MECHANICAL PROPERTIES OF SOME
METASTABLE AUSTENITIC ALLOYS

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ABSTRACT

The present investigation was initiated to gain additional information about certain metastable austenitic alloys which undergo a strain-induced martensitic transformation. The mechanical properties of these alloys were investigated as a function of the temperature of prior deformation and nickel content at various test temperatures. The properties were correlated with the progress of the transformation during straining.

The results indicated that prior deformation in the neighborhood of M_D was beneficial if the temperature of deformation was conducive to the formation of an optimum amount of precipitation of carbides. Prior deformation below M_D resulted in the formation of some martensite during deformation and this increased the yield strength without considerable loss in ductility if the amount of martensite was limited and the M_D was high. It was also found that austenite stability was an important factor that affected the mechanical properties. When the stability was very low, premature yielding occurred at low stresses because of stress induced martensite formation and this was accompanied by a low elongation and extraordinarily high strain hardening rate. Nickel additions stabilized the alloys and prevented premature yielding.

INTRODUCTION

In recent years there has been a great demand for high strength materials, particularly steels. The mechanical properties of steel, including the yield strength and the elongation, are primarily governed by the composition and heat treatment. In recent years, another dimension has been added to the traditional processing of steel, i.e., mechanical working in conjunction with heat treatment.

Conventional heat treatment of steel with a view of increasing the strength results in low ductility. A departure from conventional heat treatment has been suggested as a means for obtaining high strength levels in hardenable steels. Lips and Van Zuilen¹ reported that unusually high strength could be developed in steel by warm working metastable austenite prior to quenching to room temperature. The work of Schmatz and Zackay² confirmed the feasibility of this thermal-mechanical treatment as a means to improve the properties of high strength steels. Further studies in this direction^{3,6} have proved that the high strength obtained is due to the fine carbide precipitation and high dislocation density as well as reduction in the martensite plate size.

The limited elongation of most high strength steels is due to the tensile instability which begins at low levels of strain, leading to necking and fracture. The criterion for necking; viz., $\frac{d\sigma}{d\epsilon} = \sigma$ postulates that the strain hardening rate must increase in direct proportion to the yield strength if necking is to be prevented¹⁴. Barriers stronger than dislocation tangles must be formed during plastic straining if the strain-hardening rate is to be increased. Martensite plates can act as strong barriers to dislocation motion. Metastable austenitic stainless steels are

found to strain-harden at high rates when martensite forms during straining.⁸ The effect of strain-induced martensite in increasing the strain hardening rate is still greater in steels of higher carbon contents. This has led to the development of a new series of steels which are known as TRIP steels (TRIP stands for TRansformation Induced Plasticity).

Preliminary investigations on these steels have indicated that high strength levels could be obtained with excellent ductility and fracture toughness.^{7,9} The effects of carbon and manganese contents, amount and temperature of prior deformation and test temperature have been investigated further.^{10,11} The high yield strength of these steels is attributed to the dislocation tangles and precipitation of carbides that occur during the prior deformation. After the prior deformation treatment in the austenitic state, the steels are rendered susceptible to the strain induced transformation to martensite on further straining below the M_D temperature. The formation of strain induced martensite in local regions of high stress or strain prevents necking in those regions and the neighboring regions start to yield. Thus the transformation proceeds with the propagation of the Luder's band along the gage section of the tensile specimen. This considerably hardens the gage section and prevents fracture until the whole section has been transformed.

The present status of knowledge about TRIP steels encompasses a fairly good understanding of the basic strengthening mechanism, but there is still scope for the development of explicit theories that can predict the behavior of these steels under various experimental and service conditions. The present investigation was initiated with a view of studying factors such as the effects of austenite stability,

temperature of prior deformation, rate of transformation during straining and also the so called phenomenon of "premature yielding".

Stability of austenite is controlled by composition and test temperature. The present investigation studied the effect of additions of nickel. The effect of austenite stability on the mechanical properties has been investigated before in austenitic stainless steels.^{8,12,13}

Stability governs mechanical properties because mechanical properties depend on what stage in the straining the transformation to martensite begins and this in turn depends on how stable the austenite is. In this investigation, the austenite stability was varied by varying the nickel content and test temperature. Temperature of prior deformation, as a factor affecting mechanical properties has been studied before.¹⁰

The present study included prior deformation temperatures close to and below the M_D temperature. Temperature of prior deformation controls the mechanical properties by influencing the extent of precipitation and the resulting dislocation configuration. The formation of martensite during prior deformation, as happens when the prior deformation is given below the M_D temperature, can result in an increase in yield strength.

One of the most important factors is the progress of the transformation during straining. The rate at which the transformation proceeds

determines the rate at which the steel strain hardens. It was proposed to study the progress of the transformation as a function of strain at different test temperatures and with different temperatures of prior deformation. Another phenomenon that was investigated in the present study was the yielding at low levels of stress that occurs when

austenite is highly unstable. This has been referred to by Kuling, et al.

as the stress induced transformation to martensite and the term "premature yielding" suggested by the same authors can be used synonymously. Premature yielding is not desirable because it results in low yield strength and low elongation. Factors that can prevent this were studied. The investigation was complemented by a study of pure iron nickel alloys that did not contain carbon. The behavior of these alloys was compared with the carbon-containing alloys.

2. Experimental Procedure

A. Selection of Alloys

Two main series of alloys were selected for this investigation. The series with carbon was selected on the basis of previous work on high strength metastable austenitic steels.^{7,10,16,17} The nickel content was varied to get alloys with different stabilities of austenite. The M_D temperatures of these alloys varied from above room temperature to below -196°C . The M_S temperature was below room temperature so that the alloys were austenitic on quenching to room temperature.

The alloys without carbon were Fe-Ni compositions containing different percentages of nickel from 20% to 30%. Two of these alloys, h and i, are known to undergo strain-induced transformation to martensite.^{18,19} The less stable of the two (h) had an M_D above room temperature and an M_S below room temperature and the alloy i, which was more stable had both M_D and M_S below room temperature. The other two alloys in this series were designed to be on either side of the stability range mentioned above.

The chemical composition of the alloys is given in Table 1.

B. Processing of Alloys

Raw materials of purity 99.99% were employed in preparing the alloys by induction melting in a helium atmosphere. The molten alloys were cast into water-cooled copper molds to get 16 lbs. ingots. The ingots were homogenized for 3 days at 1100°C and forged at 1100°C to $2\frac{1}{2}'' \times 1\frac{1}{2}''$ section. Further reduction to a thickness of $\frac{1}{4}''$ was carried out by rolling at 450°C . The plates were then austenitized in stainless steel heat treating containers. An austenitizing tempera-

ture of 1200°C for 2 hours was employed for the series of alloys with carbon and 1150°C for 1/2 hour for the carbonless series. Austenitizing was followed by a brine quench.

The plates were then pickled to remove the surface scale. The aqueous pickling bath employed contained 10% nitric acid, 1-2% hydrochloric acid. The temperature of the bath was maintained between 38°C and 60°C. The time of pickling was three minutes.

The material was then deformed 70% at various temperatures by rolling. Rolling at temperature above room temperature was carried out using preheated rolls. The sheets were heated between passes to the rolling temperature. When rolling was carried out at room temperature, the sheets were cooled in water between passes. Cryogenic baths of dry ice in ethyl alcohol and ethyl alcohol in liquid nitrogen were used for rolling temperatures of -78°C and -120°C respectively. The time required for rolling was two hours or less in all cases. To avoid excessive heating during rolling small reductions were given at a time. The plates rolled at temperature above room temperatures were water quenched.

C. Mechanical Testing

Tensile specimens were prepared from the rolled sheets. The dimensions of the tensile specimens are shown in Fig. 1. Testing was performed on an Instron testing machine using a crosshead speed of 0.1 cm/min.

Room temperature tests were performed in air. For tests at -78°C the specimen was kept immersed in a mixture of dry ice and ethyl alcohol. Liquid nitrogen bath was used for tests at -196°C.

Gage marks were punched on the specimen so that the elongation could be measured by means of a traveling microscope graduated in 0.002 mm. Engineering stress could be calculated using the original cross sectional area of the specimen. To study the strain-hardening behavior, it was necessary to draw the flow stress curves. The strain hardening exponent (n) is related to true stress (σ) and true strain (ϵ) by the relation

$$\sigma = A \epsilon^n$$

where A is a constant known as the strength coefficient.¹⁴

By differentiating and solving for n

$$n = \frac{d(\log \sigma)}{d(\log \epsilon)}$$

Log-log plots of true stress and true strain were plotted, from which the strain hardening exponent was determined as the slope of the straight line in the region of strain hardening, after the Luders strain and before necking.

D. Magnetic Testing

The saturation induction of the specimen was measured before, during and after tensile testing. The permeameter employed for this purpose has been described elsewhere.²⁰ Percentage of martensite could be calculated from the saturation induction. Tensile tests at room temperature and at -196°C were accompanied by continuous magnetic tests. For tests at -78°C , the specimen had to be unloaded and removed from the grips every time a magnetic reading was desired. A specimen containing less than 1% martensite, as indicated by the magnetic testing, was assumed entirely austenitic.

Only the series of alloys with carbon could be tested in this way because in this series of alloys the austenite was non-magnetic whereas

the martensite was ferromagnetic. In the carbonless alloys under consideration the austenite was ferromagnetic because of the high nickel content and the increase in saturation induction due to martensite formation could not be detected. For this reason, optical microscopy was employed to follow the transformation where necessary.

E. Optical Microscopy

Mechanical polishing was employed since it was found that no significant amount of martensite was nucleated by polishing in this manner. The specimens were etched in a solution containing 5.0 gm of cupic chloride, 100 ml hydrochloric acid, 100 ml methyl alcohol and 100 ml water.

RESULTS

A. Alloys Containing Carbon

The tensile properties of these alloys are listed in Tables II through X. All these alloys were given a prior deformation of 70% at various temperatures and tested at 22°C, -78°C and -196°C. In Tables II through VI the properties for a particular test temperature are listed in each table, and in Tables VII through X the properties for a particular alloy are listed in each table.

The yield strength, ultimate strength and percent elongation are plotted as functions of the temperatures of prior deformation in Figs. 2 through 5, 7 through 10 and 12 through 15. The yield and ultimate strengths were found to decrease first, then increase, reach a maximum and then decrease, as the temperature of prior deformation was increased. The peak in the strength was recorded at around 250°C-350°C. Elongation values increased, then decreased, increased again and then decreased again as the temperature of prior deformation was increased.

The transformation to martensite during straining was followed by magnetic testing and some typical variations of percent martensite with percent strain are shown in Figs. 20 through 22. The transformation initiated at a low rate, then the rate gradually increased, reaching a maximum when about 10-15% martensite was formed as a result of straining alone and then decreased. The initial rates were higher for alloys deformed previously above M_D . At low temperatures of testing, the alloy b deformed previously at 450°C formed a large amount of martensite at the yield point observed in the stress-strain curve. There was no detectable transformation below this point. Alloy c showed two maxima

in the rate of the transformation.

Figure 25 shows the variation of yield strength, ultimate strength and elongations for alloy b deformed previously at 450°C and tested at 22°C, -78°C, -196°C. The yield strength was a minimum at around -80°C.

B. Alloys Without Carbon

The tensile properties of these alloys are listed in Tables XI through XIII. The yield strength, ultimate strength and percent elongation are plotted against the temperature of prior deformation for alloy i in Figs. 6, 11 and 16. Yield and ultimate strength showed a decrease with increase in the temperature of prior deformation and then stayed constant. The properties are plotted as a function of test temperature for different alloys in Figs. 17 through 19.

The transformation to martensite was followed qualitatively by metallographic studies. Figures 23 and 24 give the structure of alloys h and i respectively for some cases of interest where it was necessary to interpret the mechanical properties in terms of the martensite transformation.

DISCUSSION

A. Variation in Mechanical Properties with the Temperature of Prior Deformation

Yield strength is plotted against the temperature of prior deformation in Figs. 2 through 6; ultimate strength is plotted against the temperature of prior deformation in Figs. 7 through 11; and elongation is plotted against the temperature of prior deformation in Figs. 12 through 16.

In the alloys that contained carbon, it was found that as the temperature of prior deformation was increased, there was first a decrease in yield strength, and then an increase, with a maximum in yield strength occurring at around 250°C to 350°C, before a drop occurred at higher temperatures (Figs. 2-5). The maxima that occurred around 250°C to 350°C are believed to be due to carbide precipitation. In other investigations with similar alloys, maxima have been found at 450°C. This has been taken to mean that prior deformation at 450°C gives the optimum properties for TRIP steels. The findings of the present investigation, however, prove that this need not be the case for all the alloys that belong to this category of steels. The optimum temperature varies with the composition of the alloy. Whereas considerable variations in the percent of nickel do not show considerable variations in this optimum temperature, the addition of a new element may affect this temperature considerably.

The drop in yield strength with temperature of prior deformation at higher temperature is due to two factors. One is the direct effect of precipitate coarsening. This is similar to the effect that occurs in the age-hardening of aluminum alloys. The second effect is the one that occurs as a result of the decrease in stability of these alloys.

With greater amount of precipitation taking place at higher temperatures of prior deformation, a depletion of carbon and carbide forming elements occurs and the stability of the alloys with respect to the martensitic transformation decreases, and the M_D temperature increases. When tested at any temperature below M_D , martensite can form at low levels of stress. The stress induced formation of martensite can cause gross yielding in the tensile specimen.^{5,15} This effect is discussed further in a later section. The high yield strength with lower temperatures of prior deformation was accompanied by a decrease in the elongation. This occurred because a considerable amount of martensite formed when prior deformation was given at temperatures far below the M_D temperature. The presence of large amounts of martensite prior to testing caused a considerable loss in ductility.

In the iron-nickel alloys without carbon there was no maximum in the yield strength at higher temperatures of prior deformation and the yield strength did not vary with the temperature of prior deformation as long as this temperature was above M_D (Fig. 6). In these alloys there was no precipitation taking place, so the stability did not vary with the temperature of prior deformation. The dislocation configurations produced as a result of prior deformation can vary with the temperature of prior deformation, but the absence of difference in the mechanical properties with prior deformation temperatures up to 200°C indicated that the dislocation configurations were not different for temperatures below 200°C. The variation in mechanical properties with prior deformation temperatures below M_D followed a similar trend as with the steel compositions. Yield strength and ultimate strength increased

whereas elongation decreased. This is indicated in Figs. 6, 11 and 16 for alloy i.

In both series of alloys, it can be seen that there is no inverse relationship between strength and ductility. This behavior is due to the strain induced transformation to martensite that takes place. In this connection, Bressanelli and Moskowitz¹² have shown that the mechanical properties of alloys in which a strain induced transformation to martensite occurs during testing depends on what stage in the straining the transformation starts.

B. Transformation to Martensite During Testing and Its

Effect on Mechanical Properties

The alloys that contain carbon have their M_s temperatures suppressed to below -196°C after 70% deformation as a result of mechanical stabilization. This effect is absent in iron-nickel alloys that do not contain carbon. So, whereas the former series of alloys was always tested above M_s , some tests in the latter series were below the M_s and therefore the effect of athermal martensite should also be considered.

1. Alloys with Carbon

a. Alloy b (8% Ni) This alloy was found to have an M_D temperature slightly above 150°C . It transformed during testing at all the three test temperatures employed. The progress of the transformation during testing at 22°C is shown in Fig. 22 as a function of percent engineering strain for different temperatures of prior deformation. The variation of the amount of martensite with strain shows that the initial rate of transformation increased with strain, reached a maximum, and then decreased, as in the investigations of Angel.¹³ It was found that the initial rate of transformation was higher with prior deformation above M_D temperature than with prior deformation below M_D . This is because prior deformation below M_D causes some martensite to form during deformation. This causes some of the strain embryos to be used up as martensite nucleates at these embryos. As a result, martensite nucleates at fewer nuclei during testing. If this argument is true, then the initial rate should be lower with a prior deformation at 25°C than with a prior deformation at 150°C , since the former causes more nuclei to be used up by formation of more martensite. This is actually the case observed (Fig. 22). The initial rate of transformation with prior

deformation at 450°C was closer to the rate observed for 150°C than to the rate observed for other temperatures above M_D . This is probably because, when prior deformation is given at 450°C, carbide precipitation occurs at some strain embryos, thus inhibiting martensite formation at these embryos. This leaves fewer nuclei for initial nucleation of martensite during testing.

The decrease in the rate of transformation during testing began after the formation of about 10-15% martensite as a result of testing alone, and the initial amount of martensite present before testing did not seem to have any effect on this.

Room temperature tests for tensile properties indicated that the alloy had good properties when prior deformation is given at 150°C. Very little, if any, precipitation can be expected to take place at 150°C. The comparatively high yield strength was therefore due mainly to the small amount of martensite that was present. Another interesting result was the yield strength of 281,000 psi with an elongation of 22.6% when the prior deformation was given at 25°C resulting in the formation of about 33% martensite.

This alloy was the least stable in the series. The premature yielding due to martensite formation occurred during testing at -78°C and -196°C. Though not very obvious, this effect was probably present to a slight degree even at 22°C. Figure 21 shows the variation of the amount of martensite as a function of percent strain for the alloy previously deformed 70% at 450°C and tested at 22°C, -78°C and 196°C. At the lower temperatures there was no transformation in the straight line portion of the stress-strain curve. At the yield points of the

stress-strain curves, a large amount of martensite was formed. On the other hand, at 22°C martensite formed gradually with increasing amounts of strain.

b. Alloy c (12% Ni) The increase in nickel content to 12% caused considerable stabilization of the austenite. The alloy did not transform during testing at room temperature. However, transformation was observed when tested at -78°C. Excellent combinations of strength and ductility were obtained for all temperatures of prior deformation above 120°C. The latter formed some martensite during rolling and this considerably decreased the elongation. A yield strength of 195,000 psi with an elongation of 51.7% was observed when the temperature of prior deformation was 200°C. The stress-strain curve for this alloy with this treatment is given in Fig. 27. The variation of percent martensite with percent strain is shown in Fig. 20. The variation shows maxima in the rate of transformations at two points. After the initial increase and decrease in the rate of transformation, there is another similar variation starting at a strain of about 20%. The cause of this is not exactly known. It seems to indicate that once one kind of nuclei are used up by the transformations, another set of nuclei are activated when the strain reaches about 20%, or when the stress corresponding to this strain is reached.

This alloy seemed to have the optimum stability of austenite with 70% deformation at 200°C and at a test temperature of -78°C. The superior properties can be attributed to the fact that the transformation to martensite takes place in the most desirable manner and rate to prevent tensile instability. This should be a guide line for design of new

alloys. Under the conditions mentioned above, the alloy had an M_D temperature of about 20°C . In comparison with alloy b, it can be seen that a 4% addition of nickel had depressed the M_D by about 140°C . This is in contrast to a nickel factor of $9.5^\circ\text{C}/\text{percent nickel}$ predicted by Angel¹³ for the $M_{D(30)}$ temperature. Another factor to be noted is that for obtaining the excellent properties, the test temperature was about $80-90^\circ\text{C}$ below the M_D temperature. To obtain such properties at room temperature the M_D should be raised. This can be done by decreasing the nickel content to about 9-10%.

c. Alloys with Higher Nickel Contents Increasing nickel contents above 12% stabilized the steel considerably. The M_D was lowered to -120°C with 16% Ni and to about -196°C with 20% Ni. The alloy d with 16% Ni exhibited the best combination of ductility at -196°C , which is about 80°C below its M_D temperature. With a prior deformation of 70% at 450°C , a yield strength of 228,600 psi and an elongation of 37% were obtained, whereas prior deformation at -120°C which is close to and slightly below the M_D temperature, gave a yield strength of 348,500 psi and an elongation of 8.4%. The alloy with 20% Ni was very stable and showed very little transformation at a test temperature as low as -196°C .

2. Alloys Without Carbon

Of the four alloys in this series, the alloys f and j did not undergo a strain induced transformation to martensite. The former was martensitic as quenched and the latter was austenitic even after testing at -196°C . For the other two alloys, the transformation was followed only qualitatively.

a. Alloy h (29% Ni) This alloy was found to have an M_s of -20°C

and an M_D of 150°C . Prior deformation at 25°C caused the formation of martensite (Fig. 23a). Further transformation occurred during testing at room temperature and a high elongation was obtained. The microstructure after testing showed more martensite than before (Fig. 23b). At the test temperatures of -78°C and -196°C , athermal martensite formed on cooling to the temperature. There was still some austenite left after the formation of strain induced martensite during rolling and athermal martensite on cooling to -78°C (Fig. 23c). Since the test temperature was below M_s , this austenite was extremely unstable and transformed at a relatively low value of stress. This happened in spite of the fact that martensite present before testing strengthened the alloy to some extent. Cooling to -196°C almost completely transformed the alloys (Fig. 23d) and the mechanical properties observed were characteristic of a fully martensitic structure.

b. Alloy i (34% Ni) This alloy had its M_D below room temperature and therefore it did not transform during testing at room temperature. At lower temperatures of testing (-78°C and -196°C), the alloy could transform. The transformation prevented tensile instability, and resulted in higher elongation values than at room temperature. This alloy was susceptible to premature yielding at test temperatures above M_s . This was shown by the lower yield strength at -78°C and -196°C than at 22°C . However, unlike alloys with carbon, premature yielding in carbonless alloys did not cause a very serious loss in elongation, presumably because the iron-nickel martensite was quite ductile as compared to the iron-carbon martensite.

Martensite formed during prior deformation when the temperature of

prior deformation was below M_D . The remaining austenite did not transform during testing at -78°C . Since it was known that there was no mechanical stabilization in iron-nickel austenites, it must be concluded that the martensite formed during rolling must have been responsible for this behavior. Once some martensite had formed during rolling, it was harder to transform the remaining austenite by further straining at temperatures above M_S . However, if further straining took place at temperatures below M_S , as when testing at -196°C , the alloy was unstable and readily transformed to martensite. Figure 24b shows athermal martensite which formed on cooling to -196°C after a prior deformation of 70% at 200°C . When tested at -196°C the difference between prior deformation temperatures above and below M_D was that the latter caused a lower elongation due to the higher amount of martensite that it had before testing.

C. Effect of Prior Deformation Around and Below M_D

The effect of prior deformation was to incorporate a high dislocation density in the alloy so as to increase the yield strength. If the deformation was carried out at high enough temperatures, some precipitation took place during deformation. In the undeformed alloy, temperatures as high as 500°C may be needed to cause precipitation because diffusion rates are low. In the deformed alloys, on the other hand, temperatures as low as 200°C can be expected to induce precipitation since both nucleation and diffusion are enhanced by plastic deformation. When the prior deformation is given below the M_D temperature, these strain embryos can nucleate martensite as well.^{21,22} It is not desirable to have too much martensite formed in this way as it can cause a considerable decrease in ductility, as found in alloy "c" deformed at -120°C and tested at -78°C. A significant result found with alloy "b" was that 32% martensite formed in this manner did not cause a very large decrease in elongation. The yield strength of this alloy when tested at 22°C was 281,000 and the elongation was 22.6%.

Prior deformation in the neighborhood of M_D imparted the best combination of strength and ductility to alloy "b" at 22°C, but not at -78°C. When tested at the latter temperature, prior deformation temperature of 250°C seemed to be the best. So it follows that the stability at the particular test temperature is an important factor in determining the properties. A prior deformation temperature of 250°C caused more precipitation and hence rendered the alloy more unstable than a prior deformation temperature of 150°C. The decrease in stability tended to decrease yield strength and elongation whereas precipitation

tended to increase the yield strength. Then there was also the effect of test temperature on stability. As the test temperature was decreased, stability decreased. Decrease in test temperature also increased the strengthening due to precipitation. The overall effect of all these factors was the following: with a prior deformation at 150°C, there was little, if any, precipitation and therefore there was no factor which could resist the tendency for a decrease in yield strength with decrease in stability through a decrease in test temperature. On the other hand, prior deformation at 250°C caused some precipitation strengthening, the effect of which increased with a decrease in test temperature and this resisted the tendency for the yield drop.

The differences in mechanical properties that arise out of a difference in prior deformation temperature are due to the temperature level per se and not due to the relative difference between the temperature and M_D . When we say the differences are due to the temperature level per se, we mean the precipitation and dislocation arrangements characteristic of deformation at that temperature. However, if the temperature of prior deformation is below the M_D temperature so that a significant amount of martensite is formed during prior deformation, the properties are improved depending on the stability of the alloy and the temperature of prior deformation.

D. "Premature" Yielding

It has been observed before that the martensite transformation may be regarded as a deformation process that competes with the usual slip process. With decreasing temperature, the critical resolved shear stress for slip increases whereas the resistance to martensite formation decreases. When the alloy is relatively unstable, the transformation can be induced by tensile stresses within the elastic range.¹⁵ The onset of the transformation can cause gross deformation of a tensile specimen resulting in premature yielding at abnormally low levels of stress.

This effect was found in the alloy b. This alloy exhibited a lower yield strength at -78°C than at 22°C for all temperatures of prior deformation. The effect was particularly noticeable when the prior deformation was given at 450°C . Figure 25 shows the variation in tensile properties with test temperatures for steel b deformed 70% at 450°C . At this relatively high temperature, significant amount of precipitation must have taken place, thus depleting carbon and carbide forming alloying elements from the austenite matrix. Considerable coarsening of the precipitate also must have taken place. In this condition the alloy was relatively weak and unstable. The stability decreased with decrease in test temperatures whereas the matrix strength increased. The former caused a decrease in yield strength due to premature yielding. Accordingly the yield strength dropped from a value of 145,000 psi at 22°C to 84,200 at -78°C . If the former factor above continues to operate, the yield strength should be still lower at -196°C . However, the strengthening of the matrix with decrease in temperature can resist yielding and accordingly a slightly higher yield strength was observed

at -196°C . The presence of these effects is indicated by the curves that show the variation of percent martensite with strain (Fig. 21). There was no formation of martensite in the straight line portion of the stress-strain curve, but at the "yield point" there was formation of a large amount of martensite. Once a large amount of martensite was formed, there was rapid strain hardening and a high ultimate strength was recorded along with a low elongation. Figure 26 shows the stress-strain curves at different test temperatures for this alloy. The strain hardening exponents were as high as 1.28 and 1.1 for test temperatures -78°C and -196°C respectively.

Premature yielding was present in the alloys without carbon also. Alloys that were deformed above M_D and tested below M_D exhibited low yield strengths at -78°C and -196°C (Fig. 17). However, in these alloys the yield drop was not so serious as in the steel compositions. Moreover, no yield point resulted in these alloys due to premature yielding. There was gradual and continuous yielding. This was probably because the formation of iron-nickel martensite did not cause very high stresses in the austenite matrix and also, martensite in these alloys was more ductile and thus deformed without a sharp yield point.

In spite of a high ultimate strength and strain-hardening rate, premature yielding is not desirable as it causes considerable decreases in elongation and yield strength. The present investigation showed that there are three ways by which premature yielding can be resisted: These are (1) Increasing the stability of an alloy by increasing the nickel content - premature yielding was found in alloy b (8% Ni) whereas it was absent in alloy c (12% Ni). The addition of 4% nickel

decreased the M_D by about 140°C so that the stability at low temperatures was greatly enhanced. (2) Forming some martensite during prior deformation. The martensite that formed hardened the steel by a "dispersion hardening" effect. The amount of martensite was critical. Too little martensite was not effective (alloy b with prior deformation at 150°C) whereas too much martensite caused a considerable decrease in elongation (alloy c deformed at -120°C and tested at -78°C). (3) Causing optimum amount of precipitation during deformation. This was found in alloy b deformed at 250°C . Here premature yielding was resisted to some extent due to the strengthening of the matrix by precipitation of alloy carbides.

The first two factors are more effective than the last one.

SUMMARY AND CONCLUSIONS

In the carbon-containing alloys, the variation of yield strength with the temperature of prior deformation showed a maximum at a certain temperature of prior deformation. For the different alloys considered in this investigation the maxima occurred in the temperature range 250°C-350°C. These maxima corresponded to an optimum precipitate size and became less pronounced as the nickel content of the alloy was increased. The alloys without carbon did not exhibit the maxima.

Martensite formed when prior deformation was given below M_D . If the M_D was high, this caused an increase in yield strength on subsequent testing, without causing a large decrease in ductility. In steels with low M_D , however, there was a significant decrease in elongation when large amounts of martensite were formed during prior deformation.

The stability of austenite was an important factor that controls the mechanical properties. When austenite stability is very low, martensite can form at low stresses during testing, and formation of martensite can cause gross yielding of the tensile specimen. This is termed premature yielding and this results in a low elongation and high ultimate tensile strength with extraordinarily high strain hardening rates. Premature yielding during testing can be resisted by forming martensite during prior deformation, or by increasing the austenite stability by increasing the nickel content. With respect to the latter, it was found that a 4% increase in nickel content caused a decrease of about 140°C in the M_D temperature.

A difference of about 80°C-90°C between the test temperature and M_D was the most beneficial for mechanical properties.

The formation of martensite during testing started at a low rate initially. The rate gradually increased, reaching a maximum when about 10-15% martensite was formed, and then decreased. The initial low rate was not noticeable when the austenite stability was very low. Instead, a very high rate was observed at the "yield point" observed in the stress-strain curve.

In alloys that were deformed previously below M_D to form martensite, the variation of initial rate of martensite formation during testing showed a decrease with decrease in the temperature of prior deformation.

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Table I. CHEMICAL COMPOSITION OF THE ALLOYS

Alloy Series	Alloy No.	Alloy Designation	wt. % Cr	wt. % Ni	wt. % Mn	Wt. % C	wt. % Fe
I	691-12	b	9	8	2	0.325	Balance
	691-13	c	9	12	2	0.290	Balance
	691-14	d	9	16	2	0.292	Balance
	691-15	e	9	21.4	2	0.287	Balance
II	692-10	f	-	20	-	-	Balance
	692-12	h	-	29	-	-	Balance
	692-13	i	-	33.7	-	-	Balance
	692-14	j	-	37.8	-	-	Balance

Table II. PROPERTIES OF ALLOYS AT 22°C AFTER 70% PRIOR DEFORMATION

Alloy	% Ni	% C	Temperature of Prior Deformation °C	Y.S. 1000 psi	T.S. 1000 psi	<u>T.S.</u> Y.S.	% Elongation	Strength Coefficient A, 10 ⁵ psi	Strain Hardening Exponent, n	% Martensite Before Test	After Test
b	8	0.325	450	145	256	1.77	20.0	11.54	0.71	0.4	73.5
			300	178.5	248.5	1.39	16.6	6.50	0.44	0.6	61.5
			250	188	258.5	1.38	20.6	6.11	0.41	0.7	75
			200	165.2	232	1.41	22.3	6.48	0.52	0.8	63.5
			150	188	257	1.37	31.0	5.62	0.38	1.2	74.0
			25	281	282.5	1.01	22.6	5.90	0.34	32.8	68.0
c	12	0.29	450	180	180	1.00	5.7	-	-	0.2	1.4
			300	195	195	1.00	4.4	-	-	0.3	1.0
			250	182	182	1.00	3.6	-	-	0.2	0.9
			200	183.2	183.2	1.00	6.1	-	-	0.3	0.3
			150	182.4	182.4	1.00	3.6	-	-	0.3	0.4
			100	196	196	1.00	4.1	-	-	0.6	1.3
			25	199	199	1.00	3.8	-	-	1.8	7.5
			-120	293.8	293.8	1.00	2.4	-	-	32.5	32.9

Table III. PROPERTIES OF ALLOYS AT 22°C AFTER 70% PRIOR DEFORMATION

Alloy	% Ni	% C	Temperature of Prior Deformation °C	Y.S. 1000 psi	T.S. 1000 psi	<u>T.S.</u> Y.S.	% Elongation	Strength Coefficient 10 ⁵ psi	Strain Hardening Exponent, n	% Martensite Before Test	% Martensite After Test
d	16	0.292	450	173.5	173.5	1.0	4.0	-	-	0.1	0.2
			300	182.2	182.2	1.0	3.7	-	-	0.1	0.2
			200	171	171	1.0	3.3	-	-	0.1	0.2
			100	172	172	1.0	3.3	-	-	0.1	0.2
			25	175.5	175.5	1.0	3.3	-	-	0.3	0.3
			-78	197.5	197.5	1.0	2.3	-	-	0.5	0.7
			-120	201.5	201.5	1.0	2.9	-	-	1.6	1.8
e	21.4	0.287	450	169.7	169.7	1.0	2.7	-	-	-	0.2
			300	180.7	180.7	1.0	2.8	-	-	-	0.2
			200	153.8	153.8	1.0	2.9	-	-	-	0.2
			25	159.5	159.5	1.0	2.7	-	-	0.3	0.3
			-78	179.5	179.5	1.0	3.5	-	-	0.2	0.3
			-120	222	222	1.0	2.2	-	-	0.1	0.2

Table IV. PROPERTIES OF ALLOYS AT -78°C AFTER 70% PRIOR DEFORMATION

Alloy	% Ni	% C	Temperature of Prior Deformation °C	Y.S. 1000 psi	T.S. 1000 psi	$\frac{T.S.}{Y.S.}$	% Elongation	Strength Coefficient A, 10 ⁵ psi	Strain Hardening Exponent, n	% Martensite Before Test	% Martensite After Test
b	8	0.325	450	84.2	282.5	3.36	10.3	107.20	1.42	0.4	82.2
			300	148.8	302	2.03	21.4	16.00	0.77	0.1	87.1
			250	156.2	317	2.03	21.3	22.20	0.89	0.9	88.6
			200	128.5	283.5	2.21	14.0	27.60	0.96	0.9	85.5
			150	138.1	292.5	2.12	17.3	25.10	0.90	2.5	86.3
			25	269	328	1.22	15.2	8.07	0.37	33.7	89.5
c	12	0.29	450	196	222	1.13	42.9	6.06	0.81	0.2	69
			300	218.2	228.4	1.05	44.1	5.96	0.75	0.2	71
			250	185.4	197	1.06	33.2	5.86	0.59	0.2	73
			200	195	207.7	1.07	51.7	5.56	0.82	0.1	63.7
			150	194	205	1.06	48.6	5.46	0.79	0.2	69.1
			100	200	205.5	1.03	43.7	5.50	0.79	0.3	63
			25	213	213	1.00	36.8	5.46	0.69	1.9	60.3
			-120	294.5	315	1.07	2.5	-	-	38	41

Table V. PROPERTIES OF ALLOYS AT -78°C AFTER 70% PRIOR DEFORMATION

Alloy	% Ni	% C	Temperature of Prior Deformation °C	Y.S. 1000 psi	T.S. 1000 psi	$\frac{T.S.}{Y.S.}$	% Elongation	Strength Coefficient 10^5 A, psi	Strain Hardening Exponent, n	% Martensite Before Test	% Martensite After Test
d	16	.292	450	196	196	1.0	7.2	-	-	0.1	1.9
			300	193.8	193.8	1.0	4.2	-	-	0.1	1.1
			200	189	189	1.0	6.6	-	-	0.1	0.1
			100	178.5	178.5	1.0	5.5	-	-	0.1	1.4
			25	183.4	183.4	1.0	5.3	-	-	0.2	1.6
			-78	206.6	206.6	1.0	4.6	-	-	0.6	2.2
			-120	217	217	1.0	4.7	-	-	2.0	3.7
e	21.4	.287	450	186	186	1.0	3.7	-	-	0.2	0.2
			300	191.5	191.5	1.0	4.2	-	-	0.4	0.4
			200	171	171	1.0	3.2	-	-	0.2	0.2
			25	176	176	1.0	3.8	-	-	0.4	0.4
			-78	189.5	189.5	1.0	3.8	-	-	-	-
			-120	198	198	1.0	3.2	-	-	0.4	0.6

Table VI. PROPERTIES OF ALLOYS AT -196°C AFTER 70% PRIOR DEFORMATION

Alloy	% Ni	% C	Temperature of Prior Deformation °C	Y.S.	T.S.	T.S.	%	Strength	Strain	% Martensite	
				1000 psi	1000 psi	Y.S.	Elongation	Coefficient A, 10 ⁵ psi	Hardening Exponent, n	Before Test	After Test
d	16	.292	450	228.6	236.5	1.03	37.0	4.53	0.29	-	57
			300	233.4	233.4	1.00	25.7	3.92	0.20	-	45.3
			200	230	232	1.03	34.2	4.32	0.27	-	45.7
			25	220	224	1.02	21.4	3.43	0.15	0.8	35.8
			-78	272	272	1.00	8.6	-	-	1.4	17.9
			-120	348.5	348.5	1.00	8.4	-	-	1.1	16.6
e	21.4	.287	450	193	220.5	1.14	12.7	3.34	0.07	0.3	2.1
			300	197.2	220.5	1.12	8.3	2.56	0.03	0.3	2.1
			200	186.4	207	1.11	1.11	2.61	0.06	0.3	1.0
			25	187.8	215.8	1.15	12.5	2.75	0.06	0.3	2.5
			-78	223	236	1.06	6.7	-	-	0.2	2.2
			-120	268	296	1.11	9.7	-	-	0.2	0.9

For alloy e yield strength refers to 0.1% offset.

Table VII. PROPERTIES OF ALLOYS AFTER 70% DEFORMATION

ALLOY b		8 % Ni		0.325 % C		9% Cr		2% Mn	
Test Temperature °C	Temperature of Prior Deformation °C	Y.S. 1000 psi	T.S. 1000 psi	T.S. Y.S.	% Elongation	Strength Coefficient 10^5 ^A psi	Strain Hardening Exponent, n	% Martensite Before Test	After Test
22	450	145	256	1.77	20	11.54	0.71	0.4	73.5
	300	178.5	248.5	1.39	16.6	6.50	0.44	0.6	61.5
	250	188	258.5	1.38	20.6	6.11	0.41	0.7	75
	200	165.2	232	1.41	22.3	6.48	0.52	0.8	63.5
	150	188	257	1.37	31	5.62	0.38	3.5	74
	25	281	282.5	1.01	22.6	5.90	0.34	32.8	68
-78	450	84.2	282.5	3.36	10.3	107.20	1.42	0.4	82.2
	300	148.8	302	2.03	21.4	16.00	0.77	0.1	87.1
	250	156.2	317	2.03	21.3	22.20	0.89	0.9	88.6
	200	128.5	283.5	2.21	14.0	27.60	0.96	0.9	85.5
	150	138.1	292.5	2.12	17.3	25.10	0.90	2.5	86.3
	25	269	328	1.22	15.2	8.07	0.37	33.7	89.5
-196	450	116	334.7	2.89	9.9	55.60	1.24	0.2	85.7
	300	162	264.5	1.63	8.7	50.60	1.15	0.8	69.5

Table VIII. PROPERTIES OF ALLOYS AFTER 70% DEFORMATION

ALLOY c		12% Ni		0.29% C		9% Cr		2% Mn	
Test Temperature °C	Temperature of Prior Deformation °C	Y.S. 1000 psi	T.S. 1000 psi	T.S. Y.S.	% Elongation	Strength Coefficient 10^5 A psi	Strain Hardening Exponent, n	% Martensite Before Test	% Martensite After Test
22	450	180	180	1.00	5.7	-	-	0.2	1.4
	300	195	195	1.00	4.4	-	-	0.3	1.0
	250	182	182	1.00	3.6	-	-	0.2	0.9
	200	183.2	183.2	1.00	6.1	-	-	0.3	0.3
	150	182.4	182.4	1.00	3.6	-	-	0.3	0.4
	100	196	196	1.00	4.1	-	-	0.6	1.3
	25	199	199	1.00	3.8	-	-	1.8	7.5
	-120	293.8	293.8	1.00	2.4	-	-	34.1	32.9
-78	450	196	222	1.13	42.9	6.06	0.81	0.2	69
	300	218.2	228.4	1.05	44.1	5.96	0.75	0.2	71
	250	185.4	197	1.06	33.2	5.86	0.59	0.2	53
	200	195	207.7	1.07	51.7	5.56	0.82	0.1	63.7
	150	194	205	1.06	48.6	5.46	0.79	0.2	69.1
	100	200	205.5	1.03	43.7	5.50	0.79	0.3	63
	25	213	213	1.00	36.8	5.46	0.69	1.9	60.3
	-120	294.5	316	1.07	2.5	-	-	38	41

Table IX: PROPERTIES OF ALLOYS AFTER 70% DEFORMATION

ALLOY d 16 % Ni 0.292% C 9% Cr 2% Mn									
Test Temperature °C	Temperature of Prior Deformation °C	Y.S. 1000 psi	T.S. 1000 psi	$\frac{T.S.}{Y.S.}$	% Elongation	Strength Coefficient 10^5 A psi	Strain Hardening Exponent, n	% Martensite Before Test	% Martensite After Test
22	450	173.5	173.5	1.0	4.0	-	-	0.1	0.2
	300	182.2	182.2	1.0	3.7	-	-	0.1	0.2
	200	171	171	1.0	3.3	-	-	0.1	0.2
	100	172	172	1.0	3.3	-	-	0.1	0.2
	25	175.5	175.5	1.0	3.3	-	-	0.3	0.3
	-78	197.5	197.5	1.0	2.3	-	-	0.5	0.7
	-120	201.5	201.5	1.0	2.9	-	-	1.6	1.8
-78	450	196	196	1.0	7.2	-	-	0.1	1.9
	300	193.8	193.8	1.0	4.2	-	-	0.1	1.1
	200	189	189	1.0	6.6	-	-	0.1	0.1
	100	178.5	178.5	1.0	5.5	-	-	0.1	1.4
	25	183.4	183.4	1.0	5.3	-	-	0.2	1.6
	-78	206.6	206.6	1.0	4.6	-	-	0.6	2.2
	-120	217	217	1.0	4.7	-	-	2.0	3.7

Table IX. (Continued) PROPERTIES OF ALLOYS AFTER 70% DEFORMATION

ALLOY d 16% Ni 0.292% C 9% Cr 2% Mn									
Test Temperature °C	Temperature of Prior Deformation °C	Y.S. 1000 psi	T.S. 1000 psi	$\frac{T.S.}{Y.S.}$	% Elongation	Strength Coefficient $A \cdot 10^5$ psi	Strain Hardening Exponent, n	% Martensite Before Test	% Martensite After Test
-196	450	228.6	236.5	1.03	37.0	4.53	0.29	-	57
	300	233.4	233.4	1.00	25.7	3.92	0.2	-	45.3
	200	230	232	1.03	34.2	4.32	0.27	-	45.7
	25	220	224	1.02	21.4	3.43	0.15	0.8	35.8
	-78	272	272	1.00	8.6	-	-	1.4	17.9
	-120	348.5	348.5	1.00	8.4	-	-	1.1	16.6

Table X. PROPERTIES OF ALLOYS AFTER 70% DEFORMATION

ALLOYS e		21.4% Ni	0.287% C	9% Cr	2% Mn				
Test Temperature °C	Temperature of Prior Deformation °C	Y.S. 1000 psi	T.S. 1000 psi	T.S. Y.S.	% Elongation	Strength Coefficient 10^5 psi	Strain Hardening Exponent, n	% Martensite Before Test	% Martensite After Test
22	450	169.7	169.7	1.0	2.7	-	-	-	0.2
	300	180.7	180.7	1.0	2.8	-	-	-	0.2
	200	153.8	153.8	1.0	2.9	-	-	-	0.2
	25	159.5	159.5	1.0	2.7	-	-	0.3	0.3
	-78	179.5	179.5	1.0	3.5	-	-	0.2	0.3
	-120	222	222	1.0	2.2	-	-	0.1	0.2
-78	450	186	186	1.0	3.7	-	-	-	0.2
	300	191.5	191.5	1.0	4.2	-	-	-	0.4
	200	171	171	1.0	3.2	-	-	-	0.2
	25	176	176	1.0	3.8	-	-	0.4	0.4
	-78	189.5	189.5	1.0	3.8	-	-	-	-
	-120	198	198	1.0	3.2	-	-	0.4	0.6

Table X. (Continued) PROPERTIES OF ALLOYS AFTER 70% DEFORMATION

ALLOY e		21.4% Fe		0.237% C		0% Cr		2% Mn	
Test Temperature °C	Temperature of Prior Deformation °C	Y.S. 1000 psi	T.S. 1000 psi	T.S. Y.S.	% Elongation	Strength Coefficient 10^5 A psi	Strain Hardening Exponent, n	% Martensite Before Test	% Martensite After Test
-196*	450	193	220.6	1.14	12.7	3.34	0.07	0.3	2.1
	300	197.2	220.6	1.12	8.3	2.56	0.03	0.3	2.1
	200	186.4	207	1.11	11.1	2.61	0.06	0.3	1.0
	25	187.8	215.8	1.15	12.5	2.75	0.06	0.3	2.5
	-78	223	236	1.06	6.7	-	-	0.1	2.2
	-120	268	296	1.11	9.7	-	-	0.1	0.9

* Yield strength refers to 0.1% offset.

Table XI. PROPERTIES OF ALLOYS AT 22°C AFTER 70% PRIOR DEFORMATION

Alloy	% Ni	% C	Temperature of Prior Deformation °C	Y.S.* 1000 psi	T.S. 1000 psi	$\frac{T.S.}{Y.S.}$	% Elongation	Strength Coefficient 10^5 psi	Strain Hardening Exponent, n
f	20	-	25	160	164	1.03	2.8	-	-
h	29	-	25	104.6	108	1.03	18.4	1.8	0.19
i	33.7	-	-120	143.7	153	1.03	2.7	-	-
			-78	135	139.5	1.03	2.5	-	-
			25	97.5	103	1.06	4.4	-	-
			100	97.25	101	1.04	5.0	-	-
			200	95.9	100.5	1.04	6.2	-	-
j	37.8	-	-120	102.4	107.7	1.05	2.7	-	-
			-78	101	106.6	1.06	3.2	-	-
			25	97.2	100.8	1.04	4.2	-	-

* 0.1% Offset

Table XII. PROPERTIES OF ALLOYS AT -78°C AFTER 70% PRIOR DEFORMATION

Alloy	% Ni	% C	Temperature of Prior Deformation °C	Y.S.* 1000 psi	T.S. 1000 psi	$\frac{T.S.}{Y.S.}$	% Elongation	Strength Coefficient 10^5 psi	Strain Hardening Exponent, n
f	20	-	25	163.2	175	1.07	3.5	-	-
h	29	-	25	82.5	141.8	1.72	8.2	-	-
i	33.7	-	-120	161	165	1.03	3.0	-	-
			-78	139	142	1.02	3.6	-	-
			25	30.5	110.7	1.40	18.0	1.66	0.13
			100	90.75	109.5	1.21	17.5	1.83	0.18
			200	101.9	112.5	1.10	20.0	2.09	0.24
j	37.8	-	-120	115.5	127.2	1.10	5.6	-	-
			-78	114	124.6	1.19	4.1	-	-
			25	110	119.4	1.09	7.5	-	-

* 0.1% Offset

Table XIII. PROPERTIES OF ALLOYS AT -196°C AFTER 70% PRIOR DEFORMATION

Alloy	% Ni	% C	Temperature of Prior Deformation °C	Y.S.* 1000 psi	T.S. 1000 psi	$\frac{T.S.}{Y.S.}$	% Elongation	Strength Coefficient 10^5 A psi	Strain Hardening Exponent, n
f	20	-	25	198.5	208.8	1.05	4.1	-	-
h	29	-	25	107.5	187	1.74	9.9	-	-
i	33.7	-	-120	196	196	1.00	9.8	1.24	0.07
			-78	121.4	183.2	1.51	15.6	3.05	0.18
			25	73.5	165.8	2.26	17.7	2.69	0.16
			100	72.65	167	2.30	17.4	2.66	0.16
			200	73.2	170	2.32	17.7	2.84	0.18
j	37.8	-	-120	150.4	156.2	1.04	24.4	2.73	0.22
			-78	144	154.8	1.08	25.2	2.97	0.27
			25	144	144	1.00	25.7	2.46	0.21

* 0.1% Offset

FIGURE CAPTIONS

- Fig. 1. Tensile specimen used in the determination of tensile properties.
- Fig. 2. Variation of yield strength of alloy b at different test temperatures as a function of the temperature of prior deformation.
- Fig. 3. Variation of yield strength of alloy c at different test temperatures as a function of the temperature of prior deformation.
- Fig. 4. Variation of yield strength of alloy d at different test temperatures as a function of the temperature of prior deformation.
- Fig. 5. Variation of yield strength of alloy e at different test temperatures as a function of the temperature of prior deformation.
- Fig. 6. Variation of yield strength of alloy i at different test temperatures as a function of the temperature of prior deformation.
- Fig. 7. Ultimate tensile strength of alloy b at different test temperatures plotted against the temperature of prior deformation.
- Fig. 8. Ultimate tensile strength of alloy c at different test temperatures plotted against the temperature of prior deformation.
- Fig. 9. Ultimate tensile strength of alloy d at different test temperatures plotted against the temperature of prior deformation.
- Fig. 10. Ultimate tensile strength of alloy e at different test temperatures plotted against the temperature of prior deformation.
- Fig. 11. Ultimate tensile strength of alloy i at different test temperatures plotted against the temperature of prior deformation.
- Fig. 12. Percent elongation as a function of the temperature of prior deformation for alloy b tested at different temperatures.
- Fig. 13. Percent elongation as a function of the temperature of prior deformation for alloy c tested at different temperatures.

Fig. 14. Percent elongation as a function of the temperature of prior deformation for alloy d tested at different test temperatures.

Fig. 15. Percent elongation as a function of the temperature of prior deformation for alloy e at different test temperatures.

Fig. 16. Percent elongation as a function of the temperature of prior deformation for alloy i at different test temperatures.

Fig. 17. Yield strength of carbonless alloys as a function of test temperature. Prior deformation: 70% at 25°C.

Fig. 18. Ultimate tensile strength of carbonless alloys as a function of test temperature. Prior deformation: 70% at 25°C.

Fig. 19. Percent elongation of carbonless alloys as a function of test temperature. Prior deformation: 70% at 25°C.

Fig. 20. Variation of volume percent martensite with percent strain for alloy c with a prior deformation of 70% at 200°C. Test temperature: -78°C.

Fig. 21. Variation of volume percent martensite with percent strain for alloy b tested at different test temperatures. Prior deformation: 70% at 450°C.

Fig. 22. Volume percent martensite vs. percent strain for alloy b with prior deformation at temperatures indicated, and tested at 22°C.

Fig. 23. Optical micrographs of alloy h with a prior deformation of 70% at 25°C.

(a) Before room temperature test. Magnification 250x.

(b) After room temperature test. Magnification 250x.

(c) Cooled to -78°C after prior deformation. Magnification 250x.

(d) Cooled to -196°C after prior deformation. Magnification 250x.

Fig. 24. Optical micrographs of alloy i with a prior deformation of 70% at 200°C.

(a) After testing at -78°C. Magnification 250x.

(b) Cooled to -196°C after prior deformation. Magnification 250x.

Fig. 25. Tensile properties of alloy b as a function of test temperature. Prior deformation: 70% at 450°C.

Fig. 26. Engineering stress-strain curves of alloy b deformed previously at 450°C and tested at different test temperatures.

Fig. 27. Engineering stress-strain curve of alloy c deformed previously at 200°C and tested at -78°C.

GAGE LENGTH: 1"
THICKNESS: 0.05"

THICKNESS: 0.05"

TENSILE SPECIMEN

XBL 705-911

Fig. 1

XRL 705-912

Fig. 2

XBL 705-913

Fig. 3

XBL 705-914

Fig. 4

XBL 705-915

Fig. 5

XBL 705-916

Fig. 6

XBL 705-917

Fig. 7

XBL 705-918

Fig. 8

XBL 705-919

Fig. 9

XBL 705-920

Fig. 10

XBL 705-921

Fig. 11

XBL 705-922

Fig. 12

XBL 705-923

Fig. 13

XBL 705-924

Fig. 14

XBL 705-925

Fig. 15

XBL 705-926

Fig. 16

XBL 705-927.

Fig. 17

XBL 705-928

Fig. 18

XBL 705-929

Fig. 19

XBL 705-930

Fig. 30

XBL 705-931

Fig. 21

XBL 705-948

Fig. 22

(a)

(b)

(c)

(d)

XBB 706-2504

Fig. 23

(a)

(b)

XBB 705-2410

Fig. 24

XBL 705-947

Fig. 25

XBL 705-946

Fig. 26

XBL 705-945

Fig. 27

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