformation range, with the aid of concomitantly occurring recrystallization when the metal has been cold worked. The difference in nature of results at very high temperatures, and thus of the shapes of the curves in FIGS. **13** and **14**, from those for example in FIGS. **2**, **3**, **7**, **8** and **11**, is understood to be accounted for by the different characteristics of the alloy compositions, as indicated by a typical example of a phase diagram for these stainless grades, shown in simplified form in FIG. **15**. In this graph, which is essentially a plot of equilibrium phase conditions at various temperatures, for various weight percent contents of chromium, and which for convenience of illustration is specifically such a plot for alloys containing 0.05 percent carbon (it being understood that generally similar configurations pertain to other carbon levels, with specific differences in shape and location of phase boundary lines), it will be seen that there is a so-called gamma loop as indicated at **20**.

More particularly, in FIG. **15** the gamma loop is surrounded, in effect, by an alpha plus gamma band **21**, representing a region for occurrence of the same kind of transition as is involved in the triangular area of FIG. **12**. Regions below these areas **20** and **21**, e.g. as generally indicated at **22**, can be considered equivalent to the alpha or alpha-cementite regions at the bottom of FIG. **12**, i.e. where the ferritic type of microstructure prevails, as distinguished from the basic austenitic structure within the gamma loop **20**. It will be understood that the upper portion of the complete diagram, where the boundaries of regions **20** and **21** return to the zero chromium axis, is omitted from FIG. **15** as being of no concern here, and likewise certain special characteristics are omitted relative to regions below the band **21**, or partly within it, where, depending on temperature and chromium content, there are varying proportions of one or more iron-chromium carbides. For simplicity the latter features of composition have been merely indicated by the letter K, it being understood that at various localities, K comprises one or more of various iron-chromium carbide compositions.

As will be recognized from phase diagrams of the sort shown in FIG. **15**, ferrite-austenite transition regions can exist over a large range of temperatures for at least some alloys, thus presumably accounting for the phenomena at very high temperatures in FIGS. **13** and **14**, while nevertheless affording practice of the invention at conveniently lower temperatures for all these alloys, as explained. Moreover, although the equilibrium phase diagrams, including the position of the gamma loop, vary for different alloys of these types and specifically for different carbon levels, and the phase boundaries may also shift somewhat with heating rate, and although a complete transformation to austenite may not be attainable, alpha-gamma transition regions appropriate for present purposes are available in generally all cases, e.g. for the variations in chromium content in the 400 series and for variations in carbon level, from extremely low values up to 0.5 percent or higher. In other words, while the driving force for austenite transformation on rapid heating into the two-phase field is more complicated in these alloys than for ordinary carbon steel, this circumstance does not alter the basic applicability of the present invention.

As indicated, these results demonstrate the attainment of the desired superplastic state with the ferritic grades of stainless steel, it being understood (as explained above) that the term ferritic or "normally ferritic" is herein employed to include those particular grades which are usually martensitic in the as-produced cold state, in that the reversible ferrite-martensite transformation occurs at lower temperatures than the terminal values of the present process, which consequently involves rapid heating, in all cases, upward through the ferrite region to the point of alpha-gamma transition. As shown by FIG. **13**, advantage may be taken of the instability of occurring recrystallization in these stainless steels in the same general way as for the plain carbon grades. The extent of cold working required to effectuate this contribution to high ductility appears to vary with steel compositions, but it will be understood that the fact of such contribution and the percentage of cold work needed can be readily determined by tests of the sort illustrated in FIG. **13**, for any given steel. In other words, although the advantageous effects of recrystallization are not presently deemed essential (where major instability by phase transition can occur), they appear to be generally capable of determination and realization in desired cases.

It is to be understood that the invention is not limited to the specific operations and materials herein described, but may be carried out in other ways without departure from its spirit.

We claim:

1. A method of establishing a condition of high ductility, that provides deformability at a desired strain rate, in a steel which is of normally ferritic compositional character and has a carbon content such that as exhibited by the equilibrium phase diagram for said compositional character there is an alpha-gamma-transition temperature value where upon heating to and holding at such value microstructural instability can be temporarily attained, comprising subjecting a body of said steel to rapid heating, at a rate of at least about 10° F. per second, to the aforesaid temperature value and holding the body at said value for providing said high ductility in said body during an interval of microstructural instability while transformation occurs in the steel from alpha iron at least partially to gamma iron.

2. A method as defined in claim **1**, in which the steel is a carbon standard steel having a carbon content in the range up to about 1 percent, and said rapid heating is effected at a rate of at least about 50° F. per second from a temperature substantially below any value for transformation of alpha iron in the steel.

3. A method as defined in claim **1**, in which the steel is a stainless steel of the 400 series having a carbon content in the range up to about 0.2 percent.

4. In a method as defined in claim **1**, procedure for deforming the steel body as characterized by said high

ductility, comprising subjecting the body to rapid deformation by applying stress thereto while the body is held at the said temperature value in said interval of microstructural instability.

5. A method of establishing a condition of high ductility, that provides deformability at a strain rate greater than 0.1 per minute, in a steel which is of normally ferritic compositional character and has a carbon or alloy content such that as exhibited by the equilibrium phase diagram for said compositional character there is an alpha-gamma-transition temperature value capable of providing temporary microstructural instability, comprising cold reducing a quantity of said steel by at least about 70 percent and thereafter, without annealing said cold reduced steel, rapidly heating a body of said non-annealed, cold reduced steel, at a rate of at least about 10° F. per second, to the aforesaid temperature value and holding the body at said value for providing said high ductility in said body during an interval of microstructural instability while transformation occurs in the steel from alpha iron at least partially to gamma iron.

6. A method as defined in claim 5, in which said quantity of steel is cold reduced by at least about 90 percent.

7. A method of establishing a condition of high ductility, that provides deformability at a high strain rate, in a steel of normally ferritic composition having a carbon content providing a hypoeutectoid character such that said steel exhibits a phase field, in a range of elevated temperature values, which at equilibrium contains both ferrite and austenite, comprising subjecting a body of said steel to rapid heating, at a rate of at least about 20° F. per second, to a temperature in the aforesaid range, the rapidity of said heating and the selection of said temperature being such that after reaching and upon being held at such temperature the steel undergoes microstructural instability which does not substantially terminate before the end of an interval of at least several seconds at said temperature, and thereupon holding the body at said temperature for providing said high ductility therein during occurrence of said microstructural instability while phase transformation is occurring in the steel.

8. A method as defined in claim 7, in which the steel is a carbon standard steel having a carbon content less than 0.8 percent, said range of temperature values extending upward from about 1,330° F.

9. A method of establishing a condition of high ductility, that provides deformability at a high strain rate, in a steel of normally ferritic composition having a carbon content providing a hypoeutectoid character such that said steel exhibits a phase field, in a range of elevated temperature values, which at equilibrium contains both ferrite and austenite, comprising subjecting an essentially non-annealed, cold-reduced body of said steel, which has been cold reduced by at least 70 percent without subsequent annealing, to rapid heating, at a rate of at least about 20° F. per second, to a temperature in the aforesaid range, the rapidity of said heating and the selection of said temperature being such that substantial recrystallization of the cold-reduced steel occurs only after approaching said temperature, and such that after approaching such temperature the steel experiences microstructural instability which does not substantially terminate before the end of an interval of at least several seconds at said temperature, and thereupon holding the body at said temperature for providing said high ductility therein during occurrence of said microstructural instability while phase transformation and recrystallization are occurring in the steel.

10. In a method as defined in claim 9, procedure for deforming the steel body as characterized by said high ductility, comprising subjecting the body to rapid deformation by applying stress thereto while the body is held at the aforesaid temperature in the aforesaid condition of microstructural instability and progressing recrystallization.

11. A method as defined in claim 10, in which the steel is a carbon standard steel having a carbon content in the range up to about 0.5 percent, said range of temperature values extending upward from about 1,330° F., said rapid heating being effected at a rate of at least about 50° F. per second, and said holding and deforming procedure comprising holding the body at said temperature for at least several seconds to develop said condition of instability and recrystallization and then effecting said deformation while said last-mentioned condition continues.

12. A method as defined in claim 11, in which said quantity of steel is cold reduced by at least about 90 percent.

13. A method of establishing a condition of high ductility in a selected steel and of deforming said steel while characterized by said condition, said steel being of normally ferritic compositional character and having a carbon content such that as exhibited by the equilibrium phase diagram for said compositional character there is an alpha-gamma transition temperature value where upon heating to and holding at such value microstructural instability can be temporarily attained, comprising: subjecting a body of said steel to rapid heating, at a rate of at least about 10° F. per second, to the aforesaid temperature value, the rapidity of said heating and the selection of said temperature value being such that after reaching and being held at such value the steel undergoes temporary, phase-transforming, microstructural instability, creating said condition of high ductility permitting deformation at a strain rate of at least about 10 per minute, in an interval of sufficient duration to permit deformation at a strain rate not greater than about 1,000 per minute; and while holding the body at said temperature value, subjecting the body to deformation under stress to a selected extent, by applying said stress during an interval which substantially coincides with at least part of said first mentioned interval, so that said deformation is effected while the body is in a condition of high ductility.

14. A method as defined in claim 13, in which the steel has a carbon content less than 0.8 percent.

15. A method as defined in claim 14, in which the steel is a carbon standard steel having a carbon content in the range up to about 0.5 percent.

16. A method as defined in claim 14, in which the steel is a stainless steel of the 400 series.

17. A method of establishing a condition of high ductility in a selected steel and of deforming said steel while characterized by said condition, said steel being of such normally ferritic composition and carbon content that it exhibits a phase field, over a determinable elevated temperature range in the course of transfor-

mation between ferrite and austenite, which at equilibrium contains both ferrite and austenite, comprising: subjecting a body of said steel to rapid heating, at a rate of at least about 20° F. per second, to a temperature selected in the aforesaid range, the rapidity of said heating and the selection of said temperature being such that after reaching and upon being held at said selected temperature the steel undergoes ductility-enhancing microstructural instability over an interval long enough to permit large deformation by stress in such interval; and while holding the body at said selected temperature, applying stress to the body within said interval to deform it to a selected extent.

18. A method as defined in claim **17**, in which the steel is a carbon standard steel having a carbon content less than 0.8 percent and said rapid heating is effected at a rate of at least about 50° F. per second from a temperature substantially below any value for transformation of ferrite to austenite.

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