FOREWORD

This report was prepared by the General Electric Research Laboratory under USAF Contract No. AF 33(616)-7714, "Research on the Effects of Basic Physical Parameters on Engineering Properties of Intermetallics." This contract was initiated under Project No. 7350, "Refractory Inorganic Non-Metallic Materials," and Task No. 735001, "Ceramic and Cermet Materials Development." The work was administered under the direction of the Directorate of Materials and Processes, Deputy for Technology, Aeronautical Systems Division, with Lt. Jacobson acting as project engineer.

This report covers work conducted from December 1962 through December 1963.

The authors are pleased to acknowledge the assistance of W. Moore and J.W. Rutter for the preparation of material, of T.E. Douglas for assistance in hot-rolling, of G.E. Smith and J.F. Fleischer for analytical services, of E.M. Raviola, T.M. Brassard, and W.A. Roman for metallographic preparation, and of W. Rollins and L. Cook for mechanical tests.

ABSTRACT

Ingots of nickel-rich NiAl were prepared by induction melting under argon, starting with primary materials of technical grade. Sections cut from these ingots have been hot-rolled to sheet at 1100° to 1200°C by enclosing them in heavy-walled stainless steel containers. Rolled sheet sufficiently free of defects to enable preparation of mechanical test specimens is obtained. Cold-rolling of the hot-rolled sheet can be accomplished at 850°C but requires a heavy reduction (about 30%) in each pass to avoid fracture.

Hot-rolled NiAl undergoes a transition to brittle behavior at about 600°C. Below this temperature the mechanical properties are characterized by limited ductility, a sensitivity to surface notches, and by a flow stress that is relatively invariant in respect to temperature. Above 600°C, ductility increases markedly, sensitivity to surface conditions diminishes, and the flow stress becomes strongly temperature dependent. Single crystals do not exhibit a transition in mechanical behavior and are much weaker and more ductile than polycrystalline material at temperatures below 800°C. Single-crystal rods have been bent at room temperature to a maximum fiber strain of 25% without fracture. Polycrystalline material is of more limited ductility, the maximum fiber strain for fracture at room temperature being a few per cent.

Some few observations on the ductile-brittle transition and other low-temperature properties of the compound AgMg are also included.

This technical report has been reviewed and is approved.

W.G. Ramke
Chief, Ceramics and Graphite Branch
Metals and Ceramics Division
Air Force Materials Laboratory

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I. INTRODUCTION

Among the intermetallic materials are several compounds that are of considerable potential for high-temperature structural applications. The use of intermetallics as structural materials, however, as well as their mechanical fabrication for experimental and other purposes has been hindered by the general absence of knowledge or understanding of their mechanical behavior. This contract is directed toward filling this gap with particular attention to what is probably the most serious mechanical shortcoming of intermetallics: their lack of ductility at ordinary temperatures.

The program seeks the establishment of correlations of structural, compositional, and physical parameters of intermetallic compounds with their mechanical properties. It is hoped that such correlations will lead to an elucidation of the factors affecting the flow process in such materials, and particularly to a better understanding of the source of brittleness in polycrystalline intermetallics.

From December 1958 to December 1960, techniques were studied for producing suitable test specimens of both AgMg and NiAl, methods for testing were devised, and the tensile behavior of the CsCl structure compound AgMg was extensively documented in terms of strain, strain rate, temperature, grain size, composition, and metallurgical processing treatment.[1, 2]

From December 1960 to December 1961, a study was made of the grain boundary hardening that was found to occur in many intermetallic compounds having a stoichiometric excess of active metal component.[3] This grain boundary hardening was shown to be associated both with the anomalously high ductile-brittle transition temperatures common in these materials and with the "pest" phenomenon occurring in certain intermetallics.

From December 1961 to December 1962, the effects of metallurgical variables on the ductile-brittle transition temperature of AgMg were examined using a unique testing device. Studies of compositional effects on flow stress of AgMg and of oxygen-induced grain boundary hardening in that compound were also extended. All of these studies are reported in the last summary report.[4]

As indicated above, previous studies under this contract have been concerned with the compound Bi_2Tl and the CsCl structure compounds NiAl and AgMg. Processing studies were carried out on all three materials, and although feasibility of a hot extrusion technique was established for all of these compounds, it was not possible to obtain consistently good product from the high-melting NiAl by this method. For this reason most of the tensile property measurements

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made during the course of the investigation were on extruded AgMg. Both to establish the generality of these observations and to obtain results on a material of possible application for structural purposes, it was decided to put renewed emphasis upon the mechanical fabrication of NiAl in order to prepare material suitable for mechanical testing.

A major deterrent to the use of, or experimentation with, intermetallic phases is the limited ductility usually found at temperatures below the hot-working range. However, the ductility of AgMg was notably improved by hot extrusion and many of the refractory metals--for example, tungsten--also show improved ductility when mechanical working is continued from a hot-working range down into the cold-working range. Therefore, the consequences of such hot-cold-work were examined during the current period on NiAl. Hot-rolling, rather than extrusion, was chosen as the processing technique because of the possibility of better process control. Since the earlier work with AgMg showed that ductility is greater with nonstoichiometric compositions having an excess of the less reactive component, initial attention has been given to nickel-rich NiAl compositions. Finally, some consideration has been given to ductility improvement via the preparation of material of exceptional purity since the ductility of the more reactive metals is frequently related to purity, particularly in those instances where a wide range of solubility exists for interstitial impurities.

In addition to the above studies on NiAl during the current period, some few additional experiments have been carried out on AgMg: the effects of grain boundary hardening on transition temperatures and the effects of ternary additions on low-temperature tensile behavior. The work that has been undertaken in both of these areas will now be discussed in detail.

II. PROCESS STUDIES ON NiAl

A. Melting Procedure

Material of technical grade was prepared by induction melting in the form of a 25-lb ingot of NiAl (heat 702) having a charge composition: Ni 52, Al 48 A/o (atom %). The charge consisted of carbonyl nickel and 99.99% aluminum. The crucible was a high-purity MgO (Lava 50.395). The melting procedure consisted of the following steps: (1) the nickel was charged and melted under an atmosphere of argon, (2) the furnace chamber was evacuated to a pressure of 50 microns and the melt solidified to aid in removal of gases, (3) argon was readmitted to a pressure of 1 atm and the charge remelted, (4) the aluminum was charged and melted, (5) the melt was poured into a graphite mold 3 3/8 inches in diameter by 9 inches long with a refractory hot top. Chemical analysis of the ingot yielded:

Ni--53 A/o

Al--47 A/o

O_2--0.0018 w/o (weight %)

Additionally, several small ingots (1 inch in diameter by 2 inches long) were prepared in the following way. Vacuum remelted, carbonyl nickel and 99.999% aluminum in the proportion: Ni 52, Al 48 A/o were charged into an alumina crucible and melted under 1 atm of argon by high-frequency induction heating. The crucible was then lowered with reference to the coil so that the melt solidified directionally, upward from the bottom of the crucible. All of the ingots consisted of columnar grains extending for the length of the ingot; however, one ingot consisted of only a few large grains. Small rods of cylindrical shape, nominally 0.18 inch in diameter by 2 inches long, were cut from this ingot by electric discharge machining using a tantalum tube as an electrode. These rods frequently consisted of a single crystal and were subsequently employed in bend tests and in cold-rolling experiments to evaluate the properties of single crystals. Prior to such use the rods were hand polished with silicon carbide paper to remove ripples formed in the machining process, and were then electropolished. No other surface defects were observed resulting from electric discharge machining. Chemical analysis of this ingot yielded the following:

Ni--52% (A/o)

Al--48% (A/o)

O--0.0009% (w/o)

B. Hot-Rolling Procedure

In preparing to hot-roll NiAl, account was taken of two properties of the compound that would affect rollability: (1) the hardening and embrittlement of grain boundaries as a result of oxidation at high temperature, and (2) greatly reduced ductility at lower, hot-working temperatures. Accordingly, clad billets were prepared for hot-rolling which consisted of a section cut from the NiAl ingot enclosed in a heavy-walled steel container. The container served to isolate the NiAl billet from contact with both the ambient atmosphere and the cold rolls of the mill. In addition, the enclosure restricts lateral movement of the NiAl billet during rolling and thereby reduced the opportunity for edge cracking. Initially, the containers were constructed of mild steel. More recently, type 347 stainless has been used for the two cover plates since the flow stress of stainless in the temperature range near 1200°C is a better match for NiAl than mild steel. Figure 1 shows the form of enclosure used. Table I gives a roll pass schedule that has been found to produce hot-rolled NiAl sheet that is generally sound over most of its length.

Removal of the rolled-on stainless cladding may be accomplished by a machining operation such as surface grinding. However, it has been observed that grinding may produce a network of shallow microcracks in the NiAl surface and allowance must be made for such a contingency. Electropolishing is, therefore, preferable for removal of the last few mils of the stainless cladding. Another alternative to which future attention should be given is the avoidance of welding between the enclosure and the NiAl insert through the use of a suitable refractory interface.

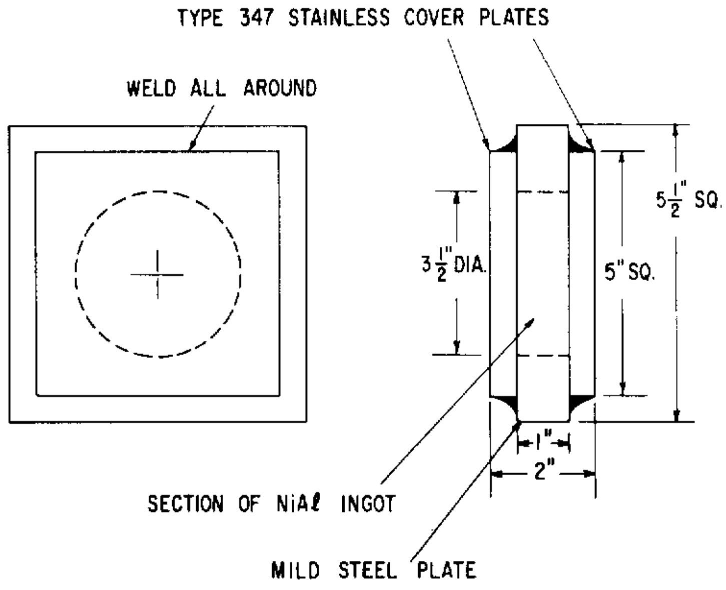


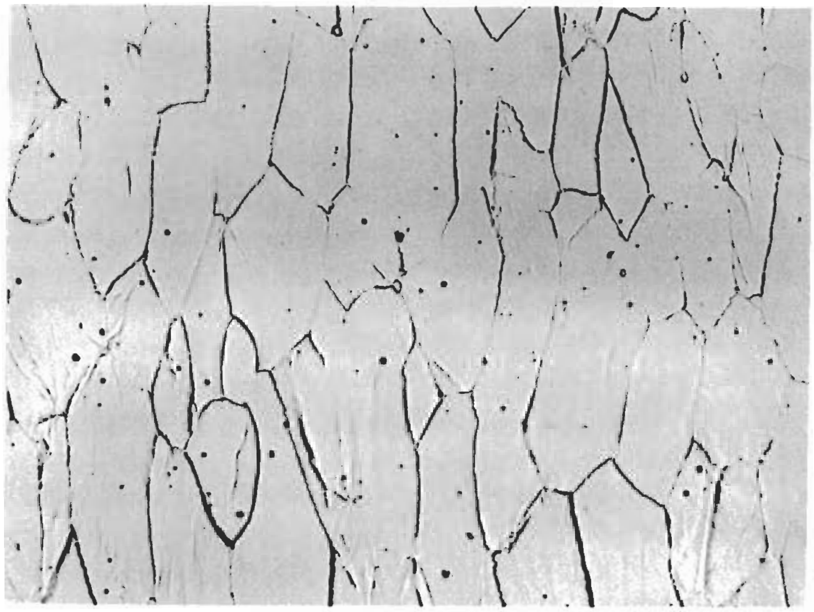
Figure 1. Enclosure for Hot-Rolling NiAl.

TABLE I
HOT-ROLLING SCHEDULE FOR NiAl IN STAINLESS TYPE 347 CONTAINER
[Birdsboro, 2-high, hot reversing mill (16-inch-diameter rolls)]

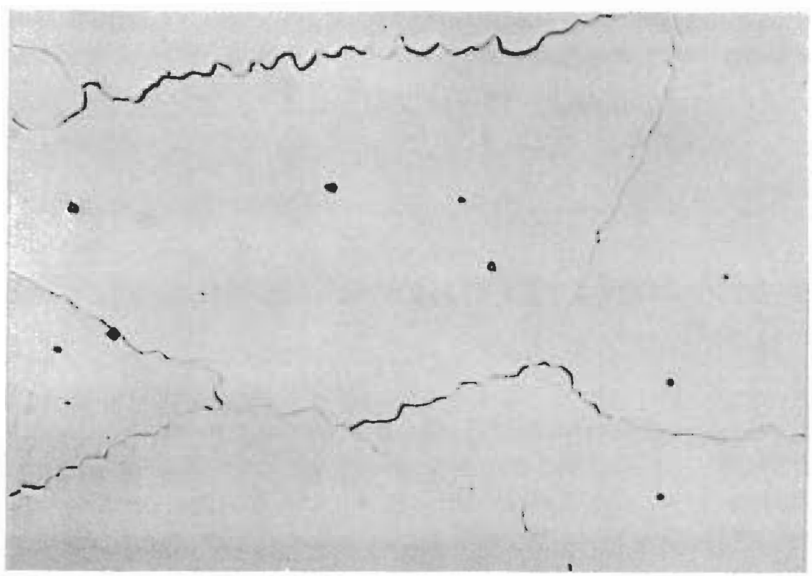
Pass No.	Roll Opening (inch)	Sheet Thickness (inch)	Reduction in Thickness (%)	Roll Speed (ft/min)	Rolling Temp (°C)
1	1.590	1.62	19	41	1200
2	1.270	1.30	20	41	1200
3	1.010	1.04	20	41	1175
4	0.750	0.78	25	87	1150
5	.560	.60	23	87	1125
6	.420	.46	23	87	1125
7	.310	.35	24	87	1100
8	.225	.27	23	136	1075
9	.180	.22	19	136	1050
10	.135	.18	18	136	1050
11	.095	.15	17	136	1050
12	.065	.12	20	136	1050

Sheet was returned to furnace and reheated following each pass except the last. Furnace atmosphere was nitrogen.

Figure 2(a) is a micrograph that shows the usual structure observed after hot-rolling at 1200°C according to the schedule of Table 1. Grains are only slightly elongated in the rolling direction and are generally free of the banding that is observed following cold deformation. Grain boundaries, however, frequently appear to be serrated on a microscopic scale as shown in Fig. 2(b). While this feature of the hot-rolled structure is not understood, it is believed that the serrations are related to the deformation process rather than to some form of boundary pinning.



(a) Hot-Rolled at 1200°C. 100X.



(b) Hot-Rolled at 1200°C. 500X.

Figure 2. Representative Microstructures of Hot-Rolled Polycrystalline NiAl.

C. Cold-Rolling Procedure

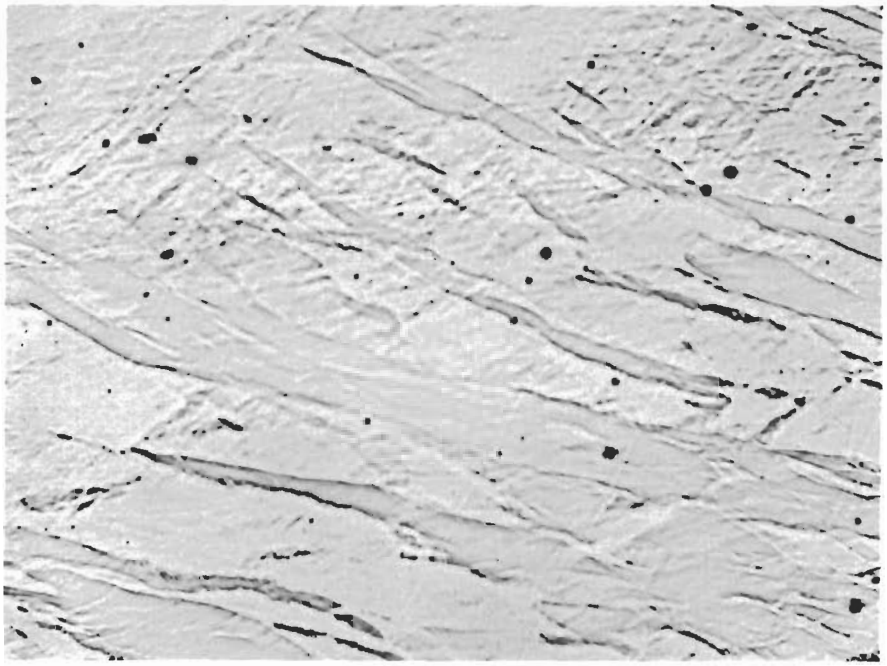
Initial attempts to cold-roll near room temperature the clad sheet resulting from the hot-rolling experiments were uniformly unsuccessful. Plastic deformation within individual grains was evident from the microstructure, but fracture occurred along grain boundaries, along the interface between deformation bands and along cleavage planes, resulting in fragmentation of the specimen. This result will be illustrated later in a comparison with other structures resulting from cold-rolling at higher temperatures.

To avoid the problem of fracture along grain boundaries, the effort was then shifted to the single-crystal rod specimens whose preparation was previously described. One of these rods was rolled through a sequence of five passes, heating the rod to 500°C before each pass. The individual reductions in thickness were uniformly about 5% and the total reduction of thickness was 25% with a 23% increase in length and an imperceptible change in width. During the fifth pass the entering end of the bar was split down the center for a short distance and rolling was discontinued. Annealing 1 hour at 800°C did not result in recrystallization nor did a second anneal of 1 hour at 1000°C. This experiment demonstrates that individual crystals of NiAl have sufficient ductility to permit rolling at temperatures not far removed from normal ambient values.

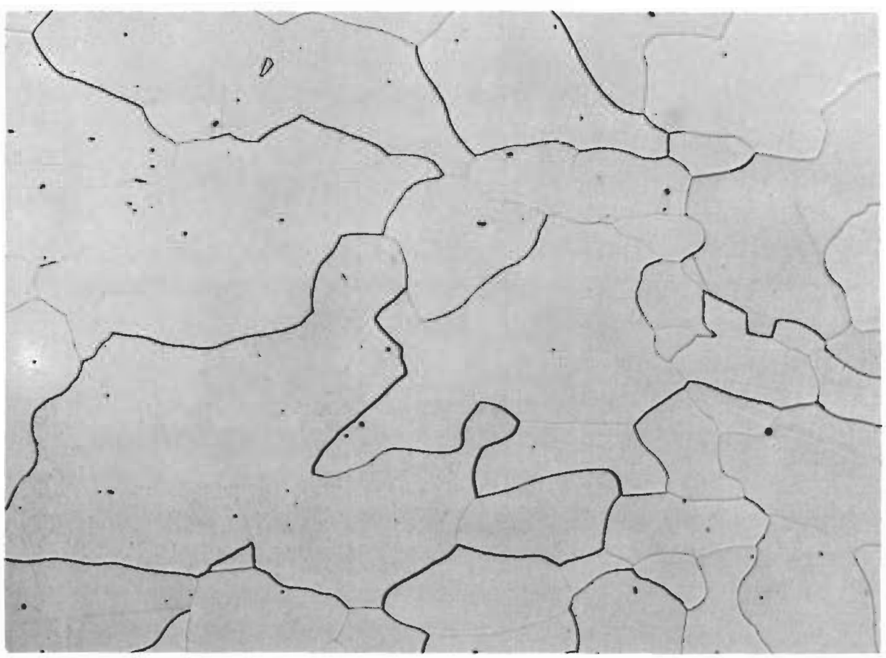
A second rod was heated to 820°C and rolled in four passes to a strip 0.069 by 0.275 inch for a total reduction of 71%. The bar was reheated before each pass. On this occasion the loss of temperature to the rolls could be estimated from visual observation, and it is estimated that the actual rolling temperature was about 500°C. Several cracks developed on the edge of the specimen, but the main body of the piece was sound. This specimen underwent recrystallization upon heating to 1000°C.

A third specimen was enclosed in a fitted stainless steel tube having a one-eighth-inch wall. It was heated to 820°C and rolled in a single pass at 10 ft/min to a strip 0.075 inch thick by 0.350 inch wide for a reduction of thickness of 57%. The rolled specimen contained no surface cracks although deformation markings were evident. One-half of this piece was then enclosed in an envelope of 0.032-inch-thick stainless steel and rolled in a second pass at 820°C to 0.036-inch thickness, a pass reduction of 52%, and a total reduction of 80%. There was, again, no evidence of fracture.

The microstructure of the piece receiving an 80% reduction is seen in Fig. 3(a). The piece having only the single pass had a microstructure very little different from that of Fig. 3(a) and the hardness was only slightly lower. Annealing 1 hour at 1000°C resulted in complete recrystallization of both specimens as shown in Figs. 3(b) and (d). The grain size is smaller for the piece having the greater reduction. Annealing 1 hour at 800°C resulted in the recrystallization of some regions and the retention of the cold-worked structure in other areas of the piece receiving the 57% reduction [Fig. 3(c)]. Thus, 800°C appears to define the lower end of the temperature range in which recrystallization of

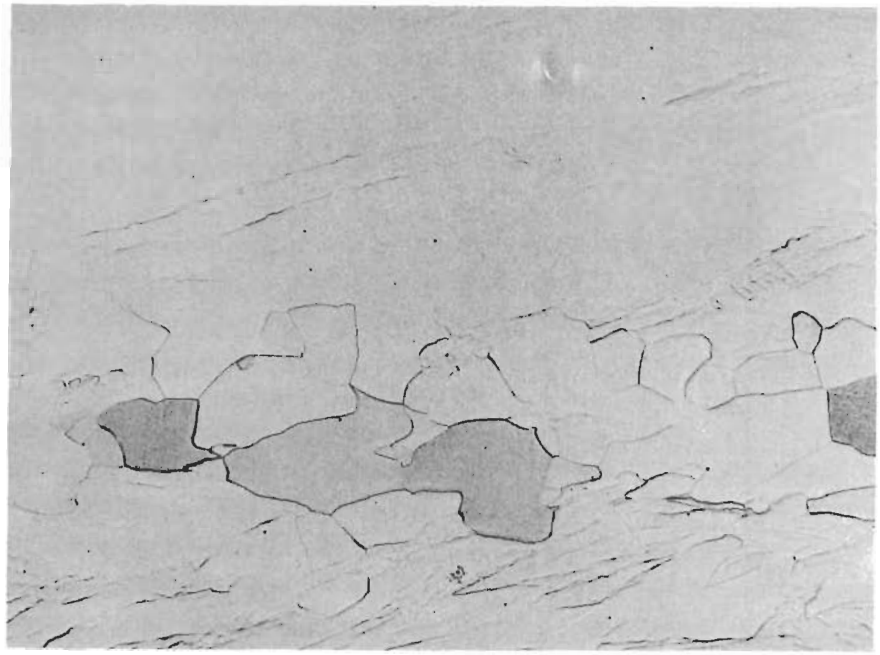


(a) Reduced 80% in Two Passes at 820°C (Hardness: 328-346 DPH). 150X.

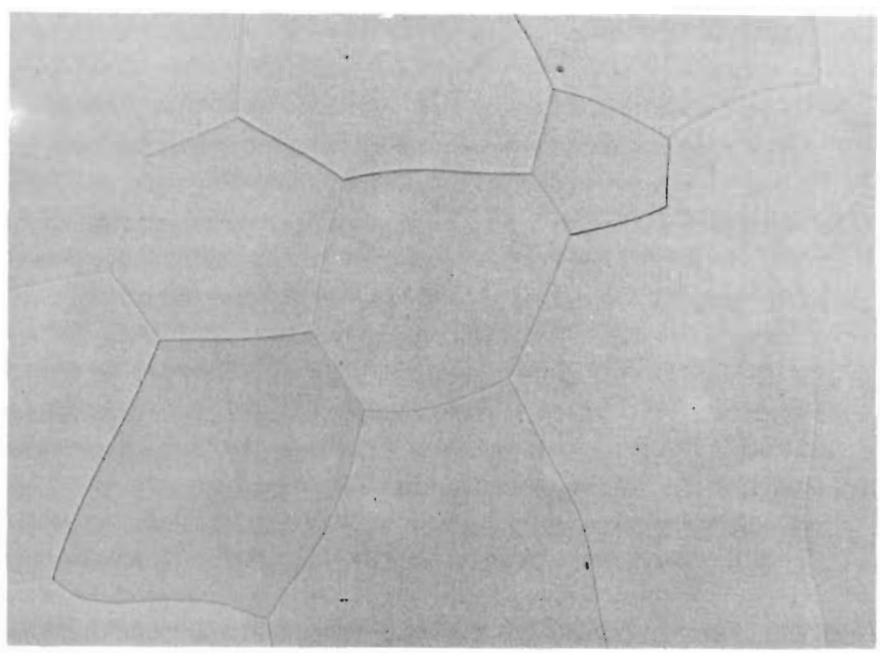


(b) Reduced 80% at 820°C, Annealed 1 hour, 1000°C (Hardness: 217-232 DPH). 150X.

Figure 3. Representative Microstructures of Rolled and Annealed NiAl Single-Crystals.



(c) Reduced 57% in One Pass at 820°C, Annealed 1 Hour, 800°C (Hardness: 241-249 DPH--Unrecrystallized Areas), (Hardness: 232-241 DPH--Recrystallized Areas). 150X.



(d) Reduced 57% at 820°C, Annealed 1 Hour, 1000°C (Hardness: 241 DPH). 150X.

Figure 3. (Continued)

severely deformed NiAl is to be expected. Recovery, as indicated by a loss of hardness without recrystallization, appears to be complete in 1 hour at 800°C.

The mechanical properties of hot-rolled NiAl sheet, which are described later in this report, make it evident that cold-working of polycrystalline NiAl must be undertaken at a temperature above 800°C. Below 800°C the ductility is insufficient for success in rolling to be probable. Moreover, the rapid rise in the flow stress with temperature below 800°C results in increased loading of the mill. The metallographic results for cold-rolled and annealed single crystals, already reported, show that recrystallization of severely deformed NiAl occurs at a temperature as low as 800°C. Thus cold deformation must be undertaken at a temperature at which recrystallization will ultimately take place as deformation proceeds. The amount of cold-working that is possible is limited, therefore, to that which causes recrystallization unless the rolling temperature is lowered as deformation proceeds. This is a possibility that has not yet been explored in this work.

When stainless-clad, hot-rolled NiAl sheet was cold-rolled in the temperature range 850° to 900°C, two forms of mechanical failure were encountered. One of these consisted of the rupture of the NiAl portion of the sheet along surfaces essentially normal to the rolling direction. The second consisted of a fracture of the NiAl along a plane parallel to the plane of rolling, and positioned, usually, at mid-thickness. This occurred after the rolled piece had been cooled to near room temperature, and frequently caused splitting of the piece into two halves. These forms of fracture are referred to as "transverse rupture" and "delamination," respectively.

Transverse rupture was encountered during the initial attempts to cold-roll, using small reductions in each mill pass. In one trial at 850°C, pass reductions varied between 5% and 15% for a total of 44% in seven passes. In another trial at 900°C, reductions were held to about 5% per pass for a total of 28% in eight passes. Reheating was employed with each pass to maintain control of temperature. Rolling speed was 120 ft/min and the stainless cladding was, initially, 0.06 inch thick thereby insulating the NiAl element from heat loss to the roll. The variability of the pass reductions when rolling with reheating to 850°C suggests that recrystallization occurred periodically during reheating, but not between every pass. When rolling with reheating to 900°C, softening seemed to occur uniformly during each reheat period, for the pass reductions were uniform.

In both of these trials the NiAl element was ruptured at intervals of about 1/4 inch along the length. The cladding was also separated from the NiAl sheet over a part of the length making direct examination of the surface of the NiAl possible. Examination of this surface at 25 diameters magnification showed that local separation, or fracture, along grain boundaries was very common, occurring particularly frequently at the junction of three boundary surfaces. Rupture of the sheet appeared to have occurred by the linking up of isolated grain boundary fissures with some elements of cleavage and shear fracture added. In subsequent rolling experiments it was found that rupture of the NiAl portion of the sheet becomes less frequent with heavier pass reductions and does not occur with reductions of 30% or more per pass. The benefit of the large pass reduction in this instance appears to be from the increased roll pressure that ensues. Deformation at the rate that is encountered during rolling at 120 ft/min evidently leads to rupture at small elongations in the absence of a large transverse compression. Whether the requirement of a high reduction in the first pass applies to subsequent passes, or whether, once a worked structure is attained, more moderate reductions are possible, has not been determined. With the necessity of lowering the rolling temperature as deformation proceeds, to avoid recrystallization, it would become increasingly difficult to maintain a large reduction in passes subsequent to the first.

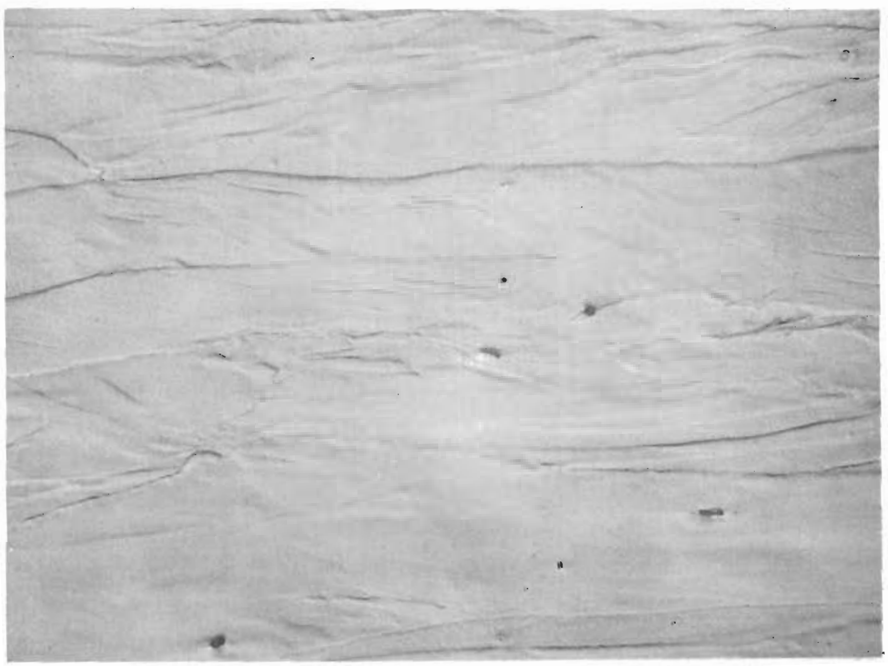
Delamination of stainless-clad NiAl sheet was first observed in hot-rolled sheet where the thickness was reduced by rolling to 0.10 inch. It was also encountered during cold-rolling of this sheet. The cause of delamination was shown to be the difference in thermal contraction of the stainless cladding and NiAl during cooling from the rolling temperature. When delamination occurred, the two sheets of half-thickness were curved so as to place the stainless layer on the inside surface. Heating to rolling temperature caused these strips to straighten and become flat. Removal from the furnace and cooling in still air, while positioned on one edge to permit equal cooling of the two large surfaces, resulted in resumption of the curved shape. Holding the rolled piece at an intermediate temperature during cooling from rolling reduces the tendency to delaminate. It is expected that the use of a ferritic steel as the cladding would reduce the tendency to delaminate through a better match of expansion coefficients. More interestingly, it has been found that with cold deformations of 40% or more the tendency to delaminate becomes very much less; whether this is due to the reduced thickness of the cladding or to improved fracture strength on the part of the NiAl has not been determined.

Figure 4(b) shows the microstructure of cold-rolled. polycrystalline NiAl, reduced 56% in thickness in a single pass at 800°C. The contrast between this structure and that of Fig. 4(a) where rolling was done at 200°C with light passes is entirely evident. Figure 4(c) shows the microstructure of another piece cold reduced in two passes of 37% and 30%, respectively, The piece was heated to 850°C before each pass, and it is evident that recrystallization has occurred in some regions during the intermediate heating. No mechanical property measurements have been made as yet on cold-rolled NiAl, although it appears that this should now be possible.

III. MECHANICAL PROPERTIES STUDIES ON NiAl

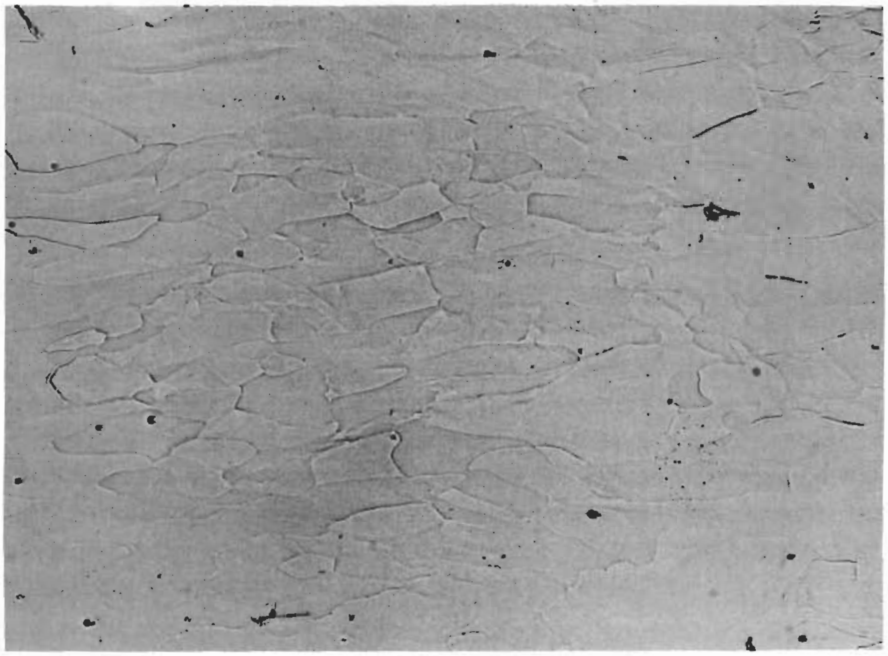
Mechanical property measurements have consisted of simple, three-point bend tests, and tensile tests. Bend tests have utilized specimens of two shapes: bars of rectangular cross section that were cut from hot-rolled sheet with an abrasive saw, and cylindrical bars that were prepared by "spark-cutting" from

(a) Reduced 50% in 14 Passes at 200°C. 500X.



(b) Reduced 56% in One Pass at 800°C. 500X.

Figure 4. Representative Microstructures of Cold-Rolled Polycrystalline NiAl.



(c) Reduced 56% in Two Passes at 850°C. 150X.

Figure 4. (Continued)

one small ingot. The rectangular specimens were nominally 1/4 inch wide and 0.06 inch thick, the thickness dimension being that of the hot-rolled sheet after removal of the clad layer by surface grinding. The cylindrical specimens were nominally 0.18 inch in diameter. All bend tests were made on a 1 1/2-inch span with as-ground specimen surfaces unless stated otherwise. The tensile specimen is shown in Fig. 5 and utilizes a roll clad sheet to facilitate loading of the specimen at temperatures where the NiAl element is of limited ductility. All mechanical tests were performed in an Instron tester.

A. Bend Tests

Bend tests have been used widely in mechanical investigations of brittle materials. The specimens are simple to prepare and loading presents a minimum of difficulty. Results of bend tests may be used to locate the transition temperature between brittle and ductile behavior and to determine the influence of testing conditions and of processing variables upon this temperature. Owing to the complexity of the stress distribution in bending once the elastic range is exceeded, one can determine stresses rigorously only for the elastic range. However, for small excursions into the plastic range the maximum fiber stress, determined with the assumption of elastic deflection, provides an approximation to the yield, or flow, stress. In this report we are using the term "proof stress in bending" for the maximum fiber stress calculated in this way for a small plastic deflection of the specimen, usually

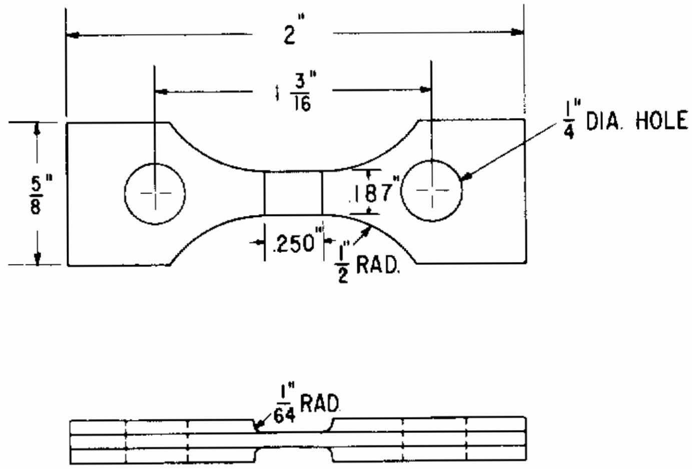


Figure 5. Tensile Test Specimen.

0.0035 inch. This is a small, but identifiable, plastic strain and the stress so determined is in good agreement with the 0.2% proof stress obtained from tensile tests at equivalent strain rates. Bend test data for rectangular specimens that are relatively long and thin in shape were also used to calculate values of the elastic modulus. For these specimens the elastic deflections are large at small values of bending loads, and machine deflection (which must be subtracted from the observed total deflection) is a small correction. The elastic moduli so determined are believed to be reliable.

Bend data for the rectangular test specimens are summarized in Tables 2 and 3. The data of Table 2 refer to specimens in the hot-rolled condition and those of Table 3 to specimens that received a high-temperature heat treatment. This heat treatment was patterned after that used by Seybolt and Westbrook[5] to reduce grain boundary hardening effects in NiAl, and consisted of 16 hours at a temperature of 1400° to 1500°C in high-purity hydrogen (dewpoint below -60°F). The treatment resulted in a marked grain coarsening, individual grains becoming one to two mm in diameter and extending through the thickness of the test specimens. The anneal also resulted in formation of a scale on the bar surfaces, presumably aluminum oxide formed by reaction with water vapor or uncombined oxygen in the hydrogen. This scale consisted of a friable outer layer that was readily brushed off and an adherent subscale. The range of annealing temperatures used came about as a result of changes in practice made in seeking to eliminate the oxidizing nature of the hydrogen atmosphere: substitution of an iron for a molybdenum retort for better sealing, use of thoria rather than alumina plates to support the specimens, and the use of a zirconium hydride getter within the retort. With none of these modifications was a nonoxidizing anneal obtained.

All of these bend data, including both the hot-rolled and the annealed conditions, can be described with reference to two regimes of temperature. A lower temperature regime extending from room to 500°C is characterized by

TABLE 2
SUMMARY OF BEND TEST DATA
ROLLED NiAl, HOT-ROLLED CONDITION
(Heat No. 702--Atomic Composition: Ni 53, Al 47)

Test No.	Temp (°C)	Deflection Rate (in/min)	Plastic Deflection (inch)	Load (lb)	Flow Stress (psi)	Remarks
C5457	800	0.002	Small	3	5,700	Ductile
	800	.02	>0.01	13	24,500	Ductile
	800	.02	~ .1	15.7	--	Ductile
	800	.05	~ .12	18.7	--	Ductile
	800	.05	~ .3	22	--	Ductile
C5462	800	.05	.0035	8.8	22,800	Ductile
	800	.05	.35	13	--	Ductile
C5458	700	.05	.0035	21.5	42,000	Ductile
	700	.05	.0125	24.7	--	Fractured
C5459	700	.002	.0035	10.7	25,400	Ductile
	700	.002	.022	12.5	--	Ductile
	700	.02	.025	17	--	Ductile
	700	.02	.34	24.3	--	Ductile
C5461	650	.002	.0035	13.4	31,400	Ductile
	650	.002	.029	17.2	--	Ductile
	650	.02	.034	19.6	--	Fractured
C5460	600	.002	.0035	17.5	44,000	Ductile
	600	.002	.007	19.4	48,500	Fractured
C5737	650	.002	.0035	16	34,000	Ductile
(electro-	650	.002	.015	18.5	39,000	Ductile
polished)	650	.02	.0185	21.2	--	Ductile
	650	.02	.0285	28	--	Ductile
	650	.05	.032	32	--	Ductile
	650	.05	.054	35	--	Fractured
C5962	500	.002	.0035	21.3	69,000	Ductile
(electro-	400	.002	.005	22	71,400	Ductile
polished)	300	.002	.0075	23.5	76,200	Ductile
	200	.002	.0095	22	71,400	Fractured
A2626	200	.002	.0035	36.5	83,000	Ductile
(electro-	R.T.	.002	.0035	42	95,500	Ductile
polished)	R.T.	.002	.0155	53	--	Fractured

(One-inch test span)

NOTE: Specimens were tested in the surface ground condition excepting tests C5737, C5962, and A2626, which were electropolished. Specimen C5737 was polished only superficially while specimens C5962 and A2626 were polished until all microcracks were removed.

TABLE 3
SUMMARY OF BEND TEST DATA
ROLLED NiAl, ANNEALED CONDITION
(Heat No. 702--Atomic Composition: Ni 53, Al 47)

Test No.	Temp (°C)	Deflection Rate (in/min)	Plastic Deflection (inch)	Load (lb)	Flow Stress (psi)	Remarks
C5470	700	0.002	0.000	8	20,000	Elastic
	700	.02	.0035	17	42,500	Ductile
	700	.02	.011	18	--	Ductile
	700	.05	.0145	19	--	Ductile
	700	.05	.250	26.5	--	Ductile
C5472	650	.002	.0000	19.5	42,800	Elastic
	650	.02	.0035	23	50,500	Ductile
	650	.02	.047	27.5	--	Ductile
	650	.05	.05	29.5	--	Ductile
	650	.05	.275	35	--	Ductile
C5471	600	.002	.0035	25	54,500	Ductile
	600	.005	.0065	26.5	57,500	Ductile
	600	.02	.0085	28	61,000	Fractured
C5473	600	.002	.0035	27.5	68,800	Ductile
	600	.02	.0045	29.5	73,700	Ductile
	600	.02	.0265	35.5	--	Ductile
	600	.05	.03	38.5	--	Ductile
	600	.05	.205	43.0	--	Ductile
C5474	500	.002	.0035	40.2	78,000	Ductile
	500	.002	.006	41.5	80,500	Ductile
	500	.02	.0095	45.4	88,000	Ductile
	500	.02	.024	48.0	--	Fractured
C5475	550	.002	.0035	29.6	71,600	Ductile
	550	.002	.0056	31	75,000	Ductile
	550	.02	.009	34.5	83,500	Ductile
	550	.02	.042	40.5	--	Ductile
	550	.05	.0455	43.5	--	Ductile
	550	.05	.28	49.0	--	Ductile
C5672	600	.02	.0035	30	61,000	Fractured
C5673	650	.02	.001	27.5	54,000	Fractured
C5674	650	.002	.0035	24.7	51,000	Ductile
	650	.002	.014	25	51,500	Ductile
	650	.02	.0175	32.4	--	Ductile
	650	.02	.0285	32	--	Ductile
	600	.02	.044	42.5	--	Ductile
	600	.02	.054	42	--	Ductile
	550	.02	.069	49	--	Ductile
	550	.02	.072	49	--	Ductile
	500	.02	.084	52	--	Ductile
	500	.02	.087	54	--	Fractured

Table 3 (Continued)

Test No.	Temp (°C)	Deflection Rate (in/min)	Plastic Deflection (inch)	Load (lb)	Flow Stress (psi)	Remarks
C5675	650	0.002	0.0035	26.5	54, 500	Ductile
	650	.002	.007	28	57, 800	Ductile
	600	.002	.022	38.8	--	Ductile
	550	.002	.039	47.5	--	Ductile
	500	.002	.051	51.5	--	Fractured
C5677	650	.002	.0035	26	47, 500	Ductile
	650	.002	.0125	27	49, 400	Ductile
	650	.02	.016	32	58, 500	Ductile
	650	.02	.027	32.5	--	Ductile
	600	.02	.042	40.2	--	Ductile
	550	.02	.058	48.8	--	Ductile
	550	.02	.068	49.7	--	Fractured
C5700 (electro-polished)	650	.20	.0035	30.5	67, 000	Ductile
	650	.20	.37	42	--	Ductile
	600	.20	.40	46.6	--	Fractured
C5794 (electro-polished)	500	.002	.002	18.5	65, 300	Ductile
	450	.002	.003	19.8	70, 000	Ductile
	400	.002	.004	20.5	72, 400	Ductile
	350	.002	.0055	21.5	76, 000	Ductile
	300	.002	.008	23.2	82, 000	Ductile
	250	.002	.009	24.5	86, 100	Ductile
	200	.002	.011	25.7	90, 500	Ductile
	R. T.	.002	.013	31.5	99, 300	Ductile
	R. T.	.02	.013	32.5	102, 500	Ductile
	R. T.	.02	.016	34.1	107, 500	Fractured
C5795 (electro-polished)	R. T.	.002	.0035	27	85, 000	Ductile
	R. T.	.002	.0113	30.2	95, 200	Ductile
C5963 (electro-polished)	R. T.	.02	.0035	23.3	119, 000	Ductile
	R. T.	.02	.052	29.5	--	Fractured
C5964 (electro-polished)	R. T.	.2	.004	25.5	113, 000	Fractured
C5968 (electro-polished)	800	.002	.0035	4.8	24, 200	Ductile
	800	.02	.007	6.4	27, 800	Ductile
	800	.2	.0105	9.0	45, 500	Ductile

NOTE: Specimens of this table indicated as electropolished were polished with silicon carbide paper until the adherent scale resulting from annealing was removed. They were then electropolished lightly, removing 0.0005 to 0.001 inch in total, from each surface.

limited ductility, a sensitivity to the condition of the surface, and by a relatively high flow stress that exhibits little variation with temperature. The flow stress is strain sensitive, however, and load-deflection curves obtained at room temperature exhibit significant work-hardening effects. The high-temperature regime begins at about 500°C and continues to 800°C, which is the upper limit of temperature examined. This region is characterized by much greater ductility, reduced sensitivity to notches or other surface defects, and by a flow stress having a marked temperature dependence. The existence of two regions of differing temperature dependence is clearly seen in Fig. 6. In this figure, proof stress data points are shown only for bars in which a minimum amount of plastic deflection (0.0035 inch) was obtained. For tests in the lower temperature range, it was usually necessary to electropolish bars as described later in order to obtain such plastic behavior. It will be observed that the hot-rolled and annealed conditions are of comparable strength in the low-temperature region. Above 500°C, the strength of the hot-rolled material deteriorates more rapidly than that of the annealed.

Unpolished specimens of the material that had received the high-temperature (1400°) heat treatment when bent at temperatures below 700°C required straining at a low rate (0.002 in/min) until a small amount of plastic strain had been imposed. The rate of bending could then be increased and ductile behavior would continue. If the initial loading were at the higher rate, fracture occurred at the stress that caused the initial plastic deformation. A second characteristic feature of annealed specimens is that fractures were intergranular--both the initiation of fracture and its propagation were intergranular. However, when annealed test specimens were electropolished lightly, the fracture became transcrystalline both in origin and propagation, and the necessity for a low initial strain rate was eliminated (see tests C5700, C5963, and C5964). Electropolished bars exhibit a measurable, although limited, ductility at room temperature, and slip trace markings are evident on the polished surfaces of bent bars after room-temperature deformation.

The effects of electropolishing in this case appear to stem from the removal of a brittle layer at the surface. Whether this layer consists solely of the adherent oxide film, or includes some thickness of unoxidized metal that is enriched in oxygen, has not been determined. The electropolishing procedure consisted of mechanical polishing with fine silicon carbide paper until the scale was removed, following which the surface was lightly electropolished. Usually no more than 0.0005 inch was removed from the surface during polishing, but this is far too much to enable a distinction to be made between the oxide film and an oxygen-rich layer below the scale. Moreover, the notch effect of the oxidized surface layer is apparently marked where grain boundary surfaces intersect the external surface, since grain boundary fracture was the predominant mode. In view of the relative rates of diffusion through the lattice and along boundary surfaces, it is to be expected that oxidation and oxygen penetration would be greater along exposed boundary surfaces. The notches would

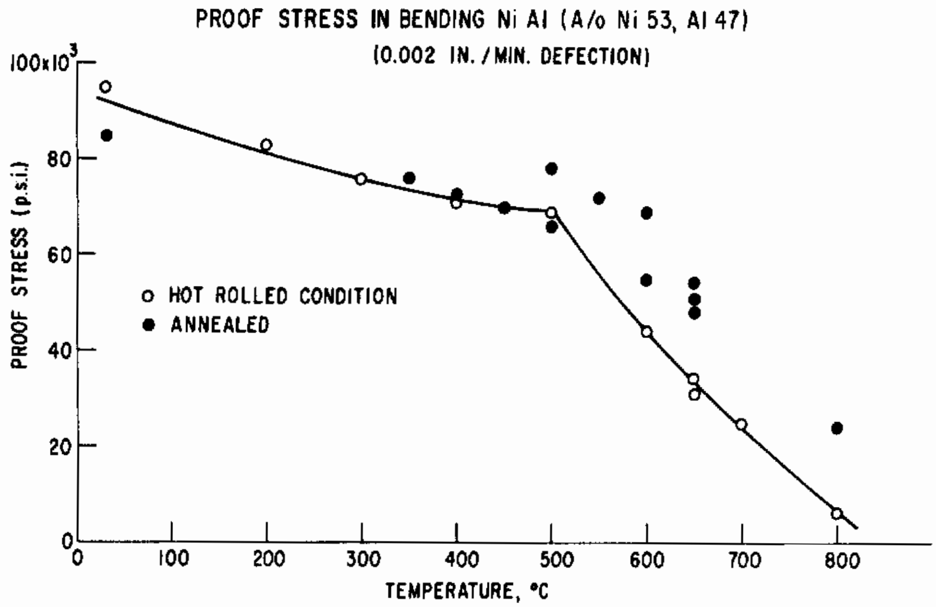


Figure 6. Proof Stress in Bending, Polycrystalline NiAl.

thus be deeper and more effective at the boundary. However, for the heat treatment used, the penetration was evidently limited to something less than 0.0005 inch.

Material in the hot-rolled condition exhibited another type of surface sensitivity. Bars that were subjected to bending in the ground condition were very ductile at 800°C, were ductile only at low strain rates at 700°C, and exhibited only limited ductility at 600°C. When the surfaces of these bend specimens were examined at 100 diameters magnification after a very light electropolish, numerous short cracks of random orientation were in evidence. By a sequence of electropolishing and inspection steps it was found that these cracks were entirely eliminated with the removal of 0.005 to 0.010 inch in thickness from the surface. The superficial nature of these cracks suggests that they are derived from the grinding operation. Bend specimens that were electropolished on the tension surface and edges so as to remove microcracks were found to exhibit limited ductility at room temperature, and slip trace markings were observed on the surface in the region of maximum bending moment. Fracture of all bend specimens tested in the hot-rolled condition, whether electropolished or not, were transcrystalline, the fracture being propagated along well-defined cleavage planes.

The results of electropolishing are less dramatic in the case of the hot-rolled condition in that no change in fracture mode is observed. However, plastic strain in tension is possible at room temperature for material in both the hot-rolled and annealed conditions in the absence of cracks or notches. NiAl is highly notch sensitive both in respect to fracture along grain boundary surfaces, as was seen for annealed bend specimens, and in respect to fracture by transcrystalline cleavage as is evident for bend specimens in the hot-rolled condition.

B. Calculation of the Elastic Modulus

Values of the elastic modulus were calculated from load-deflection plots for several of the more suitable bend tests. The best of these values is that derived from test C5795, which was loaded and unloaded through several cycles within the elastic range until a reproducible behavior was obtained. The deflection of the test machine was also measured carefully for the conditions of this test and subtracted from the total deflection, which was observed. Under the conditions of test, the machine deflection was smaller than the specimen deflection by a factor of about 50, so obtaining the specimen deflection as a difference between these two quantities does not involve a high degree of inherent error. For this test the modulus at room temperature had the value 24.5×10^6 psi, as an average of the values obtained during loading and unloading. Other values of the modulus were usually lower than this by 1 to 2×10^6 psi; however, in these cases the specimen was not cycled and local plastic deformation may have been included in the deflection measured. No significant change in the elastic modulus was observed between room temperature and 500°C. However, at 800°C a single determination under good conditions yielded an average value of 18×10^6 psi. These data may be compared with the room-temperature moduli of nickel, 30×10^6, and of aluminum, 9.9×10^6 psi.

C. Single-Crystal Bend Results

Another group of bend specimens consisted of single-crystal bars of cylindrical shape, nominally 0.18 inch in diameter by 2 inches long, obtained from the directionally frozen ingot previously described. The results of bend tests with these specimens are summarized in Table 4 and shown graphically in Fig. 7, which is another plot of proof stress (maximum fiber stress for less than 0.003-inch plastic deflection) vs temperature. Comparing these data with those for polycrystalline samples in the hot-rolled and annealed conditions, one notes the much lower level of stress necessary for plastic strain, and the absence of a temperature region where the stress has a strong temperature dependence. Although the data are extended to 1000°C in the single test C5967, there is no apparent discontinuity in the proof stress-temperature characteristic. Whether this difference is correctly ascribed to the shift in composition toward the stoichiometric proportions, or, as appears to be more probable, to the absence of grain boundaries and the constraint of neighboring grains of dissimilar orientation, remains to be determined. Although only qualitative measurements can be drawn upon, it is also clear that the ductility of these single-crystal rods between room temperature and 600°C is very much greater than that of the hot-rolled or hot-rolled and annealed polycrystalline materials.

D. Tensile Tests

A small number of tensile specimens were prepared according to Fig. 5 using an Inconel-clad sheet of hot-rolled NiAl. Removal of the cladding from the test section of the specimen is necessarily a difficult operation, and to avoid an unduly high loss by breakage it was found to be necessary to leave several mils of the

TABLE 4
SUMMARY OF BEND TEST DATA
SINGLE-CRYSTAL RODS, 0.18-INCH DIAMETER
(Composition: A/o Ni 52, Al 48)

Test No.	Temp (°C)	Deflection Rate (in/min)	Plastic Deflection (inch)	Load (lb)	Proof Stress (psi)	Remarks
C5701	640	0.02	0.002	13.5	9,500	Ductile
	640	.02	.016	17	12,000	Ductile
	640	.002	.017	17	12,000	Ductile
	610	.002	.022	16.5	11,800	Ductile
	530	.002	.035	17.5	--	Ductile
	500	.002	.045	18.5	--	Ductile
	450	.002	.055	20	--	Ductile
	400	.002	.070	22	--	Ductile
	350	.002	.085	24.5	--	Ductile
	300	.002	.095	27	--	Ductile
	230	.002	.110	33	--	Ductile
	230	.02	.112	37	--	Ductile
	230	.02	.132	44	--	Ductile
	200	.02	.140	46	--	Ductile
	200	.02	.150	49	--	Ductile
	110	.02	.170	56	--	Ductile
	110	.02	.190	67	--	Ductile
	60	.02	.210	76	--	Ductile
C5702	235	.002	.001	41.5	18,300	Ductile
	235	.02	.002	46	22,000	Ductile
	235	.02	.021	73	--	Ductile
	R.T.	.002	.030	120	--	Ductile
	R.T.	.002	.032	137	--	Ductile
	R.T.	.02	.032	144	--	Ductile
	R.T.	.02	.054	180	--	Ductile
		(Tested over a 1-inch span)				
A2625	345	.002	.001	27	13,300	Ductile
	250	.002	.003	31.5	15,500	Ductile
	R.T.	.002	.004	46	22,600	Ductile
	R.T.	.02	.004	48.5	23,800	Ductile
	R.T.	.02	.009	52.5	25,800	Fractured at flaw
		(Tested over a 1-inch span)				
A2628	R.T.	.02	.001	83	25,500	Ductile
	R.T.	.02	.0015	85.5	26,400	Fractured
		(Tested over a 5/8-inch span)				
C5967	R.T.	.002	.0004	48.7	34,100	Ductile
	300	.002	.0008	25.7	18,000	Ductile
	475	.002	.0014	19.5	13,600	Ductile
	700	.002	.0021	22.2	15,500	Ductile
	800	.002	.0026	13.7	9,600	Ductile
	900	.002	.0031	15.2	10,600	Ductile
	1000	.002	.0036	12.0	8,400	Ductile

NOTE: Tests A2625 and A2628 employed a single specimen that was rolled in five passes from a cylinder of 0.177-inch diameter to a strip: 0.132 inch thick by 0.175 inch wide, a reduction of 26%. Between roll passes the piece was reheated to 500°C, but the rolling temperature is estimated to be about 200°C because of loss of heat to the cold rolls. Following rolling, the strip was annealed 1 hour at 1000°C in hydrogen, but no recrystallization could be detected. Test A2628 utilized one-half of the broken specimen of test A2625.

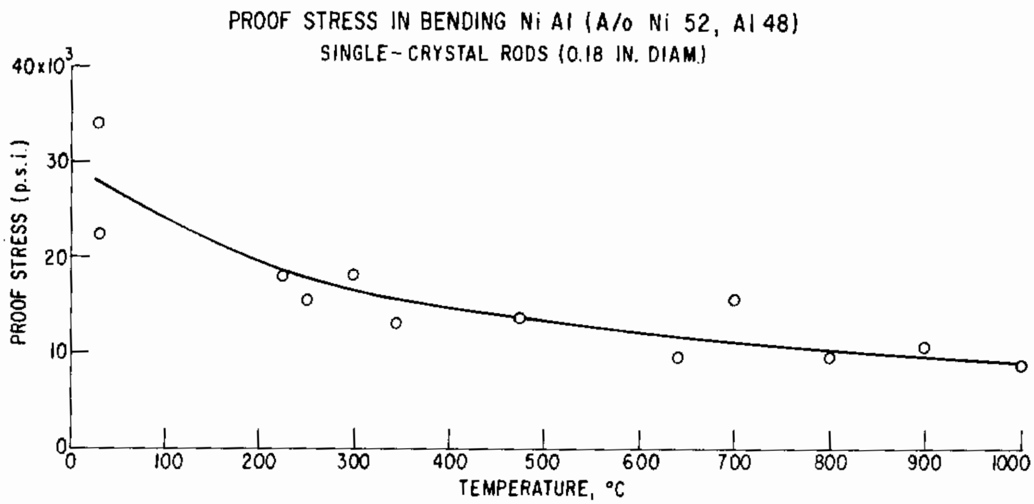


Figure 7. Proof Stress in Bending, Single-Crystal NiAl.

cladding remaining on the NiAl surface. In view of subsequent experience showing the susceptibility of NiAl to grinding cracks, this experience is not now surprising. Therefore, tensile tests were carried out with a small thickness of the cladding remaining and the full thickness of the section was used in calculating stress values. However, since the most important feature of these data is the ductile-brittle transition temperature, this is not an important departure from the intended practice.

Table 5 is a summary of the tensile data, which are also shown in a graphical representation in Fig. 8. The ductility in a test at an extension rate of 0.02 in/min changes rapidly between 700° and 800°C. Above 800°C the ductility continues to increase, but at a reduced rate. Over this same range of temperature the 0.2% proof stress also changes rapidly although in an inverse manner. These data are found to be in good agreement with bend data, for a deflection rate of 0.05 in/min (see Table 2, tests C5458 and C5462).

IV. DETERMINATION OF ACTIVE GLIDE PLANES IN NiAl

The observance of slip traces on the electropolished surface of bend specimens, which had been deformed at room temperature, raised the possibility of determining the active glide planes for NiAl. For this purpose bend bar C5795 was selected from the group that had received a high-temperature anneal. Grain growth during annealing resulted in grains that occupied the thickness of the specimen (about 0.06 inch) and were of the same dimension as the bar thickness in the plane of the surface. X-ray Laue grams were taken of 11 grains in the region of the bar where bending was to occur. A stereographic projection was prepared of each grain. The bar was then bent sufficiently to provide an abundance of slip traces on the surface of the x-rayed grains. The tension side of the bar was then examined at 200 diameters magnification using an angle-measuring eyepiece to determine the angle between observed traces and a

TABLE 5
SUMMARY OF TENSILE DATA

Test No.	Temp (°C)	Strain Rate (in/min)	0.2% Proof Stress (psi)	Ultimate Strength (psi)	Elongation at Fracture (%)
C5764	800	0.02	28,200	32,000	--
	800	.05	--	34,000	--
	700	.02	--	43,500	--
	700	.05	--	43,500	26
C5765	650	.02	--	52,300	--
C5806	900	.02	12,200	14,200	42
C5808	1000	.02	8,400	9,450	53
C5810	800	.02	24,600	25,400	32
C5811	700	.02	46,400	47,500	2

TENSILE PROPERTIES OF HOT-ROLLED Ni Al A/o Ni 53, Al 47
(0.02 IN./MIN. EXTENSION RATE)

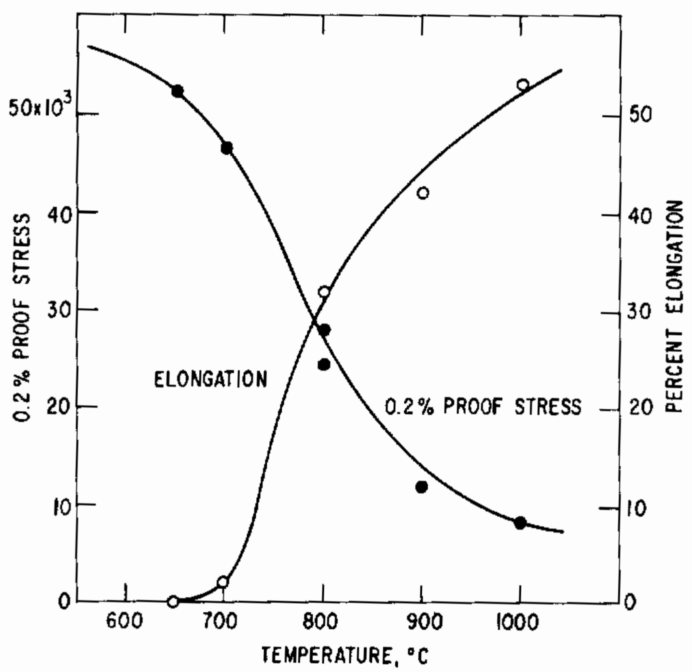


Figure 8. Tensile Properties of NiAl.

reference direction on the specimen. In a few grains only a single trace direction was found. The more usual experience was to find two prominent systems of traces. In one grain as many as five trace directions occurred.

The normals to each of the observed trace directions were plotted on their respective stereographic projections. The pole of the glide plane causing the trace must then lie somewhere along the normal. An analysis of the 11 diagrams led to the conclusion that all of the traces that had been observed could be accounted for if slip were assumed to occur on planes of the type: (110), (112), and (123), which are the glide planes found for other body-centered cubic crystals including AgMg as previously reported.[2] In one instance the normal to a trace direction when plotted on the stereographic projection intersected only the poles of planes of a [111] zone line family, that is, planes of the {110}, {112}, {123} type. In three other instances the normal lay very near to this position. Therefore, in these instances we know, without ambiguity, that glide occurred on one of these planes. In all other instances the pole of other low index planes also occurred on the trace normal, and the interpretation is indefinite.

V. STUDIES ON AgMg

During the current period a study has been made to determine the extent to which measured transition temperatures of AgMg may be affected by the presence of differing amounts of grain boundary hardening. A study has also been made of the effects of dilute ternary solute addition on the tensile behavior of AgMg compounds at low temperatures (less than room temperature) and observations have been made concerning the ductility of these ternary compounds at low temperatures.

A. Transition Temperature Studies

The effects of material and process variables on the ductility of a brittle material are usually measured in terms of a ductile-brittle transition temperature. The novel technique that has been developed for such measurements was described in a previous report[4] together with a study made of the effects of metallurgical variables on the transition temperature of AgMg. Among the results obtained at that time was the finding that the transition temperature depended strongly on the content of excess magnesium in the compound. The interpretation of this result was somewhat ambiguous in that the grain boundary hardening present in the compositions tested was not constant.

B. Effect of Grain Boundary Hardening

It was necessary, therefore, to clarify the true effect of stoichiometry on the transition temperature to first evaluate the effect of grain boundary hardening on the measured transition temperature. This was accomplished by producing various amounts of grain boundary hardening by annealing a given composition (Ag + 51.0 A/o Mg) in atmospheres of various purities. The atmospheres used were (a) argon with no getter; (b) argon with La_3Mg getter; and (c) vacuum with

La_3Mg getter. The amounts of grain boundary hardening, as measured at room temperature, both before and after testing, together with the measured transition temperatures are shown in Table 6.

TABLE 6
TRANSITION TEMPERATURES OF Ag + 51.0 A/o Mg SPECIMENS HAVING DIFFERENT AMOUNTS OF GRAIN BOUNDARY HARDENING AT ROOM TEMPERATURE

ΔH/H (%) at R.T. Prior to Testing	Transition Temp (°C)	ΔH/H (%) at R.T. After Testing
5.5	81	8.0
11.0	79	9.0
26.0	79	15.5

The data in Table 6 show that there is no shift in transition temperature with increased amounts of grain boundary hardening beyond about 6% to 8%. This is not to say, however, that the transition temperature would be insensitive to variation in amount of grain boundary hardening at very low levels (a few per cent). Of the materials tested previously, those susceptible to grain boundary hardening invariably exhibited hardening in excess of 5%. The conclusion can be drawn, therefore, that the results obtained previously on the effects of stoichiometric deviation on transition temperature are valid and reflect characteristic behavior rather than variations in grain boundary hardening.

C. Effect of Cooling Rate on Transition Temperature

With regard to the continuous bend test determination of transition temperature, the question has also arisen: Is the specimen in fact at the temperature indicated by the thermocouple? (Although the couple is submerged in the surrounding oil, it is not in physical contact with the specimen.) Furthermore, since grain boundary hardening is a temperature-dependent phenomenon, it is of interest to determine the effect of rate of cooling during the test on the measured transition temperature. The answers to both these questions are provided by the following experimental results. A specimen of Ag + 51.0 A/o Mg tested by cooling from a temperature of 140°C at a rate of 3.5 deg/min exhibited a transition temperature of 98°C. An identical specimen tested by cooling from 140°C at a rate of 14 deg/min showed a transition temperature of 100°C. The fact that the same transition temperature is obtained under very different cooling conditions indicates that (a) the specimen is, indeed, at the temperature indicated by the thermocouple; and (b) there is no error due to variations of the status of the grain boundary hardening that might accompany variations in the thermal treatment imparted by the range of cooling rates available in the test equipment. This latter conclusion is in agreement with the aforementioned observation that the transition temperature is insensitive to variations in grain boundary hardening above a level of 6% to 8%.

D. Ternary Studies

The last summary report(4) described observations on the effects of dilute ternary additions on both the flow stress and transition temperature. Further observations have been made concerning the low-temperature behavior of these ternary compounds.

Interest in tabulating the values of the flow stress at low temperature arose primarily from qualitative observation that specimens of some of the ternary compounds, although they could not be plastically deformed at room temperature, exhibited considerable ductility at liquid N_2 temperature. Outstanding in this respect was alloy No. 13, a 50.7 A/o Mg binary base containing 1.0 A/o Zn. This type of mechanical behavior may be of significant practical importance in the fabrication of such materials.

Table 7 shows the values of the flow stress of various ternary compounds, with the flow stress for some binary compounds included for comparison. It is to be noted in the table that although a binary Mg-rich compound of 50.3 A/o Mg (No. 7) cannot be tested below about 200°C, a binary base of 50.7 A/o Mg containing dilute additions of either Zn (No. 13) or Cu (No. 29) can be tested as low as liquid N_2 temperature. Furthermore, in contrast to normal behavior, the flow stresses of these ternaries at temperatures below about 0°C increase with increasing temperature.

TABLE 7
VALUES OF THE FLOW STRESS AT VARIOUS TEMPERATURES
OF AgMg COMPOUNDS CONTAINING DILUTE TERNARY ADDITIONS

	Binary	Ternary	Flow Stress (1000 psi)					
Alloy No.	Base A/o Mg	Solute A/o	-195°C	-78°C	0°C	R.T.	100°C	200°C
13	50.7	1.0 Zn	22.0	24.4	25.0	24.4	20.4	16.0
38	49.8	1.0 Zn	31.4	30.6	31.6	30.0	26.0	20.8
29	50.7	1.1 Cu	25.6	28.0	26.8	24.8	21.6	16.8
23	50.5	1.2 Al	X	36.2	--	33.3	33.0	27.8
35	48.3	1.3 Al	41.8	40.8	37.2	34.8	33.6	26.7
36	49.3	2.5 Al	X	X	X	32.4	32.4	32.8
39	50.1	--	27.5	24.0	22.0	16.5	15.2	13.7
7	50.3	--	X	X	X	X	X	18.5
4	48.0	--	48.4	48.0	--	32.4	30.5	17.0
6	49.8	--	28.0	20.4	--	17.1	14.4	9.4

X denotes too brittle to test.

Associated with the somewhat anomalous behavior of ternary compounds at low temperatures is a noteworthy feature in the stress-strain curves. Binary compounds were found in previous studies to exhibit rather normal stress-strain curves. Under appropriate conditions, typical yield points were observed in the binary compounds, but with almost negligible lower yield strain prior to the increase in stress due to strain hardening. On the other hand, the ternary compounds, both Ag-rich and Mg-rich base, exhibit gross lower yield strain (of the order of a few per cent). Extreme jerky flow occurs during the lower yield strain. The extent of the lower yield strain decreases with increasing temperature becoming nearly zero at temperature around room temperature.

VI. CONCLUSIONS

The following conclusions pertain to the intermetallic phase, NiAl, of a composition slightly nickel rich with respect to the stoichiometric composition.

1. The NiAl phase can be hot-worked by rolling in the temperature range 1050° to 1200°C while enclosed in a heavy-walled steel enclosure. This experience does not preclude the possibility of rolling without a container, but surface temperature must surely be kept above 1000°C during hot-rolling.

2. Cold-rolling of polycrystalline NiAl is made difficult by the limited ductility of the hot-rolled product below 800°C and by the rapid loss of strain hardening, and ultimate recrystallization, at 800°C and above. Cold-working must be undertaken above 800°C, but subsequent reduction steps must probably be at lower temperature as ductility considerations permit. Large reductions in the initial cold pass are necessary to avoid transverse rupture, and this requirement may also extend to subsequent passes.

3. Bend tests and tensile test data show that below 600°C polycrystalline NiAl has very limited ductility, is sensitive to surface notches or other stress raisers, and the flow stress is relatively invariant with temperature. Above 600°C, NiAl becomes a ductile metal, notch-sensitivity vanishes, and the flow stress becomes highly variant with temperature.

4. The elastic modulus of NiAl at room temperature is 24.5×10^6 psi. At 800°C, the modulus is about 18×10^6 psi.

5. Slip at room temperature in NiAl occurs on planes of the [111] zone family; that is, planes of the type {110}, {112}, and {123}. This eliminates the hypothesis that the limited ductility of NiAl at room temperature is

understandable from the Taylor principle of an insufficient number of slip systems.

The following conclusions pertain to AgMg:

6. Transition temperature increases with increasing magnesium content in excess of stoichiometry, independent of possible concomitant increases in grain boundary hardening.

7. The transition temperature of a given composition is independent of such variations of cooling rate as are experienced in the application of the continuous bend test determination of transition temperature.

8. Ternary solute additions can impart ductility at low temperatures to otherwise brittle Mg-rich bases.

VII. OTHER ACTIVITIES

Research under this contract received national recognition through two awards during the past year. A paper, "The Tensile Behavior of the Intermetallic Compound AgMg," was selected by the New England Section of the AIME as the best paper emanating from that section appearing in the 1962 Transactions of that Society (vol. 224, p. 1024). That paper was based on material contained in Summary Report, Part II.[2] Another paper, "An Apparatus for the Direct Determination of Ductile-Brittle Transition Temperatures," presented at the June 1963 meeting of the American Society for Testing and Materials, was cited as one of the three most effective and meritorious presentations at that meeting. This novel technique, developed under this contract, was reported in Summary Report, Part IV.[4]

VIII. RECOMMENDATIONS FOR FUTURE WORK

1. Document the effect of prestrain on the ductile-brittle transition temperature and examine the origin of the effect.

2. Explore other means of improving the low-temperature ductility of NiAl, e.g., through crystallographic transformation.

3. Study the mechanical behavior of a simple intermetallic without the complication of grain boundary hardening.

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