RAPID SOLIDIFICATION PROCESSING OF SUPERALLOYS USING HIGH POWER LASERS

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A new form of laser materials processing has been developed (the LAYERGLAZE™ process) whereby bulk parts can be built up from rapidly solidified (~10^4°C/s) materials. This involves the introduction of wire or powder feedstock onto a moving substrate at the point of impingement of a continuous CO_2 laser beam. LAYERGLAZE processing is currently being employed to fabricate model gas turbine disks with very fine microstructures and low defect populations. This application has required the development of high molybdenum, nickel base superalloys which are both laser-weldable, phase stable and strong at ≤760°C. Further development of the LAYERGLAZE technique has produced the concept of general "directed energy processing", which combines rapidly solidified materials fabrication with alloy composition gradients; as well as in-process nondestructive testing, defect repair and thermal treatment.

INTRODUCTION

In 1976, at the 'Third International Symposium on Superalloys', Breinan, et al, described a new method for the surface treatment of materials by controlled laser surface melting, followed by rapid self-quenching (1). Cooling rates up to 10^6°C/sec were readily attainable in appropriately thin sections. The range of microstructures achieved by this so-called

LASERGLAZE™ process included amorphous metallic solids, extended solid solution phases, ultra-fine eutectics, and refined dendritic structures.

Much work has now been done to determine the utility of LASERGLAZE processing as a means to improve the corrosion, erosion and wear properties of metal surfaces. Some good examples of improvements in all three categories have been obtained by different research groups. On the other hand, much less work has been reported on LASERGLAZE processing in conjunction with surface alloying. Two distinct approaches have been evaluated at UTRC: (1) preplacement of alloying material on the workpiece surface prior to laser melting; and (2) continuous delivery of alloying material to the laser-material interaction zone (the LAYERGLAZE™ process).

The preplacement method has been exploited as a hardfacing treatment; e.g., to harden critical regions of wear in a part. However, the continuous-feed method holds greater promise, both for surface alloying and as a means to produce bulk rapidly solidified structures in near-net shapes.

A major R&D program is now being undertaken by UTRC to exploit the continuous-feed method for the purpose of fabricating a high-performance gas turbine disk. This work has progressed to the point that sound parts, free of gross imperfections, can now be produced, using either prealloyed wire or powder feed. Some success has also been achieved in controlling the shape of the disk during fabrication. In addition, several precipitation-strengthened nickel-base superalloys have been developed which exhibit as-fabricated properties comparable to current disk alloys.

This paper constitutes a progress report on the continuous powder-feed method as applied to disk fabrication. Future directions for R&D in the area will also be discussed, including such considerations as combining incremental solidification processing with incremental thermo-mechanical treatment, as well as continuous inspection procedures.

INCREMENTAL LASER MELT PROCESSING

Although the LASERGLAZE™ process has been found to be capable of producing rapidly chilled microstructures, its applications are limited by the small section thickness required to achieve high cooling rates. Typically, an order of magnitude decrease in the melted and resolidified section thickness results in an increase in the average cooling rate of two orders of magnitude. The LASERGLAZE process is thus suited only for treatment of thin layers. In order to achieve the fabrication of rapidly-solidified thick sections (bulk parts), the LAYERGLAZE™ concept was evolved. LAYERGLAZE is essentially the sequential buildup of bulk material with controlled composition and microstructure by simultaneous material addition and LASERGLAZE processing. Although the deposition of each subsequent layer produces a high-temperature thermal transient in the layer beneath, the time duration of this transient is extremely brief. As a result, although the process will not permit the formation of bulk amorphous structures, it does allow fabrication of relatively homogeneous crystalline structures with little or no phase transformation between layers.

A schematic representation of the LAYERGLAZE process is shown in Fig. 1. Feedstock can be in the form of either wire or powder. Wire is convenient to feed but often difficult to obtain in the required alloy compositions. Powder is relatively easy to procure, and can be conveniently fed into the interaction zone. The laser beam melts both the feedstock material and some of the substrate layer, which produces good interlayer bonding, and eptiaxial solidification from layer to layer. There is a critical relationship between the location of the feedstock impingement point and the laser-heated zone. Since the substrate is rotating, feedstock impingement must occur slightly ahead of the laser beam for stable, steady-state material deposition. The feedstock is then carried into the laser beam impingement area by the rotation of the substrate. The part being fabricated is internally water-cooled during material deposition in order to prevent heat buildup, and thus maintain high cooling rates.

The LAYERGLAZE technique for incremental laser melt processing has been used to fabricate a 13.2 cm diam. x 3.2 cm thick cylindrical blank, to be subsequently machined into an ~12.7 cm diam. scale model gas turbine disk. Another model disk will be constructed in late 1980 for spin testing early in 1981. The initial trial disk fabrication produced the disk preform pictured in Fig. 2. This preform was powder-feed LAYERGLAZE fabricated from two nickel base superalloys. The core consisted of UTRC 8-12-3 (Ni-3.4Al-17.9Mo-8.4Ta, wt %), which was one of the initial alloys considered for model disk manufacture in this program. The rationale for this choice of composition and additional details of the alloy development effort are included in the following section. The core was fabricated from alloy 8-12-3 powder to a diameter of 8.4 cm, then the disk preform was completed using IN-718 powder. This test served to demonstrate both the capability of making full-sized parts and the ability to utilize different alloys sequentially in the same part by means of incremental fabrication.

The microstructure of this LAYERGLAZE-processed model turbine disk is shown in Fig. 3. The buildup process has created a uniformly layered appearance, with generally good structural integrity and interlayer bonding. Both the transverse cross-section and the disk surface display the radially-aligned, columnar grain structure and epitaxial growth which is characteristic of the LAYERGLAZE process.

In all LAYERGLAZE buildup tests, compressive deformation of the substrate was observed. This was expected, since the process involves the sequential deposition and shrinkage of successive layers. As each layer is deposited and cooled, it is stressed in tension and yields plastically. As each subsequent layer is deposited on it, the stress on the first layer gradually changes from residual tension to residual compression. Detailed calculations which describe these changes may be found in Ref. 2.

Although to date the utilization of LAYERGLAZE processing at UTRC has been concerned with the fabrication of simple axisymmetric shapes, it should be pointed out that this process is not limited to them, and in fact is capable of producing

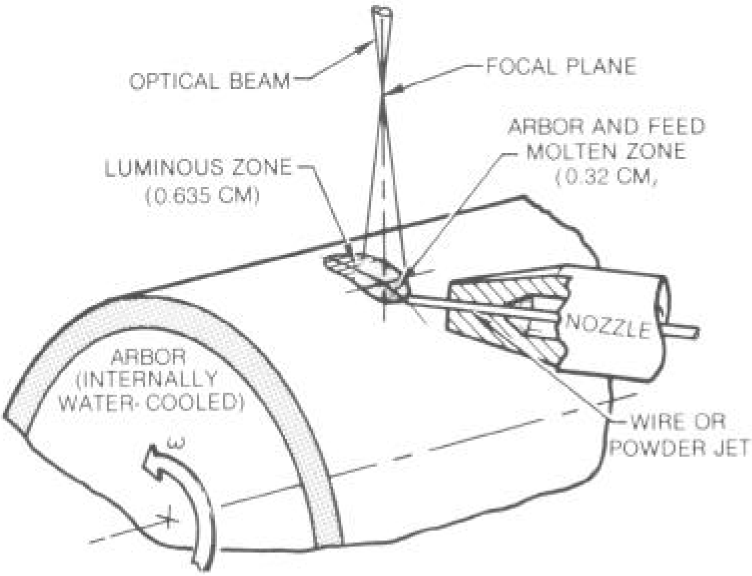

Fig. 1. Schematic Representation of the LAYERGLAZE Process.

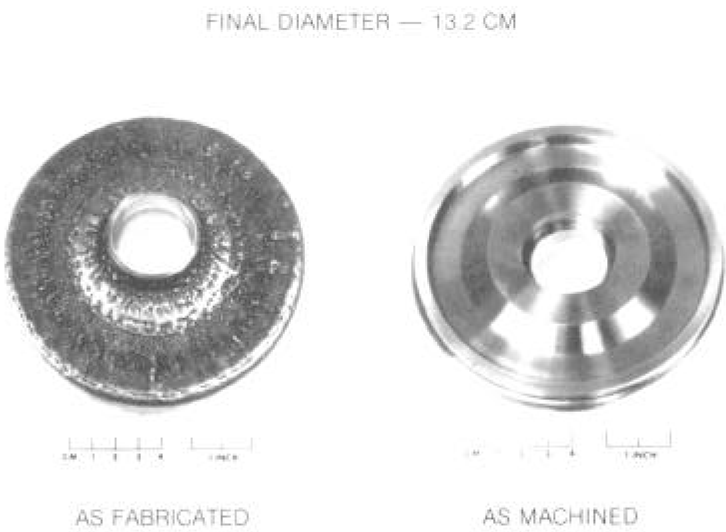

Fig. 2. LAYERGLAZE-Fabricated Turbine Disk Preform.

almost any desired shape. Some of the possibilities are illustrated in Fig. 4. As indicated, the production of more complex shapes requires the use of a numerically controlled work station, which is capable of simultaneous motion about at least three axis. However, even the simple case of translation back and forth on a straight edge has obvious implications for hard-facing of cutting tools.

ALLOY DEVELOPMENT

The development of nickel base superalloys for use as turbine disks fabricated by the LAYERGLAZE process was constrained by three sets of requirements; crack-free laser weldability, beneficial alloy response to rapid solidification ($\sim 10^4$°C/s), and mechanical properties exceeding those of current jet engine disks. The primary consideration in evaluating any alloy was that it should remain free of cracks after laser welding (2). As-chill-cast specimens of ~8 mm thickness were bead-on-plate welded at 5 cm/s with a continuous, CO_2 laser operated at 6 kW in the unstable resonator mode. The beam intensity on each specimen was approximately 1 MW/cm^2. Overlapping pass welds were made under 100% He, with the specimens mounted on a horizontal, rotating wheel. If no post-weld cracks were observed on the specimen surface or on a transverse cross-section through the entire specimen, the alloy was selected for further evaluation.

No conventional nickel base superalloys of adequate strength, including those known to be weldable by other techniques, passed the test described above. Those which cracked when laser processed included advanced disk superalloys of recent development or use, such as AF2-1DA, AF 115, IN 100, etc. Consequently, the main effort of alloy development was directed toward devising and evaluating strong nickel base superalloys which would be compatible with laser materials processing.

Research within United Technologies by Lemkey (3), Pearson (4), and Aigeltinger, et al (5) had suggested that a particular range of alloy compositions based on the Ni-Al-Mo system might be expected to possess the unique properties

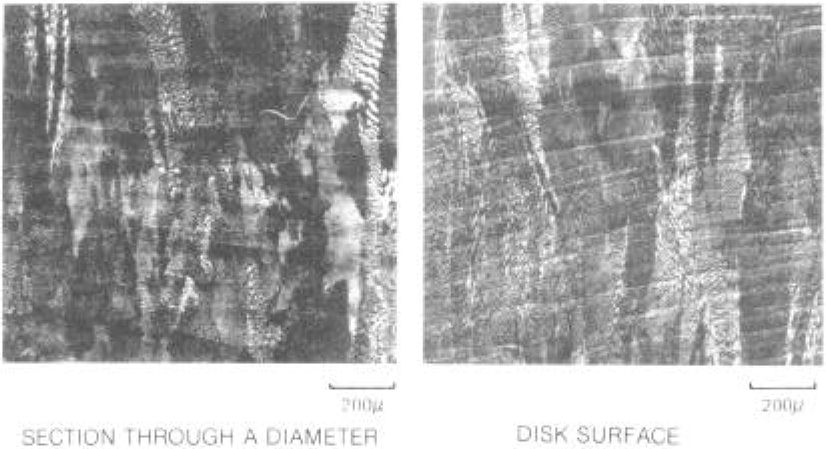

Fig. 3. Microstructure of LAYERGLAZE-Fabricated Turbine Disk.

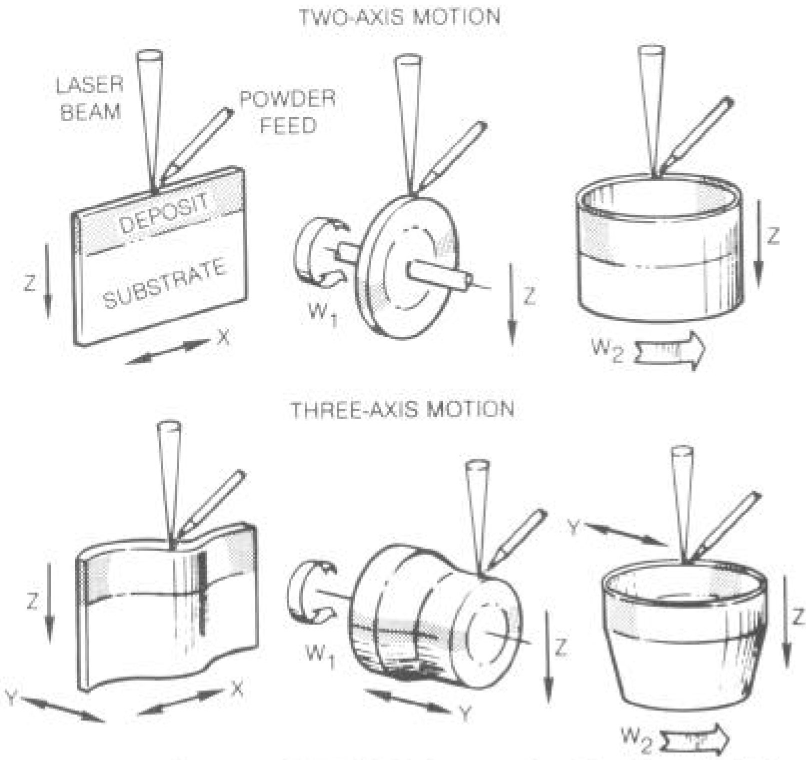

Fig. 4. Examples of LAYERGLAZE Processing for Near-Net-Shape.

required to produce a LAYERGLAZE processed turbine disk. The initial rationale for choosing this group of alloys was based on their observed resistance to laser weld cracking, lower γ' volume fraction for ductility during LAYERGLAZE processing, and potential for age hardening by Ni_3Mo and/or Ni_4Mo intermetallic compounds. Some initial consideration was given compositions intended to avoid the formation of intermetallic δ (NiMo) and to promote the strengthening of γ' by Ta.

A description of the first-generation alloy compositions and properties can be found in Ref. 2. From this group, the first alloy chosen for evaluation by LAYERGLAZE processing was UTRC 8-12-3, Ni-3.4Al-17.9Mo-8.4Ta (wt %). It was argon atomized, sieved to -150 mesh to improve flow characteristics, and utilized to fabricate several model turbine disks by the powder-feed LAYERGLAZE process; one of which is shown in Fig. 2. Extensive observation of the as-LAYERGLAZE processed microstructure of this alloy (Fig. 3) revealed very few inclusions or voids, and none larger than ~4μm. Solidification occurred by dendritic growth in all areas, with little or no observable side branching. The mean secondary dendrite arm spacing was 1.9μm, while the mean primary spacing was 3.8μm. Comparison of these data with similar observations (6) suggests that this microstructure solidified at ~10^4°C/s. Evidence that solidification was primarily epitaxial is found in the continuation of most grains through the striations which mark the successive location of the solid-liquid interface of each added layer.

Examination of as-LAYERGLAZE processed alloy 8-12-3 by transmission electron microscopy (Fig. 5) revealed that the interdendritic areas were delineated by subboundaries and by phases which have been tentatively identified as MoC. There was a high density of uniformly distributed dislocations, predominately on {111}, within the grains and subgrains. The dendrites themselves consisted of a γ matrix which contained a very fine, homogeneous dispersion of both DO_{22}-structure Ni_3Mo and γ' (Fig. 6). No coarse γ' was observed in the interdendritic regions.

As indicated by Table I, the strengths of several first generation, as-LAYERGLAZE- processed alloys were sufficiently high to merit further development. However, examination of

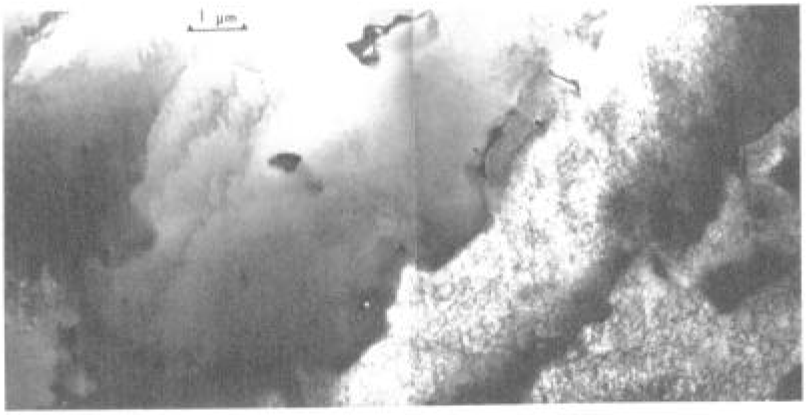

Fig. 5. As-LAYERGLAZE Processed Microstructure, Alloy 8-12-3, (Ni-3.4Al-17.9Mo-8.4Ta), which exhibits high dislocation density, interdendritic subboundaries, and some interdendritic precipitation.

Fig. 6. As-LAYERGLAZE Processed Alloy 8-12-3. Bend contour within dendrite, bright field. γ' and DO_{22} Ni_3Mo in γ matrix.

Table I. Mechanical Properties of LAYERGLAZE Process Model Turbine Disk Alloys

Alloy	Composition (wt %)	Tensile Test Data, As-Fabricated, Radial Orientation Temp (°C)	0.2% YS (MPa)	UTS (MPa)	% E
8-12-3	Ni-3.4Al-17.9Mo-8.4Ta	25	1050	1280	40
		538	965	1100	43
		704	980	1125	32
12-15	Ni-5.4Al-23.8Mo	25	1207	1393	26
		704	1180	1241	13
11-12-0-0-5	Ni-5.0Al-19.4Mo-4.4Cr	25	1100	1407	4

these alloys after annealing at 760°C for 100 hrs revealed extensive discontinuous (cellular) precipitation at both grain boundaries and interdendritic subboundaries. This caused virtually complete embrittlement by the formation of orthorhombic Ni_3Mo in a lamellar structure similar to that observed in binary Ni-Mo alloys (7) (Fig. 7). On the other hand, the dispersed γ' and DO_{22} Ni_3Mo exhibited very little coarsening, while the average microhardness increased from 420 (as-fabricatdd) to 580 VHN (200g load). Other Ni-Al-Mo and Ni-Al-Mo-Ta alloys, which required ≥18 wt % Mo to avoid cracking when laser welded, all exhibited similar cellular phase transformations when annealed at >650°C. Consequently, the program was directed toward a second generation of alloys which involved the partial substitution of Cr for Mo to avoid intermediate temperature phase instability while retaining weldability and strength.

The absence of crack formation in the experimental, high-Mo nickel base superalloys during LAYERGLAZE processing was unusually sensitive to (deliberate) compositional variation, e.g. a change of ± 1 at % Mo could be critical. The general dependence of crack formation on composition was difficult to predict, and empirical criteria for doing so were inadequate. The mechanism(s) of crack formation in Ni-Al-Mo-X alloys is not obvious, but appeared to be related to high-temperature, post-solidification ductility, and possibly to slip dispersal at grain boundaries by second phases. All alloys which contain >6 wt % Al have cracked when laser welded by means of the procedure described earlier. These problems are clearly related to, but have been confirmed not to be identical with, weld cracking in conventional nickel base superalloys.

The second generation, Ni-Al-Mo-Cr alloys have thus far shown considerable promise with respect to the program goals. Again, the absence of laser weld cracking has been difficult to predict, as incrementally increasing substitution of Cr for Mo does not result in a systematic variation of crack-free laser weldability. Nevertheless, several interesting alloys have appeared, especially alloy 11-12-0-0-10, Ni-5Al-19.5Mo-8.8Cr (wt %). This alloy can be LAYERGLAZE processed without cracking and displays no cellular transformation after annealing at 760°C for 300 hrs. Laser welded specimens annealed under these conditions were age hardened to ~R_C 54 and displayed predominately intergranular fracture surfaces when impact tested. At this time, extensive structure/property characterization of LAYERGLAZE-processed alloy 11-12-0-0-10 is planned, as well as fabrication and spin testing of a model turbine disk LAYERGLAZE-fabricated from this alloy.

DIRECTED ENERGY PROCESSING

Experience with the basic LAYERGLAZE technique has pointed the way to a potential means for controlling structure and properties more closely than has previously been possible by using sequential, in-situ processing techniques on small increments of material. The technique has been termed Directed Energy Processing and is illustrated schematically in Fig. 8. The processing starts at work station #1 with material being added by the laser melting and deposition of feedstock. This step essentially creates the part. At additional work stations, other materials processing operations can be sequentially accomplished on each incremental layer. These include inspection, mechanical deformation, and/or heat treatment. Like deposition, which derives advantages from rapid solidification of a thin section, each of these subsequent operations also derives advantage from being

Fig. 7. Alloy 8-12-3, LAYERGLAZE Processed and Annealed at 704°C, 32 hrs. Cellular precipitation nucleated at interdendritic subboundaries and involving orthorhombic Ni_3Mo.

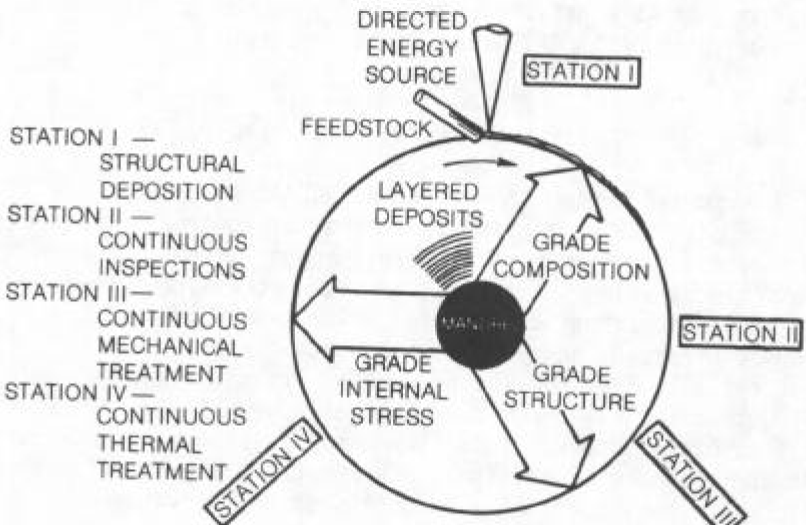

Fig. 8. Schematic Representation of Directed Energy Processing.

applied to a small layer. Incremental inspection can thus be accomplished without destructively testing the disk. Mechanical processing can be applied to small increments without resulting in significant distortion, and thermal treatments or even remelting can be accomplished immediately by interrupting the material feed.

In addition to these operations, the sequential nature of deposition and the ability to change the feedstock continuously will permit the future production of parts more specifically tailored to their applications than is presently possible. The composition, structure, and even the residual stresses in a part can be modified and controlled for optimal properties. While there exist some practical limitations on the degree to which the structure can be tailored using incremental processing, the technique clearly holds the potential for more specific structure/property control than has previously been possible.

Directed energy processing, as applied to disk fabrication, appears to offer some advantages over LAYERGLAZE processing. In the first place, the ability to detect and to remove imperfections during the actual buildup of the disk is clearly beneficial. This suggests that the process has the potential to make a virtually flaw-free disk. This in turn would enable the full benefit to be derived from the high structural strength of the material, without being compromised by concern for defect-limited fatigue behavior. Another favorable aspect of directed energy processing is its ability to erase all traces of the initial columnar grained structure in the disk by inducing in-situ recrystallization. The resulting fine-grained structure should possess higher ultimate strength and better ductility than the columnar-grained material, which would be an advantage in the bore of the disk. However, this incremental recrystallization could be halted at any desired disk diameter in order to retain a columnar-grained structure in the rim of the disk, which should be beneficial because the rim tends to run hot enough for creep to be a critical design factor.

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REFERENCES

(1) E. M. Breinan and B. H. Kear, "Surface Treatment of Superalloys by Laser Skin Melting", Superalloys, Metallurgy and Manufacture, Proceedings Third International Symposium, Claitors Publishing Div., Baton Rouge, LA, 1976, p 434.

(2) E. M. Breinan, C. O. Brown, and D. B. Snow, "Program to Investigate Advanced Laser Processing of Materials", Report No. R79-914346-4, Contract N00014-78-C-0387, United Technologies Research Center, August 1979.

(3) F. D. Lemkey, "Development of a Second Generation Ductile/Ductile $\gamma/\gamma'-\alpha$ Eutectic Alloy", Superalloys: Metallurgy and Manufacture, B. H. Kear, D. R. Muzyka, J. K. Tien, and S. T. Wlodek, eds., Claitor's Publishing Div., Baton Rouge, LA, 1976, p. 321-330.

(4) D. D. Pearson, "Ni-Al-Mo-Ta $\gamma' + \alpha$ Eutectics", Proc. Conf. In-Situ Composites-II, J. L. Walter, M. F. X. Gigliotti, B. F. Oliver and H. Bibring, eds., Ginn Custom Publishing, Lexington, MA, 1979.

(5) E. H. Aigeltinger, S. R. Bates, R. W. Gould, J. J. Hren, and F. N. Rhines, "Phase Equilibria in Rapidly Solidified Nickel Rich Ni-Mo-Al Alloys", Rapid Solidification Processing, Principles and Technologies, R. Mehrabian, B. H. Kear and M. Cohen, eds., Claitor's Publishing Div., Baton Rouge, 1978, p. 291.

(6) R. Mehrabian, "Relationship of Heat Flow to Structure in Rapid Solidification Processing", Rapid Solidification Processing, Principles and Technologies, R. Mehrabian, B. H. Kear and M. Cohen, eds., Claitor's Publishing Div., Baton Rouge, 1978, p. 9.

(7) T. Saburi, K. Komatsu, M. Yamamoto, S. Nenno, and Y. Miyutani, "A New Metastable Phase Ni_2Mo", Trans. TMS-AIME, 245, 1969, p. 2348-49.