Nov. 26, 1957 W. W. DYRKACZ ET AL 2,814,563

HIGH TEMPERATURE ALLOYS

Filed July 27, 1955 2 Sheets-Sheet 1

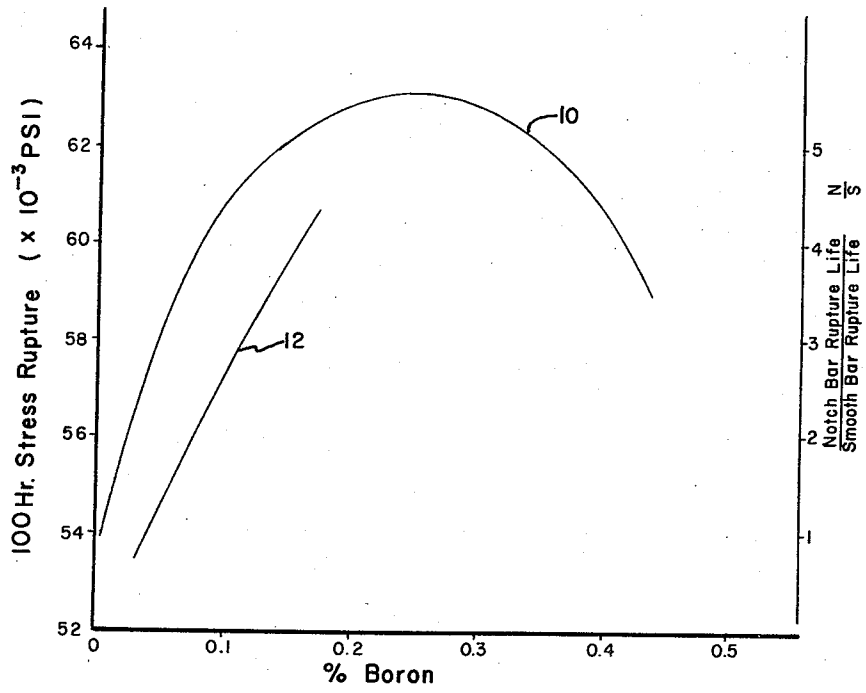



Fig. 1

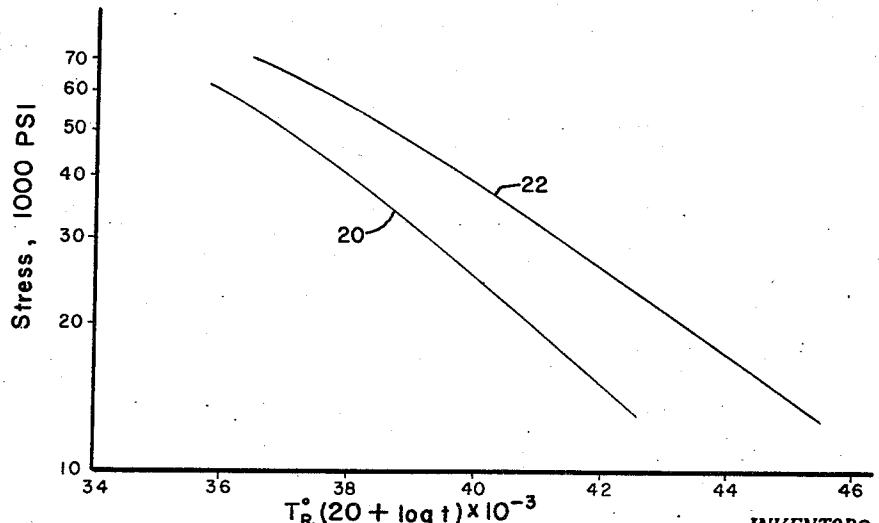



Fig. 2


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HIGH TEMPERATURE ALLOYS

Filed July 27, 1955 2 Sheets-Sheet 2

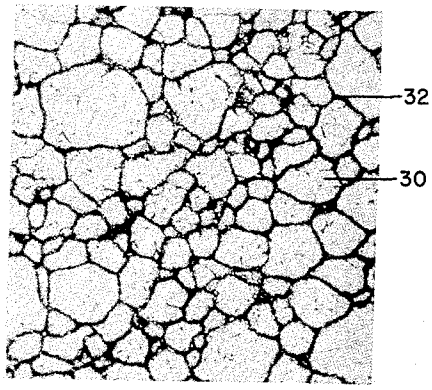



Fig. 3

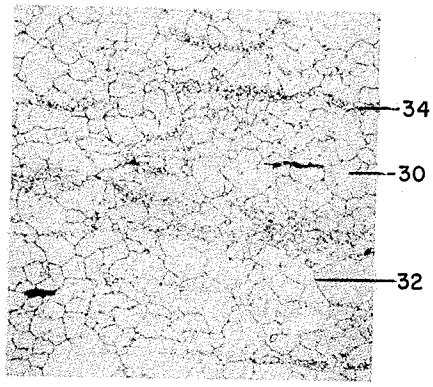



Fig. 4

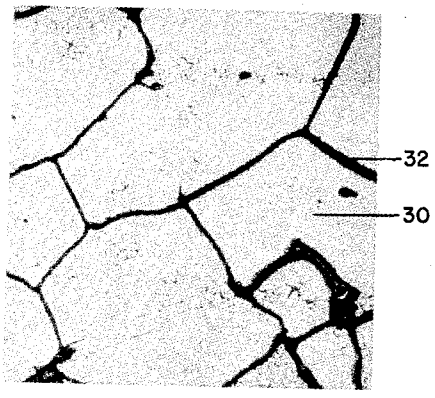



Fig. 5

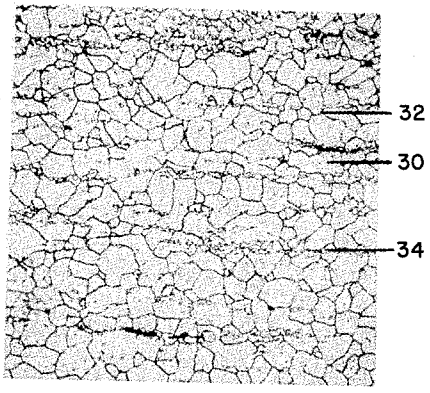



Fig. 6

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2,814,563

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2,814,563

# HIGH TEMPERATURE ALLOYS

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Application July 27, 1955, Serial No. 524,680

5 Claims. (Cl. 75—126)

This invention relates to austenitic iron base nickel-free alloys for use at elevated temperatures.

Heretofore many alloys have been made and used as turbine parts and the like where it is necessary that the alloy possess sufficient hardness, strength, and corrosion resistance to withstand the stresses and corrosive conditions encountered under operating conditions at elevated temperatures of up to 1600° F. Some prior art alloys have been made which possess the required physical properties and corrosion resistance but such alloys have been very expensive from the standpoint of their cost and manufacture. Also, such alloys, known to the trade as superalloys, usually contain high percentages of strategic alloying elements. On the other hand, other types of alloys have been made which do not contain strategic alloying elements but in most cases, the alloys free from strategic alloying elements do not have adequate strength, hardness and corrosion resistance and for this reason they cannot be used at the required operating temperatures of up to about 1600° F.

An object of this invention is to provide an austenitic iron base nickel-free alloy containing boron which is capable of withstanding high stresses at temperatures of 1200° F. up to about 1600° F.

Another object of this invention is to provide an austenitic iron base nickel-free alloy having large amounts of manganese and chromium with small amounts of carbon, silicon, molybdenum, vanadium, nitrogen and boron contained therein and which is suitable for use at high stress levels at temperatures of up to 1600° F. and which possesses a notch bar rupture life in excess of the smooth bar rupture life when tested under the same conditions of stress and temperature.

A more specific object of this invention is to produce an iron base nickel-free alloy containing critical amounts of manganese, carbon, silicon, nitrogen, chromium, molybdenum, vanadium and boron as essential alloying elements and which is suitable for use at high stresses and at temperatures of up to 1600° F.

These and other objects of this invention will become apparent to one skilled in the art when taken in conjunction with the description and the accompanying drawings in which:

Figure 1 is a graph, the curves of which illustrate the effect of boron on the stress-rupture properties of the alloy of this invention;

Fig. 2 is a graph, the curves of which illustrate the stress-rupture strength of the alloy of this invention as compared to a similar alloy which does not contain boron;

Fig. 3 is a photomicrograph taken at a magnification of 500 times of an alloy having a general composition similar to the alloy of this invention but which does not contain boron, the alloy being in the solution treated and aged condition;

Fig. 4 is a photomicrograph taken at a magnification of 500 times of an alloy of the general composition of the alloy of this invention containing boron, the alloy being in the solution treated and aged condition;

Fig. 5 is a photomicrograph taken at a magnification of 500 times of the alloy of Fig. 3 after solution heat treatment at a higher temperature and,



Fig. 6 is a photomicrograph taken at a magnification of 500 times of the alloy of Fig. 4 after solution heat treatment at a higher temperature.

In its broader aspects, the alloy of this invention comprises from about 0.20% to about 0.40% carbon, from about 16.0% to about 20.0% manganese, from about 0.15% to about 0.75% silicon, from about 11.5% to about 13.5% chromium, from about 2.0% to about 4.0% molybdenum, from about 0.60% to about 0.95% vanadium, from about 0.10% to about 0.25% nitrogen, from about 0.10% to about 0.40% boron, and the balance substantially iron with incidental impurities. It will be appreciated that where the balance is referred to as substantially all iron, incidental impurities normally found in the manufacture of steel such as copper, cobalt, nickel, phosphorus, sulfur and the like may be present in a total amount of up to 2.0% without detrimentally affecting the alloy.

Each of the elements performs a specific function within the alloy of this invention. For example, carbon imparts the requisite strength to the alloy and at least 0.20% carbon is also needed to maintain a stable austenitic structure in the alloy. Preferably, the carbon content will not exceed a maximum of 0.40% because it has been found that higher carbon contents lower the ductility of the alloy of this invention. While carbon is used as an austenitizing element, the main austenitizing element of this alloy is manganese, it being found that at least 16.0% manganese is needed in order to insure a completely austenitic structure. Not less than 11.5% chromium is used in order to impart sufficient corrosion resistance to the alloy when the alloy is used in a corrosive atmosphere and especially in atmosphere containing combustion products of leaded fuels such as the types which are used in gas turbines or other engines; however, not more than 13.5% chromium can be tolerated because it has been found that larger amounts of chromium causes serious difficulties in the manufacture and use of the alloy of this invention because of the tendency of chromium to form delta ferrite which impairs the physical properties of the alloy. Molybdenum and vanadium function to impart additional strength to the alloy of this invention by strengthening the solid solution. Vanadium also contributes to the precipitation hardening of the alloy. Nitrogen has a strengthening effect upon the alloy of the invention and contributes substantially to the austenitic stability of the alloy. It is desirable to maintain at least 0.15% silicon so that the alloy is properly deoxidized. Boron is used in the alloy of this invention in order to increase the rupture life of the alloy of this invention, boron being particularly useful in providing an alloy free from notch sensitivity. Iron comprises the balance of this alloy with the incidental impurities referred to hereinbefore.

Reference may be had to Table I in which there is listed the general range and a preferred range of the composition of the elements of the alloy of this invention.

*Table I*

[Composition.]

| Element | General range | Preferred range |
|---|---|---|
| C | 0.20–0.40 | 0.25–0.30 |
| Mn | 16.0 –20.0 | 17.0 –19.0 |
| Si | 0.15–0.75 | Up to 0.50 |
| Cr | 11.5 –13.5 | 12.0 –13.0 |
| Mo | 2.0 – 4.0 | 2.5 – 3.25 |
| V | 0.60–0.95 | 0.70–0.90 |
| $N_2$ | 0.10–0.25 | 0.15–0.22 |
| B | 0.10–0.40 | 0.15–0.25 |
| Fe | Bal. | Bal. |

The alloy of this invention may be made by any of the well known steel mill practices for example, by electric arc or induction furnace melting practice. When melted, refined and cast into ingots, the ingots may be processed in a normal steel mill manner as for example, by forging, pressing, extruding, rolling or the like to form a semi-finished mill product such as, for example, bars, billets, blooms, plate, sheet, strip and the like. The semi-finished mill product may be solution heat treated, to be referred to more fully hereinafter, and then formed into the final shape prior to aging, to be referred to more fully hereinafter, in order to obtain the optimum properties within the alloy of this invention.

Since the alloy of this invention is an austenitic alloy, it must be hardened through aging. It is preferred to solution heat treat the alloy at a temperature in the range between 1850° F. and 2100° F. for a time period ranging between 15 minutes and 4 hours in order to put the hardening elements into solid solution. In order to hold the hardening elements in the solid solution, it is preferred to cool the solution heat treated alloy rapidly for example, by quenching in water. Since the quenching of the solution heat treated alloy leaves it in a soft, ductile condition, it can be easily formed and/or machined into the desired shape. The fabricated alloy article is then preferably aged at a temperature between 1100° F. and 1500° F. for a time period ranging between 10 and 20 hours in order to obtain the optimum balance between tensile strength, yield strength, ductility and rupture life.

In order to more clearly demonstrate some of the characteristics of the alloy of this invention reference may be had to Table II which gives the chemical analysis of the alloys to be more fully referred to hereinafter with respect to Tables III and IV. It is to be noted that the alloys given in Table II are both within and outside the general range of analysis referred to in Table I so that a direct comparison can be made as to the outstanding results that are obtained with the alloys of this invention.

*Table II*

[Chemical analysis—percent by wt.]

| Heat No. | C | Mn | Si | Cr | V | Mo | $N^2$ | B | Fe |
|---|---|---|---|---|---|---|---|---|---|
| R-45 | 0.30 | 17.36 | 0.24 | 12.19 | 0.74 | 3.04 | 0.14 | ---- | Bal. |
| R-46 | 0.31 | 17.66 | 0.25 | 12.04 | 0.70 | 2.85 | 0.17 | ---- | Bal. |
| R-48 | 0.27 | 17.60 | 0.25 | 12.07 | 0.73 | 2.81 | 0.17 | ---- | Bal. |
| R-68 | 0.27 | 17.80 | 0.23 | 12.32 | 0.80 | 3.17 | 0.19 | 0.014 | Bal. |
| R-69 | 0.28 | 17.80 | 0.13 | 12.16 | 0.75 | 3.10 | 0.17 | 0.041 | Bal. |
| R-71 | 0.28 | 18.45 | 0.18 | 12.34 | 0.74 | 3.20 | 0.18 | 0.198 | Bal. |
| R-72 | 0.31 | 17.55 | 0.32 | 12.30 | 0.66 | 3.13 | 0.13 | 0.437 | Bal. |
| R-102 | 0.27 | 17.69 | 0.25 | 12.42 | 0.75 | 3.13 | 0.22 | 0.050 | Bal. |
| R-103 | 0.27 | 17.97 | 0.16 | 12.18 | 0.77 | 2.99 | 0.19 | 0.090 | Bal. |
| R-104 | 0.30 | 18.37 | 0.24 | 12.60 | 0.80 | 3.02 | 0.19 | 0.180 | Bal. |
| D-218 | 0.24 | 19.98 | 0.28 | 12.17 | 0.89 | 2.84 | 0.16 | ---- | Bal. |
| 9X-108 | 0.34 | 15.40 | 0.20 | 12.67 | 0.75 | 2.63 | 0.23 | ---- | Bal. |
| 9X-110 | 0.36 | 18.88 | 0.54 | 12.43 | 0.81 | 3.09 | 0.24 | ---- | Bal. |
| 9X-129 | 0.29 | 18.06 | 0.30 | 12.48 | 0.70 | 2.97 | 0.16 | ---- | Bal. |

Reference may be had to Table III which illustrates the stress rupture properties of some of the alloys given hereinbefore in Table II, the alloys listed in Table III



having been subjected to a solution heat treatment at a temperature of 1900° F. for a time period of one hour and then water quenched, followed by aging at 1300° F. for a time period of 16 hours and air cooled prior to being subjected to the stress-rupture test at 1200° F.

*Table III*

[Stress-Rupture Properties at 1,200° F.]

| Heat No. | Percent B | 55,000 p. s. i. | | | | 60,000 p. s. i. | | | |
|---|---|---|---|---|---|---|---|---|---|
| | | Notch bar (hrs.) | Smooth bar (hrs.) | Elong. (percent) | Red. of area (percent) | Notch bar (hrs.) | Smooth bar (hrs.) | Elong. (percent) | Red. of area (percent) |
| 9X-129 | 0 | 92 | 86 | 10 | 19.5 | -------- | 28 | 12.6 | 12.9 |
| R-68 | 0.01 | 77 | 112 | 10 | 12.8 | -------- | -------- | -------- | -------- |
| R-69 | 0.04 | -------- | -------- | -------- | -------- | 44 | 30 | 8.4 | 13.7 |
| R-102 | 0.05 | -------- | 104 | 14.8 | 20.2 | 57 | 38 | 15.0 | 29.9 |
| R-103 | 0.09 | -------- | 189 | 15.0 | 39.3 | 266 | 83 | 14.0 | 29.2 |
| R-104 | 0.18 | -------- | 292 | 30.0 | 47.6 | 559 | 146 | 18.5 | 37.4 |
| R-71 | 0.20 | 493 | 280 | 25.2 | 35.2 | -------- | 142 | 18.0 | 29.4 |
| R-72 | 0.44 | 1,481 | 160 | 21.6 | 35.8 | -------- | 90 | 22.5 | 43.2 |

It is to be noted in Table III that reference has been made to the notched bar rupture life. The test used to measure the notch bar rupture life consists of machining a "V-notch" into a standard rupture specimen at the reduced section of the specimen which, in this instance had an outside diameter of 0.275". The "V-notch" is a 60° notch positioned at the center of the reduced section, and has a 0.005" radius at the base of the notch and an 0.195" diameter across the base of the notch. The notch is produced by wet grinding and has a geometry which is such as to produce a theoretical stress concentration factor ($K_t$) of 4.2 for the notch bar in tension. This type of test is used as a measure of notch sensitivity in rupture life, the criterion used to evaluate the characteristic of freedom from notch sensitivity being that the notched bar rupture life is at least as long as a standard or smooth bar rupture life under a given stress and temperature.

From the results recorded in Table III, it can be seen by comparing alloys 9X–129, R–68, R–69 and R–102 that increasing the boron content from nil up to 0.05% while the other elements are maintained substantially constant does not produce any significant increase in the notched bar rupture life or the smooth bar rupture life when these alloys are tested at 1200° F. and at different stress levels, that is, 55,000 p. s. i. and 60,000 p. s. i. It is therefore apparent that boron contents below 0.05% do not have an appreciable effect upon the rupture life of the alloy of this invention. However, when the boron content is increased to about 0.10% in a comparable alloy such as alloy R–103 there is a significant increase in the rupture of both the smooth bar and the notched bar. The most outstanding increase in the stress rupture properties is shown when comparing alloys R–103, R–104, R–71, and R–72 wherein it is found that increasing the boron content from 0.09% to 0.44% produces a significant increase in the smooth bar rupture life and the notch bar rupture life. Thus comparing alloy R–102 with R–104 it is seen that the smooth bar rupture life has increased from 104 hours when the alloy contains 0.05% boron to 292 hours when the alloy contains 0.18% boron. A significant increase is also noted in the notched bar rupture life over and above that of the smooth bar rupture life as illustrated by the results recorded for alloy R–71 which contains 0.20% boron and as tested at 55,000 p. s. i. and 1200° F. In particular, alloy R–71 has a ratio of notch bar rupture life to smooth bar rupture life of about 1.75.

In order to more clearly demonstrate the effect of boron on the stress rupture properties of the alloys of this invention reference may be had to the graph of Fig. 1 in which curve **10** illustrates the effect of boron on the stress to produce rupture in 100 hours and curve **12** which illustrates the effect of boron on the ratio of

notch bar rupture life to smooth bar rupture life, designated as N/S ratio. Although there is a significant increase in the stress required to produce rupture in the smooth bar rupture tests in 100 hours when boron is present in amounts up to about 0.10%, outstanding results are obtained where the boron content is present in an amount of from 0.10% to 0.40% as shown by curve **10**. Referring to curve **12**, it is seen that by increasing the boron content to about 0.10% the N/S ratio is increased to greater than 1. It has been found that when the boron content is maintained within the range between 0.10% and 0.40% the N/S ratio will always be greater than 1 and in some cases 4 or greater, thus illustrating that the alloy of this invention is particularly free from notch sensitivity and has the capacity to withstand and redistribute stresses occurring in notched areas. The optimum boron content for optimum rupture life and N/S ratio is about 0.20%.

It is also apparent from Table III that there is an associated increase in ductility in the alloy of this invention when greater than 0.10% boron is contained therein. Comparing the test results obtained on alloy R–102 with the test results of alloy R–104, it is seen that the percentage elongation has about doubled by increasing the boron content from 0.05% to 0.18%. An associated increase in the reduction of area is also noted. From Table III, it is also seen that at stresses of both 55,000 p. s. i. and 60,000 p. s. i. at 1200° F., there is no significant increase in the time to produce rupture in either the notched bar rupture test or in the smooth bar rupture test when the boron content is below about 0.10%. However, upon increasing the boron content to within the range given hereinbefore in Table I, that is between 0.10% and 0.40%, a substantial increase is noted in the rupture life of the alloy of this invention. Comparing alloy R–102 with alloy R–104, it is seen that increasing the boron content from 0.05% to 0.18% has produced an increase in the notch bar rupture life of from 57 hours to 559 hours and in the smooth bar rupture life from 38 hours to 146 hours. It is apparent that the boron content of the alloy of this invention must be maintained within the range 0.10% to 0.40% in order to obtain the optimum properties in the alloy of this invention.

In order to more clearly demonstrate the overall advantages of time, temperature and stress produced by the addition of boron, references may be had to Table IV which illustrates a number of alloys of the same general composition, with the exception of the boron content, which were subjected to the smooth bar stress-rupture test at different temperatures and different stresses.

*Table IV*

[Stress-Rupture Properties.]

| Heat No. | Temp. (° F.) | Stress (p. s. i.) | Rupture time (hrs.) | Elong. (percent) | Red. of area (percent) | L-M parameter |
|---|---|---|---|---|---|---|
| R-104 [1] | 1,200 | 55,000 | 798 | 29.3 | 58.3 | 38.0 |
| R-104 [1] | 1,200 | 60,000 | 516 | 20.7 | 52.0 | 37.7 |
| R-104 [1] | 1,200 | 65,000 | 198 | 16.3 | 45.3 | 37.0 |
| R-104 [1] | 1,200 | 70,000 | 43 | 19.6 | 48.6 | 35.9 |
| R-71 [1] | 1,350 | 30,000 | 542 | 23.8 | 47.6 | 41.2 |
| R-71 [1] | 1,350 | 40,000 | 75 | 27.8 | 36.6 | 39.6 |
| R-71 [1] | 1,500 | 12,000 | 1,121 | 13.5 | 45.7 | 45.2 |
| R-71 [1] | 1,500 | 18,000 | 426 | 22.3 | 45.0 | 44.4 |
| R-71 [1] | 1,500 | 22,000 | 80 | 33.5 | 70.3 | 43.2 |
| 9X-129 | 1,200 | 60,000 | 28 | 12.6 | 12.9 | 35.6 |
| 9X-129 | 1,200 | 55,000 | 64 | 15.5 | 20.2 | 36.1 |
| 9X-110 | 1,200 | 55,000 | 90 | 8.3 | 19.5 | 36.5 |
| 9X-108 | 1,200 | 55,000 | 70 | 10.9 | 18.2 | 36.3 |
| R-48 | 1,200 | 55,000 | 90 | 9.4 | 15.2 | 36.5 |
| R-46 | 1,200 | 55,000 | 65 | 10.7 | 25.6 | 36.3 |
| R-45 | 1,200 | 55,000 | 97 | 8.7 | 14.7 | 36.6 |
| 9X-129 | 1,200 | 50,000 | 265 | 8.0 | 20.4 | 37.2 |
| 9X-110 | 1,200 | 50,000 | 80 | 26.3 | 30.0 | 36.4 |
| R-48 | 1,200 | 50,000 | 164 | 13.6 | 17.8 | 36.8 |
| D-218 | 1,200 | 50,000 | 298 | 22.3 | 44.2 | 37.3 |
| 9X-110 | 1,350 | 25,000 | 119 | 24.5 | 35.2 | 40.0 |
| 9X-129 | 1,350 | 25,000 | 147 | 24.2 | 30.0 | 40.1 |
| 9X-110 | 1,500 | 12,000 | 28 | 21.7 | 23.9 | 42.1 |
| 9X-129 | 1,500 | 12,000 | 83 | 7.5 | 13.1 | 43.0 |

[1] Alloys contain boron.



It is to be noted in Table IV that a column has a heading of L–M parameter. The L–M parameter is given by the equation $T_R°(20+\log t)\times 10^{-3}$ in which $T_R°$ is the absolute temperature and log $t$ is the time in hours. This equation can be plotted against the applied stress to obtain a master rupture curve, which will be more fully explained hereinafter. From Table IV, it is seen that the alloys of this invention containing boron are particularly adapted for use at elevated temperatures of up to 1500° F. or greater as evidenced by the tests on alloys R–104 and R–71 at different temperatures and different stress levels. The other alloys listed in Table IV are similar to alloys R–104 and R–71 but do not contain boron. The alloys which contain boron are outstanding as compared to the alloys which do not contain boron.

Referring now to the graph of Fig. 2 which is a plot of the stress vs. the L–M parameter, curve **20** graphically illustrates the results obtained on alloys which do not contain boron as tested at different temperatures and at different stresses and curve **22** is a graphic illustration of the results obtained on the alloys of this invention having the same general composition and containing boron as tested at different stresses. It is apparent that the displacement of curve **22** upwardly and to the right of curve **20** is such as to show a far superior result for the alloys of this invention containing boron as compared to the results obtained on similar alloys which do not contain boron. Thus by means of the master rupture curve **22** of Fig. 2, for any point on curve 22, a corresponding stress and parameter value can be read and the parameter solved by a means of the equation given hereinbefore. Thus for a given stress the corresponding parameter value of the alloy without boron will be considerably lower than the alloy containing boron. Conversely with a given parameter value the alloy containing boron will sustain a higher stress than the alloy without boron. Translated into terms of time and temperature, it is apparent from the master rupture curve that the alloy containing boron will sustain higher stress for any given temperature and time, or a higher temperature for any given stress and time, or a longer time for any given stress and temperature than will a corresponding alloy without boron. From Fig. 2, it is apparent that the alloy of this invention possesses far superior characteristics when the boron content is maintained within the range given in Table I.

Referring now to Figs. 3 through 6, the effect of boron on the microstructure of the alloy of this invention upon heat treatment is readily seen. In Fig. 3, which is a photomicrograph taken at a magnification of 500 times of alloy 9X–129 after having been solution treated at 1900° F. for one hour and aged at 1300° F. for 16 hours, it is noted that very large grains of austenite **30** have developed during the solution heat treatment. It is also apparent that a heavy concentration of grain boundary precipitate **32** has also precipitated during the heat treatment. Further, the austenite **30** grains are non-uniform in size, it being noted that the grain size ranges between ASTM 5 and 8. Referring now to Fig. 4, which is a photomicrograph taken at a magnification of 500 times of alloy R–104 containing 0.18% boron as solution treated at 1900° F. for one hour and aged at 1300° F. for 16 hours, the alloy of Fig. 4 is seen to have substantially uniform grains of austenite **30** corresponding to a grain size of about ASTM 8. It is also to be noted that the addition of 0.18% boron has resulted in considerably less grain boundary carbide precipitate **32**. Since there is a much lower susceptibility to grain growth in the alloy containing boron at the solution heat treatment temperature, this in effect makes possible a marked improvement in the rupture strength while at the same time maintaining the fine grain size. The small banded phase **34** occurring in this alloy is associated with the presence of boron in this as well as other alloys containing boron.

By increasing the solution heat treatment temperature up to 2050° F., the alloy of Fig. 3 which does not contain boron develops an extremely large grain as illustrated in Fig. 5. It is seen that the grain size of the austenite **30** in Fig. 5 has increased to an ASTM grain size of 3–6 and is associated with a particularly heavy concentrate of grain boundary carbide precipitate **32**. This has a definite weakening effect on the rupture properties of such an alloy. On the other hand, an increase in the solution heat treatment temperature of up to 2050° F. does not affect the fine grain size of the alloy of Fig. 4, which contains 0.18% boron, as is evidenced by the photomicrograph illustrated in Fig. 6. The austenite **30** shown in Fig. 6 has maintained substantially the same ASTM grain size of about 8 found in the microstructure shown in Fig. 4. Here also there is much less grain boundary carbide precipitate **32** than that as shown in Fig. 5. It is therefore apparent that the boron content has a definite and substantial effect on maintaining a substantially uniform austenite grain size when the alloy is heat treated within the range given hereinbefore. The refinement in the austenite grain size and the smaller amounts of grain boundary carbide precipitates resulting from the boron additions are important factors in the improvement in the rupture properties obtained.

As stated hereinbefore, the alloy of this invention is an age hardening alloy. The alloy may be hot worked to the component part. In practice, the alloy is preferably subjected to solution heat treatment at a temperature between 1850° F. and 2100° F. for a time of 15 minutes to 4 hours after which it is quenched as by water quenching to place the hardened elements into solid solution whereby the solution treated and quenched alloy is in a soft ductile condition. The alloy can then be readily fabricated by machining or grinding into the component part or a desired shape of the article of manufacture to be formed therefrom. The alloy should then be aged at 1100° to 1500° F. for 10 to 20 hours, before placing in service. It may also be machined in the solution treated plus aged condition. For certain applications requiring very high hardness, tensile and yield strengths, the alloy may be cold worked up to around 50%. This is usually done in the solution treated condition and may or may not be followed by aging. Hot-cold working in the temperature range of 1000° F. to 1500° F. may also be used to improve properties.

In producing the alloys of this invention no special skills nor apparatus are required. Further, the alloys are characterized by the absence of strategic alloying elements, the ability to withstand high stresses at elevated temperatures of up to 1500° F. or greater for long periods of time and especially its freedom from notch rupture sensitivity. The alloys of this invention can be readily reproduced by any one skilled in the art.

We claim:

1. An age hardenable austenitic iron base alloy for use at high stress levels and at temperatures of up to 1600° F. and consisting essentially of from about 0.20% to about 0.40% carbon, from about 16.0% to about 20.0% manganese, from about 0.15% to about 0.75% silicon, from about 11.5% to about 13.5% chromium, from about 2.0% to about 4.0% molybdenum, from about 0.60% to about 0.95% vanadium, from about 0.10% to about 0.25% nitrogen, and from about 0.10% to about 0.40% boron, the alloy being characterized by having a ratio of notch bar rupture life to smooth bar rupture life of greater than 1.

2. An age hardenable austenitic alloy for use at high stress levels and at temperatures of up to 1600° F. and consisting of from 0.20% to 0.40% carbon, from 16.0% to 20.% manganese, from 0.15% to 0.75% silicon, from 11.5% to 13.5% chromium, from 2.0% to 4.0% molybdenum, from 0.60% to 0.95% vanadium, from 0.10% to 0.25% nitrogen, from 0.10% to 0.40% boron and the balance iron with incidental impurities.

3. An age hardenable austenitic alloy for use at high stress levels and at temperatures of up to 1600° F. and consisting of from 0.25% to 0.30% carbon, from 17.0% to 19.0% manganese, up to 0.50% silicon, from 12.0% to 13.0% chromium, from 2.50% to 3.25% molybdenum, from 0.70% to 0.90% vanadium, from 0.15% to 0.22% nitrogen, from about 0.15% to 0.25% boron, and the balance iron with incidental impurities.

4. An age hardenable austenitic alloy for use at high stress levels and at temperatures of up to 1600° F. and consisting of about 0.28% carbon, about 18.4% manganese, about 0.18% silicon, about 12.3% chromium, about 3.2% molybdenum, about 0.74% vanadium, about 0.18% nitrogen, about 0.20% boron, and the balance iron with incidental impurities.

5. An age hardened article of manufacture for use at elevated temperatures and at high stress levels, comprising an alloy consisting of from 0.20% to 0.40% carbon, 16.0% to 20.0% manganese, 0.15% to 0.75% silicon, 11.5% to 13.5% chromium, 2.0% to 4.0% molybdenum, 0.60% to 0.95% vanadium, 0.10% to 0.25% nitrogen, 0.10% to 0.40% boron, and the balance iron with incidental impurities, the alloy being formed to the predetermined shape of the article and being in the age hardened condition resulting from quenching from a solution heat treatment at a temperature in the range between 1850° F. and 2100° F. and aging at a temperature in the range between 1100° F. and 1500° F. for a time period ranging between 10 and 20 hours, the alloy being further characterized by having a ratio of notch bar rupture life to smooth bar rupture life of greater than 1.

**References Cited** in the file of this patent

UNITED STATES PATENTS

| | | |
|---|---|---|
| 2,562,854 | Binder | July 31, 1951 |
| 2,686,116 | Schempp | Aug. 10, 1954 |

FOREIGN PATENTS

| | | |
|---|---|---|
| 884,908 | France | Aug. 31, 1943 |

OTHER REFERENCES

Stainless Iron and Steel, vol. 1, pages 138 and 139. Published in 1951 by Chapman and Hall, London, England.