The Mechanical Behavior of In-Situ NiAl-Refractory Metal Composites

J.-M. Yang

This article summarizes the mechanical behavior of the in-situ NiAl-refractory metal composites at room and elevated temperatures, with particular emphasis placed on the NiAl-Cr(Mo) composite with lamellar structure. Comparisons with other NiAl-refractory metal composites are also made wherever possible.

INTRODUCTION

Nickel aluminide (NiAl) has received attention as a promising candidate in the search for a low-density, high-strength structural alloy for high-performance turbine engine applications.[1-6] NiAl has several advantages over the currently used nickel-based superalloys, including low density (~5.95 g/cm³), which is about two-thirds the density of state-of-the-art nickel-based superalloys, and a high melting point (1,638°C). Other advantages include thermal conductivity (depending on temperature and composition), which is four to eight times that of nickel-based superalloys, and exceptional oxidation resistance up to 1,300°C. Furthermore, the ductile-to-brittle-transition temperature (DBTT) for <100>-oriented NiAl single crystal is approximately 350–400°C, and the DBTT for polycrystalline NiAl ranges from 400–600°C, which is lower than most other intermetallic compounds. With the application of low-density NiAl in high-pressure turbine blades, reductions of up to 40 percent in the turbine motor weight (blades and disk) can be accomplished.

Like most other intermetallic compounds, NiAl has its share of inherent problems. At ambient temperatures, NiAl suffers from low ductility and low fracture toughness. Furthermore, NiAl experiences poor creep resistance and low strength at elevated temperatures. In order to improve mechanical properties, extensive research has been conducted in the area of microalloying and the addition of second-phase reinforcements. Considerable progress has been made toward identifying alloying additions to improve the low-temperature ductility and high-temperature strength of NiAl. The yield strength and rupture strength of the strengthened NiAl alloys at elevated temperatures have been improved to René 80 strength levels. However, the challenge still remains in developing a NiAl alloy that has the required balance of ductility, toughness, strength, and other properties such as fatigue and impact resistance.

Another approach that has been taken to overcome the limitations of NiAl has been the addition of continuous-reinforcing fibers.[7,8] Because NiAl is intended for elevated-temperature applications in hostile environments for extended periods of time, chemical and mechanical compatibility between the matrix and reinforcement under adverse conditions are extremely important. Single-crystal aluminum oxide (sapphire) fibers are promising candidates because of their high-creep resistance and good chemical compatibility with NiAl,[7,9,10] but sev-
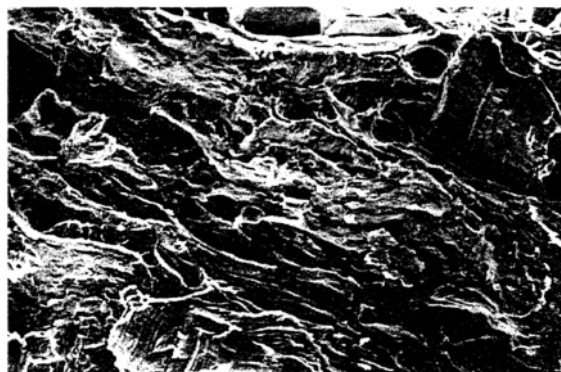

20 µm

Figure 2. Fracture morphology of NiAl-Cr(Mo) at 870°C.

eral problems have been encountered in the development of the AlO_3/NiAl composite.[8] The first and foremost problem is the degradation in strength of the Al_2O_3 fibers as a result of high-temperature exposure to NiAl during fabrication or service.[11] Sapphire-NiAl composites fall well below expectations, primarily because of fiber-strength loss, and are currently unable to compete with conventional nickel-based superalloys such as René 80.

Because of the stringent requirements of chemical and mechanical compatibility between the NiAl matrix and its reinforcement at ambient and elevated temperatures, interest has been shown in reinforcing NiAl with in-situ refractory metals by directional solidification (DS). The concept of reinforcing intermetallics with in-situ refractory metal by the DS of eutectic alloys was investigated in the early 1970s.[12-14] Recently, however, there has been a renewed interest in developing intermetallic-based DS eutectics for high-temperature structural
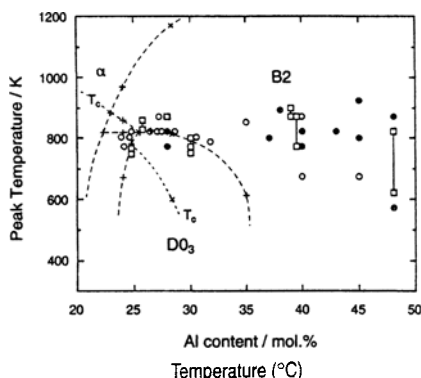

Figure 1. Tensile strengths of NiAl-Cr(Mo), NiAl-Cr, a binary NiAl, and a superalloy at room and elevated temperatures.

Table I. Microstructural Characteristics and Growth Techniques for Several In-Situ NiAl-Refractory Metal Composites

Composition (at.%)	Reinforcement Geometry	Growth Technique
NiAl-34Cr	Fibrous	Electromagnetically levitated zone process; modified Bridgeman technique
NiAl-11Mo	Fibrous	Electromagnetically levitated zone process; modified Bridgeman technique
NiAl-33.4Cr-0.6Mo	Fibrous	Electromagnetically levitated zone process
NiAl-33Cr-1Mo	Lamellar	Edge-defined film-fed growth
NiAl-31Cr-3Mo	Lamellar	Electromagnetically levitated zone process
NiAl-15.5Ta (NiAl-NiAlTa)	Lamellar	Electromagnetically levitated zone process
NiAl-33Cr-6Ta	Lamellar and Fibrous	Electromagnetically levitated zone process

applications. This is primarily motivated by the advancement of processing technology that has led to better control of processing parameters, impurity contents, and growth rates. DS eutectics can be distinguished as natural composites since eutectic alloys form a structure that consists of separate solid phases under normal solidification conditions.[15] In DS eutectics, the reinforcing phase can be either a lamellar or fibrous morphology. The advantages DS eutectics have over single-phase intermetallics include an improvement in strength and toughness due to the formation of a lamellar or fibrous morphology and a natural chemical compatibility occurring between the reinforcement and the matrix.

Recently, several NiAl-refractory metal composites have been successfully fabricated by various DS techniques such as edge-defined film-fed growth, the electromagnetically levitation process, and modified-Bridgeman techniques.[16-20] These materials include NiAl/Cr, NiAl/Mo, NiAl/V, NiAl/Cr(Mo), NiAl/NiAlTa, and NiAl-Ta-X (X = Cr, Mo, and V). The microstructural characteristics and growth techniques of several in-situ NiAl-refractory metal composites are given in Table I.[16-20]

TENSILE BEHAVIOR

The tensile properties of the NiAl-Cr(Mo) composite along the growth direction at room and elevated temperatures were investigated by Yang et al. (Figure 1).[20] The tensile behavior of the in-situ NiAl-Cr eutectic composite with fibrous structure, a binary NiAl alloy, and a superalloy were also studied for comparison.[2,14] The tensile strength of the NiAl-Cr(Mo) and NiAl-Cr composites are superior to the binary NiAl from room to elevated temperatures, although they are not as high as the superalloy. When specific properties are considered, the in-situ NiAl-Cr(Mo) and NiAl-Cr composites have a specific

Figure 3. Fatigue fracture morphology of a NiAl-Cr(Mo) composite, revealing the variation of microstructural characteristics and cracking plane.

100 μm

Table II. Fracture Toughness of In-Situ NiAl-Refractory Metal Composites

Composition (at. %)	Fracture Toughness (MPa m$^{1/2}$)	Temperature (°C)
NiAl	7	20
NiAl-11Mo	15	20
NiAl-40V	30	20
NiAl-34Cr	20.38	20
NiAl-31.4Cr-0.6Mo	19.5	20
NiAl-31Cr-4Mo	21.7	20
NiAl-28Cr-6Mo	21.6	20
NiAl-33Cr-1Mo	17.5	20
NiAl- 14.5Ta (NiAl-NiAlTa)	5.1	20
NiAl-(Cr,Al)NiTa-Cr (Ni-30.5Al-33Cr-6Ta)	14.9	20
NiAl-NiAlTa-V (Ni-28 5Al-33V-10Ta)	8.3	20
NiAl-34Cr	11.0	20
NiAl-34Cr	51 4	400
Ni-17Al-38Cr	25.4	20
Ni-17Al-38Cr	23.6	400

strength comparable to René 80. The tensile strength of the NiAl-Cr(Mo) with lamellar reinforcement is higher than the tensile strength of the NiAl-Cr composite up to 1,000°C.

At room temperature, both the NiAl/Cr(Mo) and NiAl/Cr composites display a brittle behavior with a fracture strain of less than one percent. Fractography investigations indicate that failure of the NiAl-Cr(Mo) composite tested at room temperature is initiated at the junction of grains with different lamellar orientations. However, at elevated temperatures, the NiAl/Cr(Mo) composite displayed ductile behavior with a fracture strain greater than six percent. At elevated temperatures, the fracture surface exhibited a cup-cone morphology with significant lamellar splitting along the tensile direction (Figure 2).

FRACTURE BEHAVIOR

The fracture behavior of the in-situ NiAl-Cr(Mo) composite with notch perpendicular to the growth direction has been investigated by Herridia et al.,[21] Yang et al.,[20] and Johnson et al.[18,19] The fracture-toughness values for NiAl-Cr(Mo) as well as several NiAl-refractory metal composites and binary NiAl single crystal are listed in Table II. All data were obtained using a flexural test with edge notch. The room-temperature fracture toughness with crack propagation perpendicular to the growth direction was determined to be 17.5–21.3 MPa·m$^{1/2}$. The fracture toughness of NiAl has been reported to be between 6–7 MPa·m$^{1/2}$ or 11 MPa·m$^{1/2}$ for a zone-refined single-crystal NiAl.[19] Regardless of the composition and morphology, the NiAl-Cr(Mo) composites exhibit a superior fracture-

toughness value compared to the binary NiAl single crystal. Most of the other NiAl-refractory metal composites (except the NiAl-NiAlTa system) also exhibit a significant improvement of fracture resistance over the binary NiAl alloy. In general, ductile-phase toughening is an effective approach for improving the fracture toughness of NiAl.

The principle toughening mechanisms in the NiAl-refractory metal composites have been identified as crack trapping, crack renucleation, and crack bridging.[19,21] However, the mechanisms responsible for the improved fracture resistance of the composite were dependent on the reinforcement geometry. Herridia et al.[21] showed that in a composite with fibrous structure (e.g.,
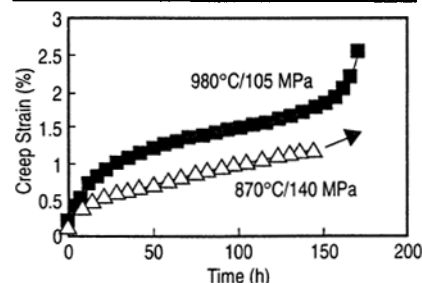

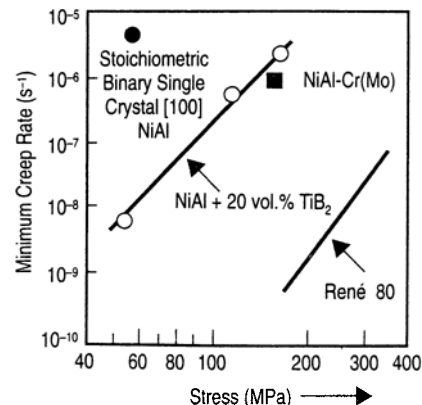Figure 4. A tensile creep curve for a NiAl-Cr(Mo) composite.

Figure 5. Minimum creep rates vs. stress plot for a NiAl-Cr(Mo) composite, a binary NiAl, a NiAl-TiB$_2$ particulate composite, and a superalloy. (Temperature = 1,145 K.)

NiAl-Cr and NiAl-Mo), crack trapping is the primary toughening mechanism. The crack front in this process is trapped (or hindered) by tough second-phase reinforcement whose fracture-toughness value exceeds the local stress intensity factor.

For composites with a lamellar structure, such as NiAl-Cr(Mo), the crack front cannot surround the tough second-phase reinforcement and must renucleate in the adjacent brittle NiAl layers. The advancing crack may be blunted or deflected by the ductile fibers or lamellae. Once a crack has initiated, unbroken fibers or lamellae may lag behind the crack front. The plastic work expended upon stretching these ductile second-phase reinforcements during crack propagation provides resistance to crack growth. This mechanism is referred to as crack bridging.[19,22] The amount of toughening provided by crack bridging depends upon the volume fraction and flow characteristics of the ductile second-phase reinforcement and the amount of debonding present at the reinforcement/matrix interfaces.

Another mechanism for the superior fracture toughness of the NiAl-Cr(Mo) and NiAl-Cr composites has been proposed by Johnson et al.[19] This includes an increase in crack-growth resistance due to linkage of microcracks, which is similar to the shear-ligament toughening observed in two-phase TiAl alloys.[23,24] Crack deflection and renucleation caused by the ductile second-phase reinforcements may result in a series of microcracks. The material between microcracks should fracture by shear as the microcracks connect with the main crack. Linkage of these microcracks with the main crack may then provide further resistance to crack growth.

FATIGUE-CRACK-GROWTH BEHAVIOR

The fatigue-crack growth of the in-situ NiAl/Cr(Mo) composite with notch perpendicular to the growth direction at room temperature has been investigated by Cheng.[25] In a typical crack-growth curve, three distinct regions can be seen—region one (1.42×10^5 cycles to 1.55×10^5 cycles), region two (1.55×10^5 cycles to 1.79×10^5 cycles), and region three (1.79×10^5 cycles to failure). Region one exhibits steady-state crack growth with an average crack-growth rate of 5.53×10^5 mm/cycle. Region two exhibits very little or no crack growth. Region three resumes steady-state crack growth until failure with an average crack-growth rate of 4.65×10^5 mm/cycle.

Fractographic analysis reveals that the steady-state crack growth in region one of the fatigue-crack-propagation curve can be attributed to fatigue-crack growth through colonies of lamellae with random lamellar orientation. The crack propagated through the Cr(Mo) and NiAl layers in each of the randomly oriented colonies with no evidence of ductile-phase bridging or interfacial debonding (delamination). However, after cycling for 1.55×10^5 cycles, the specimen exhibited no crack growth until after 1.79×10^5 cycles.

This arrest in crack growth can be attributed to the variation of lamellar orientations, where colonies of lamellae are no longer randomly oriented, but, instead, contain lamellar interfaces perpendicular to the crack-propagation direction (Figure 3). In order to propagate, the crack must travel along the path of least resistance (in this case, along the colony boundaries). This resulted in a change in the crack-growth direction from perpendicular to the loading direction to a direction that is almost parallel to the loading direction. After 1.79×10^5 cycles, the crack-propagation direction was again perpendicular to the loading direction, but in a different plane from
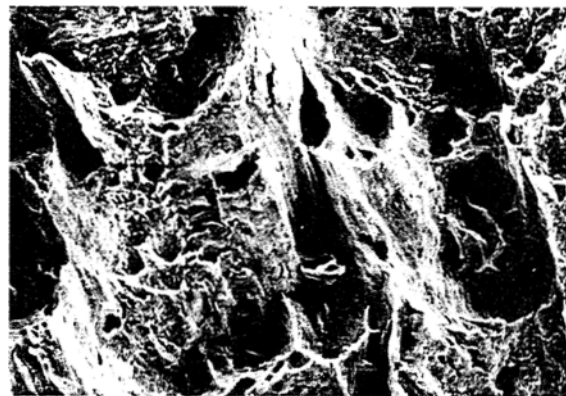

Figure 6. The fracture morphology of a NiAl-Cr(Mo) composite after creep testing at 970°C/104 MPa.

the original crack-propagation plane. Steady-state crack growth is resumed with an average crack-growth rate of 4.65×10^{-3} mm/cycle, which is less than the initial crack-growth rate of 5.53×10^{-3} mm/cycle found in region one. The decrease in crack-growth rate in region three can be attributed to the toughening mechanism of bridging. After the crack front overcame the colonies of lamellae with the same orientation, the relative movement of the crack faces of the ridge were restricted as a result of frictional contributions.

TENSILE CREEP BEHAVIOR

The tensile creep curves of the NiAl-Cr(Mo) composite along the growth direction tested at 870°C/140 MPa and 980°C/105 MPa are shown in Figure 4.[26] The composite tested at 870°C/140 MPa exhibits a transient creep regime followed by a steady-state creep regime up to 200 hours. The one tested at 980°C/105 MPa exhibited a typical three-stage creep behavior. The steady-state creep rates are calculated to be 1.39×10^{-6} s^{-1} and 3×10^{-6} s^{-1} for the composite tested at 870°C/140 MPa and 980°C/105 MPa, respectively.

The minimum creep rates versus stress for a NiAl-Cr(Mo) composite at 870°C compared to a binary single-crystal NiAl, NiAl containing 20 vol.% TiB$_2$ particles, and a superalloy are shown in Figure 5.[17] The creep resistance of the in-situ NiAl-Cr(Mo) is superior to that of the binary NiAl single crystal, although it still did not compare well against the superalloys.

Microstructural investigation of the composite reveals that the fracture morphology and the mechanisms of damage initiation in the creep-tested composite are similar to those found in high-temperature tensile fracture (Figure 6). The creep damage initiated primarily at the eutectic grains with misaligned lamellar boundary or at the growth defects of the lamellar structure.

Table III. Compressive Creep Behavior of Several In-Situ NiAl-Refractory Metal Composites

| Composition (at.%)	Stress Exponent	Activation Energy (kJ/mol)	Temperature Range (K)
NiAl [001] (Ni-50Al)	6.3	439.6	1,100–1,300
NiAl-Cr (Ni-33Al-34Cr-0.1Zr)	7.69	—	1,300
NiAl-Cr(Mo) (Ni-33Al-28Cr-6Mo)	6.38	456.6	1,200–1,400
NiAl-NiAlTa (NiAl-14.5Ta)	4.9	522	1,200–1,400
NiAl-(Cr,Al)MTa-Cr (Ni-30.5Al-33Cr-6Ta)	5.22	445.2	1,100–1,300 (low stress exponent regime)
NiAl-NiAlTa-(Mo,Ta)	6.18	—	1,200
NiAl-NiAlTa-(Mo,Ta)	3.6	317.9	1,300–1,400
NiAl-Cr (Ni-31Al-36Cr)	6.1	—	1,300
NiAl-Cr(Mo) (Ni-33Al-33Cr-1Mo)	8.2	—	1,300
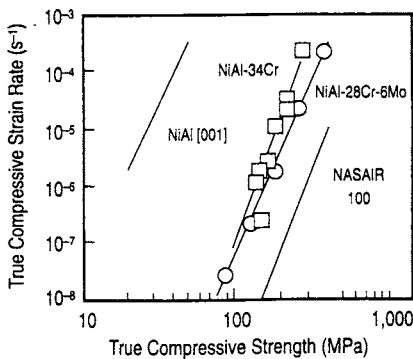
Figure 7. The compressive flow stress-strain rate behavior at 1,300 K for the NiAl-34Cr and NiAl-28Cr-6Mo eutectics compared to NiAl and a single-crystal superalloy.
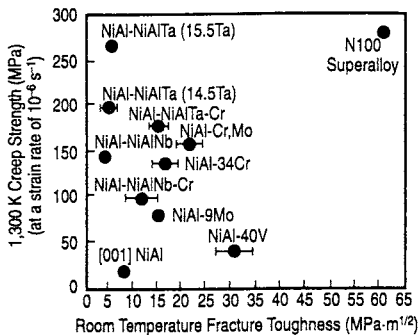

Figure 8. The performance of NiAl-based DS eutectics compared to a binary single-crystal NiAl and a single-crystal superalloy.

COMPRESSIVE CREEP BEHAVIOR

The compressive creep behavior of the in-situ NiAl-Cr(Mo) composite along the growth direction has been investigated by Johnson et al. and Pollock et al.[27] The strain rates versus stress plot for NiAl-28Cr-6Mo and NiAl-Cr eutectic composites are shown in Figure 7. Results for a binary NiAl single crystal and a single-crystal superalloy are also included for comparison. The addition of fibrous or lamellar reinforcements through DS has a significant influence on the creep behavior of NiAl. The steady-state creep rate of the composites under compression are, in general, several orders of magnitude lower than the binary single-crystal NiAl. The kinetics of creep deformation of NiAl-Cr(Mo) and several other NiAl-refractory metal composites are summarized in Table III.[18,19,27] However, Pollock et al. showed that the creep resistance of the NiAl-Cr(Mo) and NiAl-Cr composite transverse and at 45° to the growth direction are inferior to that of the longitudinal direction.

Johnson et al.[18] compared the performance of several NiAl-refractory metal composites with a binary single-crystal NiAl and a single-crystal superalloy. The compressive creep resistance at 1,300 K plotted against the room-temperature fracture toughness is shown in Figure 8.

Among the NiAl-refractory metal composites, the NiAl-Cr(Mo) system exhibits a good combination of creep resistance and room-temperature damage tolerance, although it is still not as good as the superalloys.

CONCLUSIONS

Significant progress has been made toward developing in-situ NiAl-refractory metal composites for high-temperature structural applications. The presence of well-aligned fibrous or lamellar reinforcement through DS can provide significant improvement in room-temperature fracture resistance and high-temperature creep resistance over the binary NiAl. However, these in-situ composites are anisotropic.

Much more concerted efforts are needed to further improve the performance of these materials from room to elevated temperatures. The alloy compositions and processing conditions that control the resulting microstructure and properties need to be optimized. The durability and deformation characteristics of these materials in a high-temperature oxidizing environment need to be better understood. Finally, development of design and testing methodology for components made out of low-ductility and anisotropic materials will also be required.

References

1. D.B. Miracle, "The Physical and Mechanical Properties of NiAl," *Acta Met. et Mat.*, 41 (3) (1993), p. 649.
2. R. Darolia et al., "Overview of NiAl Alloys for High Temperature Structural Applications," *Ordered Intermetallic Compound—Physical Metallurgy and Mechanical Behavior*, ed. C.T. Liu et al. (1992), p. 679.
3. R. Darolia, "NiAl for High Temperature Structural Applications," *JOM*, 43 (3) (1991), p. 44.
4. R.D. Noebe, R.R Bowman, and M.V. Nathal, "Review of the Physical and Mechanical Properties of the B₂ Compound NiAl," *Int. Mat. Rev.*, 38 (1993), p. 193.
5. D.B. Miracle and R. Daroha, "Structural Applications of NiAl," *Intermetallic Compounds: Principals and Practice*, ed. J. Westbrook et al. (New York: John Wiley & Sons, 1993).
6. R. Darolia, "NiAl for Turbine Foil Applications," *Structural Intermetallics*, ed. R. Darolia et al. (Warrendale, PA: TMS, 1993), p. 495.
7. R.R. Bowen, "Influence of Interfacial Characteristics on the Mechanical Properties of Continuous Fiber-Reinforced NiAl Composites," *Intermetallic Matrix Composites II*, 273, ed. D.B. Miracle et al. (Pittsburgh, PA: MRS, 1992), p. 145.
8. R.A. Amato and J.-M. Yang, "Status and Development of Nickel Aluminide (NiAl) Composites," *Intermetallic Matrix Composites III*, ed. J.A. Graves et al. (Pittsburgh, PA: MRS, 1994), p. 212.
9. C. Liu et al., "Processing and High Temperature Deformation of Al₂O₃ Fiber-Reinforced NiAlFe Composites," *Mat. Sci. and Engrg.*, A191 (1995), p. 49.
10. C. Ward and A. Culbertson, "Issues in Potential IMC Applications for Aerospace Structures," *Intermetallic Matrix Composites III*, ed. J.A. Graves et al. (Pittsburgh, PA: MRS, 1994), p. 3.
11. S.L. Draper and I.E. Locci, "Al₂O₃ Fiber Strength Degradation in Metal and Intermetallic Matrix Composites," *J. of Mat. Research*, 9 (6) (1994), p. 1397.
12. H.E. Cline and J.L. Walter, *Met. Trans.*, 1 (1970), p. 2091.
13. H.E. Chen et al., *Met. Trans.*, 2 (1971), p. 789.
14. J.L. Walter and H.E. Cline, *Met. Trans.*, 1 (1970), p. 1221.
15. D.M. Shah and D. Anton, "In-Situ Synthesis of Intermetallic Matrix Composites," in Ref. 7, p. 225.
16. K.M. Chang, "Fracture Mechanisms in NiAlCr Eutectic Composites," in Ref. 7, p. 191.
17. K.S. Kumar and G. Bao, "Intermetallic Matrix Composites: An Overview," *Comp. Sci. and Tech.*, 52 (1994), p. 127.
18. D.R. Johnson et al., "NiAl-Based Polyphase In-Situ Composites in the NiAl-Ta-X Systems, *Intermetallics*, 3 (1995), p. 493.
19. D.R. Johnson et al., "Processing and Mechanical Properties of In-Situ Composites from the NiAl-Cr and the NiAl-(Cr,Mo) Eutectic Systems," *Intermetallics*, 3 (1995), p. 99.
20. J.-M. Yang et al., "Microstructure and Mechanical Behavior of In-Situ Directional Solidified NiAl/Cr(Mo) Eutectic Composites," *Acta Materialia*, 45 (1) (1997), p. 295.
21. F.E. Herradia et al., "The Fracture Resistance of Directionally Solidified Dual-Phase NiAl-Reinforced with Refractory Metals," *Acta Met. et Mat.*, 41 (2) (1993), p. 305.
22. B.D. Flin, M. Ruhle, and A.G. Evans, "Toughening in Composites of Al₂O₃ Reinforced With Al," *Acta Met. et Mat.*, 37 (1989), p. 3001.
23. K.S. Chan, "Micromechanics of Shear Ligament Toughening," *Met. Trans.*, 22A, 1991, p. 2021.
24. K.S. Chan, "Understanding Fracture Toughness in Gamma TiAl," *JOM*, 44 (5) (1992), p. 30.
25. K. Cheng, M.Sc. thesis, University of California, (1996).
26. J.-M. Yang, R.A. Amato, and K. Bain, unpublished results.
27. T.M. Pollock and D. Kolluru, "Deformation of NiAl-based Intermetallics," *Micromechanics of Advanced Materials*, ed. S.N.G. Chu et al. (Warrendale, PA: TMS, 1995), p. 205.

ABOUT THE AUTHOR ▬▬▬▬

J.-M. Yang earned his Ph.D. at the University of Delaware in 1986. He is currently a professor in the Department of Materials Science and Engineering at the University of California at Los Angeles. Dr. Yang is a member of TMS.

For more information, contact J.-M. Yang, University of California at Los Angeles, Department of Materials Science and Engineering, Los Angeles, California 90095-1595; (310) 825-2758; fax (310) 206-7353; e-mail jyang@seas.ucla.edu.