# Ductility of Ferritic Stainless Weld Metal

## Weld ductility can be explained in terms of the relative values of the yield stress and fracture stress, each of which is determined by composition and cooling rate

BY B. POLLARD

ABSTRACT. The ductility of GTA welds in Fe-26Cr and Fe-35Cr alloys containing 0.005% C and 0.010% N was studied by means of tensile tests on all weld metal tensile specimens at temperatures within the range −240F to +300F. Metallographic techniques were used to obtain an understanding of the mechanisms responsible for fracture initiation in Fe-Cr welds and their effect upon weld ductility.

Weld ductility can be explained in terms of the relative values of the yield stress, $\sigma_Y$, and the fracture stress, $\sigma_F$, each of which is a function of composition and cooling rate. Increasing the Cr level of the alloy from 26 to 35% severely reduces weld ductility because it both increases $\sigma_Y$ and reduces $\sigma_F$, primarily by substitutional hardening, although increasing the Cr level also promotes "885" embrittlement. Additions of Ti, Zr or Cb to iron-chromium-carbon-nitrogen alloys reduce $\sigma_Y$ and increase $\sigma_F$ by precipitating carbon and nitrogen, thus for a given ductility level increasing the tolerance of the welds for interstitials.

The effects of postweld heat treatment can be explained by the $\alpha + \alpha'$ phase separation around 885F and the tendency to form grain boundary carbonitride films during slow cooling. A postweld anneal of ½ hour at 1560F, followed by a water quench, removes the $\alpha + \alpha'$ formed during the weld cycle, thus reducing $\sigma_Y$, increasing $\sigma_F$ and increasing weld ductility. Furnace cooling after a postweld anneal increases $\sigma_Y$ by $\alpha + \alpha'$ phase separation and reduces $\sigma_F$ by formation of a grain boundary carbonitride film, thus reducing weld ductility. Increasing the weld heat input increases $\sigma_Y$ slightly but has negligible effect upon weld ductility at room temperature.

### Introduction

It is well known that increasing the chromium level of iron-chromium alloys increases both oxidation[1] and corrosion[1-4] resistance but the application of high chromium ferritic stainless steels such as type 446 (26% Cr) has been severely limited by their poor weldability. Preheating prevents cracking as the weld cools [5, 6] but welds in type 446 made with matching weld

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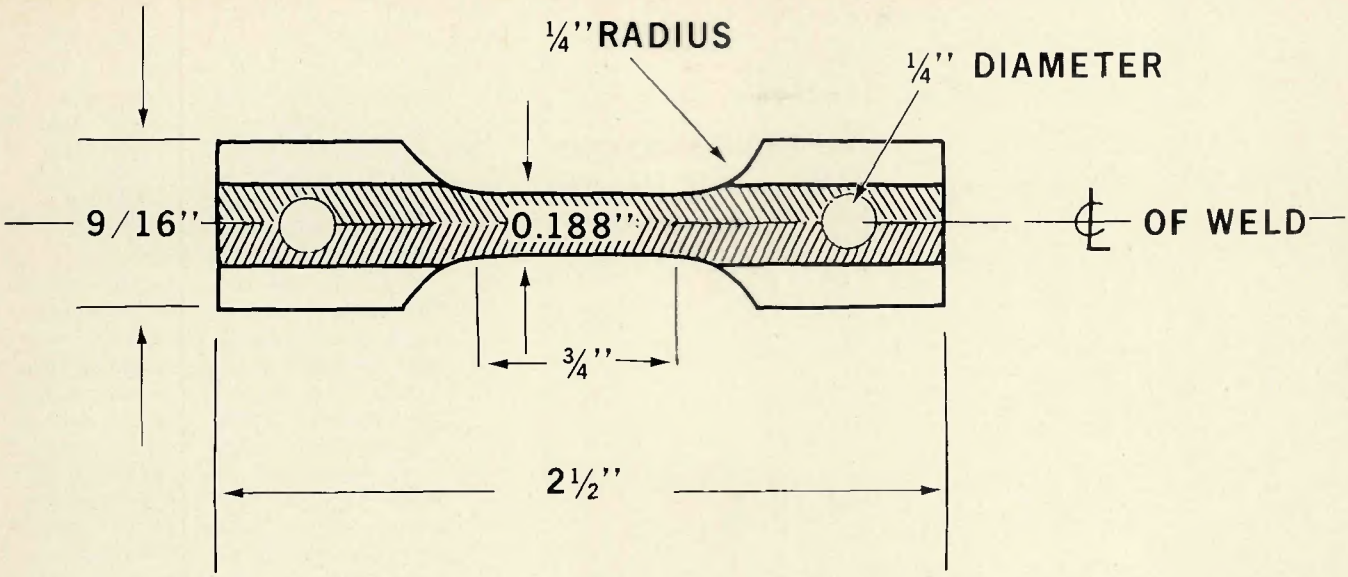


Fig. 1—Weld metal tensile specimen

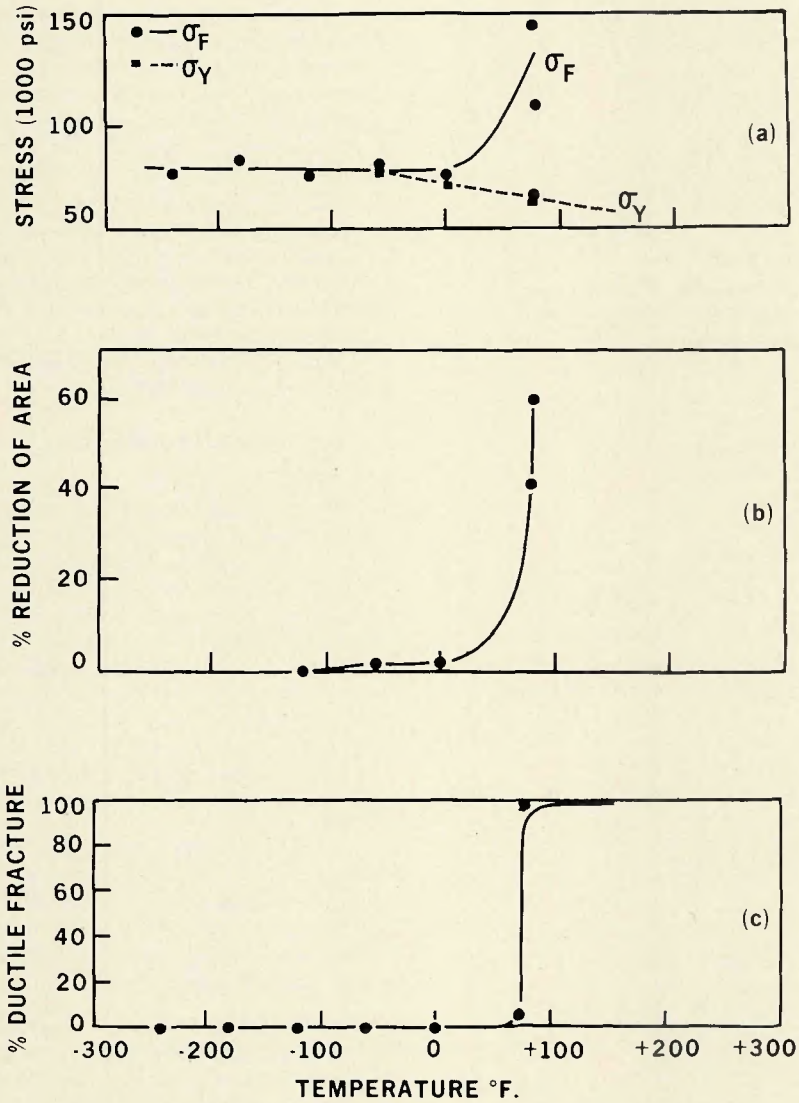


Fig. 2—Tensile properties of Fe-26Cr welds versus test temperature: (a) fracture stress ($\sigma_F$) and yield stress ($\sigma_Y$); (b) percent reduction of area; and (c) percent ductile fracture

metal are completely brittle in the "as-welded" condition and ductility is only partly restored by a postweld annealing treatment.[5-7] The low ductility of type 446 weld metal has been attributed to its large grain size[5-8] but studies of base material mechanical properties[9-11] indicate that the interstitial level could be equally important.

Steels containing about 20% or more chromium and 0.05% carbon are completely ferritic at all temperatures up to the melting point.[12] The absence of a phase transformation means that grain refinement can only be produced by working and recrystallization. However, these steels are by no means single phased, since carbides and nitrides can precipitate upon cooling,[10-12] the preferred locations for precipitates being high energy sites such as grain boundaries where they may be expected to have strong influence on the ductility of the material.

At temperatures in the vicinity of 885F another precipitate is formed which causes hardening and embrittlement. The nature of this reaction is still not completely understood but the most recent study indicates that the embrittlement results from a separation into iron-rich ($\alpha$) and chromium-rich ($\alpha'$) phases.[13] A fourth phase, sigma, can form at temperatures of 1000 to 1500F[14] but times in excess of 10 hours are required for its separation so it is of no concern in welding.

Since it is not possible to refine the grain size in iron-chromium welds by heat treatment, lowering the interstitial level appears to be the most promising approach for improving weld ductility. Several new steelmaking processes, most involving vacuum

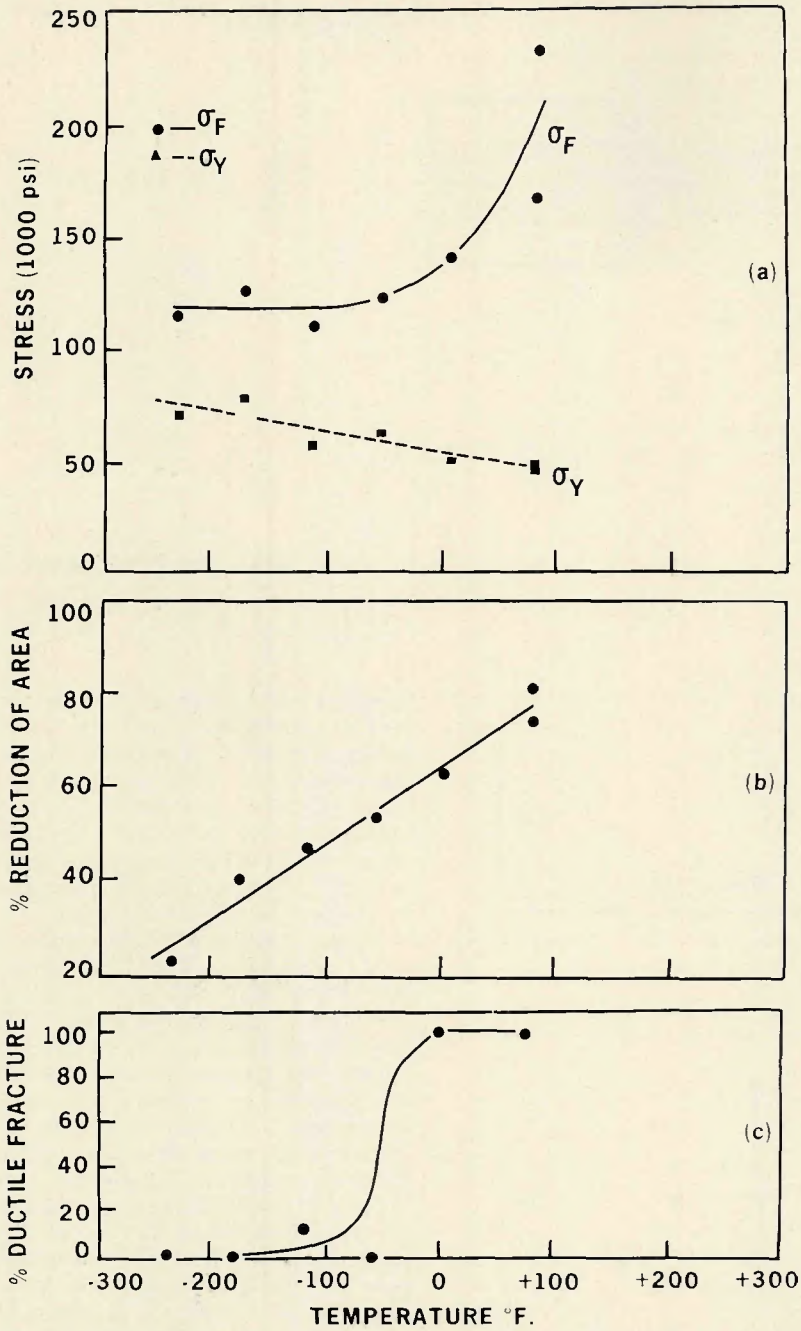


Fig. 3—Tensile properties of Fe-35Cr welds versus test temperature: (a) fracture stress ($\sigma_F$) and yield stress ($\sigma_Y$); (b) percent reduction of area; and (c) percent ductile fracture

**Table 1—Chemical Analysis of Heats for Weld Ductility Study**

| Heat No. | C | N | Cr | Mn | Si | S | P | Ti | Zr | Cb |
|---|---|---|---|---|---|---|---|---|---|---|
| 1913 | .004 | .005 | 26.00 | .09 | .40 | .012 | .005 | | | |
| 1918 | .003 | .009 | 26.62 | .08 | .33 | .007 | .006 | | | |
| 1914 | .004 | .008 | 34.91 | .18 | .56 | .014 | .013 | | | |
| 1917 | .005 | .006 | 34.88 | .08 | .76 | .012 | .010 | | | |
| 2400 | .010 | .019 | 25.65 | .22 | .015 | .018 | .012 | | | |
| 2401 | .009 | .020 | 25.76 | .22 | .013 | .018 | .012 | .17 | | |
| 2402 | .007 | .017 | 25.89 | .23 | .012 | .016 | .013 | .41 | | |
| 2403 | .007 | .019 | 25.82 | .23 | .013 | .017 | .014 | | .23 | |
| 2404 | .010 | .013 | 25.82 | .22 | .012 | .014 | .012 | | .74 | |
| 2405 | .010 | .017 | 25.68 | .21 | .014 | .021 | .014 | | | .32 |
| 2406 | .013 | .022 | 25.46 | .21 | .014 | .021 | .015 | | | .89 |

refining, are now available or under development; these will produce steels with much lower interstitial levels than are attainable using the electric furnace, thus greatly increasing the feasibility of making weldable high chromium ferritic stainless steels. However, a low interstitial level alone may not be sufficient to produce ductile welds. The objective of the present investigation was therefore to identify and explain the key parameters affecting the ductility of welds in high chromium, low interstitial ferritic stainless steels.

In general, the ductility of a metal, or the amount of plastic deformation which precedes fracture, is determined by the relative values of the yield stress, $\sigma_Y$, and the fracture stress, $\sigma_F$, each of which is a function of temperature, grain size and extent of both substitutional and interstitial alloying and microstructural features such as second phase morphology.[15] The tensile properties of welds in Fe-26Cr and Fe-35Cr alloys with very low interstitial levels (approximately 0.005 carbon and less than 0.010 nitrogen) were first measured as functions of temperature. After establishing the general behavior pattern for Fe-26Cr and Fe-35Cr welds, the effects of weld heat input, postweld heat treatment and alloying additions on weld ductility were investigated. Chemical analysis and metallographic techniques helped obtain an understanding of the mechanisms responsible for fracture initiation in iron-chromium welds and their effect upon weld ductility.

## Experimental Procedure

### Material

The chemistries of heats used in the investigation are listed in Table 1. The alloys were produced as 25-lb vacuum melts. All material was hot rolled to 0.250 in. thick hot band, cold rolled to 0.100 or 0.188 in. air annealed for 30 min at 1560F and water quenched. Each piece was grit blasted to remove oxide, leveled, and pieces 6 in. by $1^1/_2$ in. were cut with rolling direction parallel to the length of the specimen.

### Welding

Full penetration autogenous gas tungsten-arc (GTA) welds were made with sufficient width to allow 0.188 in. wide tensile specimens to be cut from the weld. Typical welding conditions are given in Table 2.

### Tensile Testing

Tensile specimens with ¾ in. gauge lengths were machined from the welds with the specimen axis coincident with the weld axis, as shown in Fig. 1. Testing was performed over the temperature range −240F to +300F at a strain rate of 0.13 in./in./min. Yield

stress, twinning stress, fracture stress and percent reduction of area were calculated. The cross-sectional area of the specimen at fracture was measured photographically. The percent ductile fracture was estimated by examining the fracture surface at 25X with a stereomicroscope. The percent reduction of area was used in preference to the total elongation or uniform elongation as a measure of ductility because it showed a stronger temperature dependence.

All testing was performed on an Instron machine fitted with a furnace for elevated temperatures and a chamber cooled with liquid nitrogen for low temperature testing.

#### Precipitated and Soluble C and N

The form of the carbon and nitrogen, whether dissolved or precipitated, was determined for welds and base metal. Precipitated carbon was determined by electrolytically dissolving the sample in 10% HCl using a current density of 0.025 amp/sq in. The residue was filtered off and its carbon content measured by Leco combustion. Total carbon was measured by Leco combustion of a solid sample. Dissolved carbon was then determined by difference.

For precipitated nitrogen determinations, the sample was dissolved in bromine and methyl acetate and the nitrogen in the residue measured by the micro-Kjeldahl method. Total nitrogen was then measured by the macro-Kjeldahl method and dissolved nitrogen determined by difference.

#### Metallography

Longitudinal sections of broken tensile specimens were prepared using conventional metallographic procedures and etched in Glyceregia. Specimens were examined at magnifications up to 1000X for microstructural features likely to have an effect upon the fracture process, such as precipitate morphology, inclusions and deformation twins. In addition, a metallographic examination was made of tensile specimens which had been strained until necking occurred and then unloaded.

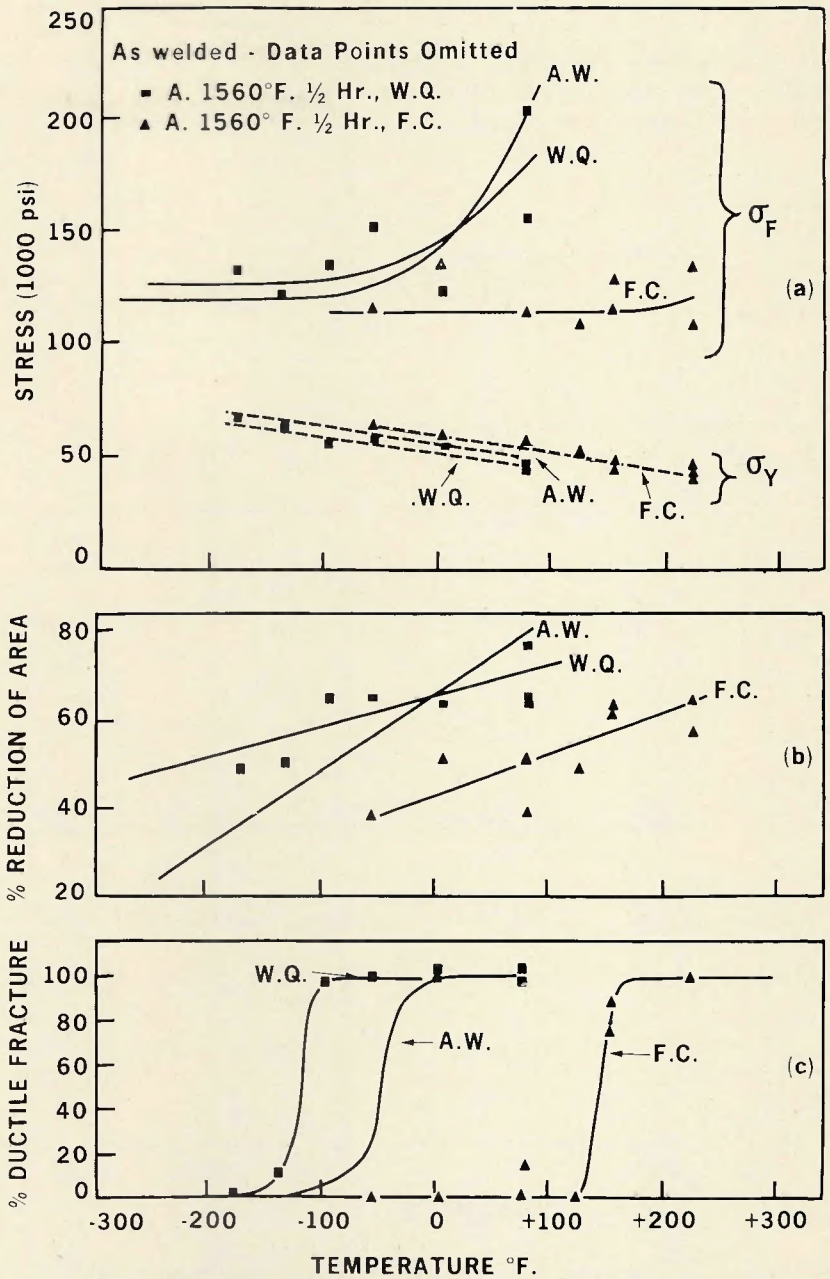


Fig. 4—Effect of postweld heat treatment on tensile properties of Fe-26Cr welds: (a) fracture stress ($\sigma_F$) and yield stress ($\sigma_Y$); (b) percent reduction of area; and (c) percent ductile fracture

## Results

Fracture stress, $\sigma_F$, yield stress, $\sigma_Y$, percent reduction of area and percent ductile fracture are plotted as functions of temperature for Fe-26Cr and Fe-35Cr alloys, as-welded, in Figs. 2 and 3 respectively. The fracture stress versus temperature curves for each alloy consist of two regions: a low temperature portion where fracture occurs in a brittle manner (the cleavage fracture stress being essentially independent of temperature) and a high temperature portion where the fracture stress increases sharply with increase in temperature (the fracture occurring in a completely ductile fashion).

The yield stress for both alloys decreased steadily with increase in temperature. The percent reduction of area for the Fe-26Cr alloy decreased steadily with decrease in temperature from 79% at +75F to 25% at −240F (Fig. 2b) whereas the percent ductile fracture went through a rather sharp transition from 100% ductile to 100% brittle fracture between zero and −60F (Fig. 2c).

Increasing the chromium level from 26% to 35% reduced the cleavage fracture stress from 120 ksi to 76 ksi and simultaneously increased the yield stress by 12 to 14 ksi. The resultant percent reduction of area fell off rapidly with decrease in temperature below +75F (Fig. 3b) and the fracture appearance showed a very sharp ductile-brittle transition (Fig. 3c). The level of chromium had only a small effect upon the temperature dependence of yield stress; $\Delta\sigma_Y/\Delta T \doteq 92$ psi/°F for Fe-26Cr and 80 psi/°F for the Fe-35Cr.

The yield stresses, plotted in Fig. 2 and 3, are for yielding by slip. Sudden

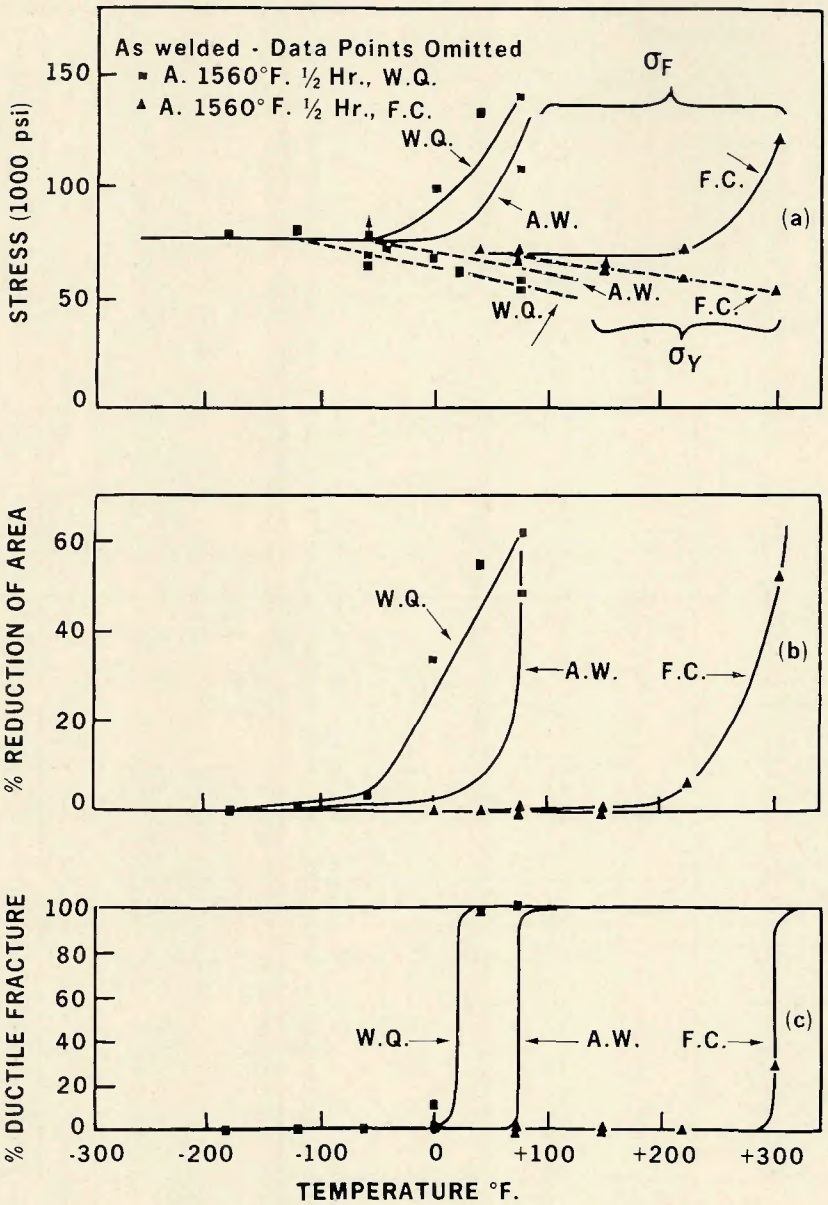


Fig. 5—Effect of postweld heat treatment on tensile properties of Fe-35Cr welds: (a) fracture stress ($\sigma_F$) and yield stress ($\sigma_Y$); (b) percent reduction of area; and (c) percent ductile fracture

load drops during tensile testing, accompanied by loud clicks, indicated that mechanical twinning was also occurring; for Fe-26Cr welds at test temperatures of −120F and lower; for Fe-35Cr welds at test temperatures of +75F and lower. When twinning occurred at stresses less than the 0.2% offset yield stress, the latter was obtained by extrapolation.

### Effect of Heat Input on Weld Tensile Properties

Room temperature tensile properties of welds made with heat inputs of approximately 10 and 25 K-joules/inch are listed in Table 3. Increasing the heat input from 10 to 25 K-joules/inch increased the yield strength of the Fe-26Cr alloy by 2 ksi and reduced the fracture stress from 209 to 131 ksi. However, the fracture was still 100% ductile and only a small decrease in reduction of area (from 79 to 73%) resulted. Similarly, the yield strength of the Fe-35Cr alloy increased with increased heat input, but fracture stress, reduction of area and ductile fracture were not significantly changed.

### Effect of Postweld Heat Treatment

For the Fe-26Cr alloy, a postweld anneal of ½ hour at 1560F, followed by a water quench, increased the brittle fracture stress by about 6 ksi and reduced the yield stress by about 3 ksi (Fig. 4a). This increased the reduction of area at temperatures below −10F, the improvement increasing as the temperature fell (Fig. 4b), while the minimum temperature for 100% ductile fracture fell from 0°F to −80F (Fig. 4c).

Annealing the welds for ½ hour at 1560F, followed by a furnace cool to room temperature (average cooling rate 180F/hr between 1560F and 600F) increased the yield stress by 2 to 3 ksi and reduced the cleavage stress (Fig. 4a). This resulted in a decrease in percentage reduction of area (Fig. 4b) and an increase in the minimum temperature for 100% ductile fracture of approximately 160F from 0°F to +160F, (Fig. 4c).

Similar results were obtained for the Fe-35Cr alloy. Welds annealed at 1560F for ½ hour and water quenched showed a reduction in yield stress of 4 to 7 ksi plus a slight increase in the ductile fracture stress (Fig. 5a). This resulted in an increase in reduction of area (Fig. 5b), and a reduction in the ductile-brittle transition temperature of approximately 60F (Fig. 5c).

Annealing and furnace cooling Fe-35Cr welds using the same thermal cycle as for the Fe-26Cr welds increased the yield stress by 7 to 8 ksi and reduced the cleavage fracture stress (Fig. 5a). This caused severe embrittlement, the temperature for 50% reduction of area and 50% ductile fracture increasing from +75 to +300F, as shown in Fig. 5b and 5c.

### Effect of Ti, Zr and Cb on Weld Ductility

Since these elements are strong carbide and nitride formers, it was felt that their addition might improve weld ductility by removing carbon and nitrogen from solution (both are strong ferrite strengtheners). The base alloy for this series of heats was Fe-26Cr containing approximately 0.01% C and 0.02% N. The room temperature weld tensile properties are listed in Table 4. Heat 2400, without any additives and with high interstitial level (0.01C + 0.02N) possessed essentially no ductility at room temperature, with 0 to 1% reduction of area and 100% cleavage fracture. Compared with the low interstitial heat (Heat 1913), the yield stress was increased by 5 ksi and the fracture stress fell to 40 to 57 ksi, the latter a reduction of 150 to 170 ksi.

The addition of 0.17 Ti (Heat 2401) reduced the yield strength by 6 to 7 ksi, increased the fracture stress 112 to 190 ksi, the reduction of area to approximately 70% and the percent ductile fracture to between 50 and 100%. Increasing the titanium level to 0.41% reduced the yield strength

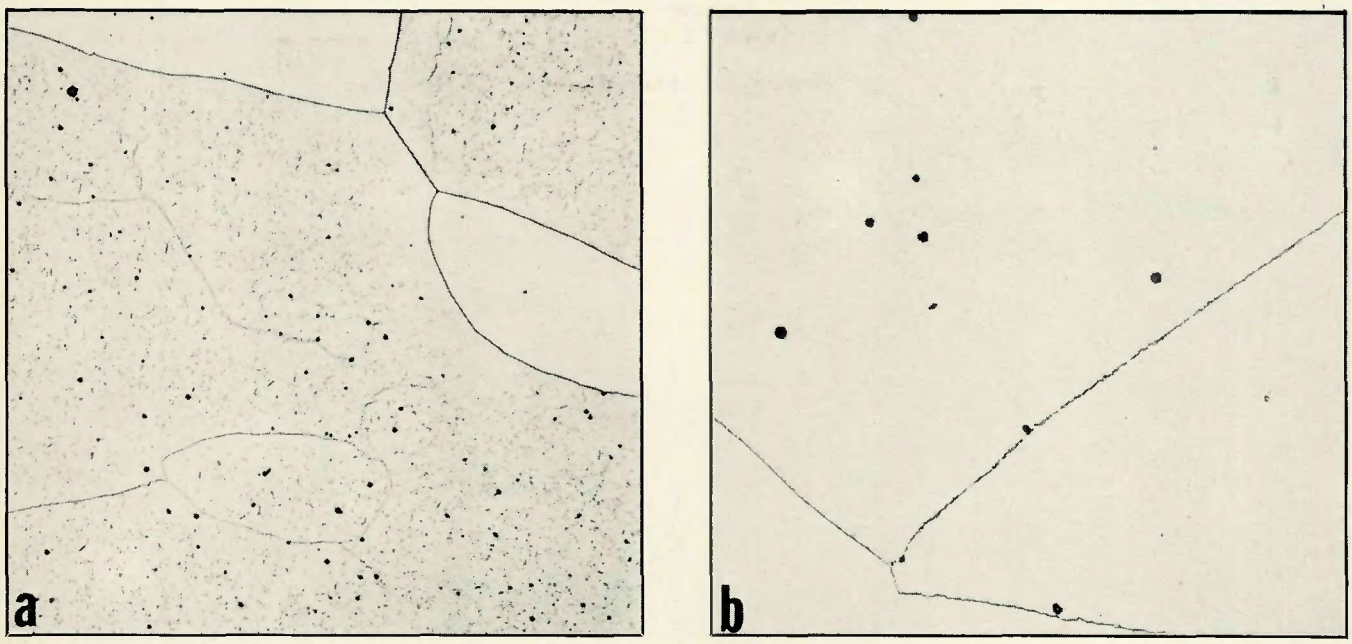


Fig. 6—Microstructure of Fe-26Cr weld metal: (a) as-welded and (b) annealed 1560F, ½ hr, FC. Etched in glyceregia. Mag: 500X

another 4 to 5 ksi but the fracture stress fell to about 70 ksi, the reduction of area decreased from 70% to between 13 and 21%, and the percent ductile fracture fell to zero.

Similar results were obtained with Zr and Cb additions. Zr at the 0.23% level (Heat 2403) reduced the yield strength by approximately 12 ksi and increased both reduction of area and the percent ductile fracture to 100%. Increasing the Zr level to 0.74% reduced the fracture stress to 37 ksi and failure occurred with zero reduction of area in a completely brittle fashion.

Adding 0.32% Cb reduced the yield stress by approximately 8 ksi while the reduction of area and ductile fracture values were similar to those obtained with the low Zr and Cb additions. With 0.89% Cb in the material, the welds cracked during cooling so tensile tests were not possible.

### Precipitated and Soluble C and N

The results of the analysis are listed in Table 5. Making allowance for the accuracy of the data (±0.001% C, ±0.0005% N), it is clear that for both Fe-26Cr and Fe-35Cr alloys in base material and welds, most of the carbon is precipitated, whereas nearly all the nitrogen is in solution.

### Metallography

The low interstitial weld metal consists of coarse ferrite grains containing a very fine homogeneous carbide precipitate plus some larger particles, which are mainly sulfides (Fig. 6a). No precipitate was visible in the grain boundaries. Annealing ½ hour at

Fig. 7 — Example of ductile fracture initiation by void formation at inclusion - matrix interfaces and subsequent coalescence of voids. Fe-26Cr weld. Annealed at 1560F, ½ hr, W.Q. and stained at +220F. Mag: 500X

**Table 2—Welding Conditions**

Electrode, ⅛ in. diam tungsten/2% thoria, 90 deg point
Stickout (distance beyond collet), 1.7 in.
Arc length, 0.1 in.
Shielding gas, argon at 30 cfh
Backing gas, argon at 30 cfh

| | Plate thickness, in. | |
|---|---|---|
| | 0.10 | 0.188 |
| Welding speed, in./min. | 10 | 4 |
| Current, amp | 150 to 175 | 135 to 160 |
| Arc voltage | 11 to 12 | 10.5 to 11.0 |
| Heat input, joules/in. | 9900 to 12,600 | 21,300 to 26,700 |

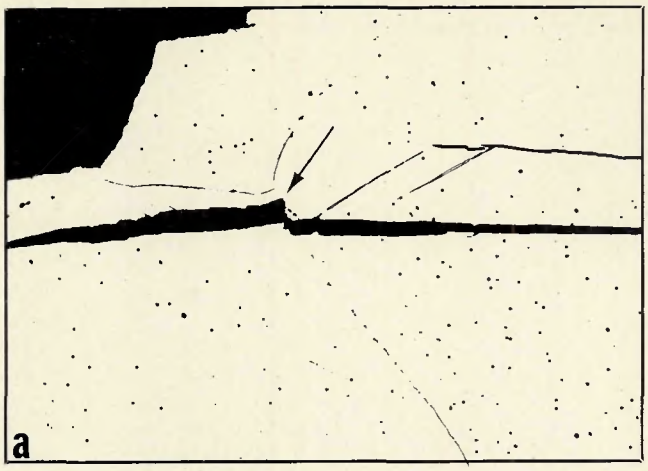



Fig. 8—Examples of cleavage fracture initiation in Fe-26Cr welds: (a) Annealed at 1560F, ½ hr, WQ and tested at —180F, showing fracture initiation at grain boundary (marked by arrow). Original Mag: 500X (enlarged 2.6 times for reproduction); (b) Annealed at 1560F, ½ hr, FC and tested at 0°F, showing fracture initiation at grain boundary carbonitrides. Original Mag: 1000X (enlarged 3.5 times for reproduction)

1560F, followed by a water quench, produced no visible change in the microstructure. Furnace cooling after annealing for ½ hour at 1560F resulted in the formation of a discontinuous grain boundary precipitate in addition to the homogeneous precipitate within the grains (Fig. 6b).

The microstructure of an Fe-26Cr weld with a nitrogen level of 0.02% (Heat 2400) was indistinguishable from those with the low interstitial level.

The microstructures of the Fe-35Cr welds were similar, with the exception that a few isolated grain boundary carbides were observed in welds which had been water quenched after annealing at 1560F. Also, the sulfide inclusions were more numerous in the Fe-35Cr welds.

Due to the variations in grain size and shape within a weld and the difficulty in distinguishing between primary and subgrain boundaries in Fe-Cr welds, precise measurements of grain size were not possible. Estimates showed that grain size was reduced by about 40% by increasing the Cr level from 26 to 35%, or by increasing the nitrogen level of an Fe-26Cr weld from 0.005 to 0.020%. Increasing the weld heat input from 10 to 25 K-joules/inch caused no significant change in grain size. Additions of 0.17% Ti, 0.23 to 0.74% Zr (or 0.89% Cb) increased the grain size in Fe-26Cr welds with 0.02% N by about 50%. However, 0.32% Cb caused no change in grain size and 0.41% Ti reduced the grain size by 33%. The effects of variations in grain size of this order are small compared with the effect of composition on solution strengthening.

### Fracture Initiation Mechanisms

Ductile fracture of both Fe-26Cr and Fe-35Cr welds in all conditions occurred by void formation at inclusion-matrix interfaces and subsequent coalescence of the voids. An example of this mode of fracture initiation is shown in Fig. 7 for an annealed and water quenched Fe-26Cr weld.

When Fe-26Cr welds, "as-welded", were strained within the temperature range where fracture is by cleavage, unloaded just short of failure and subsequently sectioned, no cracks were found. In Fe-26Cr welds, annealed and water quenched and pulled to failure, cleavage cracks nucleated at grain boundaries (Fig. 8a) but other cracks were also observed which started in the middle of the grains. In Fe-26Cr welds, furnace cooled after annealing, cracks were observed in carbonitrides, located in grain boundaries approximately parallel to the tensile axis (Fig. 8b).

Fe-35Cr welds which failed by cleavage did so with less than 4% reduction of area and no cracks were observed in specimens unloaded before fracture, even when mechanical twinning had occurred.

## Discussion

Results of tensile tests at various temperatures indicate that the effect of compositional, welding and postwelding variables on the ductility of ferritic stainless weld metal can be explained in terms of their effect upon the yield stress, $\sigma_Y$, and the fracture stress, $\sigma_F$. Brittle fracture (0% reduction of area) occurs when $\sigma_Y > \sigma_F$. Since $\sigma_Y$ has only a small temperature dependence and $\sigma_F$ for cleavage is essentially independent of temperature, small changes in $\sigma_Y$ and $\sigma_F$ can have a large effect upon the ductile-brittle transition temperature, particularly if $\sigma_Y$ is only slightly smaller than $\sigma_F$, which is the case for the Fe-35Cr welds. When testing is at a temperature close to the ductile-brittle transition, large changes in weld ductility can result from very small variations in $\sigma_Y$ or $\sigma_F$.

The yield stress, $\sigma_Y$, is expressed as a function of grain size by the Petch equation:

$$\sigma_Y = \sigma_i + K_Y\, d^{-1/2} \quad \text{(Ref. 16)}$$

where $\sigma_i$ = lattice friction stress (Peierl's stress)
$K_Y$ = constant
$d$ = half the grain diameter

Welds in ferritic stainless steels are coarse grained so that $\sigma_Y$ is largely determined by the value of $\sigma_i$. The fracture stress, $\sigma_F$, shows a similar dependence upon grain size[15] but in the present investigation the average grain size did not vary significantly between welds. However, the scatter in the tensile data could be attributed to an occasional larger than average grain in a specimen. $\sigma_Y$ and $\sigma_F$ are therefore determined by (a) the extent of solution strengthening, (b) the phase separation reaction occurring around 885F, and (c) the presence of the interstitial elements carbon and nitrogen. Solution strengthening and

**Table 3—Effect of Heat Input on Room Temperature Tensile Properties of Welds**

| | Weld Heat Input K-joules/in. | Yield Stress ksi | Fracture Stress ksi | Reduction of Area, % | Ductile Fracture, % |
|---|---|---|---|---|---|
| Fe-26Cr | 10 | 50.4 | 209 | 79 | 100 |
| (Heat 1913) | 25 | 52.5 | 131 | 73 | 100 |
| Fe-35Cr | 10 | 63.5 | 145 | 61 | 100 |
| (Heat 1917) | 25 | 68.6 | 150 | 62 | 97.5 |

**Table 4—Effect of Alloying Additions on Room Temperature Tensile Properties of Welds in Fe-26Cr**

| Heat No. | Alloying Addition | Alloying Addition* | Yield Stress ksi | Fracture Stress ksi | Reduction of Area, % | Ductile Fracture, % |
|---|---|---|---|---|---|---|
| 2400 | — | — | — | 40 | 0 | 0 |
| | | | 55.5 | 57 | 1 | 0 |
| 2401 | .17 Ti | 1.63 | 49.3 | 113 | 66 | 100 |
| | | | 48.5 | 191 | 75 | 50 |
| 2402 | .41 Ti | 4.77 | 45.2 | 77 | 21 | 0 |
| | | | 43.4 | 68 | 13 | 0 |
| 2403 | .23 Zr | 1.30 | 43.2 | — | 100 | 100 |
| | | | 44.3 | — | 100 | 100 |
| 2404 | .74 Zr | 4.60 | — | 37 | 0 | 0 |
| | | | — | 37 | 0 | 0 |
| 2405 | .32 Cb | 1.68 | 47.4 | 171 | 73 | 50 |
| | | | 47.1 | 250 | 84 | 100 |
| 2406 | .89 Cb | 3.62 | cracked after welding | | | |

* No. of times the stoichiometric equivalent

**Table 5—Chemical Analysis for Precipitated and Soluble Carbon and Nitrogen**

| | Carbon, wt. % | | | Nitrogen, wt. % | | |
|---|---|---|---|---|---|---|
| Specimen No. | Total | Pre-cipitated | Soluble | Total | Pre-cipitated | Soluble |
| 1913 base metal | .004 | .002 | .002 | .006 | <.001 | .006 |
| 1913 weld | .004 | .002 | .002 | .006 | <.001 | .006 |
| 1917 base metal | .005 | .004 | .001 | .008 | <.001 | .007 |
| 1917 weld | .005 | .004 | .001 | .009 | <.001 | .008 |

**Table 6—Effect of Alloying Additions on Ambient Temperature Flow Stress of Ferrite**

| Element | $\Delta\sigma$ (Ksi) per 1 wt % addition* |
|---|---|
| C, N (up to 0.02%) | 805 |
| P | 99 |
| Si | 12.2 |
| Cu | 5.6 |
| Mn | 4.7 |
| Mo | 1.5 |
| Ni | 0 |
| Cb (up to 0.06%)† | 224 |

* After Pickering and Gladman (measured at 18C)
† As-rolled steels

the "885" reaction both increase $\sigma_Y$ and reduce $\sigma_F$. The effect of carbon and nitrogen depends on whether they are in solution or precipitated. In solution they reduce dislocation mobility and hence raise $\sigma_Y$ and reduce $\sigma_F$. Precipitating carbon and nitrogen from solution reduces $\sigma_Y$, but the effect on $\sigma_F$ depends upon the precipitate morphology, as will be discussed later.

The effects of individual elements on solution strengthening of low carbon steel have been determined by Pickering and Gladman[17] (Table 6). Although a gross extrapolation of this data is required to apply it to ferrite containing 26 to 35% Cr, such an extrapolation is worthwhile as it gives a semi-quantitative picture of the effect of each element on $\sigma_Y$ ($\sigma_i$) and the resultant weld ductility.

It is clear from Table 6 that $\sigma_i$ will be minimized and the weld ductility maximized by reducing C, N, P, Si and Mn to the lowest possible levels. Mo has little strengthening effect so that 1% Mo can be added to Fe-26Cr for improved pitting corrosion resistance without significantly reducing weld ductility. Pickering and Gladman did not measure the solution strengthening effect of chromium but using the data in Table 6, values of $\sigma_i$ obtained from Petch plots for the Fe-26Cr and Fe-35Cr alloys and the chemical analysis in Table 1, a value of 0.6 ksi/% Cr was calculated for the contribution of chromium to $\sigma_i$.

The contribution per % Cr to $\sigma_i$ is therefore lower than most other elements likely to be present in ferritic stainless steel but its effect is significant because of the large amount present. Chromium level had more effect upon weld ductility than any other variable investigated. Increasing the Cr level increases solution strengthening and accelerates the "885" reaction, thereby raising $\sigma_Y$ and reducing $\sigma_F$. The overall effect is a sharp reduction in weld ductility.

In the as-welded condition $\sigma_Y$ and $\sigma_F$ for both Fe-26Cr and Fe-35Cr welds are determined by the Cr level, the amount of carbon and nitrogen in solution and the presence of a small amount of $\alpha + \alpha'$ which has occurred despite the fairly rapid cooling rate. Annealing at 1560F dissolves any precipitated interstitial (mainly carbon) and the $\alpha + \alpha'$ particles. Water quenching from 1560F prevents the $\alpha + \alpha'$ particles reforming but cannot prevent the carbon from partly re-precipitating, as during the original cooling of the weld.

The observed reduction of $\sigma_Y$ and increase in $\sigma_F$ must therefore be attributed to the removal of $\alpha + \alpha'$. Increasing the Cr level reduces the solid solubility of carbon and nitrogen. This explains why a trace of grain boundary carbide was found in the Fe-35Cr welds after annealing at 1560F and water quenching, whereas the grain boundaries of the Fe-26Cr welds were clean. This may have reduced somewhat the improvement in ductility obtained by annealing and water quenching the Fe-35Cr welds.

Furnace cooling from the annealing temperature precipitates nearly all the nitrogen and carbon but the reduction in solution strengthening is more than offset by the $\alpha + \alpha'$ phase separation and the formation of grain boundary carbonitride films. The net result is an increase in $\sigma_Y$, a large reduction in $\sigma_F$

and as a result, a substantial decrease in weld ductility.

**Effect of Heat Treatment.** The effects of heat treatment upon the fracture stress can be explained if the mechanisms responsible for fracture initiation are understood. The absence of cracks in Fe-26Cr welds, strained at temperatures where fracture occurs by cleavage and unloaded after extensive necking had occurred, indicates that crack nucleation rather than crack propagation is the important step in the brittle fracture of Fe-26Cr welds. Since when twinning occurred it was always within the first 2% of plastic strain, twin-twin and twin-grain boundary interactions can be discounted as mechanisms of crack initiation.

The absence of twin-induced cleavage in Fe-26Cr and Fe-35Cr welds is rather surprising since Lagneborg[18] has shown that twins can initiate cleavage fracture in coarse grained Fe-30Cr and Fe-30Cr-0.8Ti alloys. However, his tests were conducted at room temperature and cleavage fracture only occurred after previously aging for 75 hours at 885F.

The freedom from grain boundary carbonitrides rules out carbonitride cracking as a possible mechanism of fracture initiation. By a process of elimination, the fracture initiation mechanism for cleavage fracture of Fe-26Cr welds must therefore be either slip band-slip band or slip band-grain boundary interaction. Annealing and water quenching Fe-26Cr welds did not change the fracture initiation mechanism but increased $\sigma_F$ slightly, probably due to an increase in the number of mobile dislocations by removal of $\alpha + \alpha'$. The effect of heat treatment of Fe-26Cr welds was to reduce $\sigma_Y$ by about 3 ksi for specimens water quenched from the annealing temperature and to increase $\sigma_Y$ by about the same amount for specimens furnace cooled from the annealing temperature. The value of $\sigma_F$ was hardly changed by water quenching from the annealing temperature but furnace cooling caused a large decrease, about 100 ksi at +75F. Therefore, while changes in the extent of $\alpha + \alpha'$ separation can account for the small changes in $\sigma_Y$, the much larger change in $\sigma_F$ caused by furnace cooling from the annealing temperature may be attributed to another cause. Metallographic evidence indicates that in Fe-26Cr welds, annealed and furnace cooled, cleavage fracture was initiated by cracking of grain boundary carbonitrides. The stress required to nucleate cracks by this mechanism was much lower than for crack nucleation by slip band-slip band or slip band-grain boundary interaction and $\sigma_F$ was therefore sharply reduced.

The same line of reasoning, as applied to the Fe-26Cr welds, leads to the conclusion that cleavage fracture initiation in Fe-35Cr welds is also by slip band-slip band or slip band-grain boundary interaction. Cleavage fracture initiation in Fe-35Cr welds, annealed and water quenched, was also by the same mechanism, in spite of the presence of a trace of grain boundary carbonitride. Cleavage fracture initiation of Fe-35Cr welds, annealed and furnace cooled, was probably by grain boundary carbonitride cracking, as for the Fe-26Cr welds.

The addition of strong nitride formers such as Ti, Zr and Cb reduced $\sigma_I$ by precipitating nitrogen as nitrides or carbonitrides. This appears to be a promising method for increasing the interstitial tolerance of ferritic stainless steels. The stoichiometric (C + N) equivalent was calculated for each alloy, assuming that the Ti, Zr and Cb formed the simple carbides and nitrides TiC, ZrC, CbC, TiN, ZrN and CbN. Maximum weld ductility was obtained when the alloying addition was one to two times the stoichiometric equivalent. Excessive amounts of Ti, Zr or Cb reduce $\sigma_F$ with a resultant drop in weld ductility. The reasons for this are not yet known.

Increasing weld heat input slowed down the cooling rate, allowing more $\alpha + \alpha'$ phase separation to occur with a resultant small increase in $\sigma_Y$ but the loss of ductility for Fe-Cr welds, tested at room temperature, was negligible. It is to be expected that the effect of heat input on weld ductility will be more important at lower temperatures where $\sigma_F - \sigma_Y$ is small.

## Conclusions

1. The results of tensile tests conducted at temperatures within the range −240F to +300F indicate that changes in the ductility of ferritic stainless weld metal can be explained in terms of the relative values of the yield stress, $\sigma_Y$, and the fracture stress, $\sigma_F$.

2. Increasing the Cr level of the alloy from 26 to 35% severely reduces weld ductility because it increases $\sigma_Y$ and reduces $\sigma_F$, primarily by substitutional hardening. Increasing the Cr level also promotes "885" embrittlement.

3. The effects of postweld heat treatment can be explained in terms of the $\alpha + \alpha'$ phase separation which occurs around 885F and the tendency to form grain boundary carbonitride films during slow cooling. A postweld anneal of ½ hour at 1560F, followed by a water quench, removed the $\alpha + \alpha'$ formed during the weld cycle, thus reducing $\sigma_Y$, increasing $\sigma_F$ and increasing the ductility of Fe-Cr welds. Furnace cooling after a postweld anneal increased $\sigma_Y$ by $\alpha + \alpha'$ phase separation and reduced $\sigma_F$ by formation of a grain boundary carbonitride film, thus reducing the ductility of Fe-Cr welds.

4. Increasing the weld heat input increased $\sigma_Y$ slightly but had a negligible effect upon weld ductility at room temperature.

5. Additions of Ti, Zr or Cb to iron-chrominum-carbon-nitrogen alloys reduce $\sigma_Y$ and increase $\sigma_F$ by precipitating carbon and nitrogen and for a given ductility level increase the tolerance of the alloy for interstitials.

6. The most practical methods of maximizing the ductility of welds in ferritic stainless steels are: keep the level of both substitutional and interstitial alloying elements as low as possible and make additions of strong carbo-nitride forming elements.

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WRC
Bulletin
No. 167
November 1971

## "Laser Welding and Cutting"

by M. M. Schwartz

The advent of high intensity low momentum surface heating devices is having a natural impact on the field of materials processing, particularly fusion welding. The laser and electron beam offer potential capability for precise control of energy and location which cannot be approached by older sources such as arcs and flames, and the fact that under certain operating conditions they transmit little or no thrust to the material being worked is an important advantage. However, the extreme intensity of these sources, expressed in dimensions of power per unit area, presents problems as well as advantages when addressed to the objective of local melting as in fusion welding.

The newest and most exciting of the two aforementioned processes is the "laser" (sometimes called optical maser). The laser represents a virtually brand-new scientific development. The possibility of a laser was first suggested in 1958 and the first experimental model was made in 1960. Since that time, advances in laser technology have been rapid and dozens of laboratories are performing laser research. Although much of this research is of a basic scientific nature many of the newer developments can be applied to welding. Experimental laser machines are already in use. No basic scientific discovery in history has been applied so fast to metalworking.

This report, prepared for the Interpretive Reports Committee of the Welding Research Council, reviews the basic principles of the laser and describes techniques, problems, advantages and disadvantages of using the laser for welding and cutting various materials.

The price of Bulletin 167 is $3.50 per copy. Orders for single copies should be sent to the American Welding Society, 2501 N.W. 7th St., Miami, Fla. 33125. Orders for bulk lots, 10 or more copies, should be sent to the Welding Research Council, 345 East 47th Street, New York, N. Y. 10017.

WRC
Bulletin
No. 166
November 1971

## "Review of Service Experience and Test Data on Openings in Pressure Vessels with Non-Integral Reinforcing"

by E. E. Rodabaugh

During the past ten years an extensive test program on nozzles and openings with integral reinforcing has been carried out under the guidance of the Pressure Vessel Research Committee of the Welding Research Council. Integral reinforcing is defined as reinforcing which is fully continuous with the shell; as contrasted to pad or saddle reinforcements, which are welded to the shells at the inner and outer peripheries. The ASME Nuclear Vessels Code requires that reinforcing be integral; the test program was intended to provide design data for such integrally reinforced openings. The results of this test program is now reflected in the ASME Nuclear Vessels Code in the form of stress indices and work is underway to more generally revise ASME Code rules for design of integrally reinforced openings.

However, many and perhaps most openings in pressure vessels are reinforced with pads or saddles; i.e. nonintegral reinforcing. The purpose of this report is to review available data on nonintegrally reinforced openings as perhaps a first step towards improvement ni code design rules for such reinforcing.

Publication of these papers was sponsored by the Pressure Vessel Research Committee of the Welding Research Council. The price of Bulletin 166 is $3.50 per copy. Orders for single copies should be sent to the American Welding Society, 2501 N.W. 7th St., Miami, Fla. 33125. Orders for bulk lots, 10 or more copies, should be sent to the Welding Research Council, 345 East 47th Street, New York, N.Y. 10017.

## "Derivation of Code Formulas for Part B Flanges"

By E. O. Waters

New rules for bolted flanged connections appeared in the 1968 and 1970 Winter Addenda of the ASME Boiler and Pressure Vessel Code, and more recently in the 1971 edition of Section VIII, Division 1, providing for flanges with metal-to-metal contact outside the bolt circle. The development of this material came from recent work of the PVRC Design Division subcommittee that was assigned this topic.

There is considerable novelty in the formulas and charts in these rules, and it was the belief of the subcommittee members that their acceptance and understanding by Code users would be greatly aided if a concise account of their derivation were given. The paper was written with this in mind, and with the added purpose of providing a source to which reference could be made in the future, when specific inquiries have to be answered or revisions are contemplated.

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