Damascus Steel Rediscovered?*

By Oleg D. SHERBY**

Synopsis

The outstanding characteristics of Damascus steel are legend. Not only beautiful in design, they were known for their strength, toughness and retention of their cutting edge. The forgotten art of making these steels may have recently been rediscovered at Stanford University. The metallurgical key to the development of such steels is the use of ultrahigh carbon contents and the use of thermal-mechanical processing procedures for attainment of sub-micron microstructures. Such steels are superformable (superplastic) at warm temperature and strong and ductile at room temperature. The preparation and properties of fine-grained ultrahigh carbon steels will be described, including their use in making laminated composites similar to the structures noted in Japanese laminated tools and welded Damascus steel.

1. Preface

It gives me great pleasure to present this special lecture to you at the annual meeting of the Iron and Steel Institute of Japan in Toyama. In this respect I give personal thanks to Professor Tadahisa Nakamura of Tokyo Institute of Technology. He was my most gracious host who not only arranged my talk for this meeting, but also made my stay in Japan a very memorable and delightful one.

My presentation today is centered on ultrahigh carbon (UHC) steel, superplastic bonding and UHC steel laminated composites. This work has preoccupied our interests at Stanford University in the past few years.¹⁻⁹ The iron-carbon phase diagram (Fig. 1) illustrates the carbon range which we designate as ultrahigh carbon steel. It is from 1% C to 2.1% C.

Fig. 1. Ultrahigh carbon (UHC) steels are defined in this paper as steels containing 1 to 2.1% C as shown in the above iron-carbon phase diagram. Cast irons contain carbon in excess of the solubility limit of carbon in austenite ($>2.1\%$ C) and high carbon steels generally contain 0.6 to 1.0% C.

Below 1% C, are the high carbon steels (0.6~1% C), the medium carbon steels (0.25~0.6% C) and the mild steels ($<0.25\%$ C). The cast irons are above 2.1% C since cast iron, by definition, contains carbon in excess of the maximum amount that can be dissolved in austenite.¹⁰

Ultrahigh carbon steels can be processed to exhibit many desirable characteristics. As will be shown later they can be made superplastic ($>1500\%$ elongation) at warm temperature. They can be strong (1 000 MPa yield strength) and ductile ($\sim 20\%$ elongation) at room temperature. They can also be made very hard ($R_c=68$ with 4 300 MPa compression fracture strength) by heat treatment. And, they can be used as a solid state bonding agent through superplastic joining. These properties seemed unique to us at the time we discovered them. More recently, however, we concede the possibility that we may have simply rediscovered the ancient and forgotten art of making Damascus steel and Japanese laminated steel composites.

We illustrate in Photo. 1 Damascus steel swords used by the Russians in the War of 1612. Not only beautiful in design, they were known for their high toughness and fine cutting edge. Belaiew,¹¹ in a review of the history of "bulat steel" (the Russian word for Damascus steel), mentioned that many such steels were in the composition range of 1.2 to 1.8% C. They were also extensively worked mechanically. Our UHC steels are generally worked extensively in both the gamma and alpha range of temperature for attainment of fine uniform microstructures. Thus, the analogy between our UHC steels and Damascus steels is strong. [The same fine structure, however, can be obtained by methods other than extensive working such as by thermal cycling¹² or warm temperature consolidation of rapidly quenched fine powders.¹³]

Japanese blacksmiths from ancient times also worked with ultrahigh carbon steels. An example is shown in Fig. 2 where "uwagane" steel is solid state welded to a piece of soft iron and then processed and heat treated to make a cutting tool with a hard tip and a soft matrix. The "uwagane" steel is of hyper-eutectoid composition as seen in the photomicrograph. It was made from "kera" steel, a high carbon steel of variable carbon content (1.5~2.0% C) which was hammered and folded many times. Presumably the special processing of "kera" steel was done to homo-

* Special Lecture held at the 96th ISIJ Meeting, October, 1978, at Toyama University, Toyama 930. Manuscript received March 22, 1979.

** Professor, Department of Materials Science and Engineering, Stanford University, Stanford, California 94305, U.S.A.

Photo. 1. Damascus swords used by the Russians in the War of 1612. Such swords were often of carbon content in the range of 1.5% C and were extensively mechanically worked prior to heat treatment.¹¹⁾ [Swords are displayed in the Armoury Museum in Moscow.] (Photo by courtesy of Artia Publishing Company.)

genize the structure and allow some decarburization to occur.¹⁴⁾

In the following sections we describe some of the unique properties that are achievable with ultrahigh carbon steels.

II. Superplasticity in UHC Steels

When we initiated our studies on UHC steels in 1973 we knew nothing about ancient steel making. We were totally ignorant of the metallurgy of Damascus steel and early Japanese laminated tool fabrication. What we were interested in was superplasticity. This property refers to the ability of metallic alloys to elongate extensively ($>400\%$) at warm temperatures. We had done considerable work in evaluating the superplastic characteristics of aluminum-zinc¹⁵⁾ and uranium base alloys.¹⁶⁾ This work led to an understanding of the many prerequisites needed for superplastic properties.¹⁷⁻¹⁸⁾ The major prerequisite is a fine grain size which should remain stable at superplastic forming temperatures. The grain boundaries, however, should be mobile enough to minimize high stress concentration at triple points, but should be inhibited from rapid grain growth by

Fig. 2. Japanese blacksmiths of ancient times made laminated tools by solid stage bonding ultrahigh carbon (UHC) steels, known as "uwagane" to soft iron. After hot forging, the laminated tool was heat-treated to yield a hard cutting edge (adapted from E. C. Bain¹⁴⁾).

the presence of fine particles of a second phase. The second phase should exhibit ductility at the superplastic forming temperature in order to avoid inter-phase boundary separation. In addition, the grains should consist of high angle boundaries.³⁾

We have made UHC steels superplastic by utilizing the principles presented above. The basic idea is to process such steels so that the matrix consists of ultra-fine ferrite grains stabilized by the presence of sub-micron cementite particles. The composition and temperature range where superplasticity can be expected is shown in the iron-carbon phase diagram of Fig. 3. As can be seen, the temperature range of potential superplastic behavior is quite wide. For example, for a 1.6% C steel, a UHC steel can be superplastic from 600°C to 800°C. This permits considerable flexibility in temperature control during superplastic forming of such materials.

Several methods²⁾ have been developed for preparing fine ferrite grains in UHC steels. The most common one is a hot and warm working process. In this method the UHC steel is heated into the austenite range and soaked for sufficiently long time to dissolve all the carbides. The material is then deformed continuously as it cools through the austenite-cementite region. This part of the process leads to a refinement of the austenite grains, to subgrain formation and to generation of randomly dispersed dislocations, allowing uniform precipitation of fine carbides at dislocation, subgrain and grain boundary sites. When the

A_1 temperature is reached (727°C), austenite of about eutectoid composition transforms to ferrite and cementite (usually as pearlite but a divorced eutectoid reaction is also possible¹⁹). Further deformation causes the cementite plates in pearlite to deform and refine into fine spheroidized particles. Isothermal deformation at 650°C is commonly used to refine the structure further. A typical strain during working in the austenite and austenite plus cementite region is 1.5 (*i.e.*, a reduction of 4 to 1 in rolling). About the same strain is used during working in the ferrite plus cementite region. During isothermal rolling at 650°C , about 5 to 10% per pass is commonly used with 5 min reheating soaks between passes. Photograph 2 reveals optical micrographs of the as-cast UHC steel and of the hot-and-warm processed UHC

Fig. 3. The above Fe-C phase diagram illustrates the range of composition and temperature where UHC steels can be superplastic if they are made sufficiently fine grained. A high carbon content is needed in order to introduce a large volume fraction of cementite particles for stabilization of the fine ferrite grains.

steel. The cast structure reveals the presence of coarse proeutectoid cementite particles in a matrix of ferrite and cementite. Hot and warm working leads to a refined structure consisting of ultrafine ferrite grains (typically $1\ \mu\text{m}$ or smaller) with a uniform distribution of submicron cementite particles. Such a material is superplastic at warm temperature. An example of a tensile test performed on a 1.6%C steel at 650°C and at a strain rate of 1%/min is shown in Fig. 4. This material exhibits 1155% elongation prior to failure.

As can be seen in the T.E.M. photomicrograph given in Photo. 3(a), the initial ferrite grain size after hot-and-warm processing is about $1/2\ \mu\text{m}$ with a high concentration of dislocations within each ferrite grain. Submicron cementite particles are readily evident. After superplastic deformation, the UHC steel sample reveals dislocation-free ferrite grains, equiaxed in shape, containing cementite particles (Photo. 3(b)). The ferrite grains are approximately $1\ \mu\text{m}$ in size revealing that some grain growth occurred as a result of concurrent deformation. The grip region of the superplastically deformed sample revealed very little

Fig. 4. The above photograph illustrates a sample of a fine grained UHC steel (1.6%C) deformed superplastically at 650°C . The initial engineering strain rate was 1%/min.

ULTRAHIGH CARBON STEEL (1.9% C)

- (a) Reveals coarse proeutectoid cementite particles in a matrix of ferrite and cementite in the as-cast state.
- (b) Reveals fine equiaxed structure of ferrite grains containing submicron cementite particles after hot and warm working.

Photo. 2. Optical photomicrographs of a 1.9%C steel

- (a) The microstructure of the steel in the as hot-and-warm rolled condition, revealing a submicron ferrite grain size with fine cementite particles and a high dislocation density within the ferrite grains
- (b) The microstructure of after 100% superplastic deformation at 650°C at 1%/min strain rate. Dislocation-free microstructure is noted with some ferrite grain growth occurring as a result of concurrent straining.

Photo. 3. Transmission electron micrographs of hot and warm worked 1.9%C steel

grain growth (although the high dislocation density within the ferrite grains decreased considerably) attesting to the stability of the ferrite grains from the presence of the carbide particles.

III. Tensile Strength and Ductility of UHC Steels at Room Temperature

The fine sub-micron ferrite grains created by extensive hot-and-warm working of the UHC steels would suggest that high tensile strengths at room temperature may be achieved. Figure 5 reveals the stress-strain curve of a 1.3%C UHC steel in the as-processed condition and after annealing at 650°C. The yield strength of the hot-and-warm processed UHC steel is about 1 400 MPa (200 ksi) with a tensile ductility of about 4% elongation. After annealing for 20 min at 650°C, the dislocation density is reduced, and the yield strength is decreased to 1 000 MPa (~140 ksi), but the tensile elongation is in the order of 20%. This high ductility is a surprising but pleasing result. Thus, the strengths of UHC steels are about twice those for HSLA steels (~500 MPa) but with essentially the same tensile ductility. Only sheet have been prepared of the UHC steel to date (~2 mm thick) so that studies on the impact characteristics and fracture toughness have not been investigated.

IV. Heat Treatment of UHC Steels

Ultrahigh carbon steels containing fine ferrite grains with a large volume fraction of fine spheroidized carbides represent materials with unique heat treating possibilities. This is because fine ferrite grains, containing many nuclei for transformation, will transform to fine austenite grains when heated to just above the A_1 temperature (727°C). In addition, undissolved cementite will exist (10vol% in a 1.6%C steel) which will stabilize the austenite grains and inhibit their growth. Upon quenching, ultrafine martensite will form because the martensite that forms cannot prop-

Sample A was warm-worked extensively at 565°C with no annealing after warm working.

Sample B was annealed for 20 min at 650°C after warm working at 565°C.

Fig. 5. True stress-true strain curves for a 1.3%C steel at room temperature.

agate through the austenite grain boundaries. This heat treating procedure is illustrated in Photo. 4. The quenched martensite structure is shown. It cannot be resolved optically, hence the term “quasi-martensite” is used as a subscript underneath the photomicrograph (apparently, such unresolvable structures in quenched hypereutectoid steels had been noted over fifty years ago²⁰) and the product was called “hardenite”). Recently, we have resolved the structure of “quasi-martensite” through transmission electron microscopy as submicron packets of lath martensite with a small fraction (about 20%) of ultrafine ($<0.1 \mu\text{m}$) twin martensite.⁷⁾

The compression properties of “quasi-martensite” are impressive. Figure 6 illustrates the point. The stress-strain behavior of the quenched UHC steel is such that yielding does not begin until a yield strength of over 2 300 MPa (~350 ksi) is surpassed. The material then deforms plastically to a strain of about 10% before fracture occurs at stresses approaching 4 300 MPa (~650 ksi). This strength is equivalent

HEAT TREATMENT OF ULTRA-HIGH CARBON STEEL (1.6 % C)

AS ROLLED

 $R_c = 37, \sigma_y = 150 \text{ ksi}$

CEMENTITE IN FERRITE

CEMENTITE IN AUSTENITE
(schematic)

QUENCHED

 $R_c = 68, \sigma_f = 600 \text{ ksi}$

CEMENTITE IN QUASI-MARTENSITE

Photo. 4. A partially schematic diagram illustrating the expected transformation from fine ferrite grains to fine austenite grains when heating a UHC steel to just above the critical temperature (727°C). Upon water quenching, optically unresolvable martensite is obtained. Transmission electron microscopy has revealed the presence of submicron lath and twin martensites.⁷⁾

to the compression fracture strength of tungsten carbide. Of note, however, is that, whereas tungsten carbide has virtually zero compression ductility, the UHC steel can plastically deform about 10% prior to fracture. This quality, together with a high hardness ($R_c \approx 68$), would suggest that such steels may be good candidates for use as wear or abrasion resistant materials.

V. Superplastic Bonding and UHC Steel Composites

The ultrafine ferrite grains produced in the hot-and-warm processed UHC steel (Photo. 3(a)) presents a unique situation. The material contains a large fraction of grain boundaries which represent regions of high atom mobility. In fact, at 650°C , the diffusivity of iron atoms is one million times faster in the ferrite grain boundary than in the matrix of ferrite.^{20,21)} Such efficient paths for atom mobility would suggest that UHC steel may be bonded readily in the solid state at temperatures below the A_1 temperature. This concept was verified by pressure bonding experiments performed at 650°C .²³⁾ Photograph 5 shows the excellent bond developed between two UHC steel plates (1.3%C) at 650°C when pressure bonded at a stress of 70 MPa (10 ksi) for 2 hr. The bond appears perfect. The line of demarcation seen in the photomicrograph is due to the flattening

Fig. 6. Room temperature compression stress-strain curve of a fine grained 1.3%C steel heat treated to 750°C and water quenched. A high compression ductility is obtained ($\sim 10\%$) with high compression fracture strength ($\sim 4300 \text{ MPa}$).

of the grains at the interface from deformation arising from the applied compression force. The plates shown in Photo. 5 were ground flat prior to pressing at 70 MPa at room temperature. They were then

Photo. 5. The above photomicrograph illustrates that superplastic UHC steel plates (1.3%C) can be readily solid state diffusion bonded in air at relatively low temperature ($0.51 T_m$) and low pressure (69 MPa). A line of demarcation can be seen at the bond interface because the cementite particles and ferrite grains are flattened at the interface, leading to flat ferrite grain and interphase boundaries.

heated in air to 650°C where solid state bonding was achieved. We believe the grain boundaries act as sinks for impurity atoms on the mating surface, keeping the surfaces clean and thus allowing for bonding by inter-atom exchange across the interface. Recently, we have achieved ready bonding of UHC plates by roll bonding at 650°C. In this case the ground plates are welded on the edges to minimize oxidation prior to roll bonding at temperature.²⁴⁾

The success of bonding fine grain UHC steel plates led us to consider the possibility of joining superplastic UHC steel to non-superplastic steels below the critical temperature of 727°C. This possibility of ready bonding seems quite reasonable if the presence of many grain boundaries leads to establishing clean surfaces at the interface of the plates. Photograph 6 reveals similar success was achieved when UHC steel was alternately layered with mild steel (1005) and pressure bonded at 650°C at about 70 MPa (10 ksi). A low magnification photomicrograph of the laminated composite is shown in the upper portion of Photo. 6 and a high magnification photomicrograph of one of the interfaces is shown in the lower portion of the figure. The bond appears metallurgically perfect. The ferrite in the mild steel appears in intimate contact with the ferrite in the UHC steel. No major interdiffusion due to chemical gradients has occurred because the carbon content in solution in both steels is the same (~ 0.01 carbon) and lattice diffusion of substitutional atoms would be very slow at 650°C ($D \approx 10^{-15} \text{ cm}^2/\text{sec}$).²⁰⁾ Laminated composites of the type shown in Photo. 6 have also been prepared by the roll bonding technique described earlier.

Photo. 6. (a) A twenty five layer laminated composite of UHC steel-1005 steel is shown after pressure bonding (69 MPa) at 650°C.
(b) A high magnification optical photomicrograph of the interface between a UHC steel-1005 steel layer, shown below the laminated composite, indicates that a good metallurgical bond is present,

Three predictions can be made with respect to the laminated composite depicted in Photo. 6. One relates to an applied mechanics argument regarding the possibility of making the laminated composite superplastic at warm temperature [The argument goes as follows: if the UHC steel is superplastic but stronger than mild steel at warm temperature, it will control the plastic flow behavior of the composite when tested in the layered direction. Thus, the composite as a whole will behave superplastically barring unexpected effects arising at the interface of the dissimilar steels]. Second, the laminated composite should not experience delamination since a true metallurgical bond is developed involving an indistinguishable and continuous ferrite phase. Third, the laminated composite is capable of "selective heat treatment." That is, when the laminated composite is heated just above the A_1 temperature, the UHC steel will transform to austenite plus cementite whereas the mild steel will remain untransformed. Upon quenching, the UHC steel will become martensite of Rockwell "C" 68 perfectly bonded to the unaffected mild steel. The result should be a strong and fracture tough material. The next three photographs illustrate the possible validity of the three predictions outlined above.

Photograph 7 shows the possible superplastic tendency in the laminated UHC steel-mild steel composite. This eleven layer laminated composite was heated to about 600°C then bent with pliers to a zero T bend. The uniform deformation of each layer and the total absence of any tendency to crack would suggest that the laminated composite may be superplastic.

Photograph 8 represents a severe bend performed at room temperature of a two layer UHC steel-mild steel composite in the as-pressed condition. A photomicrograph of the interface at the most severely deformed region indicates no sign whatsoever of any delamination. Another sample was bent at right angles to that shown in Photo. 8. Again no sign of delamination was observed. Both observations attest to the integrity of the metallurgical bond between the UHC steel-mild steel plates.

Photograph 9 illustrates in a yet more dramatic way the integrity of the bond between a UHC steel-mild steel composite prepared by warm pressing. In this case, the two layer composite, after warm pressing at 650°C, was "selectively heat treated." It was heated to 770°C for 20 min and then water quenched. The sample was then bent at room temperature until cracks were visually observed in the martensite portion of the composite. These cracks can be seen in the low magnification photograph (Photo. 9). At high magnification one specific large crack is shown. The crack can be seen to be arrested just as it enters the mild steel region. A fine crack is seen to propagate transversely from the main crack. This crack propagates only in the martensite region. No delamination is evident at the original interface of the two layer composite or in the narrow interdiffusion zone that formed during heat treating at 770°C (the carbon

gradient in the austenite that forms at the interface is the driving force for interdiffusion). This example attests to the potential of optimizing strength and toughness of UHC steel composites through selective heat treatment.

Lamination of steels by solid state bonding is not a new process. Mention was already made that Japanese cutting tools (Fig. 2) were prepared in this manner. Bonding appeared to have been made invariably at temperatures much above the A_1 where considerable interdiffusion occurs. Our own studies indicate that bonding is relatively easy to achieve below the A_1 temperature if one of the steels is superplastic. This concept and practice seems novel in the history of contemporary metallurgy. There is evidence, however, that over 2 000 years ago, the ancient blacksmiths of Asia Minor may have performed similar processing treatments.²⁵⁾ Photograph 10 illustrates the magnified region of an interface in an Adze blade dated 400 BC found on the coast of Turkey. The dissimilar carbon content steels appear to be metallurgically bonded, and the possible absence of an interdiffusion zone suggests that the two dissimilar steels making up the bonded blade may have been hammered at temperatures below the A_1 temperature. The dark regions near the interface may be gra-

Photo. 7.

A zero T bend is readily achieved in an eleven layer laminated composite of a 1.6% C UHC steel-1005 steel at warm temperature ($\sim 600^\circ\text{C}$). The uniform deformation of each layer, with no evidence of fracture, attests to the possible superplastic nature of the laminated composite. The thickness of the composite was 2.5 mm.

1005 - ULTRAHIGH CARBON STEEL (1.3%C) LAMINATE COMPOSITE
NOT HEAT TREATED

Photo. 8. An as warm-pressed two layer laminated composite of a 1.3%C UHC steel-mild steel is bent at room temperature. Photomicrographs of the interface at the bent region reveal no sign of delamination attesting to the sound metallurgical bond in the laminate.

Photo. 9.
A 1.3%C UHC steel-mild steel composite after " Selective-heat-treatment " (770°C for 20 min then water quenched) is bent until cracks in the martensite portion of the composite become visible. The photomicrograph shown to the left illustrates that the crack in the martensite region is arrested in the mild steel region. No delamination is noted in the interface region indicating that a good metallurgical bond is present in the laminated composite.

phitized regions due to extensive mechanical working if graphitizing elements such as silicon or nickel were present in the high carbon steel. The similarity between the Adze blade and our UHC steel composite (right side of Photo. 10) shown at the same magnification, is striking.

Superplastic bonding of steels may prove a useful

method of welding without the deleterious features of heat affected zones as in conventional liquid phase welding. In Photo. 11 we illustrate two 4140 couplings joined at 700°C by pressure bonding with a UHC steel wafer. The tensile strength of the bond was evaluated by mating two solid cylinders of 4140, square in cross-section, with a UHC steel superplastic

(a) Adze Blade, Al Mina, Coast of Turkey (About 400 B.C.)
 (b) UHC steel-mild steel composite, Stanford University, (California (1978))

Photo. 10. Photomicrograph of the interface of a laminated Adze Blade (dated 400 B.C.) attests to the metallurgical skill of the blacksmiths of ancient time (left figure). The bond is compared with the UHC steel-mild steel composite press bonded at Stanford (right figure).

(The author thanks Dr. Robert Maddin for permission to publish the Adze blade interface shown above.)

Photo. 11. Two 4140 steel couplings are joined by pressure bonding with a 1.6% C UHC steel at 700°C (Courtesy of E. Slaughter, Pratt and Whitney Corp, West Palm Beach, Florida).

wafer. After appropriate pressure bonding at 650°C, samples were machined for tensile testing at room temperature. Photograph 12 shows three tensile samples. The left specimen is as-machined, and the middle specimen is a similar one but etched to show the position of the UHC steel wafer. The specimen on the right was tensile deformed to failure. Failure occurred outside of the UHC steel-4140 region indicating the high integrity of the bond (the yield strength of the sample was 400 MPa).

VI. Summary and Conclusions

Recent research work at Stanford University has led to the development of submicron ferrite grains containing very fine cementite particles in ultrahigh carbon (UHC) steels. Much of the work is centered on plain carbon steels containing 1 to 2.3% C (15 to 35 vol% cementite, respectively). These UHC steels have been shown to exhibit at least four unusual and

Photo. 12. Tensile test samples of 4140 steel solid state bonded with a 1.6% C UHC steel.

The two samples on the left are illustrated in the as-machined condition, unetched and etched respectively.

The sample on the right is shown after testing at room temperature.

The sample failed in a region away from the welded zone attesting to the soundness of the UHC steel-4140 steel interface.

desirable characteristics:

(1) They are superplastic at warm temperatures (elongations exceeding 1 500% have been achieved).

(2) They are both strong and ductile at room temperature after appropriate thermomechanical treatment ($\sim 1\,000$ MPa yield strength with 10–20% ductility are readily achievable).

(3) They can be made exceptionally hard ($R_c = 68$) by appropriate heat treatment with high compression fracture strength ($\sim 4\,300$ MPa) and good compression ductility (~ 7 –10%).

(4) They can be bonded superplastically to other ferritic steels at temperatures below the lower critical temperature ($< 727^\circ\text{C}$). Laminated composites made by such bonding techniques exhibit metallurgically sound bonds, are superplastic at warm temperature, and can be selectively heat treated to result in strong and tough material.

Many of the features of the ultrahigh carbon steels described herein appear to be similar to those attributed to the legendary Damascus steels and to ancient Japanese laminated tools. The blacksmiths of centuries ago were known to work with ultrahigh carbon steels and the thermal mechanical processing techniques used to make them could well have been similar to those utilized in our studies.

Acknowledgements

The author is indebted to many friends and colleagues for the results described in this paper. Dr. Conrad Young and this author jointly conceived, in 1973, the idea that steels could be made superplastic when the carbon content was in the range 1 to 2% C. Mr. Eldon Cady and Dr. Bruno Walser were instrumental in the early work that lead to the successful development of these superplastic steels. Dr. Sabri Kayali helped establish the importance of obtaining high angle boundaries as a prerequisite for superplasticity and Professor H. Sunada determined the unique properties of heat treated UHC steels. Dr. Robert Caligiuri, Mr. Robert Whalen, Mr. L. Eiselstein and Mr. Barry Snyder did the early tests on superplastic bonding and ultrahigh carbon steel composites. Dr. Jeffrey Wadsworth has worked on all phases of the UHC steel program since 1976 and has contributed extensively to the ideas presented in this paper. The author is deeply grateful to all these colleagues and participants, with whom he is still in intimate contact on the subject of UHC steels. The research work was sponsored by the Defense Advanced Research Projects Agency and by the Office of Naval Research and special thanks are extended to Drs. Arden Bement, Bruce MacDonald and Edward Van Reuth for their helpful discussions and guidance.

The author is deeply indebted to Mr. R. Perkins and Dr. T. Tietz for use of the metal workings facilities at Lockheed Laboratories in Palo Alto.

Professors S. Umekawa and T. Suzuki of the Tokyo Institute of Technology assisted in the preparation of this manuscript for publication in the journal of the

Iron and Steel Institute of Japan and the author gratefully acknowledges this assistance.

REFERENCES

- 1) O. D. Sherby, B. Walser, C. M. Young and E. M. Cady: *Scripta Met.*, **9** (1975), 569.
- 2) O. D. Sherby, C. M. Young, B. Walser and E. M. Cady: "Superplastic Ultrahigh Carbon Steels", U. S. Patent No. 3 951 697, April 20, 1976.
- 3) B. Walter, E. S. Kayali and O. D. Sherby: 4th International Conference on the Strength of Metals and Alloys, Nancy, Sept. 1976, Vol. I, 456.
- 4) J. Wadsworth and O. D. Sherby: *J. Mechanical Working Tech.*, **2** (1978), 53.
- 5) J. Wadsworth and O. D. Sherby: *J. Mat. Sci.*, **13** (1978), 2645.
- 6) B. Walser and O. D. Sherby: *Met. Trans.*, (1979), in press.
- 7) H. Sunada, J. L. Lin, J. Wadsworth and O. D. Sherby: *Int. J. Mats. Sci and Engr.*, **38** (1979), 35.
- 8) T. Oyama, J. Wadsworth, M. Korchynsky and O. D. Sherby: Fifth International Conference on the Strength of Metals and Alloys, Aachen, August, 1979.
- 9) J. Wadsworth and O. D. Sherby: *Foundry Management and Technology*, **106** (1978), 59.
- 10) Metals Handbook, VIII, American Society for Metals, Ohio, (1973).
- 11) N. Belaiew: *JISI*, **97** (1918), 417.
- 12) S. Kayali, H. Sunada, T. Oyama, J. Wadsworth and O. D. Sherby: *J. Mat. Sci.*, (1979), in press.
- 13) L. E. Eiselstein: Ph. D. Thesis, Department of Materials Science and Engineering, Stanford University, Calif. 94305, 1979.
- 14) E. C. Bain: *JISI*, **200** (1962), 265.
- 15) C. M. Packer and O. D. Sherby: *Trans. ASM*, **60** (1967), 21.
- 16) O. D. Sherby, D. L. Bly and D. H. Wood: Physical Metallurgy of Uranium Alloys, J. J. Burke, D. A. Colling, A. E. Gorum and J. Greenspan, ed., Brook Hill Publishing Company, Mass., (1976).
- 17) O. D. Sherby: *Science J.*, **5** (1969), 75.
- 18) O. D. Sherby, R. D. Caligiuri, E. S. Kayali and R. A. White: Proceedings of the Twenty Fifth Sagamore Conference, New York, July, (1978); Recent Advances in Metal Processing, ed. by J. J. Burke, V. Weiss and R. Mehrabian, Syracuse Univ. Press, New York, in press.
- 19) J. H. Lin: Ph. D. Thesis, Department of Materials Science and Engineering, Stanford University, Calif. 94305, August, 1979.
- 20) S. L. Hoyt: Metallography, II, McGraw-Hill Book Company, Inc., New York, (1921), 167.
- 21) F. S. Buffington, K. Hirano and M. Cohen: *Acta Met.*, **9** (1961), 434.
- 22) D. W. James and G. M. Leak: *Phil. Mag.*, **12** (1965), 491.
- 23) R. Caligiuri: Ph. D. Thesis, Department of Materials Science and Engineering, Stanford University, Calif. 94305, August, 1977.
- 24) J. Wadsworth, R. Caligiuri, R. Whalen, L. E. Eiselstein, B. C. Snyder and O. D. Sherby: presented at Annual AIME Meeting, New Orleans February 20–22, (1979).
- 25) R. Maddin, J. D. Muhly and T. S. Wheeler: *Scientific American*, **237** (1977), 122.