

**A TRIBUTE TO PROFESSOR WALTER SHEPHERD OWEN –
ONE OF THE PIONEERS IN
THE FIELD OF MARTENSITIC TRANSFORMATION**

Mica Grujicic, Ph.D.

Wilfred P. and Helen S. Tiencken Professor
Department of Mechanical Engineering
241 Fluor Daniel Building Clemson University
Clemson, SC 29634-0921
Phone: 864-656-5639 Fax: 864-656-4435 mica@ces.clemson.edu

Frederick J. Schoen, M.D., Ph.D.

Executive Vice Chairman, Department of Pathology,
Brigham and Women's Hospital;
Professor of Pathology and Health Sciences and Technology (HST),
Harvard Medical School
75 Francis Street Boston, MA 02115
Phone: 617-732-5709 Fax: 617-232-9820 fschoen@partners.org

ABSTRACT

This manuscript is a tribute to Professor Walter Shepherd Owen, e, from two of his former students. Professor Owen died on October 10, 2007 in Boston, MA. He was 87. The world lost an internationally recognized scientist, an accomplished academic, an effective administrator, a great communicator, a prolific researcher and a pioneer in the field of physical metallurgy. To both of us he was a beloved friend and mentor. From his early days at Liverpool University to his retirement from MIT, he was an inspired research worker, a dedicated adviser and guide, a distinguished international scientist, and an effective University planner and administrator. The present manuscript provides a brief biography of Professor Owen and an overview of his main contributions to metallurgy and particularly the field of martensitic transformation, to which he contributed greatly over nearly half a century.

1. A SHORT BIOGRAPHY



Professor Owen was born and grew up in post-World War One Liverpool in West Derby. While his father provided Professor Owen with the Welsh blood and the family name, his mother instilled in him strong intellectual skills and deep socialist beliefs. Upon completing the elementary and high-school education, he enrolled in the Metallurgy Department at the University of Liverpool from where he graduated with a First Class Honors B. Eng degree in 1940. Although the Second World War interrupted his academic career he was still able to use his skills in the research laboratories at the English Electric Company and was able to earn his M. Eng degree from the University of Liverpool in 1942.

After the war Professor Owen returned to the Liverpool University as Lecturer in the Metallurgy Department. He received a PhD from the same university in 1950 and soon joined his academic “hero” Professor Morris Cohen, on a one year post doctoral Commonwealth Fund fellowship, in the Department of Metallurgy at the Massachusetts Institute of Technology (MIT). While at MIT, he developed a keen interest into bainitic, massive and martensitic transformations. He returned briefly to the Liverpool University to be called back by MIT in 1954 to assume the position of a research scientist and to lead a team investigating the brittle fracture of steels. But by 1957 he returned to the Liverpool University as Head of the Metallurgy Department and the Henry Bell Wortley Professor of Metallurgy.

During his 9-year tenure as Head of the Metallurgy Department, he transformed the Department from a small academic program housed in an old workhouse to a large research and teaching institution. During this time, the department moved to the fully-refurbished George Holt Laboratories, which had been home to the Physics Department, and Professor Owen had a new office which was once used by Nobel Prize winner James Chadwick, who discovered the neutron. During this time, the department truly flourished as talented teachers, researchers and students joined the department while Professor Owen used his industry contacts to secure funding for many large and ambitious research projects. Another of his major achievements was the introduction of new undergraduate courses to attract students from the departments of Physics and Chemistry to study metallurgy in their final year. Professor Owen also took an important leadership role within the University of Liverpool, as Dean of Engineering. Also while at Liverpool he joined Robert Maxwell’s Pergamon Press to stimulate publishing in the fields of metallurgy and materials science.

After spending nine years at the University of Liverpool, Professor Owen returned to the United States in 1966 to assume the position of Chair of the Department of Metallurgy at Cornell University later moving on to Northwestern University where he became the Dean of the Technological Institute and Vice President for Science and Research. He returned to MIT in 1973, to lead what is now known as the Department of Materials Science and Engineering as its Chair. One of his first challenges was to build consensus on renaming his department from Metallurgy and Materials Science to the Department of Materials Science and Engineering, a change that he accomplished in his second year as department head. The change was not universally popular, especially by some who were displeased that the word “*metallurgy*” had been dropped from the department’s title after first appearing in 1889. Soon afterwards, other top-ranked departments followed. In his decade of tenure as Head of the Department of Materials Science and Engineering at MIT, Professor Owen led a major effort to diversify and revitalize the faculty. This resulted in significant growth in the department’s ceramics and polymers programs while its historical strength in metallurgy was maintained. His stewardship of the department was critical to sustaining its top-ranked position in the succeeding decades.

For his long and distinguished role in the materials science and engineering community and for "*Leadership in Research on Structure/Property Relationships in Metals and in the Extension of Such Concepts to the Educational Basis of Materials Science and Engineering*", he was elected to the National Academy of Engineering in 1977.

Professor Owen retired from MIT in 1982 and for the last 20 years of his life he divided his time between Cambridge, Massachusetts and his beloved Welsh retreat in Porthmadog. In his later years he loved nothing more than walking along the cliff paths and chatting to the Welsh locals. In commemoration of his 80th birthday, the local authority in Porthmadog placed a plaque on a cliff top bench “*To Celebrate the Life of Professor Walter S. Owen*”.

After retirement, Professor Owen maintained a deep commitment to research, was both active in consulting and deeply committed to the education of materials scientists and engineers and to furthering the evolution and development of the field of materials science and engineering, particularly in the areas of conservation, recycling, environmental concerns, public policy, and materials availability.

2. OVERVIEW OF SELECTED CONTRIBUTIONS TO THE FIELD OF MARTENSITIC TRANSFORMATION

Professor Owen's research career focused on the mechanisms and consequences of martensitic transformation in ferrous systems. In an effort to illustrate the high level and longevity of contribution and his impact, we have chosen to highlight specific papers and the logic of his research themes over his entire career. Of necessity, some important contributions are omitted. Since Professor Owen worked on various aspect of martensite at three different academic/research institutions: the University of Liverpool, Cornell University and MIT, this section is structured accordingly.

2.1 The University of Liverpool Years

Strength of Virgin Ferrous Martensite: The extreme hardness of quenched steel is a striking phenomenon which has been the subject of speculation from earliest times. In 1939, Carpenter and Robertson [1] discussed a number of possible explanations, including age hardening and the high internal strain of freshly quenched martensite, but concluded that "*the ultimate cause of the hardness remains to be discovered.*" In 1962, Winchell and Cohen provided the answer [2-5]: The fundamental cause is interstitial solid-solution hardening. Fe-Ni-C alloys with a transformation temperature M_s below about 250K transform to internally twinned plate martensite with an irrational habit in the vicinity of $\{3,10,15\}_\gamma$. If the M_s temperature is near or above room temperature, lath martensite with a dislocation substructure and a habit plane near $\{111\}_\gamma$ is formed. An important distinction is that the former is tetragonal but the latter is cubic. Both are hardened substantially by carbon in solution. Winchell and Cohen [2] found that tetragonal martensite readily age hardens at temperatures near room temperature. This presented them with a problem of determining the effect of carbon content on strength of the un-aged ("virgin") martensite in Fe-Ni-C alloys. Consequently, they had to correct their experimental results in order to remove the effect of aging that had occurred in "*as-quenched*" specimen during holding before testing.

Roberts and Owen [6,7] extended the experimental studies of Winchell and Cohen [2-5] in two significant aspects: (a) To avoid the necessity for back extrapolation, they devised a method of rapidly quenching ($>5 \times 10^3$ K/s) and reheating thin-foil specimens to eliminate effects due to aging or tempering during specimen preparation and testing; and (b) by measuring the temperature and strain-rate dependence of the flow stress of both tetragonal twinned plate martensite, (similar to that studied by Winchell and Cohen [2-5]), and cubic lath martensite with a dislocation substructure. That this procedure was successful, even when applied to the Fe-21.3Ni-C series of alloys which formed cubic martensite at temperatures just below 373 K, is convincingly demonstrated by comparing the measured flow stresses in the two studies (Figures 1 and 2). The effects of aging above 213 K are evident in the results of Winchell and Cohen but are totally absent from those reported by Roberts and Owen.

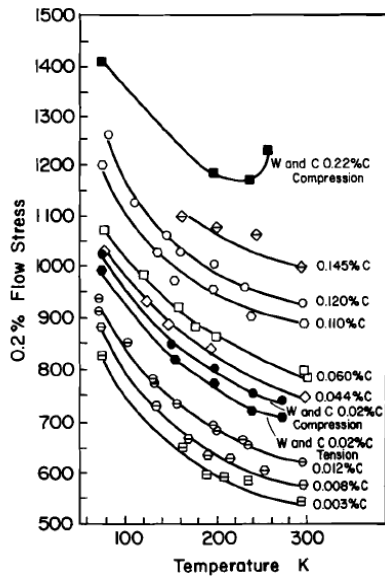


Figure 1. Flow stress as a function of test temperature. Open symbols: Fe-21.3Ni-C cubic martensite; Roberts and Owen [7]. Solid symbols: Fe-Ni-C tetragonal martensite; Winchell and Cohen [2].

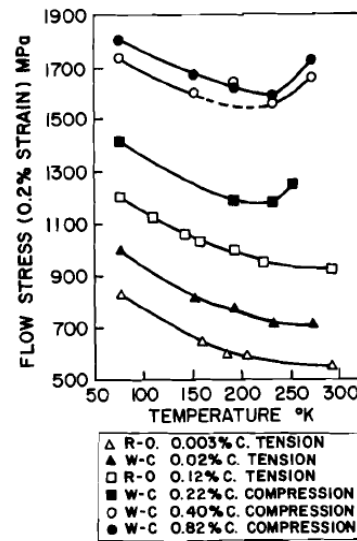


Figure 2. Flow stress as a function of test temperature, showing the effects of aging above 200 K. W-C: Winchell and Cohen [2]. R-O: Roberts and Owen [7].

Thermally-activated Plastic Flow in Virgin Martensite: Roberts and Owen [7] made a significant contribution to utilizing stress relaxation and strain-rate change experiments over sufficient ranges of strain rate and carbon concentration to determine the thermal activation parameters controlling the kinetics of deformation of ferrous martensite. These measurements, together with the measurements of the temperature dependence of the flow stress at a constant strain rate, allowed them to define and parameterize one of the first dislocation-based models for thermally-activated plastic flow in martensite [7].

Chemical Hardening of Virgin Martensite: Early attempts to model the plastic flow of virgin martensite [3,5] assumed a random distribution of carbon atoms on the assigned octahedral sites of the martensite. If, however, the carbon is not arranged randomly, the possibility arises of significant strengthening resulting from an increase in the free energy of the system by a change in configurational entropy brought about by the redistribution of carbon atoms by a dislocation sweeping a slip plane. This effect, often called “*chemical hardening*” and its role in plastic deformation of virgin martensite was first discussed by Roberts and Owen [7]. Their findings were subsequently confirmed by the observation that plastic deformation decreases the tetragonality of virgin martensite suggesting that a moving dislocation redistributes the carbon among the three types of octahedral sites or perhaps into tetrahedral sites [8], thereby adding to the incremental change in energy and, consequently, to the increase in resistance to plastic flow.

2.2 The Cornell University Years

The Role of Solute Drag in the Kinetics of Martensitic Interfacial Motion: The most significant research dealing with the martensitic transformation carried out by Professor Owen

and his students at Cornell University dealt with the role of solute drag in the kinetics of martensitic interfacial motion [9,10]. They assumed that drag is a result of the elastic interaction of the strain fields of dissolved solute atoms in the martensite lattice with the stress field of the moving martensitic interface. Only interstitial solutes were considered, because their associated anisotropic strain field is much larger than that of substitutional solutes. As a result of this interaction, the energy of the crystal may be reduced, under proper conditions of interfacial velocity and temperature, by one of two processes: (a) rearrangement of solute atoms by single jumps into energetically favorable, crystallographically equivalent neighboring sites, the Snoek model [11,12], or (b) solute diffusion over many atomic distances, the most favorable sites being chosen at any distance from the interface, the Cottrell model [13,14]. The Snoek and Cottrell models are each simplified versions of the solute drag phenomenon. During the time required for redistribution of the solute, by either process, the interface moves a short distance so that the solute atoms are no longer in minimum-energy positions corresponding to the new position of the interface. The result is that the solute redistribution is out of phase with the instantaneous low-energy configurations of the system and energy dissipation occurs, causing a drag on the interface.

Owen et al. [9,10] adopted the solute drag model to rationalize the slow isothermal growth of $\{111\}_\gamma$ martensite in Fe-10Ni-C alloys. The elastic interaction of the strain fields of dissolved carbon atoms in the martensite lattice with the stress field of the moving martensitic interface was determined for a specific dislocation model of the interface, consistent with the particular transformation crystallography. Only the anti-coherency interfacial dislocations were considered. It was found that the drag force produced by the Snoek interaction is quite insignificant compared with that produced by the Cottrell interaction.

According to the latter model, the long range diffusion of solute atoms to an interface moving at a constant velocity gives rise to a steady-state concentration profile, known as the Cottrell atmosphere. This profile is obtained by solving the governing differential equation for the velocity of the interface, v_R , relative to the solute atoms in the atmosphere:

$$v_R = -\frac{D}{kT} \frac{\partial w}{\partial x} - \frac{D}{C_0} \frac{\partial C_0}{\partial x} - v \quad (1)$$

where D is the diffusion coefficient, w is the solute/interface elastic interaction energy, C_0 the solute concentration, k the Boltzmann constant, T the temperature, x a spatial coordinate, and v is the interface velocity. Equation 1 is subjected to the appropriate boundary conditions and the continuity condition $(\partial / \partial x)(C_0 v_R) = 0$. The total solute drag force is then obtained by integrating the solute/interface interaction forces over the volume of the Cottrell atmosphere. An exact, analytical, closed-form solution for the drag force is obtainable only in the limits of low and high interfacial velocities. However, an approximate solution for the drag force fitting all velocity regimes can be developed [15].

An example of the calculated drag force vs. interfacial velocity results at 723K for the Cottrell interaction of carbon atoms with the martensite interface in a series of Fe-10Ni-C alloys with different carbon levels is shown in Figure 3. It is interesting to note that, for a typical magnitude of the thermodynamic driving force at the M_s of about 1090 J/mol, none of the alloys investigated experience an instability in the interfacial motion. Consequently, a stable-equilibrium condition corresponding to the balance between the driving force and the drag forces is attained. This finding then predicts in a stable slow-rate martensite growth, Figure 4, which

was confirmed experimentally [9,10]. Furthermore, except for the alloys with the least amount of carbon, the drag force is predicted to be a linear function of the interfacial velocity, Figure 3.

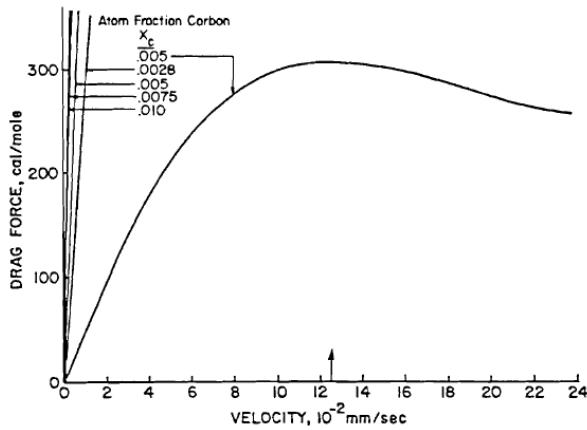


Figure 3. Calculated force-velocity curves for the Cottrell interaction of carbon atoms with the massive martensite boundary at 723 K for various carbon levels [10].

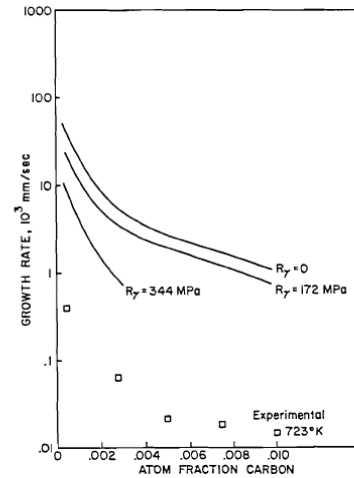


Figure 4. Comparison of experimental and theoretically calculated width growth rates for Fe-10Ni-C alloys at 723 K [10].

A comparison of the calculated interfacial velocities and measured growth rates (in the width direction) for Fe-10Ni-C alloys at 723K is given in Figure 4. The calculated curves are for different values of R_γ , and correspond to different approximations regarding the additional frictional forces opposing the interfacial motion. These were assumed to be independent of the velocity of the interface and, hence, to simply lower the net driving force. The model predicts a steep drop in growth rate with the initial small addition of carbon, in excellent agreement with the experimental observations. However, the growth rates predicted by the solute drag model [9,10] were one or two orders of magnitude larger than their measured counterparts. The extent of this discrepancy was found to be affected by several factors such as: (a) the chosen value of the additional frictional forces; (b) the choice of the interfacial-dislocation core cutoff radius; and (c) the assumptions made regarding the diffusion coefficient within the Cottrell atmosphere. While these factors can be adjusted to obtain a good quantitative agreement between the model and the experiment, the assumptions required could not be deemed fully justified. Schoen and Owen [10] correctly pointed out that the main cause for the over-prediction of the martensite growth rates is the assumption of linear elastic interactions used in their model.

2.3 The MIT Years

Definition of Thermo-elastic Martensitic Transformation: One of the significant contributions of Professor Owen and his students at MIT was helping properly define the notion of thermo-elasticity. There are two sources of dislocation generation that are transformation induced. If the transformation shape change cannot be totally accommodated elastically and appreciable plastic deformation by slip takes place, an increase in the matrix dislocation density results, with possible loss of interfacial mobility due to dislocation jamming. The concurrent relaxation of the elastic field around the martensite particle also decreases the thermodynamic driving force

available during the reverse interface motion. Transformations accompanied by significant plastic accommodation are not thermo-elastic.

Dislocations may also be generated by the moving interface. In many alloys in which the transformation is thermo-elastic, the lattice invariant strain is produced by the formation of a stack of thin, parallel twins with alternating twinning directions. At the interface of such a plate, there are local accommodation strains which may be elastic when the twins are thin, but plastic if they are not [16]. When a plate of the latter type shrinks, the moving interface leaves behind trails of sessile dislocation loops. On reversing the direction of the moving interface, the interface remains stable and glissile as it moves through the dislocation debris [17]. The density of dislocations in the matrix increases with each reversal of growth direction until it reaches some limiting density [18]. Thus, in these alloys, there is local irreversible deformation produced not by the shape change but by the moving interface.

Taking all these facts into account, Ling and Owen [19,20] classified a transformation as thermo-elastic if the motion of the interface is reversible, even though there may be a significant thermal hysteresis due to a high density of interface generated dislocations in the matrix. They further identified elastic accommodation of the transformation shape change as the essential condition for thermo-elastic growth. With this definition, thermo-elastic growth is favored by a small chemical driving force at the M_s , a small transformation shape strain, and a matrix with a high flow stress [21].

Most thermo-elastic alloys, such as the inter-metallic β -phase alloys, have these properties. An exception is a group of ferromagnetic iron alloys, such as ordered Fe_3Pt , which have a shape deformation similar in magnitude to that in other ferrous alloys that do not transform thermo-elastically. The chemical driving force in ordered Fe_3Pt is also substantially higher than in β -phase alloys. Ling and Owen [19,20] recognized that this, in part, reflects the larger elastic accommodation strain energy, but also that the absence of soft elastic modes (comparable with those found in the β -phase alloys), is an additional important reason why the driving force at M_s is relatively large in Fe_3Pt .

Computer Modeling of the Growth of Thermo-elastic Martensite: Ling and Owen [19,20] developed one of the first computer models of thermo-elastic growth which enabled determination of the largest plate which can behave thermo-elastically. They were also able to calculate the associated strain energy and, thus, the maximum driving force at the M_s . By calculating the plastic-zone size of a martensite particle in equilibrium at the M_s temperature, the model was next used to assess relative propensity of a given alloy system to exhibit thermo-elastic growth.

The model of Ling and Owen [19,20] was a three-dimensional continuum model which considered contributions of both the shear and the dilatational transformation strains. The first plate of martensite to form at M_s was assumed to nucleate within a grain in a polycrystalline aggregate and to grow rapidly in the radial direction, until stopped by a grain boundary or other barrier. The plate is then assumed to thicken in a direction, normal to the plane of the thin plate. The resulting lenticular plate was represented by an oblate spheroid. The matrix and the martensite were assumed to be elastically isotropic, but with different elastic moduli.

The martensite plate was assumed to be a coherent particle embedded in an infinite matrix of the parent phase. The martensitic-transformation induced stresses in the plate and in the matrix were calculated using Eshelby's model [22]. It was further assumed that plastic flow occurs in a

volume enclosed by a surface at which the von Mises criterion for yielding is satisfied. The effect of yielding-induced stress relaxation on the size of the transformation zone was not considered since in order to understand thermo-elastic growth, it is important to know only whether or not significant plastic flow occurs.

The elastic properties of the matrix were found to play a significant role in the process of accommodation. In this regard, a favorable factor for thermo-elastic growth is pre-martensitic elastic softening of the lattice (Figure 5), which may be caused by different phenomena in different alloy systems. Alloys that transform thermo-elastically all have an anomalous positive temperature dependence of the elastic constant $C' = (C_{11} - C_{12})/2$ over a temperature range that includes the M_s temperature. Elastic softening affects the transformation in two ways: it reduces the chemical driving force at the nucleation stage, and it facilitates the elastic accommodation of the lattice shape change. Thus, a plate that grows thermo-elastically at M_s will continue this form of growth as the temperature is lowered.

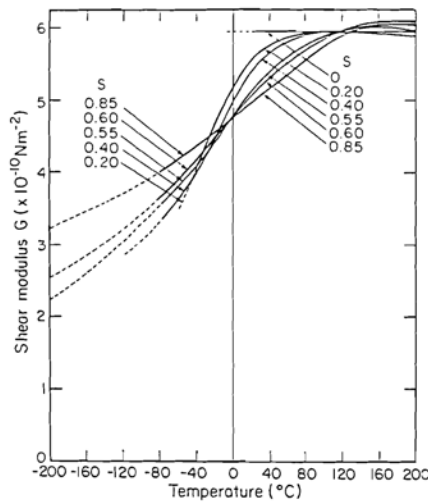


Figure 5. Temperature dependence of the average shear modulus, G , as a function of the degree of order S in Fe_3Pt [20].

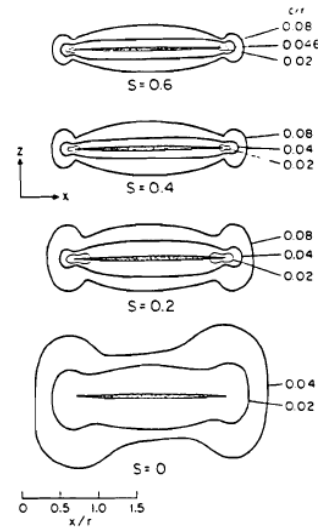


Figure 6. Matrix yield stress contours of an oblate spheroid lying in the x - y plane with the transformation shear in the x direction [20].

Most thermo-elastic alloys possess long-range order which increases their yield strength and, thus, inhibits plastic yielding. Alloys that are ferromagnetic at the M_s temperature, such as Fe_3Pt , are further strengthened by a magnetostriction effect known as "Invar strengthening," which is a function of the degree of order, S [23]. Ling and Owen [19,20] made a major contribution to our better understanding of the role of long-range order in accommodation of the macroscopic shape change accompanying martensitic transformation. An example of the results obtained by Ling and Owen [19,20] pertaining to the size of the yield envelopes associated with an isolated martensite plate as a function of the degree of long-range order S is shown in Figure 6. The change in the yield-envelope size demonstrates the importance of long-range order. A thin plate (aspect ratio = 0.02) in a disordered matrix ($S = 0$) is completely enclosed within a yield envelope. However, in all cases where $S > 0.2$, matrix yielding is confined to a small zone at the radial edge of the plate. The change is particularly striking between $S = 0$ and $S = 0.2$. Over this

narrow range, the elastic softening is dramatic (Figure 6), and the M_s temperature decreases more than 80 K . The growth is predicted to be irreversible in disordered specimens but to be thermo-elastic in those with $S \geq 0.2$. Subsequent experimental investigation of the martensite growth in Fe_3Pt confirm these predictions [19,20].

The model of Ling and Owen [19,20] revealed the following important basic fact: Even in the most favorable circumstances when a Fe_3Pt alloy is nearly fully ordered, growth of an isolated plate can proceed thermo-elastically only while the plate is thin. In sharp contrast, their experimental investigation showed that the transformation in ordered Fe_3Pt goes to completion while remaining thermo-elastic. Clearly, for this to happen, the transformation strains must be relieved by a reversible elastic deformation process. In the case of ordered Fe_3Pt , they observed that additional thin plates of martensite are repeatedly nucleated in the accommodation strain fields of previously nucleated plates.

Following the aforementioned observation, Ling and Owen [19,20] expanded their model to include the case of multiple nucleated plates. In general, as an isolated plate of thermo-elastic martensite thickens, the large shear stresses near the edge (tip) and over the broad surface of the plate assist the nucleation of adjacent plates. In the plastic zone near the edge, nucleation may be strain induced, and nonparallel crystallographic variants of the plate may form edge to edge. Alternatively, stress-assisted nucleation adjacent to the broad interface may produce a variant approximately parallel to the original plate. The change in the yield-envelope size and shape associated with putting together two parallel plates of opposite shear touching along the z -axis obtained by Ling and Owen [19,20] is displayed in Figure 7. It is seen that while the yield envelope of an isolated oblate spheroid ($c/r = 0.8$) in a matrix with $S = 0.6$ completely encloses the plate (Figure 6), for two parallel, touching plates of opposite shear, each with $c/r = 0.08$, only a small yield envelope confined to the edges of the plates is produced (Figure 7). Ling and Owen [19,20] further demonstrated that as more plates are added to the stack, the yield envelope continues to shrink.

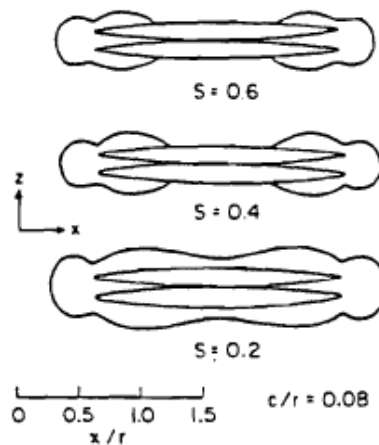


Figure 7. Matrix yield stress contours of two parallel ellipsoidal plates with opposite shear strain. Aspect ratio: 0.08 [20].

In general, as an isolated plate of thermo-elastic martensite thickens, the large shear stresses near the edge (tip) and over the broad surface of the plate assist the nucleation of adjacent plates. In the plastic zone near the edge, nucleation may be strain induced, and nonparallel crystallographic

variants of the plate may form edge to edge. Alternatively, stress-assisted nucleation adjacent to the broad interface may produce a variant approximately parallel to the original plate. The change in the yield envelope on putting together two parallel plates of opposite shear touching along the z-axis is displayed in Figure 7. It is seen that while the yield envelope of an isolated oblate spheroid ($c/r = 0.8$) in a matrix with $S = 0.6$ completely encloses the plate (Figure 6), for two parallel, touching plates of opposite shear, each with $c/r = 0.08$, only a small yield envelope confined to the edges of the plates is produced (Figure 7). Additional calculations show that as more plates are added to the stack, the yield envelope continues to shrink.

Dynamics of Thermo-elastic Martensitic Interfacial Motion: One of the last contributions of Professor Owen to the field of martensitic transformation pertains to the experimental characterization, modeling and computer simulations of the basic dynamics of martensitic interface motion. It is well-established that the analysis of martensite growth must include a consideration of the following two fundamental phenomena: (a) dynamics of interfacial motion and: (b) accommodation of the transformation strain under consideration of variant selection. The structural changes occurring during the transformation require special interfaces between the parent phase and martensite, whose dynamics holds the key to many kinetic aspects of transformation. On the other hand, accommodation of the transformation strain may control morphological development of a martensite plate. While in the case of non-thermo-elastic martensite, there is a dynamic interaction between the interfacial dynamics and plastic accommodation of the transformation strain which may govern the path of martensite growth, Owen and co-workers [24-27] recognized that, in the case of thermo-elastic martensite, the role of plastic accommodation of the transformation strain is minimal and that this transformation, hence, offers a unique opportunity for studying the basic dynamics of martensitic interfacial motion.

Owen and co-workers [24-27] further recognized that the dislocation models for martensitic interfacial structure proposed in the literature suggest that the dynamics of interfacial motion can be analyzed using the formalisms developed to describe dislocation dynamics. A general dislocation description of the martensitic interface has been developed by Olson and Cohen [28,29] and Olson [30] who introduced two types of interfacial dislocations: a) the “*coherency dislocations*” which, as they move with the interface, accomplish the lattice deformation while maintaining the continuity of crystal lines and directions across the interface and; (b) the “*anti-coherency dislocations*” which, being true lattice discontinuities, disrupt the lattice continuity and, as they move, produce the lattice invariant deformation in martensite. A simultaneous operation of the two deformation modes, the lattice deformation and the lattice invariant deformation, leave the interface plane macroscopically invariant, i.e., undistorted and un-rotated.

Within their work, Owen and co-workers [24-27] analyzed the basic dynamic behavior of martensitic interfaces using the framework of lattice dislocation dynamics. Two limiting cases of the martensitic interface structure were considered: (a) the case when the interface can be appropriately described in terms of an array of non-interacting (well-spaced) interfacial dislocations; and (b) the case when the interfacial dislocations are so closely spaced that the interface can be approximated by a continuous distribution of dislocations. In the first case, they demonstrated that the interfacial dynamics can be analyzed by including a “*chemical/transformation*” driving force in the equation of motion for a single lattice dislocation. In the second case, on the other hand, while the lattice dislocation dynamics is still relevant, several parameters in the equation of motion had to be redefined to reflect the fact that the interface now acts as a planar defect. For both the aforementioned dislocation models of

martensitic-interfacial structure, the two basic modes/regimes of interfacial motion were considered: (a) the continuous mode in which the motion is sustained and controlled by various energy-dissipative processes (e.g., phonon and electron drag) and; (b) the discontinuous or jerky mode in which the motion is controlled by thermal activation of the interface/obstacle interactions.

The continuum regime of martensitic interfacial motion was analyzed using the framework initially developed in Professor Owen’s research group at Cornell University [9,10]. On the other hand, the thermally activated regime of martensitic interfacial motion carried out by Owen and co-workers [24-27] at MIT was a new direction in the research and is briefly overviewed below.

The basic concept of thermally activated dislocation motion can be understood by analyzing the schematic dislocation/obstacle interaction force-distance profile depicted in Figure 8.

Representing the position of a dislocation segment by the area swept during its motion, a , the variation of the internal friction stress with the position defines a τ vs. a profile of the form given in Figure 8(a). For an applied stress τ^A , the dislocation is in stable equilibrium at point a_1 . To reach the point a_2 of unstable equilibrium, the dislocation must overcome its activation energy Q_1 defined by the shaded area under the curve in Figure 8(a). During such a thermally activated event, the dislocation line sweeps an “activation area” a^* , whose magnitude is a function of τ^A , as shown in Figure 8(b). The activation energy can be obtained by the following integration:

$$Q(\tau^A) = b \int_{a_1}^{a_2} (\tau - \tau^A) da = \int_{\tau^A}^{\hat{\tau}} a^* d\tau \quad (2)$$

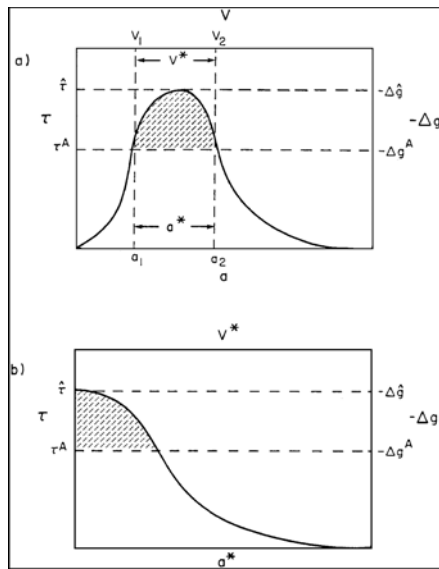


Figure 8. Schematic dislocation/obstacle interaction profiles. (a) Force-distance profile. (b) “Sheared profile” representing stress vs. activation area or driving force vs. activation-volume relation [25].

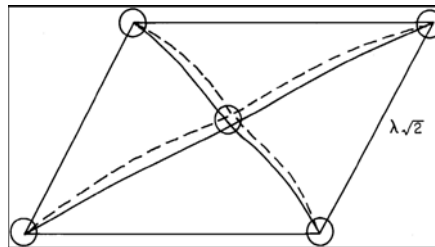


Figure 9. Schematic representation of the interaction between an interface and regularly distributed discrete particles with average planar spacing [25].

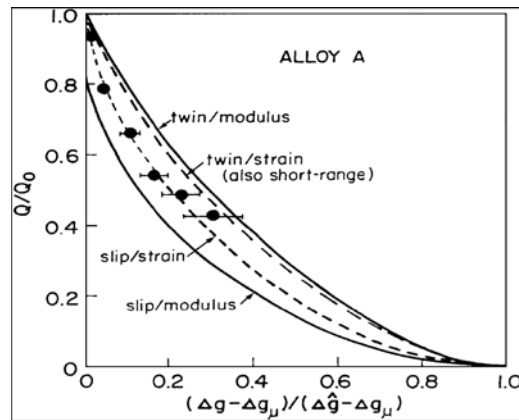


Figure 10. Normalized activation energy vs. driving force plot showing a comparison between different model predictions and the experimental mobility data in a Cu-14.3Al-4.0Ni single crystal [26].

The overall velocity of the dislocation in the thermally activated region, v , is governed by the rate at which the dislocation overcomes the barrier and is defined as:

$$v = v_0 \exp\left(-\frac{Q}{kT}\right) \quad (3)$$

where the pre-exponential factor v_0 is proportional to the corresponding attempt frequency. The two activation parameters, the activation area and the activation energy and their dependence on the applied stress can be determined from experimental dislocation mobility data using the following relations:

$$Q = \left. \frac{\partial \ln v}{\partial (-1/kT)} \right|_{\tau^A} \quad (4)$$

$$a^* = \left. \frac{kT}{b} \frac{\partial \ln v}{\partial \tau^A} \right|_{\tau} = - \left. \frac{1}{b} \frac{\partial Q}{\partial \tau^A} \right|_{\tau} \quad (5)$$

When an interface consists of an array of well-spaced dislocations, the τ vs. a^* profile for the interfacial dislocations can be used to describe the interface/obstacle interaction. Conversely, when the interface is more appropriately described in terms of a continuous distribution of dislocations, the thermally activated event is three-dimensional in nature. For this case Owen and co-workers [24-27] introduced an “activation volume” V^* as:

$$V^* = \left. \frac{\partial Q}{\partial \Delta g} \right|_{\tau} \quad (6)$$

and demonstrated that the thermal activation analysis of the interfacial motion can be formulated in terms of the corresponding (driving force) Δg vs. V^* profile. For an interface consisting of well separated dislocations, V^* can be defined in terms of a^* through the relation:

$$V^* = da^* \quad (7)$$

When an obstacle profile such as the one shown in of Figure 8 pertains to long-range interface/obstacle interactions, V^* becomes very large and consequently, the role of thermal activation becomes less important. If the interface can be described in terms of an array of well-spaced dislocations, the interface/obstacle interactions result in a velocity-invariant friction stress τ_μ which oppose the interface motion by lowering the net applied stress. In other words, the “thermal” component of τ^A that can assist thermal activation of the dislocation over short-range obstacles now becomes $\tau^A - \tau_\mu$. When several short-range obstacles simultaneously interact with an interface, the overall obstacle profile can be defined by the root sum of the squares of the individual profiles. Again, when the interface is more appropriated represented by a continuous distribution of dislocations, the “athermal” and the “thermal” components of the interfacial force should be defined in terms of Δg rather than τ .

As a dislocation moves through a perfect crystal, its core energy varies in a periodic manner due to the translational symmetry of the lattice. The resulting periodic lattice friction force, the Peierls barrier, can become the rate controlling mechanism of dislocation motion at lower temperatures. When an interface can be described in terms of well separated dislocations, the interfacial dislocations may be considered to interact individually with the periodic lattice potential in much the same manner as the lattice dislocations. For the case when interfacial

dislocations are closely spaced and the interface is more properly treated as a 2-dimensional entity, Owen and co-workers [24-27] developed a Peierls-type interfacial friction model.

While the effect of the Peierls barrier may be important in the growth of microscopic martensite particles under weak driving forces, Owen and co-workers [24-27] recognized that the motion of an interface of macroscopic dimensions, just like the motion of lattice dislocations, is more likely controlled by its interaction with crystal defects. For example, it is well established that cutting of repulsive forest dislocations by other dislocations, which involves generation of elementary jogs, gives rise to a significant strain hardening of metals at moderate temperatures. Likewise, the interaction of dislocations with attractive forest dislocations gives rise to long-range “athermal” friction stresses which oppose dislocation motion. Similar line-line interactions are expected to control the interfacial dynamics when the interface structure is dominated by well-spaced anti-coherency dislocations. The interactions of closely spaced coherency dislocations with forest dislocations, can also be important but this interaction, was treated as a line-surface problem.

Since the interface can interact with remote forest dislocations, the resulting interfacial friction stress was recognized to be predominantly of the long range character and, thus, athermal. Similar conclusions were drawn for surface-surface interactions involving the martensitic interface and other surface defects, such as anti-phase boundaries. It was also noted that besides giving rise to additional (athermal) friction stresses, the line-surface and surface-surface interactions may be of a potential importance in assisting the generation of coherency dislocations by a pole mechanism during thickening of small martensitic particles.

Discrete defects which take part in point-surface interactions were identified as the defects that most likely control thermally activated interfacial motion. These interactions can be both a long-range (athermal) character as well as of the short-range (thermal) character. The interaction of an interface with regularly distributed localized discrete particles, with average planar spacing λ , (Figure 9), has been analyzed by Owen and co-workers [24-27]. For a simplified dislocation model of interfacial structure, and using the analytical expressions for the corresponding interfacial stress field, Owen and co-workers [24-27] derived the driving force vs. activation volume obstacle profiles for long- and short-range strain and modulus interface/ obstacle interactions. Furthermore by applying an integration procedure analogous to the one indicated by Eq. (2), the corresponding Q vs. Δg relations were derived. The results of this procedure are shown in Figure 10. Labels “*slip*” and “*twin*” were used in Figure 10 to indicate the nature of the lattice-invariant deformation. It is customary to represent the Q vs. Δg relations as the ones shown in Figure 10, by the following expression:

$$Q = Q_0 \left[1 - \left(\frac{\Delta g - \Delta g_\mu}{\Delta \hat{g} - \Delta g_\mu} \right)^p \right]^q \quad (8)$$

where Q_0 is the maximum activation energy, Δg_μ is the athermal part of Δg , and the exponents p and q describe, respectively, the long-range (small Δg) and the short-range (large Δg) interfacial/obstacle interaction behavior. The characteristic values of p and q for the various interface/particle interactions are summarized in Table 1. The results shown in Figure 10 and Table 1 suggest that the interaction of a “*slip*” interface with the same array of discrete obstacles, is significantly different than that of a “*twin*” interface. Owen and co-workers [24-27] also found that the magnitudes of Q_0 and $\Delta \hat{g}$ are both substantially larger for a “*slip*” interface, than a

“twin” interface. The last finding suggests that a “twin” interface is intrinsically more mobile than the “slip” interface.

Lattice invariant deformation	Obstacle	Interaction	p	q
Twinning	Discrete particle	Strain	1	2
		Modulus	1	9/5
	Peierls barrier	—	1	2
		Strain	1/2	3/2
Slip	Discrete particle	Modulus	4/5	2
		Short range	1	2
	Peierls barrier	—	1	2
Twinning	Experimental measurement of single interface mobility in Cu-14.3Al-4.0Ni [16, 17]		0.39	1.35

Table 1. Values of p and q exponents in Eq. (3) for various types of interface/particle interactions.

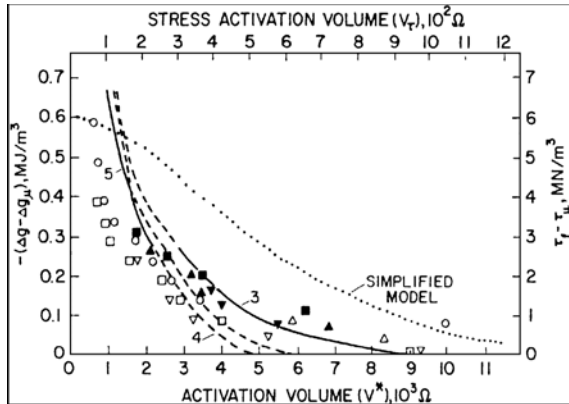


Figure 12. Comparison of the model predictions for the strain-interaction obstacle profile (solid lines) and the experimentally measured obstacle profile in a Cu-14.3Al-4.0Ni single crystal [27].

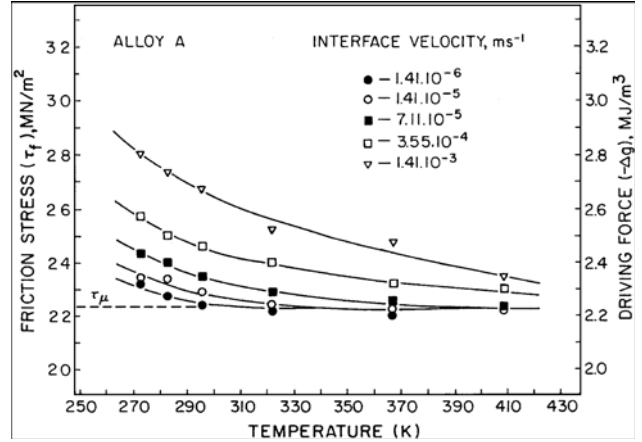


Figure 11. Temperature and velocity dependence of the friction stress (driving force) in a Cu-14.3Al-4.0Ni single crystal [26].

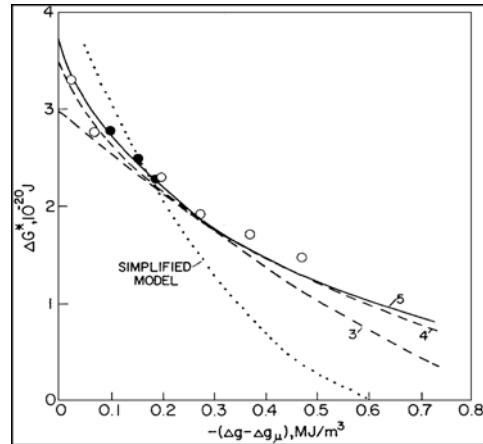


Figure 13. Comparison of the model prediction (solid curves) for the strain-interaction activation energy vs. driving force relation and the experimentally measured relations between activation energy and driving force in a Cu-14.3Al-4.0Ni single crystal [27].

To validate their model for thermally-activated martensitic interfacial motion, Owen and co-workers [24-27] carried out a comprehensive experimental investigation of stress-assisted single-interface martensitic transformation in single crystals of thermo-elastic Cu-Al-Ni alloys. Figure 11 shows the measured temperature dependence of the interfacial friction stress and the corresponding driving force for the interfacial velocities in the range between 10^{-6} and 10^{-3} m/s in a Cu-14.3Al-4.0Ni single crystal. As indicated by the dashed line in Figure 11, there is a well-defined athermal interfacial friction stress, τ_{μ} . The corresponding Δg vs. V^* and Q vs. Δg

obstacle profile associated with the thermal part of the applied stress are shown in Figures 12 and 13, respectively. The Q vs. Δg relation was found to be consistent with Eq. (8) with $Q_0 = 0.305$ eV, $-\Delta g^{\wedge} = 0.9$ MJ/m³, $\Delta g_{\mu} = 2.21$ MJ/m³, $p = 0.39$, and $q = 1.35$. These values for p and q were found to be in good agreement with the interfacial motion model of Owen and co-workers [24-27] when an interfacial dislocation model specific to the twinned β_1 - γ_1 martensitic interface was used. By analyzing both the interfacial mobility data and the transmission electron microscopic results of the parent β_1 -phase, Owen and co-workers [24-27] determined that the thermal-activated motion of the β_1 - γ_1 interface is controlled by the “*tweed*” structure, consisting of high-modulation displacement fields. The athermal frictional stress, on the other hand, was attributed to be a long-range modulus interaction of the interface with discrete 2H precipitated in the β_1 -phase.

3. A PERSONAL TRIBUTE

The following comments were made by one of us (FJS) at a memorial event at MIT in November 19, 2007, attended by his family, friends and colleagues over many years. These comments especially exemplify the impact that Professor Owen had on the developing careers of young people.

“I was one of the first doctoral students doing graduate study under Professor Walter Owen’s tutelage at Cornell University, beginning in September 1966, shortly after he arrived from Liverpool. It was clear that Professor Owen derived immense satisfaction from his teaching, mentoring and association with young people, as well as his research in physical metallurgy. It was also apparent that he had a high commitment to excellence in a model of learning that shared responsibility between students and teachers, and he was eminently respectful and caring of the professional and personal needs of both students and his faculty colleagues. I remember best his friendly, encouraging, and optimistic manner that gently pushed me and my colleagues to do our best work, as much of it as possible, and as rigorously as possible. He was extremely generous with his time and energy, and a key contribution he made to my career and I’m sure the careers of my fellow students, accrued from the relaxed but focused technical sessions he held for his entire group of graduate students and postdoctoral fellows at his home in Ithaca, one evening per month. At these work-in-progress sessions we learned how to thoughtfully prepare a technical talk, communicate ideas to a scholarly group of peers, and vigorously defend our thinking on our feet, skills I continue to benefit from today. As we had learned the art of preparation for a scientific presentation through these sessions, I was appalled one day when I saw him depart for a major out-of-town talk and grab his box of slides (in those days a several pound mass of glass), not yet organized for the talk, and proceeded to the airport. Little did I realize that that was often the “*modus operandi*” of a seasoned professional (and a style that I and many colleagues regrettably adapted later as our careers matured).

Walter was truly a good mentor, and he taught me how to think critically about science in general. I shall always be grateful that he supported and followed (and I believe even relished) the evolution of my career from martensite metallurgy (the area he loved) as a PhD student, into biomaterials, biomedical engineering and medicine, areas he knew much less about. Only in preparation for these remarks did I uncover an article in the Cornell Engineering Quarterly in 1970 in which he predicted that the field of biomaterials would be a “growth area” to which materials science would contribute immensely (and how correct he was).

Walter was the Thomas R. Briggs Professor and Director of the Department of Materials Science and Engineering at Cornell from 1966 to 1970 when he undertook his new appointment as Dean of the Technological Institute at Northwestern University. When he came to Cornell in 1966 he took over a young department, which became a separate Materials Science and Engineering Department only the previous year. Within a year there were approximately forty students doing graduate materials study at Cornell. Through his leadership, Professor Owen established MS&E at Cornell as one of the few leading departments of its kind, composed of a distinguished group of materials specialists and strong academic programs. He developed excellent and synergistic collaborations with groups doing physics, chemistry and mechanics, thereby creating a highly interdisciplinary and intellectually stimulating environment. Through his leadership of the Department at Cornell, Walter contributed greatly to the development of materials science as a systematized science and academic discipline. Owing to his contributions then and subsequently, materials science has flourished, increasing in sophistication and contributing to many important scientific and engineering advances.

It was a privilege and an honor to be his colleague and friend for over 40 years.”

4. FINAL REMARKS

With his devotion to family, friends, research, teaching and professional service, Professor Owen has left behind a vast number of grieving people whose lives were touched and improved by his presence. He was a role model for many of us over the years, and he will live on in the memories of those who knew him. We will always remember Professor Owen.

5. REFERENCES

1. H. Carpenter and J. M. Robertson, *Metals*, Oxford University Press, 1939, 913, 1083.
2. P. G. Winchell and M. Cohen, *Trans. ASM*, 55, 1962, 347.
3. P. G. Winchell and M. Cohen, in *Electron Microscopy and the Strength of Metals*, G. Thomas and J. Washburn, Ed., Interscience, 1963, 995.
4. M. Cohen, *Trans. AIME*, 224, 1962, 638.
5. M. Cohen, *J. Iron Steel Inst.*, 201, 1963, 833.
6. M. J. Roberts and W. S. Owen, "Physical Properties of Martensite and Bainite," Iron and Steel Institute Special Report No. 93, 1965, 171.
7. M. J. Roberts and W. S. Owen, *J. Iron Steel Inst.*, 206, 1968, 375.
8. F. E. Fujita, T. Moriya, and H. Ino, in *Proc. Int. Conf. Sci. Technol. Iron Steel*, Iron and Steel Institute of Japan, 1971, 1273.
9. W. S. Owen, F. J. Schoen, and G. R. Srinivasan, in *Phase Transformations*, American Society for Metals, 1970, 157.
10. F. J. Schoen and W. S. Owen, *Metall. Trans.*, 2, 1971, 2431.
11. J. Snoek, *Physics*, 8, 1941, 711.
12. G. Schoeck and A. Seeger, *Acta Metall.*, 7, 1959, 469.
13. A. H. Cottrell and B. A. Bilby, *Proc. Phys. Soc.*, 62, 1949, 49

14. A. H. Cottrell and M. A. Jaswon, *Proc. Roy. Soc. Lond.*, A199, 1949, 104
15. J. W. Cahn, *Acta Metall.*, 10, 1962, 789.
16. S. Kajiwara and W. S. Owen, *Scr. Metall.*, 11, 1977, 137.
17. S. Kajiwara and W.S. Owen, *Metall. Trans.*, 4, 1973, 1988.
18. G. D. Sandrock, A. J. Perkins, and R. E. Hehemann, *Metall. Trans.*, 2, 1971, 2769.
19. H. C. Ling and W. S. Owen, *Scr. Metall.*, 15, 1981, 1115.
20. H. C. Ling and W. S. Owen, *Acta Metall.*, 29, 1981, 1721.
21. J. W. Christian, *The Theory of Transformations in Metals and Alloys*, Pergamon Press, Oxford, 1965, 816.
22. J. D. Eshelby, *Prog. Sol. Mech.*, 7, 1961.
23. M. J. Schmutz, Doctoral thesis, Massachusetts Institute of Technology, 1978.
24. M. Grujicic, G. B. Olson, and W. S. Owen, *J. Phys.* (Suppl. 12) 43, 1982, C4-173.
25. M. Grujicic, G. B. Olson, and W. S. Owen, *Metall. Trans.* 16A, 1985, 1713.
26. M. Grujicic, G. B. Olson, and W. S. Owen, *Metall. Trans.* 16A, 1985, 1723.
27. M. Grujicic, G. B. Olson, and W. S. Owen, *Metall. Trans.* 16A, 1985, 1735.
28. G. B. Olson and M. Cohen, *Acta Metall.* 27, 1979, 1907.
29. G. B. Olson and M. Cohen, in *Proc. Int. Conf. Solid-Solid Phase Transformations*, edited by H. I. Aaronson, TMS-AIME, 1982, 1209.
30. G. B. Olson, *Acta Metall.* 29, 1981, 1475.