Low Temperature Ductility of Ultrafine Grain α Iron and Its Substitutional Solid Solutions*

By Masato ENOMOTO** and Ei-ichi FURUBAYASHI**

Synopsis

A study was made of the effect of substitutional solute elements on the ductility of α iron with varied ultrafine grain sizes which was prepared by cold rolling and subsequent annealing to various stages of recovery and recrystallization. The addition of Si or Ti was found to have a remarkable effect of improving the low temperature ductility of ultrafine grain recrystallized structure which in an unalloyed condition showed a very small uniform elongation owing to plastic instability; Al or Ni was not very effective. Ni was found to suppress the intergranular fracture which was one of the most important causes of low temperature failure of iron having relatively coarse grain structure.

Lüders strain was also measured; it was found that the critical grain size, d_c , at which Lüders strain exceeded uniform elongation and the transition from stable to unstable plastic flow occurred, was reduced by substitutional solute elements.

Possible causes for reducing Lüders strain or d_c were briefly discussed in terms of slip character and interstitial distribution which were both expected to undergo a considerable change by the presence of those substitutional elements.

I. Introduction

Grain refinement is known to be a profitable way to improve strength and toughness of structural steels. It is also considered to be efficient in suppressing grain boundary embrittlement caused by the segregation of impurities by increasing the area of boundary region relative to a given amount of impurities. However, the ductility of ultrafine grain materials is exceedingly poor.¹⁾ The principal cause of the worsening of ductility is that ultrafine grain materials tend to undergo an inhomogeneous deformation, and its local strain

readily exceeds the maximum strain endurable by work hardening in the initially deformed region. This plastic behavior of ultrafine grain materials is generally called “plastic instability”, or “pseudo-brittle fracture”, since the reduction of area is usually large.²⁻⁴⁾

It is a well-known fact that the motion of individual dislocations under stress, or the mode of slip in α iron is profoundly influenced by the presence of substitutional solute atoms. Since the slip character has been found by some authors as one of the important factors affecting the inhomogeneous deformation,⁵⁾ the addition of substitutional elements is expected to have some notable effects on the ductility of ultrafine grain α iron. Numerous data are now available on the strength of substitutional iron alloy.⁶⁻⁸⁾ However, as to the ductility, no systematic studies^{7,9)} have been conducted yet.

In the present study, ultrafine grain structure were developed in ferrite with various kinds of substitutional elements and with various carbon contents by cold rolling and subsequent annealing. Mechanical properties measured were discussed in terms of the improvement of ductility of ultrafine grain α iron.

II. Experimental Procedure

A number of substitutional iron alloys with various kinds of solute element such as Ni, Si, Ti or Al were melted in a vacuum induction furnace. Chemical compositions of these alloys are given in Table 1. Each alloy in the series (A) contains ~2at% of solute

Table 1. Chemical composition of alloys (wt%)

Alloy	Ni	Si	Al	Ti	Mn	S	C	N	O
Fe (unalloyed iron)	0.005	0.009	0.005	0.002	0.004	0.008	0.010	0.002	0.0017
(A)	Fe-2%Ni	2.05	0.01	0.005	0.002	0.003	0.008	0.010	0.0015
	Fe-1%Si	0.01	1.01	0.005	0.002	0.003	0.005	0.005	0.0020
	Fe-1%Al	0.01	0.01	1.06	0.002	0.003	0.005	0.007	0.0007
	Fe-2%Ti	0.01	0.01	0.05	1.96	0.003	0.005	0.009	0.0024
	Fe-0.5%Si	0.01	0.49	0.005	0.002	0.003	0.005	0.010	0.0040
(B)	Fe-2%Si	0.01	2.00	0.005	0.002	0.003	0.005	0.007	0.0025
	Fe-2%Ni-1%Si	1.98	0.95	0.005	0.002	0.003	0.005	0.009	0.0029
	Fe-1%Si-0.05%C	0.01	1.00	0.005	0.002	0.003	0.005	0.048	0.0029
(C)	Fe-1%Si-0.1%C	0.01	1.03	0.005	0.002	0.003	0.005	0.095	0.0032
(D)	Fe						<0.001		
	Fe-2%Ni	Wet-hydrogen treated					0.002		
	Fe-1%Si						0.002		

* Originally published in *Tetsu-to-Hagane*, 63 (1977), 469, in Japanese. English version received February 23, 1977.

** National Research Institute for Metals, Nakameguro, Meguro-ku, Tokyo 153.

atoms. The content of Si was varied in the series (B), which also includes one alloy containing $\sim 2\text{at}\%$ each of Ni and Si simultaneously. In order to maintain grain boundary cohesion at low temperatures, the carbon content of alloys in the series (A) and (B) was adjusted to $\sim 0.01\text{wt}\%$ by adding an amount more than required to deoxidize the electrolytic iron on melting.

Carbon content was varied in the series (C) and (D). In the series (D) the carbon content was reduced by a wet-hydrogen treatment at 700°C for 3 days of the hot-rolled sheets of alloys in the series (A).

The 10 kg heats of every alloy were, after homogenization at 1200°C for 2 hr, forged and hot-rolled into 3.5 mm thick sheet, then air-cooled. They were cold-rolled by a reduction from 55 to 90% and annealed for 1 hr at various temperatures between room temperature and 950°C , by which various microstructures were developed ranging from deformed to recrystallized states. In order to avert possible effects of sub-boundaries formed on γ - α transformation during cooling, annealing was confined to α region except for the alloys in the series (C), in which annealing was carried out in the $(\alpha+\gamma)$ two-phase region for some specimens. After the annealing specimens were air-cooled and held in a bath of dry ice-alcohol to minimize the effect of aging by interstitial solute atoms.

The sheet tensile specimens with 3 cm gauge length, the thickness of which was always greater than 6 to 7 times the grain size, were strained using an Instron-type machine at a strain rate of $3 \times 10^{-4} \text{sec}^{-1}$ at 23° , -78° and -196°C .

III. Results

1. Microstructure

In Fig. 1 areal fractions recrystallized and sizes of recrystallized grains and subgrains of 90% cold-rolled Fe and Fe-2%Ni are plotted against the annealing temperature. The annealing time was 1 hr. Completely recrystallized structures were obtained at temperatures higher than 700°C for Fe and 600°C for Fe-2%Ni. A typical appearance of partially recrystallized structure of Fe-2%Ni which was cold-rolled by 90% and annealed at 480°C for 1 hr is shown in Photo. 1, the areal fraction recrystallized being about 60%. The microstructure is composed of subgrains smaller than $1 \mu\text{m}$ and recrystallized grains of $2\sim 3 \mu\text{m}$ size. Below 400°C no recrystallized grains were observed. When the areal fraction reached 100%, the mean sizes of recrystallized grains were increased to $12\sim 13 \mu\text{m}$ for Fe and $7\sim 8 \mu\text{m}$ for Fe-2%Ni, respectively.

Microstructures consisting of grains smaller than $10 \mu\text{m}$ are customarily called ultrafine grain structure. In this report, partially recrystallized fine structures containing many sub-boundaries will also be designated as an ultrafine structure, although the contribution of such sub-boundaries to plastic flow may be considerably different from that of well-defined high angle boundaries in a completely recrystallized structure.

Fig. 1. Recrystallized grain size, subgrain size and areal fraction recrystallized of 90% cold-rolled Fe and Fe-2%Ni as a function of annealing temperature

Photo. 1. Transmission electron micrograph of partially recrystallized structure which was obtained by annealing the 90% cold-rolled Fe-2%Ni alloy at 480°C for 1 hr

2. Load-elongation Curves

In Figs. 2 (a) to (f) typical examples of nominal stress (load divided by original area) *vs.* elongation curves of Fe, Fe-2%Ni and Fe-1%Si tested at 23° and -196°C are shown. The curve marked with "D" corresponds to deformed state, the curve "P" to partially recrystallized state, the curve "F" to a state completely recrystallized into fine grains ($\leq 20\sim 30 \mu\text{m}$), and the curve "C" to that completely recrystallized into relatively coarse grains ($\geq 20\sim 30 \mu\text{m}$).

The three alloys showed a similar tensile behavior at 23°C ; at -196°C , however, they presented several distinctively different features. First of all, Fe-1%Si specimens of partially recrystallized (P) or completely recrystallized structure of fine grain size (F) had an elongation far larger compared with that of the other two alloys in the same structure. The load-elongation curves of Fe and Fe-2%Ni specimen of the corresponding state were typical of plastic instability, *i.e.*, continuing decrease of load due to rapid decrease in cross sectional area after the yielding. On the other

Fig. 2. Nominal stress-elongation curves at various stages of recovery and recrystallization. D, P, F and C denote deformed state, partially recrystallized state, completely recrystallized state of fine ($\leq 20 \sim 30 \mu\text{m}$) and relatively coarse grain size, respectively.

hand, the Fe-1%Si specimen of (F) state exhibited normal tensile behavior, *i.e.*, sharp yield drop, Lüders deformation and subsequent work hardening.

Secondly, Fe and Fe-1%Si specimen of completely recrystallized state of coarse grain size (C) fractured in a brittle manner before yielding occurred, whereas Fe-2%Ni specimen of the corresponding state was of a large elongation. The fracture surface of Fe and Fe-1%Si specimen was mostly intergranular with a small fraction of grains exposing cleavage facets.

Fig. 3. Strength-ductility relationships of Fe, Fe-2%Ni and Fe-1%Si with heavily deformed ($\times$), partially ($/$) and completely recrystallized structure. The brittle specimens are marked with a horizontal line ($—$).

3. Relationship between Strength and Ductility

In Fig. 3 the relationships between strength and ductility of Fe, Fe-2%Ni and Fe-1%Si alloys are shown, taking the lower yield strength and the uniform elongation as the measure of strength and ductility, respectively. Both vary according to the annealing temperature. In the figure, the partially recrystallized structures (P) are marked with a short oblique line ($/$) and the deformed structures (D) with a small cross ($\times$), while those specimens which fractured in a brittle manner before yielding are marked with a short horizontal line ($—$), all superimposed on the corresponding data points.

At 23°C, recovery and recrystallization caused a rapid increase in uniform elongation as well as a decrease in lower yield strength in a similar manner for the three alloys. At -196°C, however, the strength-ductility relationships were varied depending on the solute element. Namely, whereas Fe specimens showed substantially no ductility in any state from deformed to completely recrystallized, Fe-2%Ni and Fe-1%Si specimens gave rise to a large elongation over a certain range of the annealing temperature. For Fe-2%Ni, for example, specimens in the completely recrystallized state with coarse grain size had a large uniform elongation, although specimens in the state from deformed to partially recrystallized showed very poor ductility. For Fe-1%Si, on the other hand, a substantial elongation was obtained when in a fine grained state from partially to completely recrystallized, though the ductility was reduced, finally to zero, owing to grain boundary embrittlement as the recrystallized grain size was increased. A more desirable combination of strength and ductility was obtained at low temperatures for Fe-1%Si than for Fe-2%Ni.

The grain size dependences of uniform elongation of Fe, Fe-2%Ni and Fe-1%Si at 23°, -78° and -196°C are given in Figs. 4 (a) to (c). Although the points in Fig. 3 represent only those specimens that were cold-rolled by more than 75%, in Fig. 4 the points

Fig. 4. Uniform elongation as a function of ferrite grain size. Dotted marks correspond to wet-hydrogen treated materials.

of the specimens with lower rolling reduction are included: grain size ranges larger than 50 μm for Fe or Fe-1%Si and larger than 25 μm for Fe-2%Ni were obtained from relatively low rolling reductions. The points corresponding to deformed or partially recrystallized states are omitted. Dotted marks are for decarburized materials in completely recrystallized state obtained by cold rolling by 90% and annealing at 700°C after the wet-hydrogen treatment.

For Fe, specimens of a grain size smaller than 23 μm (at this grain size the specimen exhibited ~10% elongation) fractured by necking, while specimens of a grain size larger than this fractured intergranularly. The addition of 2%Ni was found to suppress the intergranular fracture in coarse grained specimens, although it has very little effect of prevent-

ing the occurrence of unstable necking in fine grained specimens. On the contrary, the addition of 1%Si had a remarkable effect of improving the low temperature ductility in fine grained specimens, while coarse grained specimens fractured intergranularly as Fe specimens of coarse grain size did. Fe specimens of 11 to 14 μm grain size exhibited strong plastic instability even at -78°C. This unstable flow was improved by 2%Ni addition, not to mention the 1%Si addition.

The well-known effect of Ni on the ductility and toughness of steels has generally been attributed to the suppression of the initiation of cleavage fracture at low temperatures. The present result is, however, that the improvement effect was achieved by the suppression of intergranular fracture in relatively coarse grained ferrite. The present authors intend to discuss this observation in a separate paper.

Since the addition of Si to an amount in excess of that required for deoxidization on melting has long been believed to have a detrimental effect on the ductility of ferrite, very few investigations are reported¹¹⁻¹³⁾ on the basis of an expectation that Si should exert a valuable effect on the ductility of ferrite. In this study, it has been found that Si actually has an effect of improving the ductility of fine grained ferrite through preventing the plastic instability, and that it manifests its good effect especially at low temperatures.¹⁴⁾

4. Lüders Strain

Plastic instability is caused by an accumulation of strain in a narrow localized region. The amount of an accumulated strain can be adequately described by Lüders strain. Dependences of Lüders strain, *i.e.*, the strain at the termination of constant load extension, on ferrite grain size at 23° and -78°C are shown in Fig. 5. It should be mentioned that the load was not always constant during Lüders deformation in large grained specimens. The formation of additional Lüders bands, which readily occurred in large grained specimens, caused a stress drop, and the junction of Lüders fronts propagating in opposite directions caused a sudden increase in stress. This is because the change in the number of Lüders bands caused by the junction or annihilation may cause the change in the number of dislocations which can participate in deformation, as has already been pointed out.^{15,16)} The Lüders strain has been described to be a function of the stress at which Lüders front propagates. Consequently, special care should be taken to restrict the nucleation of additional bands for an accurate determination of Lüders strain. In the present study, no correction was made with regard to this, since the change in the propagation stress was observed only in a few specimens of large grain size.

As is obvious from Fig. 5, both the decrease in ferrite grain size and in deformation temperature caused the increase of Lüders strain. It should be noted that the addition of Ni or Si had the effect of reducing the Lüders strain of ferrite over a range of fine grain size. While the difference among the alloys was small

Fig. 5. Dependences of Lüders strain on ferrite grain size

at 23°C, it became large as the deformation temperature was decreased, *i.e.*, the Lüders strain was decreased in the order of Fe, Fe-2%Ni and Fe-1%Si when compared at the same grain size.

Figure 6 shows the relationships between the Lüders strain and the lower yield strength (or the stress at which Lüders front propagates) of the three alloys. The addition of Ni or Si was also found to cause the decrease in Lüders strain when compared at the same strength.

This effect of substitutional elements on Lüders strain seems to be closely related to the effect of improving the low temperature ductility of fine grained materials mentioned earlier.

5. Effect of Si Content

The strength-ductility relationships of the alloys with 0.5%, 1% and 2%Si are shown in Fig. 7. All the specimens of Fe-0.5%Si exhibited plastic instability except for the specimen of the largest grain size. Consequently, it may be concluded that the addition of 0.5%Si is almost inefficient in improving the low temperature ductility. The addition of 2%Si was found to have an improvement effect to nearly the same extent as the addition of 1%Si.

In Fig. 7 the points of Fe-2%Ni-1%Si are also included. The simultaneous addition of Ni and Si did not give rise to any injury on the ductility at low tem-

Fig. 6. Relationships between lower yield strength and Lüders strain. Experimental points are confined to completely recrystallized structure.

Fig. 7. Strength-ductility relationships of Fe-Si alloy with various Si contents. Data of the alloy added with 2at%Ni and Si simultaneously are also included.

peratures. On the contrary, the strength at 23°C was rather higher than the Fe-2%Si alloy; they are almost equal in total solute content ($\sim 4\text{at}\%$).

Several investigations have already been documented on the effect of substitutional solute elements on the ductility and toughness at low temperatures. The common interpretations are that the lowering of ductile-to-brittle-transition-temperature (D.B.T.T.) was caused by the strong interaction between substitutional and interstitial solute elements.^{17,18)} According to those results, the minimum in transition temperature occurred at a solute content of $\sim 1\%$ or less, and the D.B.T.T. was increased as the solute content exceeded $\sim 1\%$. Such solute content dependence is apparently different from that of the present findings, suggesting that some basic effect of solute atoms on dislocation mechanism as well as substitutional and interstitial solute interaction is responsible for the effect of Si mentioned above.

6. Effect of C Content

Strength-ductility relationships of Fe-1%Si alloy with various carbon contents from 0.002 to 0.1% are

given in Fig. 8. The specimens of partially recrystallized structure with 0.002% C were of a large elongation. The ductility was almost maintained up to 0.1% C. A few specimens with higher carbon contents annealed at temperatures higher than 700°C showed uniform elongation of less than 10%.

The microstructures with 0.05 and 0.1% C were mostly equiaxed single phase structure, whose grain size did not exceed 10 μm , much finer than the materials with 0.01% C. Cementite and other transformation products from austenite were observed at ferrite grain boundaries in the specimen annealed at 900°C.

7. Effect of Al and Ti

In order to see if any other element having the same effect as Si exists, tensile tests were conducted in the alloys with each 2at% Al or Ti. The results are shown in Fig. 9. The low temperature ductility of α iron was scarcely improved by the addition of 1% Al; the uniform elongation of Fe-1% Al specimen was as small as that of Fe specimen in the state from partially to recrystallized. Unstable necking was the main cause

of the poor ductility even in coarse grained specimens. The specimen with a maximum of 112 μm grain size was fractured after the repetition of large stress drop two or three times by twinning. In the figure, the specimen in which a stress drop by twinning was observed on load-elongation curve is marked with a short vertical line (|).

For Fe-2% Ti, flow stress at 23°C continued to increase with strain until the ultimate tensile stress was attained. At -196°C, a small yield drop was observed. A few of those specimens exhibited constant load extension over a range of initial strain. On the whole, Lüders strain of Fe-2% Ti was so small that the inhomogeneity of deformation was hardly recognized. As is shown from the figure, the effect of Ti addition is conspicuous especially in recovered or partially recrystallized state.

It should be noted that the uniform elongation of Fe-2% Ti specimen of partially recrystallized structure at 23°C is a little larger than that of Fe or Fe-1% Al specimen of the corresponding structure. The same is the case with Si; the room temperature ductility of partially recrystallized structure of Fe-1% Si is better than that of Fe or Fe-2% Ni. In this way, the effect of Si or Ti on the ductility of ferrite was found to appear even at ambient temperature, becoming more distinct as the deformation temperature was decreased.

Fig. 8. Strength-ductility relationships of Fe-1%Si alloy with various carbon contents

Fig. 9. Strength-ductility relationships of Fe, Fe-1%Al and Fe-2%Ti

IV. Discussion

As indicated above, the deformation of ultrafine grain materials is characterized by very poor ductility owing to plastic instability, which is considered to occur when

$$\varepsilon_L \geq \varepsilon_U \quad \dots\dots\dots (1)$$

here, ε_L denotes Lüders strain which usually increases as the grain size of materials decreases and ε_U denotes uniform elongation.¹⁾

As was shown in Fig. 5, grain size dependence of Lüders strain is profoundly influenced by deformation temperature and by the presence of substitutional solute element. In Fig. 10, the uniform elongation and Lüders strain of Fe-1% Si is shown as a function of ferrite grain size. It is obvious from the figure that Eq. (1) is always satisfied at a grain size smaller than a certain critical value denoted as d_c . At d_c , a ductility transition occurs from the normal tensile behavior exhibiting sharp yield drop, Lüders deformation and homogeneous work hardening, to the unstable plasticity behavior which causes necking immediately after yielding takes place. It is apparent that d_c shifts to coarser grain size side as the deformation temperature is decreased. This means, in other words, that the critical deformation temperature at which the ductility transition occurs is lowered as the grain size of materials is decreased, as has already been pointed out by W. B. Morrison.²⁾

The ε_U was found to be almost independent over the whole grain size range studied, as indicated in Fig. 4. In Fig. 10 linear extrapolation was adopted to determine d_c . Some authors have shown that the

Fig. 10. Schematic diagram for the determination of critical grain size at which a ductility transition occurs. The ϵ_L and ϵ_U denote Lüders strain and uniform elongation, respectively. The region in which plastic instability occurs is extended to larger grain size as deformation temperature is decreased.

uniform elongation rapidly decreases as the grain size is reduced to ultrafine grain size range ($\leq 10 \mu\text{m}$).¹⁾ This may be attributed to the fact that for ultrafine grain materials, in which the mean free path of dislocations can reasonably be assumed to be equal to grain diameter, more dislocations have to be stored in individual grains to undergo a given amount of strain compared with materials of large grain size. The decrease in uniform elongation over a fine grain size range apparently causes the shift of the intersection of the two extrapolation lines to a larger grain size. However, the correction of d_c seems to be rather small.

The d_c values of the present alloys are given in Table 2. It will be seen that d_c is decreased by the addition of substitutional element. In particular, the decrease in d_c of alloys with Si or Ti is remarkable.

The value of d_c was also found to increase as the strain rate was raised. In fact, the Fe-2%Ni specimen of $7 \mu\text{m}$ grain size exhibited plastic instability at the strain rate of $3 \times 10^{-1} \text{ sec}^{-1}$ at 23°C .

As stated above, occurrence of unstable plasticity which ultrafine grain ferrite readily exhibits at low temperatures and its suppression by solute elements are both conveniently described by the change of critical grain size, d_c .

In the following, a few points will be discussed about a factor affecting Lüders strain or d_c . The first point to be considered is that in the presence of substitutional solute element the distribution of interstitials such as C and N, which are known to have a great influence on Lüders deformation,¹⁹⁾ is changed. The observed homogeneous deformation of Fe-2%Ti specimen may be due to the so-called scavenging effect of interstitials by Ti atoms. As for Si, a few specimens of Fe-1%Si alloy showed no yield drop, while an initiation of constant load extension was well defined on the load-elongation curve. The similar behavior is often observed with single crystals of Fe-Si alloy.²²⁾

Table 2. Estimated values of critical grain size, d_c (μm)

Alloy	23°C	-78°C	-196°C
Fe	3	27	Very large
Fe-2%Ni	0.5	2	20
Fe-1%Si	0.4	0.9	6
Fe-1%Al	0.6	—	≥ 30
Fe-2%Ti	~ 0	—	2.5

Although detailed information on the interaction between Si atoms and interstitials is not available yet, some authors have pointed out that the presence of Si atoms in ferrite raises the temperatures at which stable and quasi stable carbides are formed.²³⁾ Therefore, it is not unreasonable to assume that Si causes retardation of the segregation of carbon atoms to dislocations, namely the formation of Cottrell atmosphere, and thus the decrease in Lüders strain.

Accordingly, it can be considered that the d_c values of the alloys in Table 2 are in the order of weakening interaction between interstitial and substitutional atoms.

However, it is not to be expected that large uniform elongation at low temperatures was attained only by the removal of interstitial atoms from matrix. This is suggested by the fact that decarburized Fe specimen having partially recrystallized structure was of a small elongation owing to plastic instability, and that the addition of Al which ties up N improved the ductility of ferrite only to a very limited extent. In addition, Fe-1%Si alloy with 0.1%C was of nearly the same elongation as Fe-1%Si alloy with 0.005%C.

It is well-known that at low temperatures a critical resolved shear stress of iron and its substitutional alloys varies according to the operative slip system, and that the ease of slip on (112) planes tends to decrease. On the macroscopic scale this is reflected on the appearance of slip pattern.²⁵⁾ As the deformation temperature is decreased the slip pattern is changed in a different manner according to the kind of the solute element added to ferrite. In particular, the change of Fe-Si alloy is reported to be remarkable, suggesting almost complete restriction of slip on (112) planes at low temperatures.

In the three dimensional polycrystalline aggregates individual grains have to deform under the constraint of surrounding grains. In this circumstance, if the number of operative slip systems is decreased, it will be more favorable to transmit strain to the neighboring grains than to accumulate it in the initially deformed grains. Suzuki⁵⁾ has pointed out that the limited number of available slip systems is one of the important reasons why Lüders band does not appear in f.c.c. metals. Therefore, it is naturally expected that Lüders strain in b.c.c. metals will also be decreased, or even disappear, as in f.c.c. metals if the availability of slip systems are restricted for some reason or other. However, a further investigation of the slip mode of the alloys such as Fe-2%Ti is required to make a definite conclusion.

As stated above, the worsening of ductility inherent

in grain refinement of materials can efficiently be avoided by the addition of a certain substitutional element such as Si or Ti.

V. Conclusion

(1) Unalloyed α iron ($\sim 0.01\%$ C) having various microstructures developed by conventional cold rolling and subsequent annealing showed exceedingly poor ductility at low temperatures. At -196°C , specimens of grain size smaller than $\sim 20\ \mu\text{m}$ fractured by necking immediately after the yielding occurred, while specimens of grain size larger than that fractured intergranularly.

(2) The addition of 2% Ni was found to be efficient in suppressing the intergranular fracture of specimens recrystallized into coarse grain size ($\geq 20\sim 30\ \mu\text{m}$), although it was not efficient in suppressing the unstable necking for specimens of fine grain size ($\leq 20\sim 30\ \mu\text{m}$).

(3) The addition of 1% Si was found to be efficient in averting the plastic instability, thereby giving an appreciable amount of ductility to fine grained specimens. The addition of 2% Si was as effective as 1% addition, while 0.5% addition was almost ineffective. The addition of Si was found to have no effect of suppressing the intergranular fracture for coarse grained specimens.

(4) Good ductility of Fe- 1% Si at low temperatures was maintained over a wide range of carbon content, at least from 0.002 to 0.1% . A weak tendency of ductility decrease was observed as carbon content was increased.

(5) Ti was found to have a similar effect of ductility enhancement as Si, while Al was not.

(6) The Lüders strain of substitutional iron alloys studied was found to be smaller than that of unalloyed iron over a fine grain size range when compared at the same grain size and also at the same strength.

(7) The critical grain size (d_c), at which Lüders strain exceeded uniform elongation and a transition occurred from stable to unstable plastic flow, was found to be increased as the deformation temperature was decreased. It was decreased by the addition of substitutional element. In particular, the decrease by Si or Ti was remarkable.

(8) Possible causes for the decrease in Lüders strain or d_c were discussed. The change of slip mode which might be imposed by the presence of substitutional atoms is considered to be more important than the change of the distribution of interstitial atoms.

REFERENCES

- 1) W. B. Morrison and R. L. Miller: *Ultrafine Grain Metals*, ed. by J. J. Burke and V. Weiss, Syracuse University Press, (1970), 183.
- 2) W. B. Morrison: Preprint of 2nd Int'l Conf. Strength of Metals and Alloys, ASM, Asilomar, USA, (1970), 295.
- 3) V. Ramachandran and E. P. Abrahamson: *Script. Met.*, **6** (1972), 287.
- 4) R. H. Goodenow and J. H. Bucher: *Trans. ASME*, **91** (1969), 603.
- 5) H. Suzuki: *Tetsu-to-Hagané*, **50** (1964), 2351.
- 6) S. Takeuchi: *J. Phys. Soc. Japan*, **27** (1969), 929.
- 7) C. E. Lacy and M. Gensamer: *Trans. ASM*, **32** (1944), 88.
- 8) W. C. Leslie: *Met. Trans.*, **3** (1972), 5.
- 9) W. P. Rees, B. E. Hopkins and H. R. Tipler: *JISI*, **177** (1954), 93.
- 10) R. Honda and H. Taga: *Metal Sci. J.*, **2** (1968), 172.
- 11) J. Imamura, H. Hayakawa and T. Hayami: *Tetsu-to-Hagané*, **60** (1974), A159.
- 12) K. Ota: *Tetsu-to-Hagané*, **28** (1941), 969.
- 13) K. Ota: *Tetsu-to-Hagané*, **56** (1970), 110.
- 14) M. Enomoto and E. Furubayashi: *Script. Met.*, **10** (1976), 717.
- 15) J. F. Butler: *J. Mech. Phys. Solids*, **10** (1962), 313.
- 16) J. F. Butler: *Acta Met.*, **10** (1962), 258.
- 17) P. Abramowitz and R. A. Moll: *Met. Trans.*, **1** (1970), 1773.
- 18) J. L. Jellison and N. S. Stoloff: *Mater. Sci. Eng.*, **13** (1974), 231.
- 19) J. R. Low, Jr. and M. Gensamer: *Trans. AIME*, **158** (1943), 207.
- 20) W. C. Leslie and R. J. Sober: *Trans. ASM*, **60** (1967), 99.
- 21) H. D. Solomon, C. J. McMahon, Jr. and W. C. Leslie: *Trans. ASM*, **62** (1969), 886.
- 22) T. Taoka, S. Takeuchi and E. Furubayashi: *J. Phys. Soc. Japan*, **19** (1964), 701.
- 23) W. C. Leslie: *Acta Met.*, **9** (1961), 1004.
- 24) A. Matsuda and K. Nakajima: *Tetsu-to-Hagané*, **62** (1976), 652.
- 25) S. Takeuchi, T. Taoka and H. Yoshida: *Trans. ISIJ*, **9** (1969), 105.