

March 27, 1973

C. D'A. HUNT
IRON BASE ALLOYS HAVING LOW LEVELS OF
VOLATILE METALLIC IMPURITIES

3,723,101

Filed June 15, 1970

2 Sheets-Sheet 1

FIG. 1

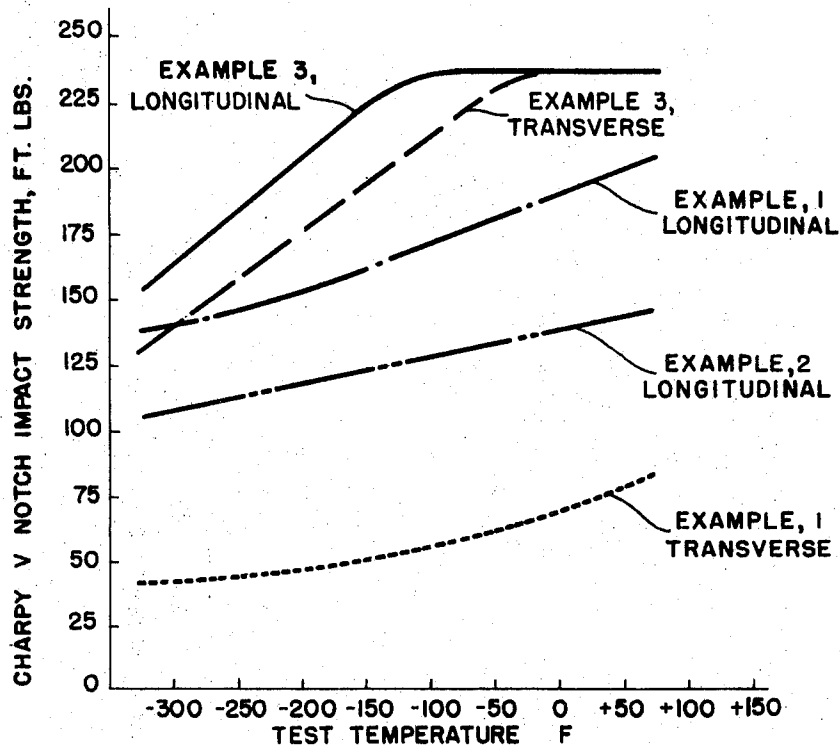
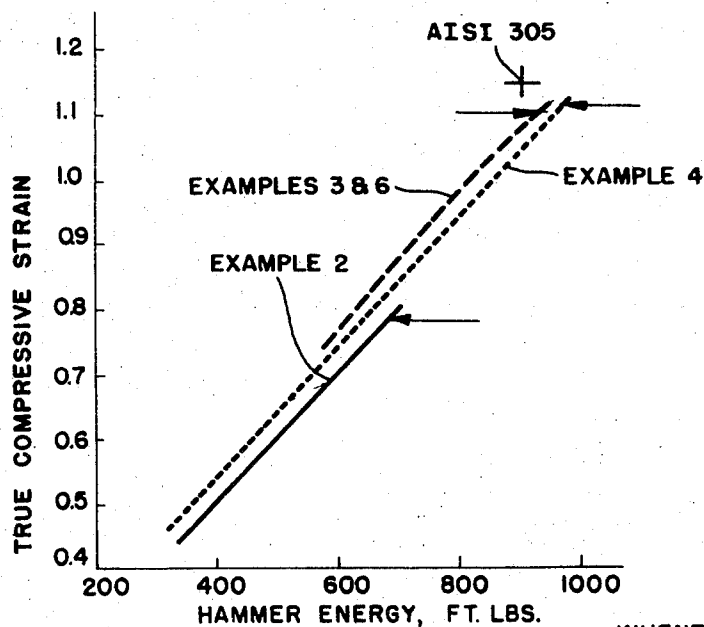


FIG. 2



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2 Sheets-Sheet 2

FIG.3

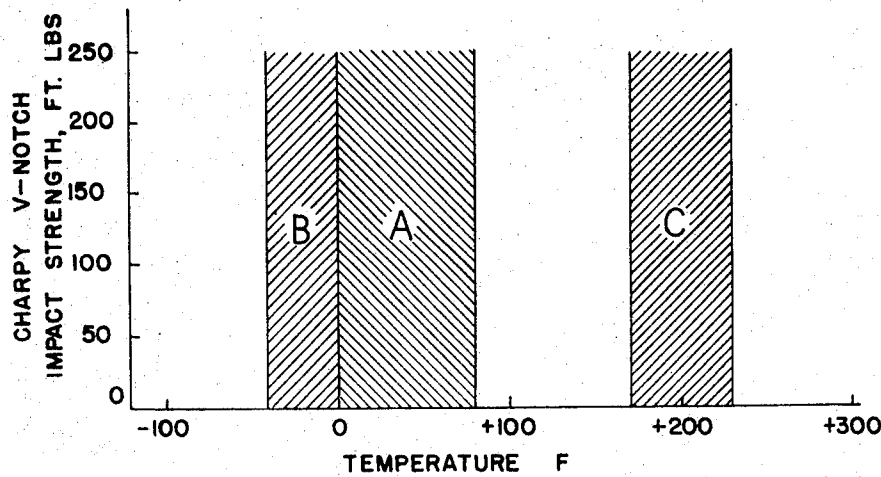
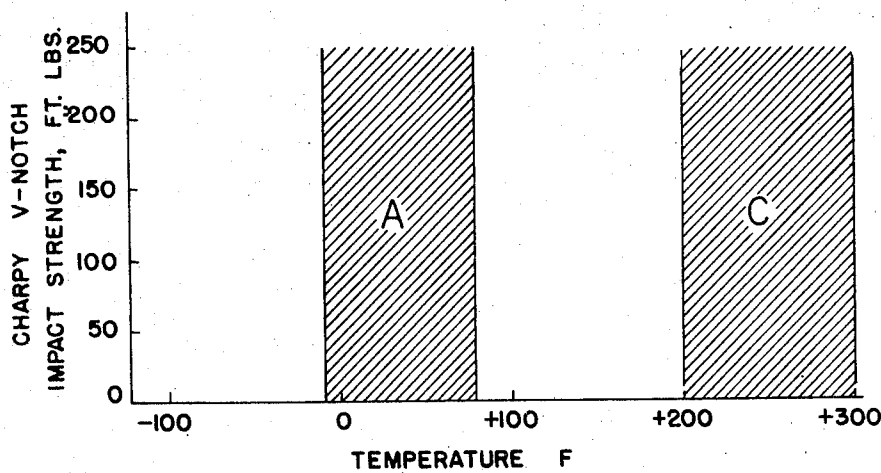


FIG.4



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1

3,723,101

IRON BASE ALLOYS HAVING LOW LEVELS OF VOLATILE METALLIC IMPURITIES

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Continuation-in-part of application Ser. No. 722,640, Apr. 19, 1968, which is a continuation-in-part of application Ser. No. 636,666, May 8, 1967, both now abandoned.
This application June 15, 1970, Ser. No. 46,443

Int. Cl. C22c 39/14, 39/20

U.S. Cl. 75—126 R

13 Claims

ABSTRACT OF THE DISCLOSURE

Iron base alloys having substantially improved physical and mechanical properties containing between about 4 and about 40% chromium, between 0 and about 15% nickel, low levels of carbon and nitrogen such that the carbon and nitrogen are in solid solution at room temperature, not more than about 6 p.p.m. total of the insoluble volatile metallic impurities lead, bismuth, cadmium, sodium, potassium, silver, calcium, magnesium, and barium, and not more than about 20 p.p.m. of the soluble volatile metallic impurities zinc and antimony.

This invention relates generally to iron base alloys, and more particularly it relates to iron-chromium alloys having substantially improved physical properties. This application is a continuation-in-part of application Ser. No. 722,640 filed Apr. 19, 1968, now abandoned, which was a continuation-in-part of application Ser. No. 636,666 filed May 8, 1967, now abandoned.

Iron-chromium alloys containing in excess of about four percent chromium are commonly referred to as stainless steel. As the name implies, a principal use of stainless steel is in corrosive environments. The corrosion protection in stainless steel is imparted mainly by the presence of chromium, and increasing chromium content generally provides increasing corrosion protection. Other alloying agents, principally nickel, are added to stainless steel to obtain desired properties, and the type and amount of alloying agents must be carefully controlled to obtain the desired properties. As used herein, the term "iron-chromium alloy" is intended to include those alloys commonly referred to as stainless steel and, in addition, includes iron-chromium alloys which are not commonly referred to as stainless steel.

The crystal structure of iron-chromium alloys has a marked effect upon the properties of the alloy. Three different crystal structures, austenitic, ferritic and martensitic, or mixtures thereof may be formed upon cooling to room temperature depending upon the amount of chromium, nickel and other alloying agents. Nickel, cobalt, manganese, copper, carbon and nitrogen are referred to as austenite formers, because the presence of these materials in stainless steel tends to stabilize the retention of an austenitic crystal structure. Chromium, silicon, molybdenum, tungsten, vanadium, tin, columbium, phosphorus, aluminum and titanium are known as ferrite formers, since these materials tend to stabilize ferritic grain structure in steels. Martensitic crystal structures are formed by the presence of appropriate combinations of both ferritic formers and austenitic formers in the stainless steel. By balancing the relative amounts of austenite formers and the ferrite formers in the alloy, in accordance with generally known principles, a given iron-chromium alloy can frequently be suitably controlled so that it will be austenitic, ferritic or martensitic, or a combination of these types.

Although iron-chromium stainless steel alloys have desirable corrosion resistance, the addition of the alloying agents necessary to obtain corrosion resistance causes stainless steel alloys to be more difficult to fabricate than

2

normal mild or carbon steel. In this connection, austenitic, ferritic and martensitic iron-chromium alloys are all susceptible to surface disruptions, in the form of edge cracking, and streaking when worked to high reduction in areas and also become work hardened. Some ferritic alloys are susceptible to a particularly undesirable surface disruption generally referred to as "surface roping." In order to remedy these defects, commercial practices involve several anneals and surface conditioning operations in reducing slab stock to sheet. Ferritic iron-chromium alloys are also notch sensitive, particularly when they contain in excess of 20% chromium, requiring careful cold working conditions to prevent brittle fracture. Austenitic iron-chromium alloys have a higher rate of work hardening than do the ferritic or martensitic alloys and are susceptible to transformation upon cold working to martensite.

Because of the widespread use of stainless steel alloys in environments where surface corrosion, and surface decoration are important, extensive investigations of stainless steel have been made. The principal aim of these investigations has been to improve corrosion resistance, generally through the development of new combinations of alloying agents. Little commercial effort has been directed to improving the fabricability of existing alloys, while retaining the desired corrosion resistance.

It is generally known that stainless steels which have low interstitial content, that is, low carbon and nitrogen content, have improved corrosion resistance, and are generally easier to fabricate at given tensile strengths than are similar stainless steels containing higher carbon and nitrogen contents. The improved fabricability which is achieved at low interstitial content is believed to result from the relative freedom from carbides and nitrides at the grain boundaries.

Other materials are present in iron-chromium alloys as impurities, generally at low levels, e.g., below about 0.1 percent by weight. One class of impurities present in all commercially available stainless steels are the "volatile metallic impurities" which are defined herein as lead, bismuth, cadmium, sodium, potassium, silver, calcium, magnesium, barium, zinc and antimony. The volatile metallic impurities may be divided into two classes, lead, bismuth, cadmium, sodium, potassium, silver, calcium, magnesium, and barium which are essentially insoluble in iron, and zinc and antimony which are slightly soluble.

Conventional iron-chromium alloys contain these impurities in varying amounts, depending upon the raw material source and the manufacturing process. Of the total volatile metallic impurities present in commercial iron-chromium alloys, there is generally between about 25 and 75 p.p.m. of the insoluble volatile metallic impurities, and between about 100 and 200 p.p.m. of the soluble volatile metallic impurities. Although the presence of these impurities in iron-chromium alloys was heretofore not considered to affect the corrosion resistance and physical properties of the alloys, their presence in iron-chromium alloys has been generally overlooked, particularly in respect to the fabricability of the alloys. Further, electron beam microprobe analysis of fractures in bearing steels shows significant zinc concentrations in the region of the fracture indicating that the presence of soluble volatile metallic impurities in iron-chromium alloys may be a contributing factor in causing failure of the alloys.

The relative ease of fabrication of a metal may be determined by the ease in which it may be deformed, for example, by rolling or swaging. Another indication of the ease of fabrication is the ductility and the toughness of the metal. One method of determining the toughness of iron-chromium alloys is to determine the ability of the alloy to withstand impact loading. In addition, the impact strength which is apparently related to the ability to withstand high stress, low cycle strains, is important

in environments where shock loads or violent impacts might be encountered.

The impact strengths of commercial austenitic iron-chromium alloys do not change drastically over wide temperature ranges, for example, from -320°F. to 200°F. , and the longitudinal impact strength of commercial austenitic alloys is generally sufficient for most purposes. However, the transverse impact strength of commercial austenitic alloys is substantially less than the impact strength in the longitudinal direction. Similarly, the transverse impact strengths of commercial martensitic and ferritic alloys are substantially less than the longitudinal impact strength.

In addition, the ferritic alloys typically exhibit a drastic reduction in both longitudinal and transverse impact strength at temperatures around room temperature or higher (the impact transition temperature) and at temperatures below room temperature have very low impact strengths and are very brittle. Thus, the ferritic alloys cannot be employed where they might be subjected to impact loading at temperatures below the transition temperature, i.e., much below room temperature. This has eliminated the use of ferritic iron-chromium alloys in many applications where, although the corrosion resistance of the ferritic alloy is sufficient for the environment, the possibility of impact loading requires the use of the relatively more expensive austenitic alloys. This is particularly true in welded structures, where the ferritic welds are quite susceptible to cracking due to poor ductility at room temperature, as well as carbide segregation in the weld region which occurs during heating.

It is an object of the invention to provide iron-chromium alloys having improved mechanical and physical characteristics. A further object of the invention is to provide iron-chromium alloys of improved ductility, thereby permitting easier fabrication of the alloys. Another object is to provide ferritic iron-chromium alloys which exhibit superior ductility and resistance to impact loading at room temperature. A still further object is to provide iron-chromium alloys having improved impact strengths, and particularly improved impact strengths in the transverse direction.

These and other objects of the invention will be more readily understood from the following detailed description and from the drawings of which:

FIG. 1 is a graph depicting impact strength versus temperature for two commercial austenitic stainless steels and an austenitic iron-chromium alloy of one embodiment of the invention.

FIG. 2 is a graph depicting true compressive strain versus hammer energy for several austenitic alloy embodiments of the invention.

FIG. 3 is a graph depicting the longitudinal impact transition temperature range for a commercial cold worked and annealed ferritic stainless steel and the longitudinal transition temperature, both unworked and cold worked and annealed, of an embodiment of a ferritic iron-chromium alloy of the invention.

FIG. 4 is a graph, similar to FIG. 3 depicting the transverse impact temperature range of the commercial ferritic alloy and the ferritic alloy of FIG. 3.

Very generally, the present invention is directed to iron-chromium alloys containing between about 4 percent and about 40 percent chromium, between about 0 percent and about 14 percent nickel, carbon and nitrogen in amounts such that the carbon and nitrogen will be in solid solution at room temperature, not more than about 6 p.p.m. of insoluble volatile metallic impurities, i.e., lead, bismuth, cadmium, sodium, potassium, silver, calcium, magnesium, and barium, and not more than about 20 parts per million soluble volatile metallic impurities, i.e., zinc and antimony. The iron-chromium alloys have substantially improved physical and mechanical properties, and have increased impact strengths, as compared to

conventional iron-chromium alloys. Austenitic iron-chromium alloys have increased impact strengths at all temperatures, e.g., at room temperature and above down to liquid nitrogen temperatures (-320°F.). The transverse impact strength of the austenitic and martensitic iron-chromium alloys in accordance with the invention is more than double the transverse impact strength of similar conventional commercial alloys containing volatile metallic impurities. Ferritic alloys in accordance with the invention have substantially improved transverse and longitudinal impact strengths at elevated temperatures (200°F. – 300°F.) and, in addition, have transition temperature ranges generally below room temperature. Desirably, the total carbon plus nitrogen content is below 250 p.p.m., and generally between 70 and 100 p.p.m. Larger amounts of carbon plus nitrogen can be tolerated in the austenite alloys, e.g., up to 600 p.p.m., due to their greater solubility in austenitic alloys than in ferritic or martensitic alloys.

The substantially improved impact strengths, in both the transverse and longitudinal directions, evidences the substantially improved fabricability and workability of the described alloys as compared to similar conventional iron-chromium alloys. It is believed, although the invention is not considered to be limited thereto, that the improved physical and mechanical properties of the described alloys is achieved as the direct result of the relatively complete absence of volatile metallic impurities, as well as the low carbon plus nitrogen content of the alloys. As indicated, there are two classes of volatile metallic impurities which are present in iron-chromium alloys, and in accordance with the disclosed invention, the alloys contain less than 6 p.p.m. of those volatile metallic impurities which are defined as "insoluble" and less than 20 p.p.m. of those volatile metallic impurities which have been defined as "soluble." Preferably, the insoluble volatile metallic impurities are present in an amount of not more than about 2 p.p.m., and the soluble volatile metallic impurities are present in an amount of not more than about 10 p.p.m.

It is to be understood that the terms "insoluble" and "soluble" as used herein, are relative terms and are intended to denote the difference between the two classes of volatile metallic impurities. These volatile metallic impurity substances tend to be concentrated at, or in the vicinity of, the grain boundaries of the alloy, or at the dendritic interfaces within the grains themselves, and it is believed that it is the concentration of these impurities at or in the vicinity of these regions which causes the iron-chromium alloys to have lesser mechanical properties in the transverse direction. The insoluble volatile metallic impurities, lead, bismuth, cadmium, sodium, potassium, silver, calcium, magnesium, and barium are essentially completely insoluble in iron-chromium alloys at all concentrations, and the total content of these impurities in the iron-chromium alloy tends to be concentrated at the last-to-solidify regions as precipitates. It is therefore quite important that the alloys be essentially completely free of, i.e., contain less than 6 p.p.m. of the insoluble volatile metallic impurities.

Zinc, and antimony, referred to herein as "soluble" volatile metallic impurities are somewhat more soluble in iron-chromium alloys than are the insoluble volatile metallic impurities and are not as deleterious as the insoluble impurities because they are not present as precipitates. However, the soluble volatile metallic impurities also tend to be concentrated at the grain boundaries, or at the intragranular dendritic interfaces, and it is believed that their presence at or in the vicinity of these regions may cause the alloys to have lesser mechanical properties in the transverse direction. Accordingly, the iron chromium alloys should be substantially free of the soluble metallic impurities, i.e., contain less than about 20 p.p.m.

It is to be understood that measurement of the extremely low levels of volatile metallic impurities in the

iron chromium alloys is, at best, a very difficult analytical problem. The limits of the volatile metallic impurities set forth herein are considered to be approximate in nature, rather than absolute, and alloy compositions differing slightly from the alloys set forth herein, but having the improved impact strengths and fabricability of the described alloys, are considered to be within the scope of the invention.

As indicated, iron-chromium alloys, and particularly stainless steel alloys, are generally identified by their

loys prepared in accordance with conventional manufacturing techniques, e.g., basic oxygen furnace, electric arc furnace, or vacuum induction furnace processes. A particular class of austenitic alloys to which the invention is directed are those comprising between about 14 percent and about 20 percent chromium and between about 8 percent and about 14 percent nickel. It has been determined that the austenitic alloys of the invention may be readily cold and hot worked and welded with better results than conventional austenitic alloys.

TABLE I

Ex.	Percent by weight										Parts per million					Volatile metallic impurities
	Cr	Ni	Al	Mn	Mo	Si	Ti	V	W	Other	C	N	O	S	P	
1.....	17.2	10.2	NA	<0.3	.14	.55	<.001	<.01	<.1		220	NA	NA	70	180	50/150.
2.....	18.3	9.1	0.1	1.54	.3	.4	NA	NA	NA		650	NA	NA	240	270	>300 total.
3.....	18.6	9.7	.003	NA	.10	.55	<.001	<.04	<.1		140	110	9	80	170	<6/20.
4.....	17.6	9.8	.003	NA	.12	.55	.082	<.04	<.1		170	50	9	80	170	<6/20.
5.....	16.8	10.0	.003	NA	.11	.53	<.001	<.04	<.1	.16Zr	190	NA	NA	60	170	<6/20.
6.....	18.5	9.9	.003	0.34	.12	.65	<.001	<.04	<.1		200	98	12	70	180	<6/20.
7.....	20.5	10.3	.003	<.03	.11	106	<.001	<.04	<.1		60	35	12	60	180	<6/20.
8.....	16.4	13.4	NA	NA	2.2	.08	NA	NA	NA		100	NA	NA	90	110	<6/20.
9.....	17.3	13.0	NA	NA	2.0	.08	NA	NA	NA		100	NA	NA	90	350	<6/20.
10.....	16.3	9.1	.003	NA	.1	.08	.06	<0.4	<.1		100	NA	NA	60	340	<6/20.

crystalline configuration, austenitic, ferritic, martensitic or combinations thereof. Many stainless steel alloys are mixtures of the different crystal configurations, and the properties of the alloys are, to a large extent, dependent upon the relative proportion of the different crystal configurations present in the alloy. Because of these different properties, and the fact that the end uses of heretofore available commercial alloys were dependent upon the crystal configurations of the alloys, each of the three types of crystal configuration is discussed separately. However, it is to be understood that in many instances the alloys will contain portions of the other crystal configurations, and the designation of an alloy as austenitic, ferritic or martensitic is intended only to designate the predominant crystal configuration present in the alloy.

AUSTENITIC ALLOYS

The austenitic iron-chromium alloys are those alloys which contain sufficient amounts of austenite stabilizers to cause the alloy to have an austenite crystal configuration at room temperature. The principal austenite stabilizer is nickel, and substantially all commercial austenitic iron-chromium alloys contain nickel, usually in an amount between about 6 percent and about 12 percent by weight and chromium, generally in an amount between about 17 percent and about 22 percent. Other austenite stabilizers, such as carbon, nitrogen, manganese, cobalt and copper may replace a portion of the nickel. In the AISI numbering system, the 300 Series stainless steels are commonly austenitic. The presence of nickel in stainless steel iron-chromium alloys is principally to obtain sufficient ductility that the stainless steel alloy can be fabricated into desired shapes, and to permit welding of the alloy.

The austenitic alloys described herein, and manufactured in accordance with the disclosed method, have substantially improved physical and mechanical characteristics when compared against conventional austenitic al-

loys. Examples 1 and 2 are commercially available austenitic stainless steels which correspond to AISI grade 304. Example 1 is a vacuum induction melt, representing the latest in commercially practicable processing, and Example 2 is an electric furnace melt. Examples 3 to 10 have normal compositions comparable to commercially available AISI 300 Series stainless steels, except that Examples 3 to 10 each have less than about 6 p.p.m. of insoluble volatile metallic impurities, and less than about 20 p.p.m. soluble volatile metallic impurities. Further, the alloys of Examples 3 to 10 have carbon and nitrogen contents such that the carbon and nitrogen is in solid solution at room temperature, i.e., below about 300 p.p.m. total carbon plus nitrogen. Accordingly, the described alloys have somewhat lower carbon plus nitrogen content than commercial stainless steels of equivalent nominal composition.

The substantially improved properties of the austenitic iron-chromium alloys described herein is illustrated by the graph of FIG. 1 in which there is plotted Charpy V-notch impact strengths (ASTM No. A370-67), in both the transverse and longitudinal directions, of Examples 1, 2 and 3. It can be readily seen that at room temperature the described austenitic stainless steel of Example 3 has substantially higher transverse and longitudinal impact strengths than does either of the commercial stainless steels of Examples 1 and 2. Further, the transverse impact strength of Example 3 is equal to the longitudinal impact strength at room temperature and at -320° F., exceeds about 100 foot pounds. The substantial improvement in the impact strength of Example 3 will also be seen from the fact that at -320° F. the transverse impact strength is greater than the longitudinal impact strength of the commercial stainless steels of Example 2, and is substantially equal to the longitudinal impact strength of Example 1.

TABLE II

Example	Charpy V-notch impact strength at room temp., ft. lb.		Charpy V notch impact strength at -320° F. ft. lb.		Reduction in area, percent		R.A.T. ratio
	Long.	Tran.	Long.	Tran.	Long.	Tran.	
1.....	205	80	140	60	77	57	.74
2.....	150		108				
3.....	235	235	166	128	84	80	.95
4.....	230	215	134	107	82	78	.96
5.....	210	75	85	30	79	76	.96
6.....	235	215	158	130	82	76	.93
7.....	230	230	140	100	86	83	.97
8.....	215	110	170	65	75	72	.96
9.....	180	120	130	40	75	73	.97
10.....	160	100	90	40	81	80	.99

There is set forth in Table II the longitudinal and transverse impact strength at room temperature and at -320° F. of Examples 1 to 10. It will be seen that the described austenitic alloys containing less than about 6 p.p.m. of insoluble volatile metallic impurities and less than about 20 p.p.m. soluble metallic impurities have substantially improved longitudinal and transverse impact strengths as compared to the conventional stainless steel alloys of Examples 1 and 2. Examples 4 to 10 have nominal compositions corresponding to several AISI 300 series stainless steels, and, in addition include additional alloying agents. Example 5 contains 0.16 percent zirconium and it can be seen that the transverse impact strength is reduced by the presence of zirconium in the alloy. However, the impact strengths of Example 5 are superior to conventional zirconium containing stainless steels of similar nominal composition. Examples 4 and 6 contain titanium and manganese alloying agents respectively, and it will be seen that these alloying agents do not affect the transverse impact strength of the alloy. Accordingly, it is possible to add manganese, conventionally added to stainless steels to mask the sulphur content thereof, and still obtain decidedly superior transverse and longitudinal impact strengths.

Table II also includes data in respect of the longitudinal and transverse percent reduction in area and the R.A.T. ratios of Examples 1 to 10. The R.A.T. ratio is a ratio of the transverse percent reduction in area (ASIM No. A370-67) to the longitudinal percent reduction in area expressed as a decimal. The R.A.T. ratio is an indication of anisotropy, and high R.A.T. ratios indicate low anisotropy, and therefore are an indication of ductility in both the longitudinal and transverse direction. It can be seen that the R.A.T. ratio of the austenitic alloys of Examples 3 to 10 are all in excess of .90, and most are in excess of .95. Conventional austenitic stainless steels have much lower R.A.T. ratios of .75 or lower, as exemplified by the R.A.T. ratio of Example 1.

In addition to the substantially improved impact strengths and R.A.T. ratios, the substantially improved properties of the austenitic alloys described herein is evidenced by other tests and comparisons with conventional austenitic stainless steel alloys. For example, a $2\frac{1}{2}$ inch square billet of the vacuum induction processed stainless steel of Example 1 (representing the latest advances in commercial stainless steel manufacture) was cold forged in an open die forge in an attempt to forge the $2\frac{1}{2}$ inch square billet to a 1 inch round billet. After about a 40 percent reduction in area the billet sheared completely through at about a 45° angle. Examination of the partially forged billet of conventional stainless steel showed edge cracks and stressing within the interior of the billet as evidenced by dislocation stress concentration lines on the surface, visible to the naked eye. A $2\frac{1}{2}$ inch square billet of Example 3, having substantially identical chromium and nickel content, but containing less than 6 p.p.m. insoluble volatile metallic impurities and less than 20 p.p.m. soluble metallic impurities was easily cold forged in the same machine under identical conditions from a $2\frac{1}{2}$ inch square to a 1 inch round billet with no evidence of embrittlement.

In simulated hot rolling operations 3 inch long pieces cut from a $2\frac{1}{2}$ inch square billet of each of Examples 3, 4 and 6 were hot rolled to 0.12 inch thick hot band. Soaking temperatures of 2350° F. were employed. The hot working included a low finishing temperature (1400° F.) as compared to a normal 1800° F. finishing temperature. However, in spite of the low finishing temperature, which would have the tendency to accentuate edge cracking, none of the austenitic alloys of Examples 3, 4 and 6 showed any cracking.

In other hot rolling tests it was found that the austenitic alloys disclosed herein require lower mill loading than do conventional austenitic stainless steels, and it was possible to take extra passes with the disclosed austenitic

alloys and finish at lower temperatures without incurring excessive mill loading problems or coiling problems. Hot rolling of Examples 3, 4 and 6 with a 1650° F. finishing temperature was found to produce mill loadings similar to conventional AISI 304 steels finished at 1800° F. This permits production of thinner hot band of the disclosed austenitic alloys, which in turn has a definite advantage in eliminating subsequent cold rolling and annealing operations.

In cold rolling investigations, 1 inch wide strips of Examples 3, 4 and 6 were cold rolled from 0.12 thick hot band. Reductions of area greater than 80 percent were achieved without feathering or cracking in any of the samples.

The cold heading performance of Examples 2, 3, 4 and 6 were tested by cutting cylindrical slugs from cold drawn and annealed bar samples and compressing the slugs with a 120-pound hammer dropped from various heights. The results of the cold heading test are illustrated in FIG. 2, where the true compressive strain produced in the sample is plotted against the energy imparted by the hammer. The curves for Example 3 and Example 6 were identical. FIG. 2 also indicated, by means of a horizontal arrow, the strain at fracture of each of the examples. In addition, there is illustrated in FIG. 2 by means of a cross the strain at fracture of a conventional AISI 305 austenitic stainless steel containing 18 percent chromium and 13 percent nickel.

In the cold heading test the test slugs reach temperatures in excess of 600° F. at high strain levels, and the energy required to produce given strains provides qualitative measures of relative strengths at temperatures of 600° F. or more.

It can be seen that the stainless steel alloys of Examples 3, 4 and 6 provide equivalent fracture strains to that of conventional AISI 305 stainless steel containing higher nickel contents. This indicates that the described austenitic alloys may be substituted for conventional higher nickel content alloys to obtain equal cold heading properties. For example, the disclosed austenitic steel alloys having nominal compositions equal to AISI 305 stainless steel could replace conventional AISI 384 alloys for use in cold heading operations, e.g., screw and rivet manufacture, and as illustrated in FIG. 2, the disclosed alloys having nominal compositions equal to AISI 304 alloys can replace conventional AISI 305 alloys.

The cold drawing performance of Examples 2, 3, 4 and 6 were determined by the Swift cupping test and by the Olsen cupping test (ASTM No. A344). The austenitic alloys of Examples 3, 4 and 6 were found to be superior by 15 percent or more than the conventional alloys of Example 2.

Corrosion tests were also conducted and it was found that the described austenitic alloys were generally better in stress corrosion resistance, intergranular corrosion resistance and massive corrosion than were equivalent conventional stainless steels. Standard ASTM Huey Tests (No. A262-64) in which the corrosion rate in boiling 65 percent nitric acid is measured, indicated that the austenitic alloy of Examples 3, 4 and 6 have superior corrosion resistance in the sensitized form when compared to Example 2 and would compare favorably with stabilized conventional AISI 321 and 347 grade stainless steel for welding operations.

FERRITIC ALLOYS

It has been seen that the austenitic alloys described herein have substantially improved impact strengths in both the longitudinal and transverse directions, and that these impact strengths are improved over the entire temperature range of -320° F. to room temperature and above. It will be seen from FIG. 1 that the impact strengths decline as the temperature is lowered, but this decline is generally linear. Ferritic alloys which contain relatively small amounts of nickel, generally less than 2

percent and most often less than 1 percent, exhibit wholly different impact strength curves than do austenitic alloys. Conventional ferritic stainless steels exhibit impact strengths generally similar to those of conventional austenitic stainless steels at elevated temperatures, i.e., above 200° F. However, ferritic stainless steels generally exhibit a drastic reduction in impact strengths within a temperature range of about room temperature up to 150° F., and at temperatures below room temperature conven-

fact data in respect of ferritic stainless steels described herein is significant, and illustrates the improved fabricability and ductility of these alloys as compared to conventional ferritic alloys. The comparison of impact strengths and the transition temperature range provides a qualitative correlation of the relative notch toughness of fabricated articles under service conditions, and of the ability to employ the disclosed ferritic alloys in welded structures at normal ambient conditions.

TABLE III

Example	Percent by weight									Parts per million					Volatile metallic impurities
	Cr	Ni	Al	Mn	Mo	Si	Ti	V	W	C	N	O	S	P	
11-----	16.7	.28	.03	.35	.06	.35	NA	NA	NA	700	NA	NA	190	200	>300 total
12-----	16.3	NA	NA	.35	NA	.33	NA	NA	NA	700	NA	NA	110	180	>200 total
13-----	10.4	.01	.03	NA	.01	.04	<.002	.02	NA	20	50	NA	90	310	<6/20
14-----	14.3	.02	.01	NA	.01	.06	<.002	NA	NA	20	8	21	90	140	<6/20
15-----	16.2	.22	NA	NA	NA	.06	NA	NA	NA	20	40	NA	70	150	<6/20
16-----	24.5	NA	NA	NA	NA	NA	NA	NA	NA	80	50	NA	100	130	<6/20
17-----	11.3	.06	.7	.03	.01	.07	.54	.02	.1	80	20	NA	60	120	<6/20
18-----	10.8	.05	2.0	.03	.01	.07	.53	.02	.1	60	20	NA	60	100	<6/20
19-----	11.4	NA	3.7	.06	.02	.04	.56	NA	NA	80	30	NA	20	130	<6/20
20-----	10.6	.04	5.6	.03	.01	.05	.52	.02	.1	80	40	NA	80	110	<6/20
21-----	24.5	NA	NA	<.02	1.08	.25	NA	NA	NA	22	95	<.02	80	120	<6/20
22-----	25.1	NA	NA	NA	2.0	NA	NA	NA	NA	10	60	.01	110	NA	<6/20

TABLE IV

Example	Charpy V-notch impact strength at 200° F., ft. lb.		Transition temperature, ° F.		Percent red in area		R.A.T. ratio
	Long.	Tran.	Long.	Tran.	Long.	Tran.	
11.....	100	50	175/225	250	-----	-----	-----
12.....	50	50	150/225	150.225	-----	-----	-----
13.....	235	235	0/50	-40/0	88	92	.96
14.....	235	235	0/50	-40/20	89	89	1.0
15.....	235	2.5	a-40/0	0/70	85	84	.99
16.....	2.5	-----	a-40/0	-----	79	81	1.03
17 ^b	235	235	100/150	75/140	84	80	.95
18 ^b	235	235	a 40/100	150/210	61	75	.81
19 ^b	235	235	175/225	200/250	54	13	.24
20 ^b	-----	-----	275/325	275/325	60	6	.09
21 ^c	235	240	-125/-50	-125/-50	85	85	1
22 ^c	235	240	-60/15	-60/15	85	85	1

^a Cold worked and annealed at 1,450° F.

^b Impact strengths measured at 300° F.

^c Impact strengths measured at room temperature.

tional ferritic stainless steels have negligible impact strength.

The temperature or temperature range over which the impact strength of ferritic steel is drastically lowered, is referred to as the transition temperature. At temperatures below the transition temperature, ferritic stainless steels are brittle and are susceptible to fracture, while at temperatures above the transition range the ferritic stainless steels are considered to be tough. The transition temperature range of a ferritic alloy is of importance in considering the usefulness of the alloy. For example, in welded articles such as ship hulls and pressure vessels rapidly propagating brittle fractures caused by impact loading at temperatures below the transition temperature can result in complete structural failure. It is therefore important that the transition temperature of ferritic alloys be below room temperature if this type of material is to find widespread use in welded structures.

It is well known that it is difficult to designate a specific temperature as the transition temperature in ferritic alloys. Testing of several samples from the same ingot generally results in varying transition temperatures. This is believed to be due to the difference in the machining of the samples, the particular grain orientation at the point of fracture, and other mechanically induced errors which lead to a spread in transition temperature values for similar ferritic alloys. The ferritic alloys described herein have even wider transition temperature ranges due to the substantially complete absence of volatile metallic impurities and the very low carbon and nitrogen content. Accordingly, the impact strength data set forth herein, although representative of a large number of samples, should not be considered absolute. Higher and lower values, and slight shifts in the transition temperature range may be encountered.

Nevertheless, it is believed that the Charpy V-notch im-

The ferritic alloys of the present invention comprise between about 10 percent and about 40 percent chromium, preferably between about 10 percent and about 28 percent chromium, and contain less than about 250 p.p.m. combined carbon plus nitrogen.

There is set forth in Table III the composition of a series of ferritic alloys having nominal compositions corresponding to typical AISI 400 series stainless steels. Example 11 is a conventional electric furnace melt and Example 12 is a conventional vacuum induction furnace melt. Examples 13 to 22 are ferritic alloys as described herein containing less than 6 p.p.m. insoluble volatile metallic impurities and less than 20 p.p.m. soluble volatile metallic impurities. Table IV sets forth the Charpy V-notch longitudinal and transverse impact strength of Examples 11 to 16 at 200° F., Examples 17 to 20 at 300° F., and Examples 17 to 20 at room temperature, and the transition temperature range in both the longitudinal and transverse directions. As indicated above, the transition temperature has been determined from a number of samples, but it can only be a relative measurement and Table IV is not intended to set forth the absolute transition temperature range. However, comparison of the impact strengths and transition temperature ranges of the ferritic alloys described herein with those of conventional stainless steel alloys clearly indicates the substantially improved physical and mechanical properties which are achieved by the described alloys.

Example 11, an AISI 430 grade stainless steel has a longitudinal impact strength of less than 100 at 200° F. and a transverse impact strength at 200° F. of less than 50. Example 12, a further sample of a conventional AISI 430 stainless steel, has transverse and longitudinal impact strengths of about 50 to 200° F. Examples 13 and 16 which have nominal compositions corresponding to AISI 400 series ferritic alloys have longitudinal and transverse

impact strengths at 200° F. which are significantly greater than the impact strengths of Examples 11 and 12.

Further, and of major importance, the alloys of Examples 13 and 16 have substantially lower transition temperature ranges than do the conventional ferritic stainless steels of Examples 11 and 12. Example 11, an electric furnace melt, has a longitudinal transition temperature range of between approximately 175° F. and 225° F., and a transverse transition temperature range in excess of 250° F. Example 12, a vacuum induction melt, has longitudinal and transverse transition temperature ranges of between approximately about 150° F. and 225° F. Example 14 has a longitudinal transition temperature range of between approximately 0° F. and 50° F. and a transverse transition temperature range of between approximately -40° F. and 20° F. Example 15 has a longitudinal transition temperature range of approximately -40° F. to 0° F. and a transverse transition temperature range of between approximately 0° F. and 70° F. The downward shift of the transition temperature range of Example 15 as compared to Examples 14 is brought about by cold working to at least about 50 percent reduction followed by grain refinement annealing, e.g., at 1450° F., and is further illustrated in FIG. 3 where the shaded area A corresponds to the longitudinal transition temperature range of the hot rolled material of Example 15 and the shaded area B corresponds to the longitudinal transition temperature range of Example 15 after cold working and annealing at 1450° F. The shaded area C corresponds to the longitudinal transition temperature range of Examples 11 and 12.

FIG. 4 illustrates the transverse transition temperature range of a hot rolled and unannealed sample of Example 15 in the shade portion A and the transverse transition temperature range of Examples 11 and 12 in the shaded portion C. Accordingly, when transition temperature ranges below 0° F. are desired such can be obtained by simple cold working and annealing.

It can be seen from Table IV and from FIGS. 3 and 4 that the transition temperature range for the ferritic alloys described herein are substantially lower than conventional alloys having similar nominal compositions. This property of the disclosed ferritic alloys is quite important, particularly where the alloy is to be welded. As previously indicated, conventional ferritic alloys have transition temperature ranges at or above room temperature, and are subject to weld cracking under impact loading. Conventional iron-chromium alloys containing 26 percent chromium cannot be welded without cracking. Accordingly, the conventional ferritic alloys are seldom used in welded structures. However, the ferritic alloys described herein, containing from 10 percent to about 26 percent chromium, less than about 6 p.p.m. insoluble volatile metallic impurities, and less than 250 p.p.m. carbon plus nitrogen have longitudinal and transverse transition temperature ranges generally below room temperature and are not subject to weld cracking at ambient conditions.

The ferritic alloys described herein are capable of use in environments where weldability and/or impact strength at room temperature is important, and may replace austenitic alloys in the fabrication of structures which are welded or are subject to impact loading. Since corrosion resistance depends primarily upon the chromium content of the alloy, the ferritic alloy described herein can be employed in place of the higher cost nickel containing austenitic alloys, thus lowering manufacturing cost without sacrificing either corrosion resistance or impact strength.

Table IV also contains data in respect of the percent reduction in area of the ferritic alloys, and indicates that R.A.T. ratios of approximately 1.0 are readily obtained. This is a further indication of the excellent ductility of the ferrite alloys, and evidences the extremely low anisotropy of these alloys.

The improved ability to work and fabricate the dis-

closed ferritic alloys is further illustrated in hot rolling, and the ferritic alloys disclosed herein may be rolled to a 50 percent reduction in area at lower mill loadings than required for 30 percent reduction in area of equivalent conventional ferritic alloys. As indicated above in connection with the austenitic alloys, the ease in hot rolling has a twofold advantage, reduction of the number of passes through the rolling mill, and reduction of the loading of the mill.

In cold rolling of the ferritic alloys disclosed herein, it has been found that the disclosed alloy can be cold rolled to high reductions without box annealing (1400° F. for 15 hours) which is necessary in cold rolling of conventional ferritic alloys to prevent intergranular corrosion during pickling and to enhance formability. Cold rolled ferritic alloys, as described herein, when rolled from unannealed hot band are softer after a 40 percent reduction than conventional ferritic alloys rolled from annealed hot band. Accordingly, it is possible to eliminate box annealing of the ferritic alloys disclosed herein. Samples of unannealed hot band of ferritic iron-chromium alloys containing 10 percent, 16 percent and 26 percent chromium were cold rolled from 1 inch to 0.006 inch (99.4 percent reduction) without edge cracking. A sample of Example 12 delaminated at about 50 percent reduction.

Grain refinement annealing of the ferritic alloys described herein may also be carried out at lower temperatures. Cold rolling to 80 percent reduction followed by annealing at 1400° F. for five minutes (75° F. lower than that required for conventional AISI 430 alloys) has produced strip having an ASTM range grain size number of 7-8, which is equal to that of conventional AISI 430 steel alloys.

Rolling of conventional ferritic alloys, and particularly AISI 430 grades, may result in a surface imperfection known as "roping" in which surface ridges or corrugations form when the strip is stretched or drawn into a part. This is quite detrimental and causes conventional AISI 430 grade ferritic stainless steel to be undesirable in certain uses, for example, in the automobile industry. Rolling and forming of the disclosed ferritic alloys of the present invention, including those equivalent to AISI 430, does not result in undesired roping, thus permitting use of the disclosed alloys where conventional AISI 430 alloys are not suitable.

In addition to the more typical AISI 400 ferritic alloys, the present invention is also directed to specialty ferritic alloys. Examples 17 to 20 of Tables III and IV are directed to ferritic alloys containing aluminum. The alloys of Examples 17 to 20 are low chrome alloys, the aluminum thereof causing hardening and strengthening of the low chrome alloy and improved oxidation resistance. These alloys have excellent corrosion and oxidation resistance and are useful in many moderately high temperature applications. However, the presence of aluminum in the alloy has caused problems in the manufacture of these alloys heretofore, and commercially available ferritic alloys containing one percent or more aluminum have generally been limited to nondecorative uses, such as automobile mufflers and furnace burners. The aluminum containing ferritic alloys disclosed herein are easily manufactured, and can readily be produced in sheet form for use in decorative applications where superior corrosion resistance is desired.

It can be seen from Table IV that the effect of the addition of aluminum is to raise the transition temperature range of the alloys in both the longitudinal and transverse directions to above room temperature. However, these transition temperature ranges are still well below the usual transition temperatures of conventional aluminum containing ferritic alloys. At 300° F., Examples 17, 18 and 19 containing 1 percent, 2 percent and 4 percent aluminum have impact strengths in both the longitudinal and transverse directions which are in excess of 235. At 6 percent aluminum (Example 20), the additional

13

aluminum has driven the transition temperature range upwardly to a point where 300° F. is within the transition temperature range and the impact strength at this temperature is not readily ascertained.

The effect of cold working and annealing the aluminum containing alloys in accordance with the invention can be seen from Examples 17 and 18 where annealing has the effect of reducing the longitudinal transition temperature range about 50° F.

The effect of increased aluminum content of Examples 17 to 20 also can be seen to have a direct effect upon the R.A.T. ratio, and at higher aluminum contents the alloys become quite anisotropic. Although this results in somewhat lesser ductility than the alloys of Examples 13 to 16, the alloys of Examples 17 to 20 are capable of fabrication, either by hot or cold working, and exhibit improved ductility and toughness as compared to conventional aluminum containing alloys.

A particularly useful ferritic alloys of the invention is one containing between one-half and 2 percent by weight molybdenum. Ideally, such an alloy comprises about 26 percent by weight chromium and about 1 percent by weight molybdenum. An alloy of this latter type is indicated in Tables III and IV as Example 21, Example 22 being roughly similar but with approximately twice the molybdenum content. Alloys of this type have the excellent toughness, formability and weldability of the other ferritic alloys of the invention and exhibit extremely high resistance to a wide variety of corrosive environments. As did the other ferritic alloys of the invention, Examples 21 and 22 exhibit a relatively low transition temperature, and exhibit yield and tensile strengths comparable to conventional 430 stainless steel.

With specific reference to weldability, alloys of the composition of Examples 21 and 22 can, as can the other examples of alloys of the invention, be welded satisfactorily without the special before and after heat treatments required for welding conventional ferritic stainless steel. Sound ductile weldments have been obtained using both the TIG process and resistance (spot) welding. Autogenous TIG weld passes have been made in fully annealed sheet parallel to the rolling direction and the bend ductility of transverse face bend samples determined by means of a guided bend test. The alloy showed excellent ductility, being capable of IT bends in the as-welded condition without cracking. In another evaluation of weldability, one inch by 4 inch sheet coupons were spot welded together following recommended practices for spot welding stainless steels. The alloy was easily spot welded and easily withstood shearing loads of 2,000 lbs.

Alloys of the invention containing between one-half and 2 percent molybdenum, by weight, provide excellent corrosion resistance in a variety of environments. Generally, such alloys are equal to or better than chrome nickel austenitic stainless steels in oxidizing solutions and in many reducing solutions. Unlike the chrome nickel austenitic stainless steels, ferritic stainless steels of the invention containing between one-half and 2 percent molybdenum are substantially immune to chloride stress corrosion cracking. Results of "Huey" tests using boiling 65% HNO₃ and five 48-hour exposures indicated that an alloy of a nominal composition substantially corresponding to Example 21 was less than half the penetration of commercial type 304 (austenitic) stainless and roughly one-third less penetration than commercial type 304L stainless. The penetration of type 430L (ferritic) stainless for the same test was over seven times higher. Similar spectacular results have been obtained for the same alloy in tests with boiling ferric sulphate, 50% H₂SO₄ for 120 hours. Even more spectacular results have been obtained in tests of exposure to ferric chloride, boiling sodium sulphate, and boiling aluminum sulphate. Corrosion resistance comparable to that of 304L stainless and substantially better than type 430 stainless from commercial sources has been exhibited in tests with formic acid, phos-

14

phoric acid, aerated salt water immersion, and cyclic immersion in a solution of sodium chloride, calcium chloride, sodium sulphate, sodium sulphite, and sodium thiosulphate. Thus, generally, the resistance to corrosion of alloys of the invention containing one-half to 2 percent molybdenum is superior to that of conventional ferritic or austenitic stainless steels.

The susceptibility of alloys of the invention containing one-half of 2 percent molybdenum toward intergranular corrosion in the "Huey" test is substantially lower than that of conventional ferritic or austenitic stainless steels. Unlike conventional ferritic stainless steels, molybdenum steels in accordance with the invention are not susceptible to intergranular corrosion when water quenched from temperatures in the 1600° F. to 2050° F. range. Although the sensitivity to intergranular corrosion can increase when the material is slowly cooled, it still shows superior properties to that of conventional 316 stainless steel in the fully annealed condition.

MARTENSITIC ALLOYS

The martensitic iron-chromium alloys are generally hardenable by heat treatment, exhibit high tensile strengths and good ductility and toughness and are desirable where high strength and good corrosion resistance is desired. However, from a commercial standpoint, these materials have largely been ignored in the past, principally because of difficulties in fabricating, relatively low impact strengths as compared to austenitic alloys, and poor weldability.

The martensitic iron-chromium alloys have low chromium content, generally between about 10 percent and about 20 percent, usually between about 10 percent and about 14 percent and contain a sufficient amount of austenite stabilizer to cause the alloy to have a martensitic crystalline configuration at room temperature. Depending upon the relative amounts of chromium and austenitic stabilizers, the martensitic alloys may be wholly martensitic, or may be mixtures of martensite plus ferrite, martensite plus austenite plus ferrite, or martensite plus austenite. In order to obtain good weldability it is generally preferred that the martensitic alloys have a relatively low carbon plus nitrogen content and contain less than about 10 percent of ferrite crystals.

One class of commercial martensitic alloys are those which contain between about 10 percent and about 14 percent chromium and between about 2 percent and about 7 percent nickel. Such alloys have a substantially 100 percent martensitic crystalline configuration. These alloys have tensile strengths of about 100,000 p.s.i., and when the carbon content thereof is reduced to below about 0.03 percent by weight have good weldability. Higher carbon content, up to about 0.06 percent by weight may be tolerated in welding applications if 0.01 percent by weight titanium is present.

Because the martensitic alloys are air hardenable, measurements of ductility and toughness, such as impact strength and R.A.T. ratios, are significant only when measured at equal strength levels. Accordingly, comparison of the described martensitic alloys with conventional alloys is somewhat more difficult than are the austenitic and ferritic alloys. Nevertheless, martensitic alloys containing less than about 6 p.p.m. of insoluble volatile metallic impurities and less than about 20 p.p.m. of soluble volatile metallic impurities have been found to have substantially improved physical and mechanical properties as compared to the conventional martensitic alloys.

The impact strengths of the martensitic alloys are generally intermediate those of the austenitic alloys and ferritic alloys. The martensitic alloys have somewhat lower impact strengths at room temperature than do the austenitic alloys, and the martensitic alloys have a transition temperature range between that of the ferritic alloys and the austenitic alloys. However, the transition temperature range of the martensitic alloys is well below room temperature. Conventional 13 percent chromium, 4 percent nickel martensitic alloys have longitudinal impact

strengths of between about 30 and 40 foot pounds at -95° F. Alloys having similar nominal compositions but containing less than 6 p.p.m. insoluble volatile metallic impurities and less than 20 p.p.m. soluble volatile metallic impurities, and less than 250 p.p.m. total carbon plus nitrogen content, have longitudinal impact strengths in excess of 100 foot pounds at -95° F. In addition, the transverse impact strengths of the martensitic alloys as described herein are substantially improved, and are in excess of 100 foot pounds at room temperature.

A further example of a martensitic alloy is an alloy containing about 16 percent chromium and about 9 percent nickel, less than 6 p.p.m. insoluble volatile metallic impurities and less than 20 p.p.m. soluble volatile metallic impurities. This alloy is a mixture of austenite, martensite, and ferrite crystals and contains less than about 10 percent ferrite. This alloy has tensile strengths in excess of 100,000 p.s.i., and at room temperature, has a longitudinal impact strength of about 150 foot pounds and a transverse impact strength of about 90 to 100 foot pounds. At -320° F. the longitudinal impact strength is about 90 foot pounds and the transverse impact strength is about 40 foot pounds. The R.A.T. ratios of the described martensitic alloys are in excess of .90 and usually in excess of .95 demonstrating the extremely low anisotropy and excellent ductility of martensitic alloys which are free of volatile metallic impurities. This alloy is disclosed and claimed in the copending application of James H. C. Lowe, Ser. No. 46,393, filed June 15, 1970.

Similar to the ferritic alloys, the weldability of the described martensitic alloys is excellent when the total carbon plus nitrogen content of the alloys is reduced to fairly low levels, i.e., below about 250 p.p.m. The ability to weld the martensitic alloys without fear of weld cracking and embrittlement during service renders the described martensitic alloys attractive structural materials for use at temperatures below 0° F. where high tensile strengths are required. Further, the excellent ductility of the described martensitic alloys permits ready rolling and fabrication of the alloys into desired shapes.

METHOD OF MANUFACTURE

It will be seen from the foregoing that iron-chromium alloys containing less than about 6 p.p.m. total of insoluble volatile metallic impurities and less than about 20 p.p.m. soluble volatile metallic impurities exhibit substantially improved physical and mechanical properties as compared to conventional iron-chromium alloys. The removal of volatile metallic impurities from iron-chromium alloys at commercially attractive rates, and particularly the manufacture of these alloys from low cost commercial raw material sources, is advantageously effected by exposure of a molten iron-chromium alloy to electron beam bombardment at a reduced pressure of below about 10^{-1} torr. The power of the electron beam bombardment is controlled so that the power input is in excess of about 20 kilowatts per square foot of the molten alloy surface, and the duration of the exposure of the molten iron-chromium alloy to the electron beam is controlled to between about 100 and about 2000 pounds per hour per square foot of the molten alloy surface. This method of manufacturing the disclosed iron-chromium alloys is more fully described in the copending application of Charles d'A. Hunt, Ser. No. 46,156, filed June 15, 1970, which is also a continuation-in-part of abandoned application Ser. No. 722,640.

The described method can be carried out in an electron beam furnace having a single pressure zone or in a multipressure zone refining apparatus such as is described in U.S. Pat. No. 3,343,828. The use of multipressure zone refining apparatus is particularly advantageous when the iron-chromium alloy is manufactured directly from ferrochrome raw materials.

It will be seen from the foregoing that iron-chromium alloys of substantially improved physical and mechanical

properties have been described, as well as a preferred process for manufacturing such alloys. The superior toughness, ductility, isotropy, corrosion resistance and fabricability of the described alloys is such that they may be employed in environments and in structures where conventional iron-chromium alloys have been heretofore considered to be inappropriate.

It is to be understood that the invention has been described with respect to specific iron-chromium alloys in order to sufficiently describe the invention. Alternative embodiments and equivalent practices are considered to be within the scope of the invention.

Various of the features of the invention are set forth in the following claims.

What is claimed is:

1. An iron-chromium alloy obtained from a vacuum purification process, consisting essentially of between about 4 and about 40 percent by weight chromium, between 0 and about 14 percent by weight nickel, carbon and nitrogen in amounts such that essentially all carbon and nitrogen will be in solid solution in the iron-chromium alloy at room temperature, not more than about 6 p.p.m. total of the insoluble volatile metallic impurities, lead, bismuth, cadmium, sodium, potassium, silver, calcium, magnesium and barium, not more than about 20 p.p.m. total of the soluble volatile metallic impurities zinc and antimony, and the balance iron.

2. An iron-chromium alloy in accordance with claim 1 having a R.A.T. ratio in excess of 0.90.

3. An iron-chromium alloy in accordance with claim 1 having a predominantly austenitic crystalline configuration at room temperature consisting essentially of between about 14 and about 20 percent by weight chromium, and between about 8 and about 14 percent by weight nickel.

4. An iron-chromium alloy obtained from a vacuum purification process having a predominantly austenitic crystalline configuration at room temperature, consisting essentially of between about 14 and about 20 percent by weight chromium, between about 8 and about 12 percent by weight nickel, carbon and nitrogen in amounts such that essentially all carbon and nitrogen will be in solid solution in the iron-chromium alloy at room temperature, said alloy being essentially free of volatile metallic impurities, and the balance iron, said alloy exhibiting the physical properties of: (a) a R.A.T. ratio in excess of 0.90; (b) a longitudinal and transverse Charpy V-notch impact strength above about 200 foot-pounds at room temperature; and (c) a longitudinal and transverse Charpy V-notch impact strength above about 100 foot-pounds at -320° F.

5. An iron-chromium alloy in accordance with claim 4 consisting essentially of, between about 16 and about 20 percent by weight chromium, and between about 8 and about 10 percent by weight nickel.

6. An iron-chromium alloy in accordance with claim 1 having a predominantly ferritic crystalline configuration at room temperature consisting essentially of between about 10 and about 40 percent chromium, and less than about 250 p.p.m. combined carbon plus nitrogen content.

7. An iron-chromium alloy in accordance with claim 6 having a predominantly ferritic crystalline configuration at room temperature consisting essentially of between about 10 and about 28 percent by weight chromium.

8. An iron-chromium alloy in accordance with claim 6 having a predominantly ferritic crystalline configuration at room temperature consisting essentially of between about 10 and about 14 percent by weight chromium, less than 250 p.p.m. combined carbon plus nitrogen content, and up to about 6 percent by weight aluminum.

9. An iron-chromium alloy in accordance with claim 1 having a predominantly martensitic crystalline configuration at room temperature consisting essentially of between about 8 and about 14 percent by weight chromium, between about 2 and about 10 percent by weight nickel,

17

and less than about 250 p.p.m. combined carbon plus nitrogen content.

10. An iron-chromium alloy in accordance with claim 6 having a predominantly ferritic crystalline configuration at room temperature, consisting essentially of about 26 percent by weight chromium and about 1 percent by weight molybdenum.

11. An iron-chromium alloy obtained from a vacuum purification process having a predominantly ferritic crystalline configuration at room temperature consisting essentially of between about 10 and about 40 percent chromium, less than about 250 p.p.m. combined carbon plus nitrogen, said alloy being essentially free of volatile metallic impurities and the balance iron, said alloy exhibiting the physical properties of: (a) a R.A.T. ratio in excess of 0.90; (b) a longitudinal and transverse Charpy V-notch impact strength in excess of about 200 foot-pounds at 200° F.; and (c) a longitudinal and transverse impact transition temperature below room temperature after cold working to at least 50 percent reduction and annealing.

12. An iron-chromium alloy in accordance with claim 11 including between about ½ and about 2 percent by weight molybdenum.

13. An iron-chromium alloy in accordance with claim

18

11 consisting essentially of about 26 percent by weight chromium and about 1 percent by weight molybdenum.

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