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HIGH INTEGRITY CASTINGS

Edited by
C. V. White, D. Eylon, F. H. Froes

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High Integrity Castings

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Edited by
C. V. White, D. Eylon, F. H. Froes

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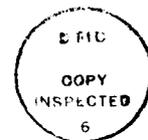
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FOREWORD

The casting industry has historically been the producer of high volume durable cast products to meet the demand of a growing industrial community. Since the end of the Second World War, the development of new cast metals and technology have helped to alter the industrial demand to a more sophisticated product requiring higher strength, greater toughness and closer tolerances.

As an example, the development of ductile cast iron in 1948 has lead to dramatic changes in both the casting technique and the design philosophy in the use of cast irons. The previous concept of a high compressive and low tensile strength material has evolved into ADI, a high strength, tough competitor to wrought products with the design flexibility of a casting.

Solidification models to aid in decreasing the lead time from concept to production are being implemented in foundries along with advanced process control techniques to improve quality and decrease costs.

In titanium alloy castings, the development of extremely fine alpha precipitate in a beta matrix has greatly improved the strength to density applicaitons in air craft and space applications. Improvements in magnesium ailloys in both strength and corrosion resistance have lead to the increased use of light weight magnesium in a wide variety of structural applications, and because of its good wetting characteristics, it has found application in metal matrix composites.

Melting technology using induction furnaces have greatly improved the metal quality and significantly decreased the production cost and time. New techniques of induction melting of magnesium have lead to improved safety and environmental conditions with a cost savings. The development of the monolithic ceramic shells in investment casting has lead to significant reduction in casting size limitation.

This symposium is organized to illustrate the level of technical sophistication and the advances in technology which has brought the casting industry up to a highly comparative level in the production of parts for today's industry.

Our thanks to the authors and speakers in this symposium for their patience and hard work in making this symposium come together. A special thanks is also due to Denise Loeffler for her secretarial assistance in the preparation of this compendium and to the Conference staff of ASM INTERNATIONAL for their participation in the development of this conference and the publication of this proceedings.

C. V. White,
D. Eylon,
F. H. Froes

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Titanium

IMPROVEMENT OF HCF PROPERTIES IN Ti-6Al-4V CASTINGS VIA THERMOCHEMICAL TREATMENT

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ABSTRACT

Two advanced thermal processing methods were applied to investment cast and hot isostatically pressed Ti-6Al-4V alloy in an effort to improve selected mechanical properties. The microstructure, room temperature tensile, and high cycle fatigue behavior of both beta solution treated (BST) and constitutionally solution treated (CST®)* castings were compared to those of conventionally annealed material. The CST® process is a thermochemical treatment (TCT) developed by Howmet Corporation to transform and refine the microstructure of Ti-6Al-4V castings through temporary alloying with hydrogen. The results revealed that both the BST and CST® processes produce substantial tensile and high cycle fatigue improvements over conventionally annealed Ti-6Al-4V. Moreover, CST® processed material exhibits a fatigue endurance limit in excess of levels reported for wrought titanium alloys. The general benefits of microstructural modification of cast titanium alloys are discussed.

TODAY'S AIRCRAFT AND ENGINE DESIGNERS are increasingly being challenged with building higher performance aircraft at lower costs than ever before. Titanium, with excellent specific strength, is playing an important role in new designs. As advancements are made in important areas such as extractive metallurgy (1) and melting techniques (2-6), both the quality and cost benefits of titanium improve. One of the most significant advancements has been in titanium net-shape technologies (7) leading to reduced material waste and costly machining (8). The most cost effective and widespread titanium net-shape technology is investment casting (9).

*CST® is a registered trade name of Howmet Corporation

Titanium alloy castings have long been accepted for chemical and marine applications due to their good general corrosion properties (10). Also many cast titanium static aerospace components have been introduced due to their high strength-to-weight ratio and desirable fracture and creep resistance properties (11). With the push to improve performance, while lowering costs, titanium alloy investment castings are currently being evaluated for dynamic applications where properties such as fatigue crack initiation resistance are critical. Cast titanium traditionally has been restricted to static applications due to unfavorable microstructural features and associated properties. The microstructure of a typical investment cast titanium alloy such as Ti-6Al-4V is a product of slow cooling from very high temperatures, through the beta transus, to room temperature. As a result, it consists of coarse prior beta grains with boundaries lined with continuous alpha and large alpha platelet colonies or packets in the intragranular regions. It has been demonstrated that both these microstructural features are deleterious to fatigue properties through the premature initiation of fatigue cracks and preferred propagation routes (12-13). By using hot isostatic pressing (HIP), detrimental defects such as gas and shrinkage porosity are eliminated (14).

Wrought titanium alloys generally exhibit better tensile ductility and fatigue properties than cast alloys of the same composition (15). This is due to the fine equiaxed alpha grain morphology achievable through thermomechanical processing. Since net-shapes like investment castings cannot be hot worked to modify the microstructure and properties, many novel heat treatments, both conventional and thermochemical, have been developed (16-21).

One form of microstructural modification can be performed via beta solution treatment (BST) (21) typically in a vacuum furnace at a tempera-

ture slightly above the beta transus. Castings are then gas fan cooled at the fastest rate possible to promote transformation to a fine alpha plate structure. The solution treatment is followed by an age to complete reprecipitation of a fine secondary alpha phase and stabilize the structure. The BST treatment has been shown to improve mechanical properties (21).

In an effort to improve the fatigue properties of titanium investment castings to compete with wrought titanium in dynamic applications, Howmet Corporation has developed a thermochemical process (22) known as Constitutional Solution Treatment, or CST[®]. Hydrogen is introduced reversibly at elevated temperature to manipulate the solutioning and reprecipitation of the alpha phase. Since hydrogen is a strong beta stabilizer, the beta transus of the alloy is reduced during the hydrogenation process, allowing for beta heat treatment at a temperature low enough to prevent grain growth. The hydrogen is then removed to low allowable levels in a vacuum annealing cycle. The resulting microstructure is characterized by fine, Widmanstätten alpha platelets in prior beta grains and the absence of coarse continuous grain boundary alpha.

The objective of the present work was to characterize and compare the microstructure and tensile and fatigue properties of investment cast and HIP'ed Ti-6Al-4V given the BST and CST[®] treatments relative to the baseline conventional anneal treatment.

EXPERIMENTAL PROCEDURES

MATERIAL - The test material was investment cast at Howmet's Whitehall Titanium Casting Unit. A Ti-6Al-4V standard grade alloy heat, TA 276, was used to cast oversize test bar (OSB) blanks that were 16mm (0.63 in.) in diameter by 125mm (5 in.) long. Conventional vacuum consumable arc melting practice was used to melt the alloy billet. The alloy billet was consumable arc remelted and centrifugally cast into a Mono-Shell[®] ceramic mold. After knock-out and cleaning, all of the OSB's were hot isostatically pressed (HIP'ed) at 900°C (1650°F) and 105MPa (15 ksi) for two hours. A typical chemical analysis of a TA 276 OSB is shown in Table I.

HEAT TREATMENT - The HIP'ed Ti-6Al-4V OSB castings were divided into three groups subjected to a conventional anneal, or a beta solution treatment (BST), or a Constitutional Solution Treatment (CST[®]), Table II. The two vacuum heat treatments were performed in an Ipsen VFC-924 vacuum furnace; the CST[®] cycle was performed in Howmet's pilot production CST[®] furnace.

The conventionally annealed bars were heat treated at 845°C (1550°F) for two hours, and gas fan cooled in argon gas with a pressure just below one atmosphere maintained throughout the

Table I - Chemistry of Ti-6Al-4V investment castings for this work (Heat No. TA 276)

	Ti (wt%)	Al (wt%)	V (wt%)	Fe (wt%)	Mo (wt%)	Sn (wt%)	Cu (wt%)	Y (ppm)	C (wt%)	O (wt%)	N (wt%)	H (wt%)
TA276	Bal	6.12	1.88	0.20	<0.02	<0.02	<0.02	<10	0.031	0.177	0.010	0.0009
Ti-6Al-4V specification*	Bal	6.75 (max) 5.5 (min)	4.5 (max) 3.5 (min)	0.30	0.10	0.10	0.10	50	0.10	0.200	0.05	0.0150

* per AMS 4991 "Titanium Alloy Castings, Investment 6Al-4V, Annealed"

Table II - Heat treatments for HIP'ed Ti-6Al-4V OSB cast test material

Designation	Parameters
Annealed	845°C (1550°F) / 2 hr / GFC
Beta Solution Treatment (BST)	1025°C (1880°F) / 30 min / GFC in He* + 620°C (1150°F) / 2 hr / AC
Constitutional Solution Treatment (CST [®])	Proprietary to Howmet

* Average cooling rate of 135°C (245°F) per minute from 1025°C (1880°F) to 480°C (900°F)

Notes HIP'ed at 900°C (1650°F) / 105 MPa (15 KSI) / 2 hr
All temperatures listed are ± 14°C (25°F)

cooling cycle. The BST treated bars were heated in vacuum to 1025°C (1880°F) for 30 minutes followed by a gas fan cool with helium gas at a pressure just below one atmosphere. The control thermocouple indicated cooling from 1025°C (1880°F) to 480°C (900°F) in four minutes for an average cooling rate of 120°C (245°F) per minute within that temperature range. The same bars were then heated in vacuum to 620°C (1150°F) and held at this soak temperature for two hours and gas fan cooled in argon gas. CST[®] processing was performed on a third group of OSB's. The production CST[®] cycle was used which was developed specifically for optimizing HCF properties and tensile ductility in standard grade Ti-6Al-4V castings (22).

MECHANICAL TESTING - OSB's in the three heat treated conditions were subjected to optical metallography, room temperature tensile testing, and high cycle fatigue (HCF) testing. The OSB's were machined into tensile and HCF bars with smooth, constant diameter gage sections.

Testing of the HCF bars was performed on servohydraulic MTS equipment at the US Air Force Wright Aeronautical Laboratories, Materials Laboratory under axial loading at room temperature with a triangular waveform, R (minimum/maximum load) = 0.1, and frequency = 10Hz.

RESULTS & DISCUSSION

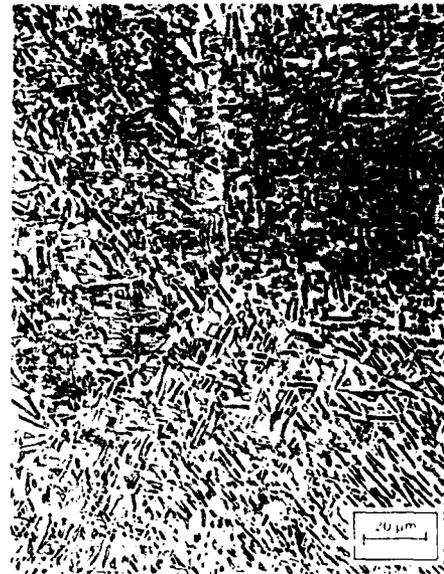
MICROSTRUCTURE - Microstructures of the annealed, BST, and CST[®] conditions are shown in Figure 1. The microstructure in the annealed bars typically consisted of relatively coarse alpha platelets arranged in colonies with prior beta grain boundaries outlined by continuous alpha phase. The BST structure was much finer than the annealed structure and comprised fine



Annealed



BST



CST®

Fig. 1 - Photomicrographs showing typical microstructures in Ti-6Al-4V bars subjected to three heat treatments: Annealed, BST and CST®. The CST® processed material has the finest structure and essentially no grain boundary alpha, 600X magnification, 2% HF etch.

alpha platelets in a basketweave or colony morphology. Some prior beta grain boundaries were outlined with alpha phase. The CST® structure was the finest of the three processes, with the alpha platelets being less lenticular than in the other conditions and arranged predominantly in a fine Widmanstätten morphology. The CST® structure was essentially free of grain boundary alpha.

TENSILE RESULTS - The room temperature tensile results, three tests per condition, are presented in Table III. The BST treatment resulted in 8% increase in average tensile strength over the annealed condition with the same average tensile elongation. The CST® treatment resulted in 13% increase in average tensile strength over the annealed condition with the same average tensile elongation.

FATIGUE RESULTS - The high cycle fatigue results are listed in Table IV and shown in Figures 2-5. Both the BST and CST® heat treatments yielded significant improvement in HCF behavior over that of the annealed OSB's. Comparison with AFML data for wrought annealed material showed that the CST® results are equivalent or superior in HCF. The CST® treatment resulted in the highest HCF strength of about 825 MPa (120 ksi) at 3×10^5 cycles. The BST treated OSB's yielded a fatigue strength of about 620 MPa (90 ksi) compared to about 550 MPa (80 ksi) for the annealed castings at 3×10^5 cycles. The improvement in tensile and fatigue properties can be attributed to refinement of the alpha platelets, modification of platelet morphology from colony to basket-weave (Widmanstätten), and reduction or elimination of the grain boundary alpha phase. Some of the ob-

Table III - Room temperature high cycle fatigue results for Ti-6Al-4V cast test bars in various heat treat conditions

Condition	Specimen No	0.2% YS		UTS		Elongation (%)	R of A (%)	E x 10 ³
		MPa	KSI	MPa	KSI			
Annealed ↓	CHHA - 1*	-	-	889	129	-	22	-
	CHHA - 2	841	122	917	133	10	19	18
	CHHA - 3	862	125	945	137	7	22	18
	Average	855	124	931	135	9	21	18
BST ↓	CHHT - 1	883	128	979	142	10	20	18
	CHHT - 2	910	132	1000	145	9	16	17
	CHHT - 3	917	133	1007	146	9	15	17
	Average	903	131	993	144	9	17	17
CST® ↓	CHST - 1	1000	145	1076	156	9	15	17
	CHST - 2	1014	147	1062	154	8	14	17
	CHST - 3	938	136	1034	150	10	16	17
	Average	986	143	1055	153	9	15	17

* Extensometer slipped during testing (results not used in average)

served improvement in fatigue properties also may have been related to the improvement in tensile strength. To separate the effect of microstructure from the increase in tensile strength, the ratio of fatigue strength at 3×10^5 cycles to tensile strength was calculated (Table V). The annealed condition has a similar ratio to the BST condition. The CST® process yielded a ratio as high as 0.78 which is an indication that the improvement in fatigue strength was predominantly the result of microstructural refinement.

Table IV - Room temperature tensile test results for Ti-6Al-4V cast test bars in various heat treat conditions (10 Hz, R = 0.1, Triangular Wave Form, $k_T = 1.0$)

Condition	MAXIMUM STRESS		Cycles to Failure
	MPa	KSI	
Annealed	724	105 0	72,100
	620	90 0	1,038,900
	603	87 5	1,034,300
	586	85 0	928,200
	579	84 0	1,426,200
	567	82 3	1,098,200
	565	82 0	1,413,400
	558	81 0	963,900
	558	81 0	433,300
	552	80 0	2,924,100
BST	793	115 0	31,400
	758	110 0	87,500
	724	105 0	229,500
	703	102 0	700,000
	690	100 0	870,200
	676	98 0	508,500
	672	97 5	639,600
	655	95 0	1,819,500
	638	92 5	5,500,000
	621	90 0	3,650,000 R.O
	621	90 0	2,162,200
	607	88 0	2,557,400
CST*	965	140 0	22,100
	931	135 0	46,200
	896	130 0	56,800
	896	130 0	13,200
	876	127 0	76,800
	869	126 0	804,200
	862	125 0	1,568,800
	855	124 0	2,724,700
	842	122 0	1,863,500
	834	121 0	3,049,300
827	120 0	2,672,000	
690	100 0	4,900,000 R.O	

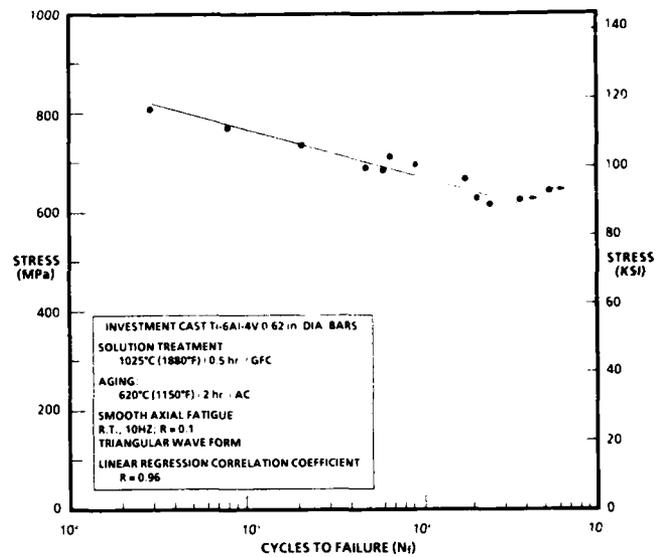


Fig. 3 - HCF data for investment cast Ti-6Al-4V test bars in the BST condition

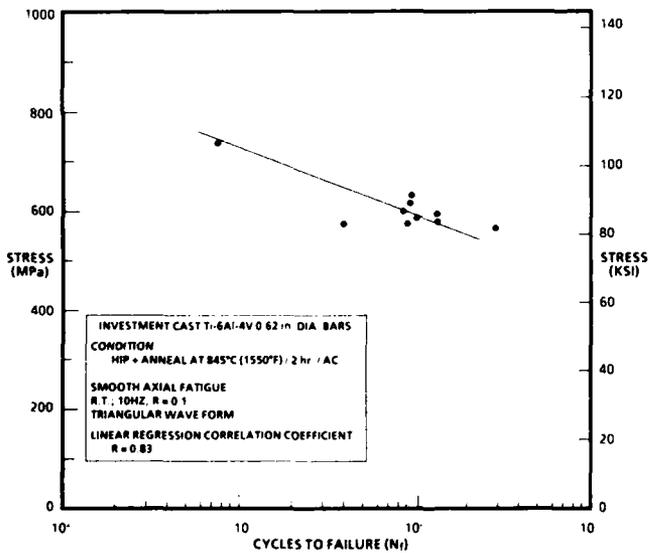


Fig. 2 - HCF data for annealed investment cast Ti-6Al-4V test bars

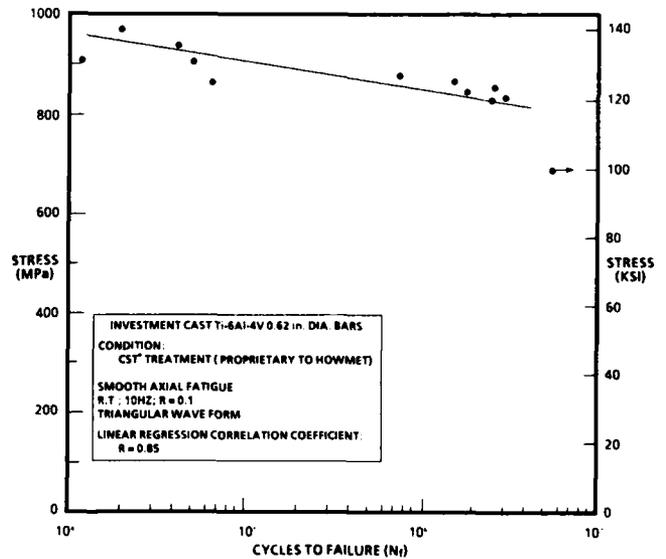


Fig. 4 - HCF data for investment cast Ti-6Al-4V test bars in the CST* condition

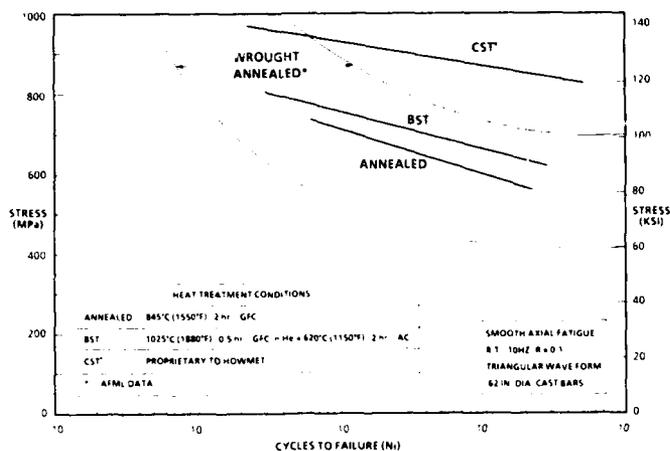


Fig. 5 - Comparison of high cycle fatigue properties of annealed, BST- and CST[®]-processed cast Ti-6Al-4V with Wrought Alloy

Table V - Comparison of fatigue strength vs. tensile strength for Ti-6Al-4V cast test bars with various heat treatments

Heat Treatment	σ_{UTS} MPa (KSI)	$\sigma_f(3 \times 10^6)$ MPa (KSI)	Fatigue Ratio $\left(\frac{\sigma_f(3 \times 10^6)}{\sigma_{UTS}} \right)$
Annealed	931 (135)	545 (79)	0.59
BST	993 (144)	607 (88)	0.61
CST [®]	1055 (153)	827 (120)	0.78

SUMMARY

A study was conducted to examine the effect of two advanced heat treatments on the microstructures of investment cast and HIP'ed Ti-6Al-4V and to correlate those microstructures with the room temperature tensile and high cycle fatigue characteristics of the test material. A beta solution treatment (BST) and the Howmet CST[®] process were compared to a 845°C (1550°F) anneal. Both the BST and CST[®] treatments produced improved tensile and fatigue behavior of test bars compared to annealed bars as a consequence of microstructural refinement, with the CST[®] processed bars demonstrating HCF strengths at 3×10^6 cycles in excess of those nominally achieved in wrought Ti-6Al-4V.

CONCLUSIONS

The Howmet CST[®] process exhibited a 275 MPa (40 ksi) improvement in HCF strength over the baseline annealed Ti-6Al-4V cast bars. The BST treatment resulted in a smaller 70MPa (10 ksi) improvement over the annealed bars. The improved HCF performance of the CST[®] processed

material is believed to be a result of a finer Widmanstätten alpha microstructure, along with the absence of continuous alpha at prior beta grain boundaries.

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TITANIUM CASTINGS FOR DEMANDING APPLICATIONS (SPACE SHUTTLE ATTACHMENT FITTINGS)

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ABSTRACT

The Space Shuttle System composed of the External Fuel Tank (E.T.), the Solid Rocket Boosters (S.R.B.), and the Shuttle Orbiter are fastened together using cast titanium fittings. This paper describes the evolution of the titanium alloy selected and comparison of tensile and fracture toughness results of Ti6Al-4V castings made by two foundries. Although one foundry produces higher strength parts from material of the same interstitial alloy level, the resultant cryogenic toughness is well above design requirements. The higher strength standard chemistry parts although lower in fracture toughness tended to have higher tensile ductility.

FEW CAST COMPONENTS IN ANY METAL are used in more critical primary structures than the family of titanium castings which link together the major components of the Space Shuttle System. Affixed to the large main fuel tank or External Tank (ET) are a group of eleven titanium fittings which attach not only the Solid Rocket Boosters (SRB), but also the Shuttle Orbiter into a single entity for lift-off by the 7 million pounds thrust of the engines.

The castings are in seven different configurations with four of them used in pairs to total eleven parts. (Fig. 1 - Group Photo.)

Fig. 2 (Schematic Location of Titanium Fittings on ET) shows the arrangement of the Orbiter attach fittings on the External Tank and Fig. 3 (ET/SRB Aft Attach Fitting Orientation) shows a cross section schematic of the SRB attachments at the aft end of the tank.

Seven of the fittings are exposed to cryogenic temperatures being attached to the liquid oxygen and liquid hydrogen sections of the tank.

Titanium alloys were the construction materials of choice because of their high strength and superior fracture toughness at cryogenic service temperatures.

This report discusses the evolution of the alloy selection and the evaluations and comparisons made on material made by different molding methods and by two different vendors.

Acknowledgement and thanks are extended to Pete Hinkeldey, Chief of Structural Materials at Martin Marietta, New Orleans, for the generous permission to use his previously unpublished data.

The first cast parts used in this application were of the Extra Low Interstitial (ELI) formulation of the Ti5Al-2.5Sn alloy and were made using the rammed graphite molding process. Although several sets of fittings were successfully produced, the casting vendor did experience manufacturing problems and urged consideration of Ti6Al-4V ELI. Problems with the ELI formulation of Ti5Al-2.5Sn were related to surface quality and weldability. This alloy has a history of being difficult to process even as a wrought product as the extra low interstitial level and normally very low iron content result in a decided propensity to be hot short, i.e. brittleness in the hot forming range. It was felt that this phenomenon also was a contributor to the problems associated with the castings.

The Ti6Al-4V ELI castings were made using rammed graphite and cut-up mechanical properties were very similar to the Ti5Al-2.5Sn ELI.

When Martin Marietta evolved a new lighter weight External Tank, the fittings were redesigned to lighter and more dimensionally tight configurations prompting a conversion from rammed graphite type castings to precision lost wax type castings. A different vendor (B) was selected and soon it became apparent that even though the Ti6Al-4V ELI chemistry was the

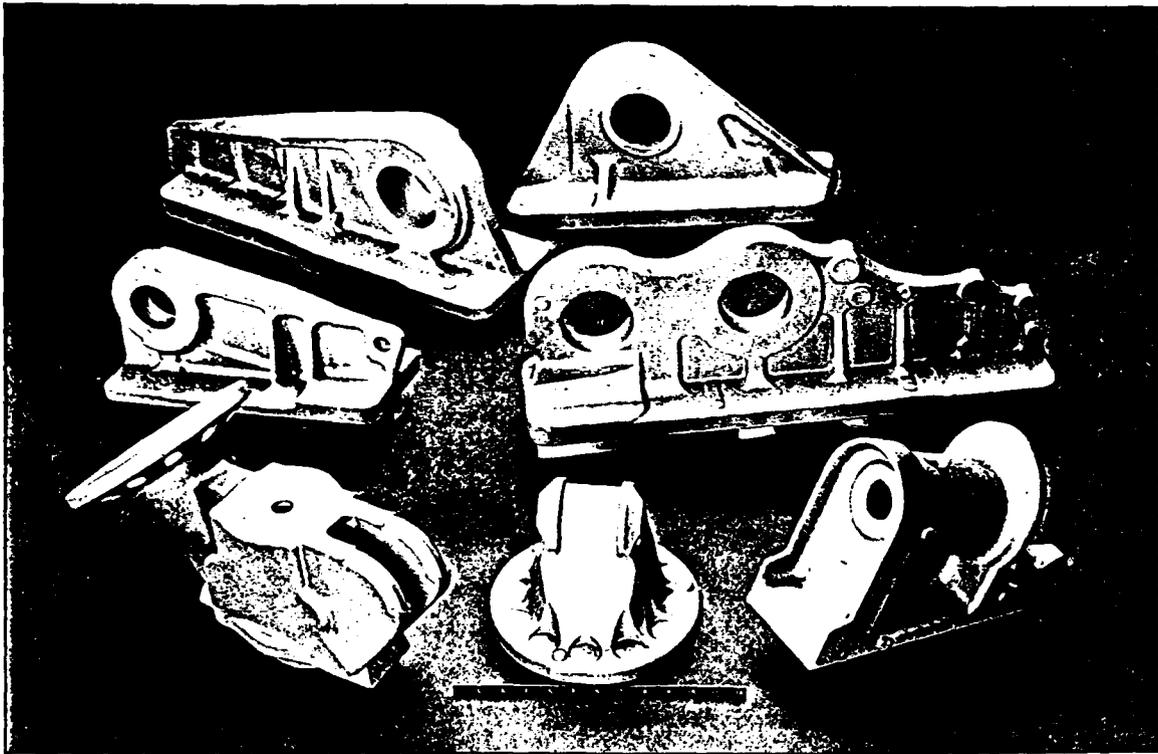
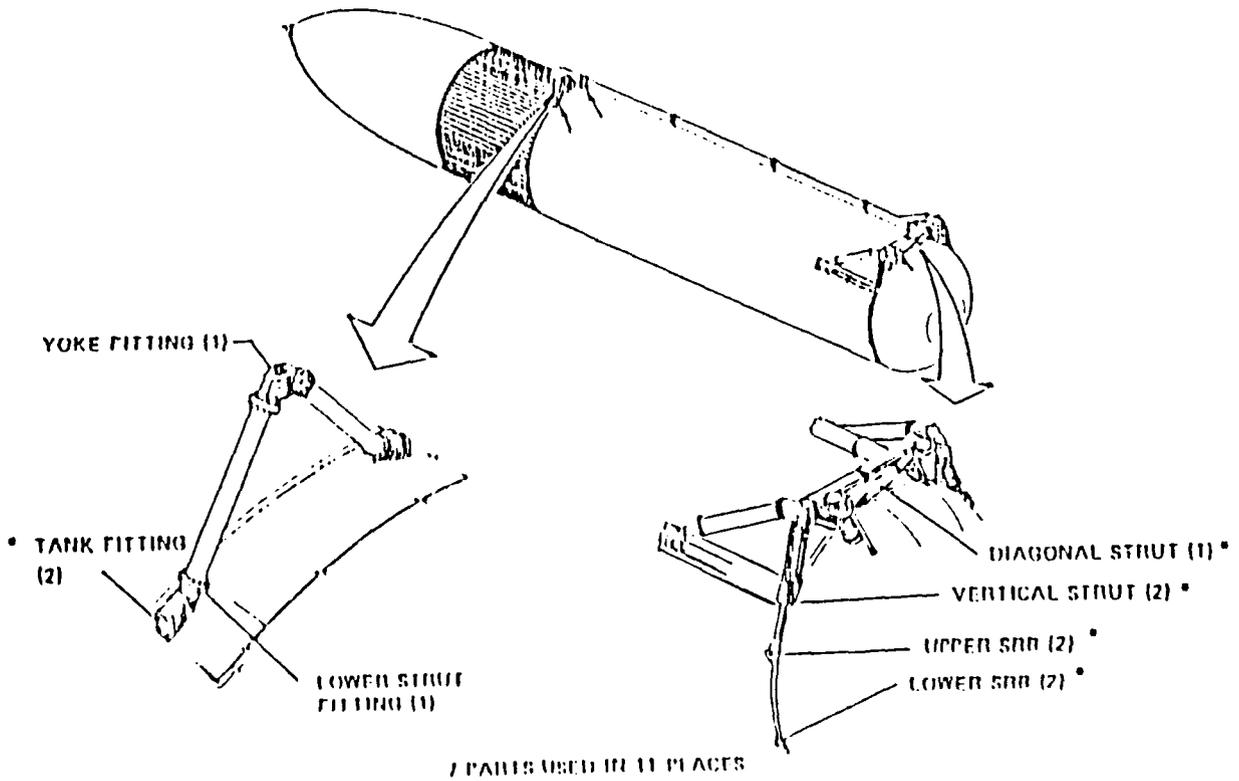
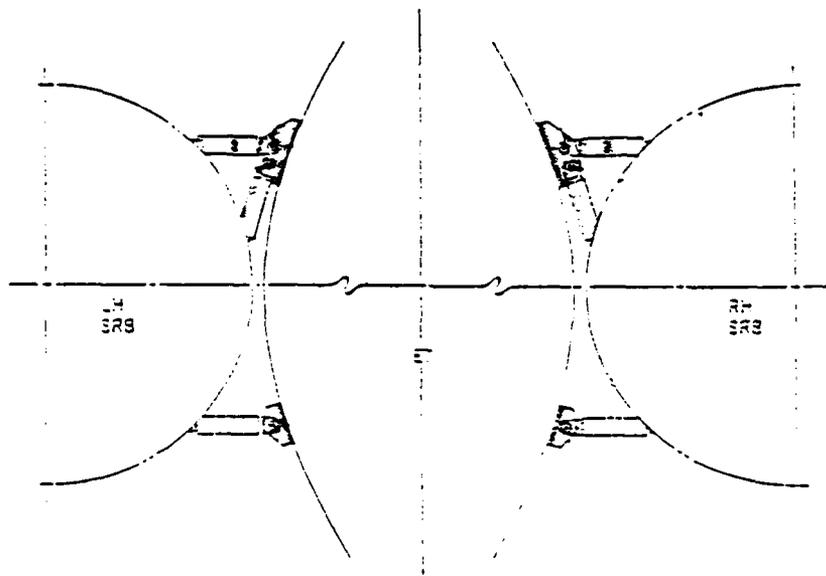


FIG. 1: FAMILY OF CAST TITANIUM EXTERNAL TANK FITTINGS



• DENOTES CRYOGENIC FITTINGS

FIG. 2: LOCATION OF TITANIUM FITTINGS ON EXTERNAL TANK



AFT ET/SRB ORIENTATION (LOOKING FWD)

FIG. 3: ET/SRB AFT ATTACH FITTING ORIENTATION

same, there was a significant difference in the tensile properties of rammed graphite type parts from one vendor (Vendor A) compared to investment parts from a second vendor (Vendor B). The investment cast ELI castings from Vendor B were so low in strength that the formulation was switched to standard grade oxygen which brought the properties back up to the level of the ELI rammed graphite parts from Vendor A. The original Vendor A was later requested to make the same parts using the investment process and the same standard Ti6Al-4V formulation. The resultant tensile properties were substantially higher than from Vendor B.

Engineers at Martin set about to run complete concurrent studies of the several conditions to sort out what was happening, and to ascertain whether the higher strength standard grade investment castings would have adequate cryogenic toughness.

TEST PROCEDURE -

Investment castings were obtained from two vendors as follows:

Vendor A

- Three Ti6Al-4V ELI annealed at 1550°F
- Two Ti6Al-4V Standard Grade
 - One Annealed at 1350°F
 - One Annealed at 1550°F

Vendor B

- Two Ti6Al-4V Standard Grade

Specimens, both tensile and fracture toughness, were excised per Fig. 4 (Layout of Test Specimens).

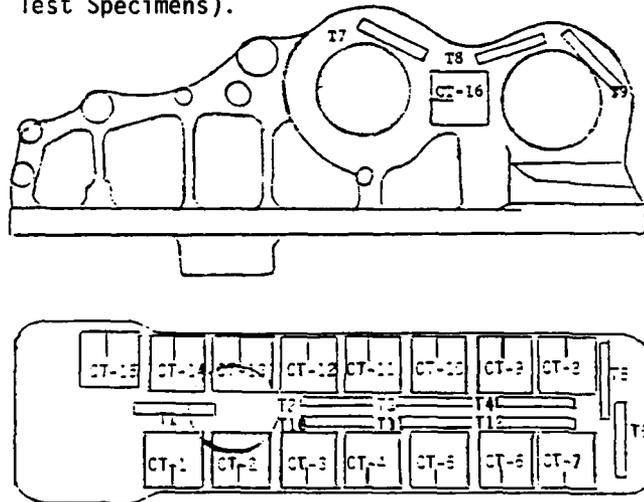
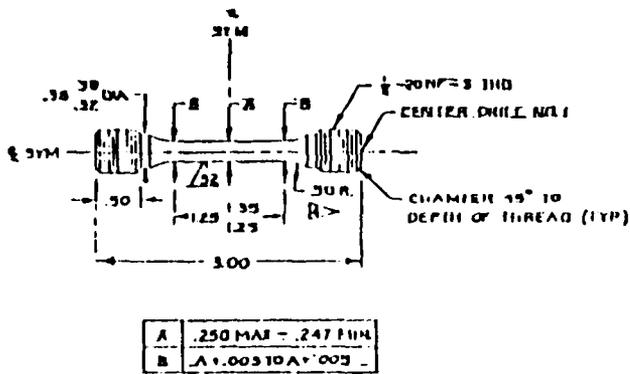


FIG. 4: TEST SPECIMEN LAYOUT

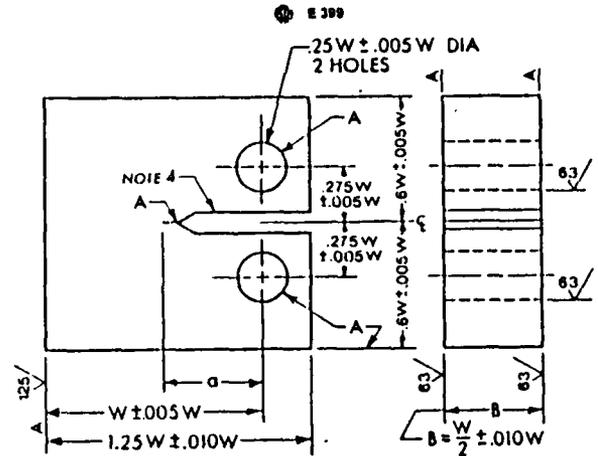
Specimens were machined per the configuration requirements of ASTM E8 for tensiles (nominal .250 inch diameter) and ASTM E 399 for the fracture toughness coupons. The fracture toughness specimen was a 1.00 inch thick compact tension specimen with a machined notch precracked in uniaxial fatigue. Fig. 5 (Test Specimen Configuration) gives details.

Tests were conducted in ambient air, liquid nitrogen (-320°F), and liquid helium (-423°F).



- ▷ THE REDUCED SECTION MUST BE SYM ABOUT THE LONGITUDINAL ξ TO WITHIN ± 0.1 . THE "B" WIDTH MUST DECREASE UNIFORMLY & SMOOTHLY INTO THE "A" WIDTH AT THE CENTER.
- ▷ BLEND FILLET RADIUS SMOOTHLY INTO REDUCED SECTION
- ▷ MAT'L PER PARTS RELEASE
- ▷ DIRECTION OF GRAIN LONGITUDINAL
- ▷ DIRECTION OF GRAIN TRANSVERSE

Round High Strength Tensile Specimen



- NOTE 1 - Surfaces shall be perpendicular and parallel as applicable to within 0.002 in. TIR.
- NOTE 2 - The orientation of the crack must match up with the two specimen portions that is square drawn from the top and bottom edges of the specimen within 0.001 in.
- NOTE 3 - Integral or detachable back edge for clip gage attachment to the crack mouth may be used (see Fig. 1 and 2).
- NOTE 4 - For starter notch and fatigue crack configuration see Fig. 7.

Ref: ASTM E399-83, Standard Test Method for Plane-Strain Fracture Toughness of Metallic Materials - Annex 4

Compact Tension Test Specimen

FIG. 5: TEST SPECIMEN CONFIGURATIONS

Fig. 6 is a matrix of the various tests and their frequency.

Titanium Castings - Test Requirements
Number of Tests

Item	Tests	Mechanical Properties			Fracture Toughness Properties		
		RT	-320°F	-423°F	RT	-320°F	-423°F
1)	Foundry A Ti6Al-4V ELI 2 Lower Aft SRB Attach Castings	5	5	5	8	8	8
2)	Foundry B Ti6Al-4V Std. Grade 2 Lower Aft SRB's or 1 Upper Aft SRB	5	5	9	3	10	15
3)	Foundry A Ti6Al-4V Std. Grade Full Annealed 2 Simulated Upper Aft SRB Attach Castings - One at 1350°F - One at 1550°F	3 3	3 3	3 3	3 3	5 5	5 5

FIG. 6: TEST MATRIX

RESULTS AND DISCUSSION -

Table 1: Tensile Test Results Summary (Average)

Test Group	Test Temp. °F	Ultimate Tensile Strength		Yield Strength		Percent Elongation
		(KSI)	(MPa)	(KSI)	(MPa)	
<u>Investment Castings</u>						
Foundry A ELI	RT	128.52	886	123.58	852	7.5
	-320	201.28	1388	186.76	1288	2.7
	-423	225.3	1553	210.67	1453	2.0
Foundry B STD	RT	129.23	891	118.33	820	6.16
	-320	206.77	1426	197.20	1360	4.0
	-423	221.30	1526	213.40	1471	2.75
Foundry A STD 1350°F Anneal	RT	135.83	937	129.38	892	9.3
	-320	211.09	1455	201.33	1388	6.0
	-423	241.99	1669	223.61	1542	4.5
Foundry B STD 1550°F Anneal	RT	133.61	921	123.93	854	7.67
	-320	210.76	1453	197.39	1361	6.0
	-423	236.27	1629	217.18	1497	5.3
STM 5636 (Allowed Minimum)	RT	110.0	758	100.0	690	4.0

Table I and Table II show the summary (averages) of the tensile and fracture toughness data respectively. All test groups exceed the minimum ultimate tensile strength for Martin specification MMS 5636 of 110.0KSI (758 MPa) at room temperature.

Table 2: Fracture Toughness Results Summary (Average - KQ)

Test Group	Test Temp. °F	K_Q
		$KSI \sqrt{in}$
Foundry A ELI	RT	102.44
	-320	76.76
	-423	67.49
Foundry B STD	RT	102.86
	-320	70.91
	-423	67.99
Foundry A STD 1350°F Anneal	RT	94.39
	-320	60.21
	-423	54.84
Foundry A STD 1550°F Anneal	RT	86.73
	-320	63.95
	-423	56.95

Tensile data are plotted in Fig. 7 and ductility in Fig. 8. The data shows that Foundry A ELI material is comparable to Foundry B standard grade and that Foundry A standard grade is definitely stronger than

Foundry B material. This points out that a probable basic process variation between the two foundries is responsible for the strength difference. The scope of this work did not research further into the cause of the difference in strength level for comparable input material between the two foundries.

Regarding ductility, it is interesting to note the unusual, and opposite from expected, trend of greater ductility with greater strength for a given test temperature. Fig. 9 (UTS vs % ELI) shows the relationship.

In summation, the two standard grade parts which had the highest strength level had the highest ductility at all three test temperatures. All three standard grade parts had better ductility than the ELI part.

Fracture toughness is plotted in Fig. 10 (Comparison of Fracture Toughness) as a function of test temperature. The lower strength parts, i.e. Foundry A ELI and Foundry B standard had the best toughness values being within a point of each other at room temperature and at -423°F. The standard parts from Foundry A, regardless of annealing cycle, were in a separate grouping about 16 to 17% below the lower strength other parts.

A comparison was made comparing fracture toughness as a function of ultimate tensile strength and is shown in Fig. 11 (Ultimate Tensile Strength vs Fracture Toughness). The trends are as expected, i.e. the toughness at a given temperature is degraded with higher strengths.

COMPARISON OF ULTIMATE TENSILE STRENGTHS

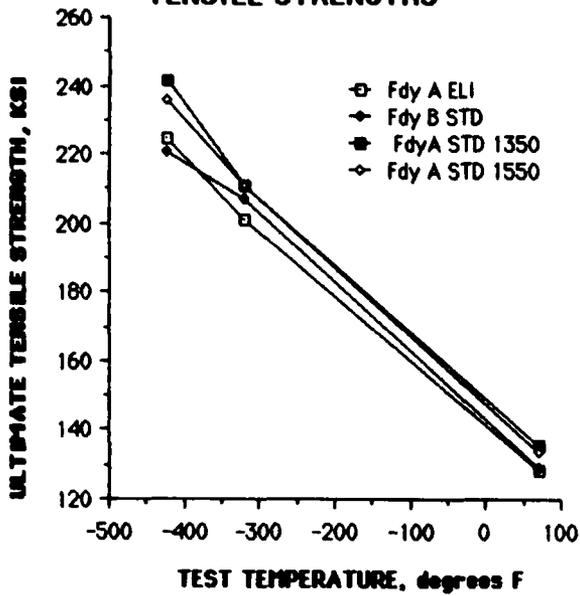


FIG. 7

COMPARISON OF PERCENT ELONGATION

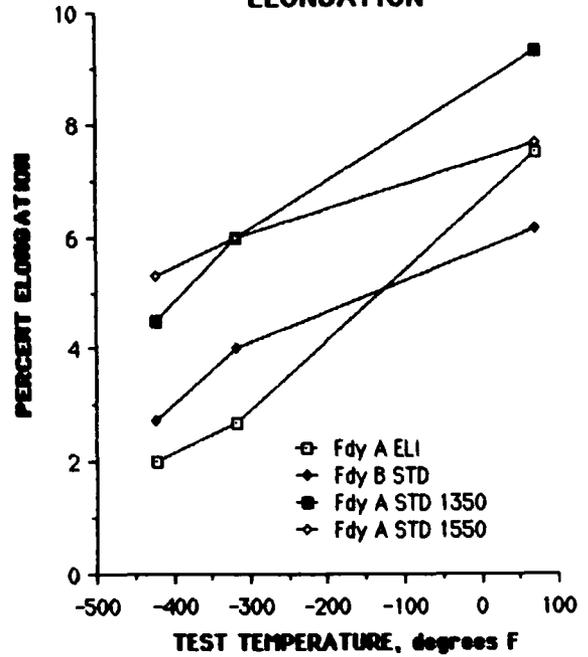


FIG. 8

ULTIMATE TENSILE STRENGTH vs. TENSILE DUCTILITY

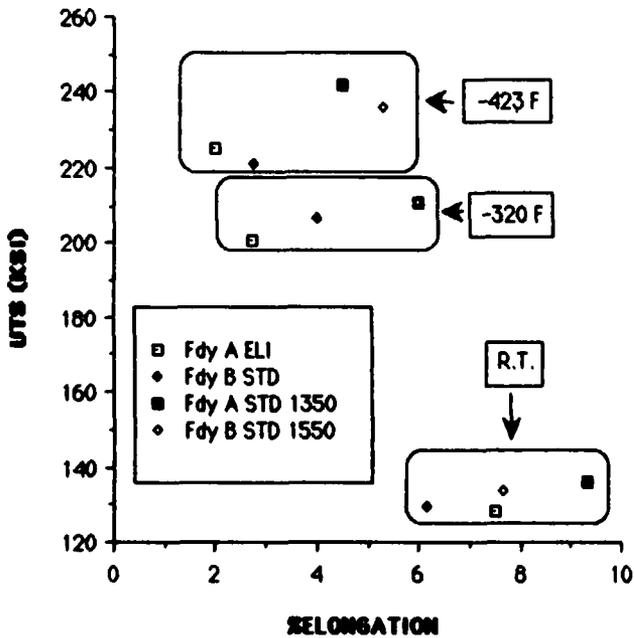


FIG. 9

COMPARISON OF FRACTURE TOUGHNESS

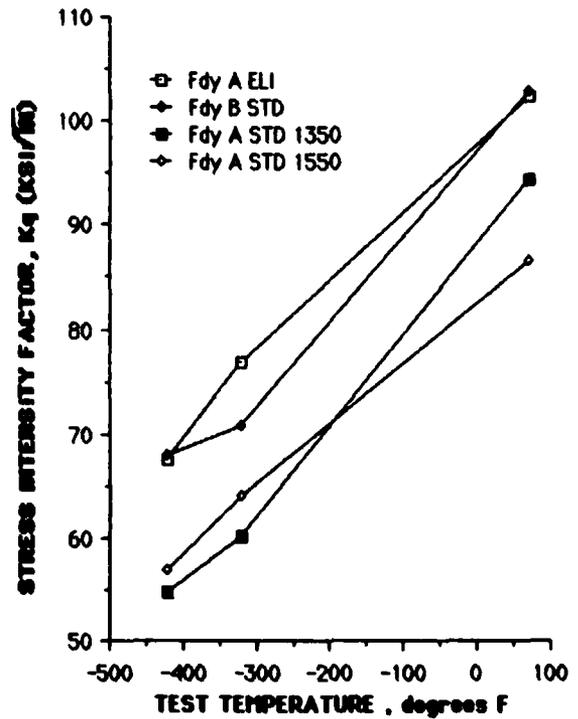


FIG. 10

The final chart, Fig. 12 (% Elongation vs Fracture Toughness) shows the inverse, unexpected relationship of better toughness with lower tensile ductility for a given test temperature. However, when the data is combined on a given set of castings, the normally expected trend of increased toughness with increased tensile ductility is observed.

In spite of the lower K_Q values from the standard grade parts from Foundry A, minimum values exceeded requirements with a high safety factor.

CONCLUSIONS - Tensile properties from investment cast ET fittings of standard grade from Foundry B and standard grade and ELI grade from Foundry A exceed the minimum requirements.

Fracture toughness, even at the lower levels associated with the higher tensile strength parts of Foundry A safely exceed minimum requirements.

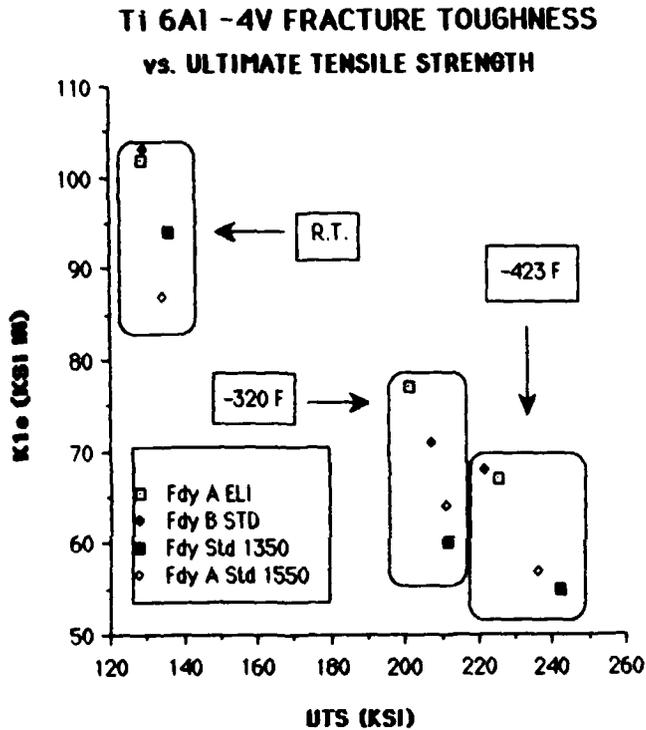


FIG. 11

Reference

Martin Marietta report ETTR-285 "Comparison of Ti6Al-4V Castings - ELI and Standard Grades Mechanical and Fracture Toughness Properties", by L. L. Johnston.

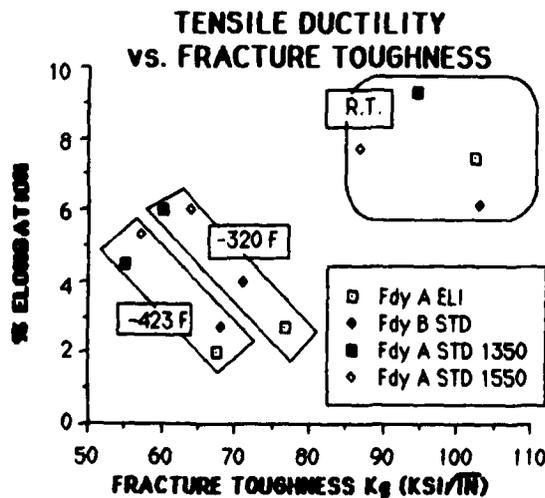


FIG. 12

IMPROVED PROPERTIES IN BETA TITANIUM ALLOY CASTINGS

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ABSTRACT

Beta or near beta titanium alloys are good candidates for applications requiring very high strength-to-density ratios and high fatigue strength. These alloys are capable of very high strength levels by precipitation of a very fine alpha phase in a beta matrix in the solution treated and aged condition. The use of the metastable beta alloys Ti-10V-2Fe-3Al (Ti-10-2-3), Ti-15V-3Al-3Sn-3Cr (Ti-15-3), and Ti-3Al-8V-6Cr-4Zr-4Mo (Beta C) for net-shape casting technology was studied in this work. By combining good casting practices, hot isostatic pressing (HIP), and heat treatments it was possible to produce fully dense castings with tensile strength and elongation as high as 1240MPa (180 ksi) and 11%, respectively, and fatigue strength as high as 980MPa (142 ksi). The combination of high tensile and fatigue strength makes these alloy castings strong candidates for application in airframe, engine, and missile components.

THE HIGH COST OF TITANIUM alloys can limit their use to applications where other lower cost alloys, such as aluminum and steels, cannot be utilized. As a result, in recent years a substantial effort has been concentrated in developing net-shape or near net-shape technologies, in an effort to make titanium alloy components more competitive [1]. Of all titanium near net-shape approaches, such as powder metallurgy (PM), superplastic forming, and precision forging, casting is by far the most developed and the most widely used. The annual shipment of titanium castings [2] has increased by 240% between 1978 and 1986 making casting the fastest growing titanium technology.

The term "castings" represents for most conventional alloy systems, such as copper, iron, and aluminum, a range of products with properties

inferior to wrought material. However, in titanium metallurgy, cast products are almost as good as wrought products and quite often exhibit fracture and creep resistance superior to wrought alloys. Therefore, titanium is almost the only alloy system in which cast parts could reliably be substituted for forged and machined parts in demanding applications [3, 4]. This is due to two characteristics of titanium alloys. The first is the microstructure developed; an elongated alpha structure which is similar to that of wrought material after beta processing. The second novel characteristic is the ability of titanium to dissolve surface contaminants, particularly the oxide, at elevated temperatures leading to complete healing of casting porosity during the HIP process. The elimination of porosity and the desirable microstructures lead to a good combination of mechanical properties in titanium alloy castparts.

To date, most of the work on titanium castings has been carried out on the alpha+beta alloy Ti-6Al-4V [3, 4] which is the most commonly used titanium alloy. The objective of the present work was to study the use of beta titanium alloys in the cast form. Beta or near-beta alloys have the potential of developing tensile and fatigue strengths higher than alpha+beta titanium alloys [5] by precipitation of fine alpha phase in a beta phase matrix through solution treatment and aging.

EXPERIMENTAL PROCEDURES AND RESULTS

MATERIAL - Oversized tensile smooth bar blanks 75 mm (3 in.) long x 6 mm (0.25 in.) gage diameter, were investment cast in clusters by conventional vacuum arc skull melting practice. The nominal chemical composition and the solution heat treatments of all three (3) alloys are listed in Table 1. All samples were HIP'd at 955°C (1750°F)/105MPa (15 ksi)/2 hr prior to solution heat treatment.

Table 1 - Chemical Composition and Solution Treatment of Conditions Studied

Condition	Alloy	Heat Treatment, °C (°F)/hr/ArC ^a	Major Alloying Elements, weight percent
1	Ti-10-2-3	845 (1550)/2 + 595 (1100)/6	10V, 2Fe, 3Al
2	Ti-10-2-3	845 (1550)/2 + 620 (1150)/6	10V, 2Fe, 3Al
3	Ti-15-3	845 (1550)/½ + 530 (990)/16	15V, 3Al, 3Sn, 3Cr
4	Beta C	900 (1650)/1/GFC ^b + 550 (1025)/8	3Al, 8V, 6Cr, 4Zr, 4Mo
5	Beta C	940 (1725)/1/GFC + 550 (1025)/12	3Al, 8V, 6Cr, 4Zr, 4Mo
6	Beta C	940 (1725)/1/GFC + 455 (850)/12 + 540 (1000)/12	3Al, 8V, 6Cr, 4Zr, 4Mo

^aUnless marked, all heat treatments were followed by argon cool.

^bGFC = gas fan cool, 150°C (300°F)/minute.

The microstructure of all six (6) conditions are shown in Figs. 1a-1f.

compared to the Ti-6Al-4V wrought annealed scatterband data in Fig. 3.

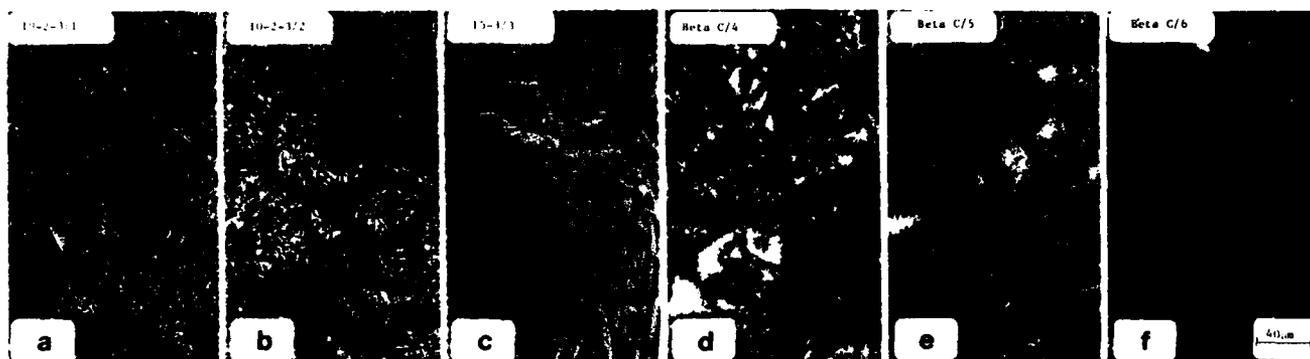


Fig. 1 - Photomicrographs of: (a) Ti-10-2-3, Condition 1; (b) Ti-10-2-3, Condition 2; (c) Ti-15-3, Condition 3; (d) Beta C, Condition 4; (e) Beta C, Condition 5; and (f) Beta C, Condition 6.

Mechanical Testing

Room temperature tensile tests were conducted on 5 mm (0.2 in.) gage diameter x 30 mm (1.25 in.) gage length specimens. Tests were done at 0.005 mm·mm⁻¹·min⁻¹ strain rate to yield, followed by 0.05 mm·mm⁻¹·min⁻¹ to failure. The results are shown in Table 2.

Room temperature smooth bar axial load control high cycle fatigue (HCF) tests were conducted on a 20 ton MTS servo-hydraulic machine with a triangular load-time waveform, R (minimum/maximum load) = 0.1, at 5 Hz frequency to 5 x 10⁵ cycles and at 10 Hz to failure. At 5 x 10⁶ cycles specimens which had not failed were considered as runout tests. The individual fatigue results of all six conditions are shown in Figs. 2a-2f with all data points so that the data scatter could be observed. The average fatigue life curves of all six (6) conditions are

DISCUSSION

The tensile data in Table 2 shows that the Ti-10-2-3 alloy has the lowest tensile strength, the Ti-15-3 is intermediate, and the Beta C is the highest. The 1241MPa (180 ksi) strength of Beta C Condition 4 along with 9% elongation is a remarkably good tensile behavior. Due to the high strength level, the fatigue strength (at 5 x 10⁶ cycles) of the Beta C conditions was the highest of all three alloys (Fig. 3) with Condition 6 displaying the best results (140 ksi runout stress). The comparison of the fatigue results to Ti-6Al-4V wrought annealed data shows that the Beta C results are well above, while the other two alloys fit within the Ti-6Al-4V scatterband (Fig. 3).

To demonstrate the good fatigue performance of the beta alloy castings, the ratio of fatigue strength (at 5 x 10⁶ cycles) to tensile strength was calculated in Table 3.

Table 2 - Tensile Test Results^a

Alloy and Specimen No.	0.2% YS		UTS		Elongation, %	R of A, %
	MPa	ksi	MPa	ksi		
<u>Ti-10-2-3</u>						
1-1	896	130	905	140	11	36
1-2	903	131	979	142	10	24
Average	903	131	972	141	11	30
2-1	855	124	951	138	14	30
2-2	855	124	945	136	13	33
Average	855	124	938	137	14	32
<u>Ti-15-3</u>						
3-1	1076	156	1131	164	8	27
3-2	903	131	965	140	13	29
Average	993	144	1048	152	11	28
<u>Beta C</u>						
4-1	1145	166	1234	179	8	15
4-2	1151	167	1241	180	9	18
Average	1151	167	1241	180	9	17
5-1	1117	162	1200	174	13	19
6-1	1151	167	1269	184	6	10
6-2	1200	174	1296	188	7	19
Average	1179	171	1282	186	7	15

^aTwo tests for each group except one test in Group 5.

Table 3 - Ratio of Fatigue Strength at 5×10^6 Cycles to Ultimate Tensile Strength

Condi- tion	Alloy	UTS, ksi	σ_f (5×10^6), ksi	$S = \sigma_f / \sigma_{UTS}$
1	Ti-10-2-3	141	102	0.72
2	Ti-10-2-3	137	100	0.73
3	Ti-15-3	152	98	0.64
4	Beta C	180	124	0.69
5	Beta C	174	127	0.73
6	Beta C	186	142	0.76

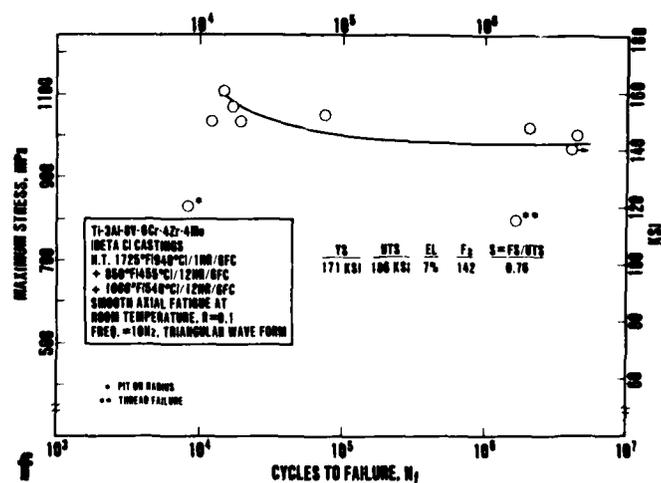
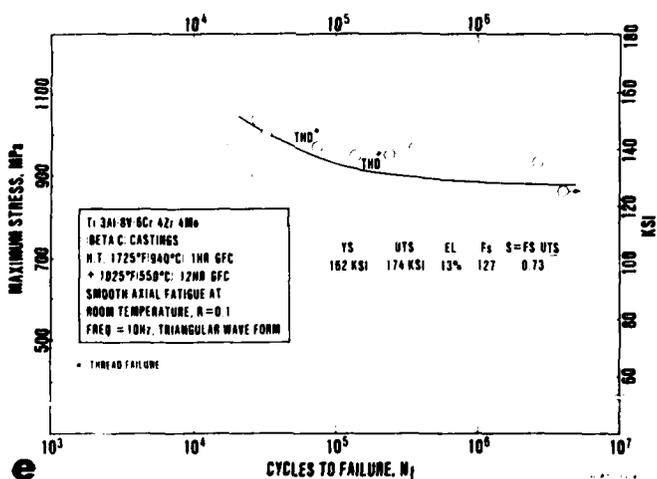
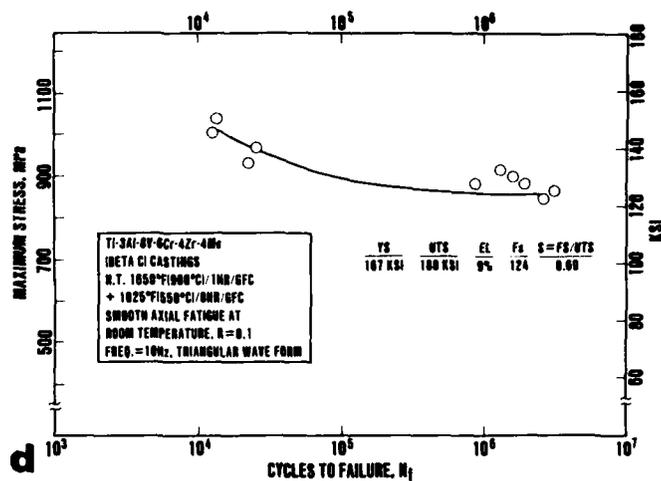
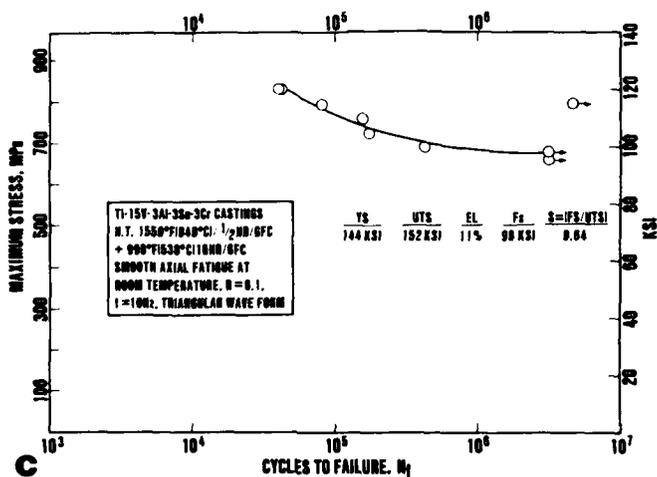
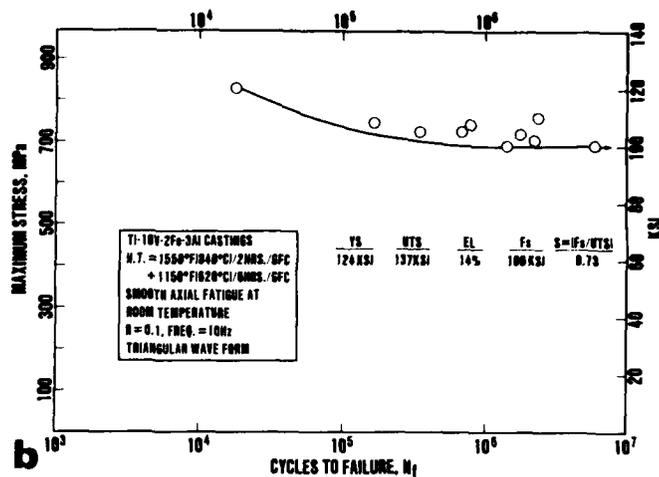
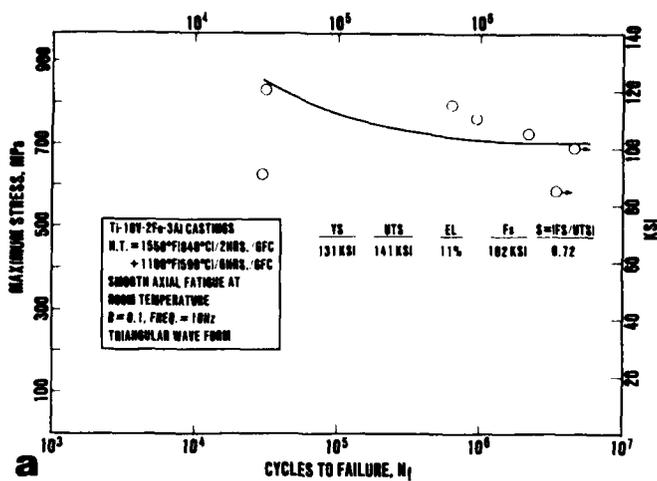


Fig. 2a-2f - Maximum stress vs. fatigue life (S-N) curves and data points of all six conditions. (a) Condition 1; (b) Condition 2; (c) Condition 3; (d) Condition 4; (e) Condition 5, and (f) Condition 6.

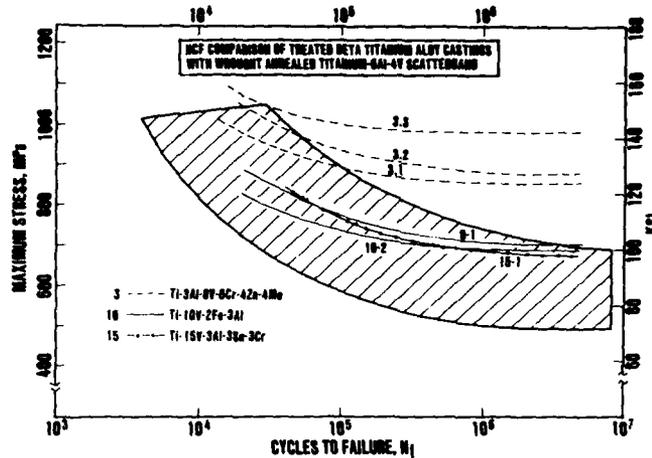


Fig. 3 - Comparison of S-N curves of all six (6) conditions to Ti-6Al-4V wrought annealed scatterband data.

For all three alloys the fatigue ratio was in the range of 0.64 to 0.76. However, Condition 6 (Beta C) not only displayed the highest fatigue strength, but also the highest fatigue ratio of 0.76. The high fatigue ratio is an indication of high levels of microstructural refinement (Fig. 1f) and integrity which retards fatigue crack initiation. However, in the case of Conditions 5 and 6 (Beta C), a few thread failures have occurred in the fatigue samples (Figs. 2e and 2f). This is not the result of material defects, but rather due to an increased notch sensitivity in the higher strength conditions. This high notch sensitivity is of a concern, since it indicates that these conditions have low tolerance to notches and material defects. At the same time, the high fatigue ratio values obtained in this study are an indication that thus far no microstructural or casting defects existed in the tested coupons. The fatigue ratios obtained are higher than those measured on a similar study done on Ti-6Al-4V castings [6], where a lower ratio range of 0.55 to 0.71 was obtained.

SUMMARY AND CONCLUSIONS

1. The influence of various solution treatment and aging cycles on the tensile and fatigue strengths of three beta titanium alloy castings has been studied. The alloys investigated were Ti-10V-2Fe-3Al, Ti-15V-3Al-3Sn-3Cr, and Ti-3Al-8V-6Cr-4Zr-4Mo (Beta C).

2. Of the three alloys tested, the Beta C demonstrated the highest tensile strength (up to 1282MPa [186 ksi]) combined with excellent fatigue strength (965MPa [140 ksi]).

3. The combination of high tensile strength, fatigue strength, and low density indicates that these alloy castings are excellent candidates for use in demanding aerospace applications.

ACKNOWLEDGMENTS

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Aluminum Castings

GRAIN REFINING OF ALUMINUM ALLOY A206 AND ITS EFFECT ON MECHANICAL PROPERTIES

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ABSTRACT

Owing to excellent mechanical properties of A206 aluminum alloy, its applications in aerospace are expected to increase continuously. The object in this study is to find the best way to add Al-5%Ti-1%B grain refinement for improving the strength of aluminum alloy A206.

In this study, a series of different amounts of master alloy were added to aluminum alloy A206 at different temperatures for various lengths of holding time. The results show that the grain size and mechanic strength are related according to Hall-Petch's rule which is the smaller the grain size the higher the the strength it is. The grain size is decreased by Al-5%Ti-1%B addition from 0.1% to 0.4%, then followed by a steady state between 0.4wt% and 0.6%, but the refining effect fades from 0.6% to 1.0%, reflecting the mechanic strength. The grain size first decreases gradually with the degassing time when the melt is at a constant temperature, and then grain size increases after holding a period of time.

The optimum condition to refine the grain size of cast aluminum alloy A206 is that 0.4% Al-5%Ti-1%B is added to melt at 700°C-720°C for 35 min, the resulting have a tensile strength of 42.8kg/mm² and elongation of 7.7%.

THE EFFECT OF HIP ON A206 ALUMINUM CASTINGS

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ABSTRACT

In this experiment, it is intended to use HIP (Hot Isostatic Pressing) process to close pores in A206 aluminum castings for improving soundness. Specimens of various grades as classified by X-ray examinations, were detected after HIP treatment. Also, the optimum combinations of pressure, temperature and time in operation of HIP to obtain optimum casting were studied. Moreover, the effects of heat treatment of post-HIPped casting were discussed.

Typically, three kinds of porosity defect appear in A206 castings, which are gas hole (or trapped gas), microshrinkage and hydrogen porosity. These defects are diminished after HIP treatment (15000 psi-460°C-2hrs), and castings are upgraded simultaneously. The average mechanical properties can be achieved at about: T.S.: 44.0 kg/mm² (62.5 ksi), Y.S.: 37.5 kg/mm² (52.2 ksi) and E.: 8%.

A206 IS A NEWLY DEVELOPED aluminum casting alloy, which has much higher strength and elongation than A356 and A357¹. For the sake of poor castability and difficult welding-repair ability, the development of A206 is limited strongly. Recently, HIP technology is used for upgrading castings except in two main processes, which are consolidation of metal powders and solid-state joining^{2,3,4,5}. Three important mechanisms in powder metallurgy have been suggested for the densification process of HIP: (1) yield criterion⁶, (2) power law creep⁷, (3) diffusion mechanism⁸, and each has its own theoretic and experimental evidence. In 1983, E. Arzt⁹ combined these three mechanisms to explain the densification process consistently. Not only the casting soundness was improved, but also the prediction of densification was realized in this experiment.

EXPERIMENTAL PROCEDURE

- (1) Preparation of specimens - A206 ingot was melted in electric furnace. After N₂ degassing treatment for about ten minutes and somewhat grain refinement by adding AlTiB₂ master alloy, the molten A206 was cast in a CO₂ sand mold at 750°C. A whole casting, including the gating and risering system is shown in Fig.1, and from which all the test bars were cut.



Fig.1 The whole casting where all the test bars were cut.

Table 1 lists the result of chemical analysis and Table 2 shows the classified specimens which are grouped according to treatment process.

Table 1 Chemical Compositions of A206 Specimens

Composition	Cu	Mg	Mn	Ti	Fe	Si	Al
wt.%	5.06	0.34	0.36	0.24	<0.08	<0.05	Bal.

Table 2 Specimen Classifications

Specimen	Treatment process
Group A	As Cast
Group B	Cast → HIP
Group C	Cast → HIP → H.T.
Group D	Cast → H.T.

(2) HIP treatment - as shown in Fig. 2.

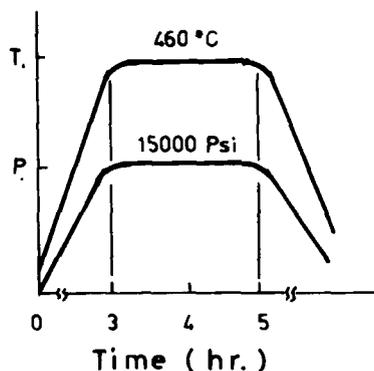


Fig. 2 HIP conditions

(3) Heat treatment - all specimens were heat treated in the same conditions, as shown in Fig. 3.

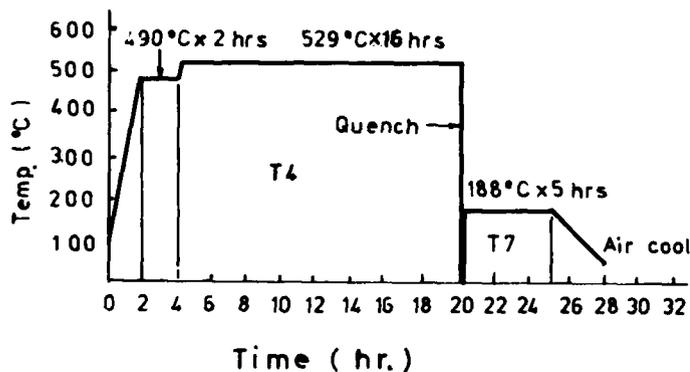


Fig. 3 Heat treatment conditions of all specimens.

(4) Microstructure and mechanical properties were treated for various treated specimens.

RESULT AND DISCUSSION

Generally, three kinds of porosity defect appear in the as-cast specimen. As shown in Fig. 4(a), they are gas holes (or trapped gas), microshrinkage and hydrogen porosity. Each defect

is named with the causes of porosity formation respectively. Basically, gas holes are formed by trapped gas which generally are caused by unsuitable design of gating and risering system. Also, microshrinkage, normally appeared at the contact areas of grains, is resulted from the inherently large solidification interval (80°C) of A206. It is almost impossible to avoid their existence even though the best directional solidification design is applied. Hydrogen porosity, which is segregated within the grain, is generated by the saturated hydrogen gas mainly from wrong degassing treatment.

All these three defects were diminished thoroughly and a clean structure was obtained after HIP, as shown in the microstructure of Fig. 4(b). Instead of porosity, many precipitates apparently appeared all over the plane owing to the influence of 460°C-2hrs-15000psi treatment. More clear evidence is observed in Fig.5. Through heat treatment, a necessary process for A206 castings before application, some porosity (only gas holes and hydrogen porosity) were recovered expectantly, Fig. 4(c) is an evidence. It is natural that microshrinkage did not recover by the effect of heat, for the vacuum state in microshrinkage defect is originally different from some pressure existed in gas hole type and hydrogen porosity type defects.

Comparing with the conventional castings (group D), shown in Fig. 4(d), all microstructures are similar except the existence of microshrinkage in Group D specimens and much smaller porosity in Group C specimens. Those are the typical differences between Groups C and D by OM inspection. Another feature concerning mechanical properties is that the grain size is approximately identical to each other whether HIPped or not. Therefore, no grain growth has been found after HIP. Average mechanical properties are given in Table 3, there

Table 3 Mechanical Properties of Group C and Group D Specimens

Properties Specimen	Y.P. (kg/mm ²)	T.S. (kg/mm ²)	E. (%)
Group C	37.5	44.0	8.0
Group D	36.7	43.1	7.0

are just a little improvement by the help of HIP. One phenomenon, probably one reason, for this small increased mechanical properties is that the heat treatment of the HIPped specimens (Group B) may have not been in optimum conditions. Figs. 4(a) and 4(b) show the different microstructures of as-cast and HIPped specimens before heat treatment. Because the HIP process resembles a pre-solution treatment, some corrections about heat treatment are needed to get high quality casting.

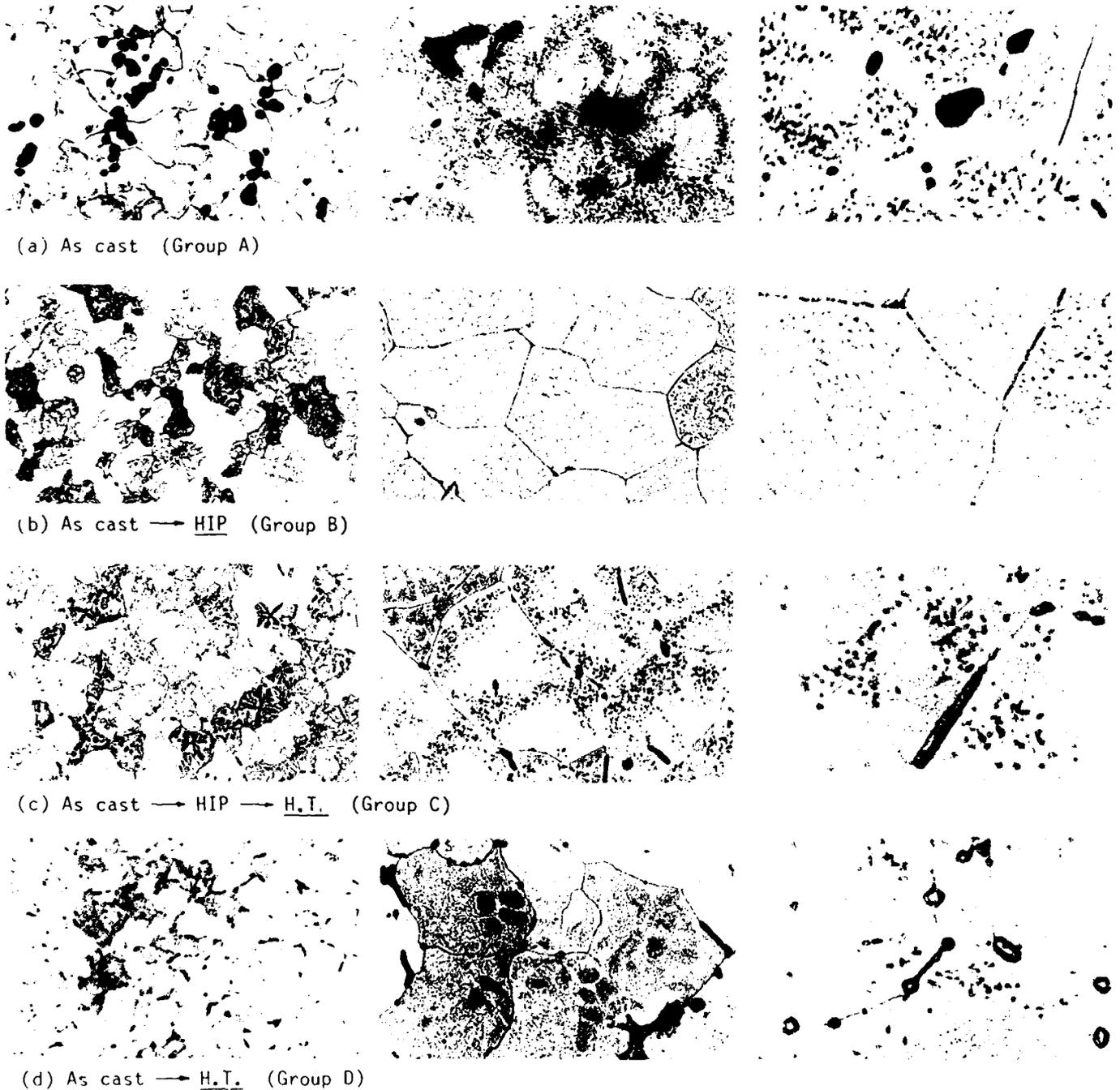


Fig. 4 The microstructures, especially the porosity defects of each group specimen were observed by OM. (75-250-500X)

Relations between mechanical properties and porosity percentage (or density) were examined and the results are shown in Fig. 6. A slightly increasing mechanical properties with decreasing porosity percentage, or increasing density were obtained with Group C specimens. The normal effect of porosity is explained as follows. Conclusively, we must point out that variation of mechanical properties not only depend on porosity percentage but also are influenced by the size and shape, even the distribution of porosity. Fig. 5 shows in more detail the porosity

defects by SEM inspections. It gives the same results, as appeared in Fig. 4 by OM. Much more and larger porosity in Group D specimens were observed than that in Group C specimens. This phenomenon also appeared in Fig. 7, which is the fracture surface. The net X-ray radiographs are shown in Fig. 8. It is clear that there is a great enhancement in casting soundness. Normally, grade D can be raised to grade C or B, meeting the ASTM standards.

A 1mm cylindrical pre-holed specimen, as shown in Fig. 9, was used to examine the process

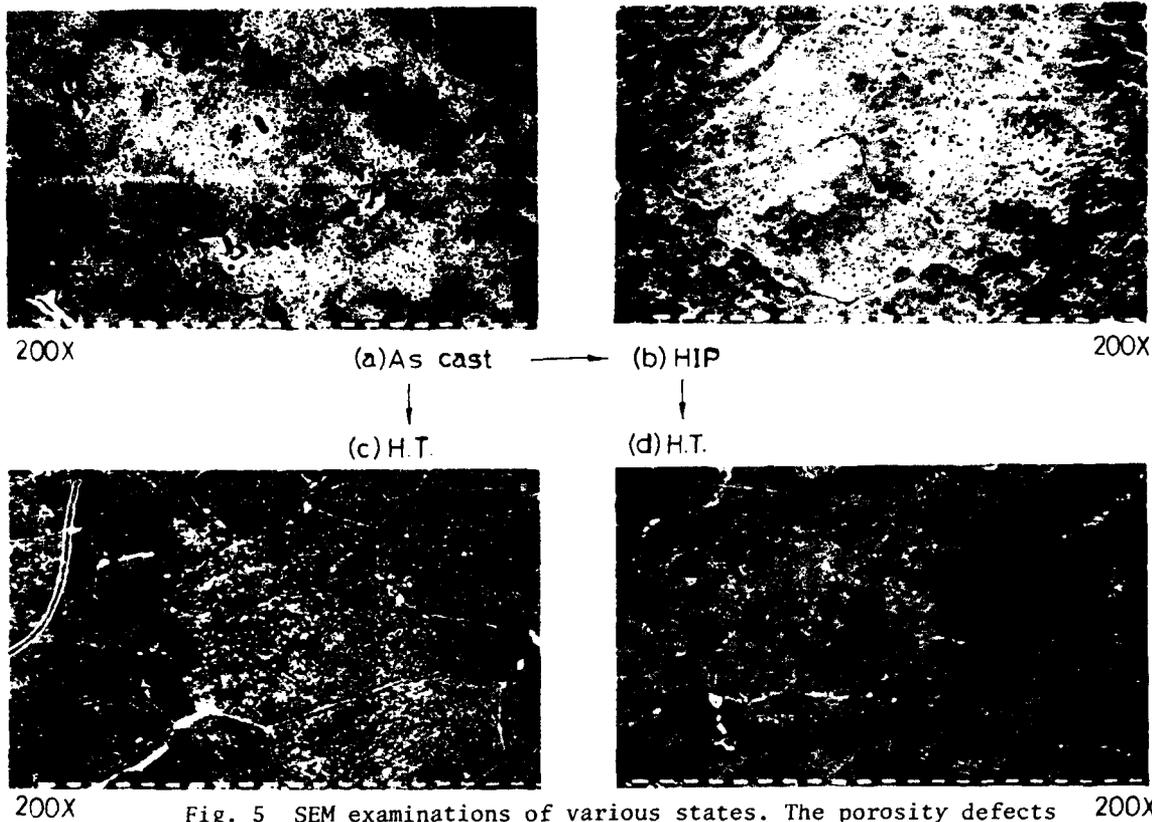


Fig. 5 SEM examinations of various states. The porosity defects appear in (a), (c) and (d) states, and are diminished by HIP, shown in (b).

of HIP treatment. From the X-ray inspection, the cylindrical hole was diminished after HIP. However, though the inspection in larger magnification by SEM, there remained a smaller pressed-hole, as shown in Fig. 10. Grains around the HIPped hole were elongated and some cracks were initiated in the stressed zone. By judging the elongated structure and the pore shape near the deformed area, it seemed that Yield Criterion can be cited as the whole deformation mechanism of HIPping castings. In view of the yield criterion, we suggest that the conditions of HIP

should be corrected to (1) shorter HIPping time because yield effect does not depend on time, and (2) less pressure, because tripling the yield strength is enough theoretically.

CONCLUSIONS

1. Normally, three porosity defects exist in A206 aluminum castings, i.e., GAS HOLE, MICROSHRINKAGE and HYDROGEN POROSITY. All these three porosity defects are diminished after HIP

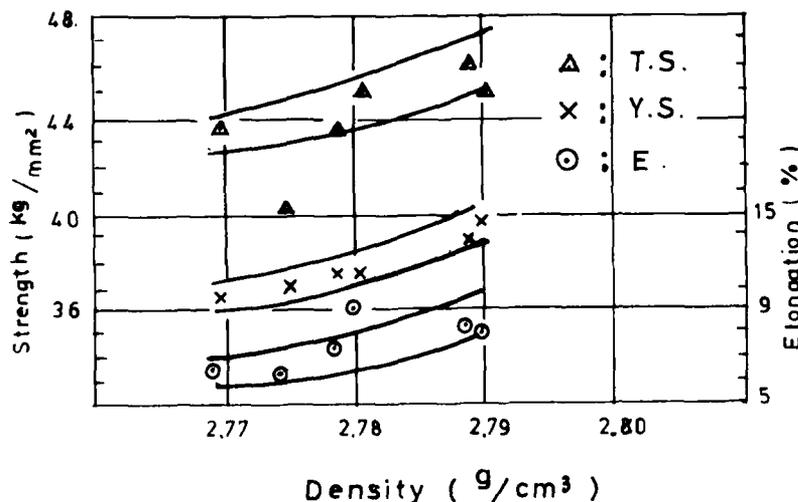


Fig. 6 Relations of mechanical properties and density, only group D specimens were experimented.

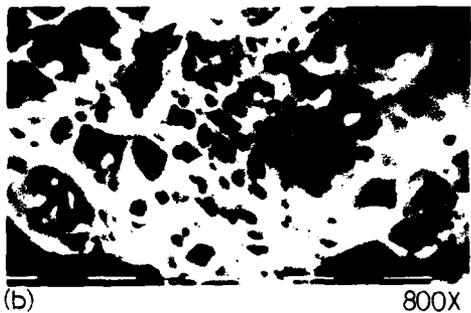
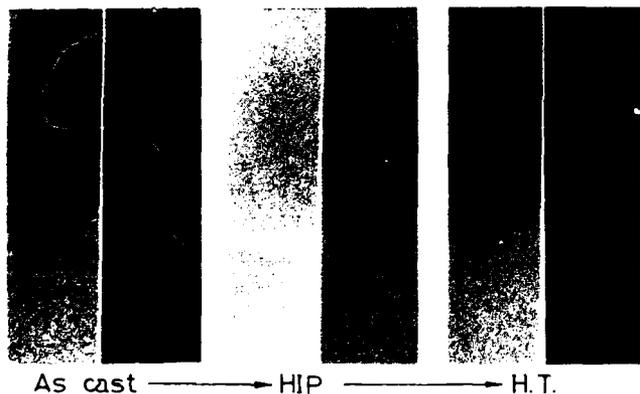


Fig. 7 Fracture surface of (a) Group D
(b) Group C specimen.



As cast → HIP → H.T.
Fig. 8 X-ray inspection of various treated specimens.

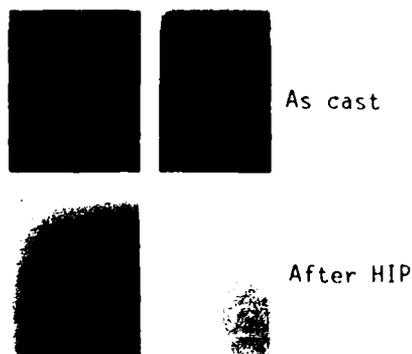


Fig. 9 From the X-ray inspection, the 1mm pre-hole was diminished after HIP.

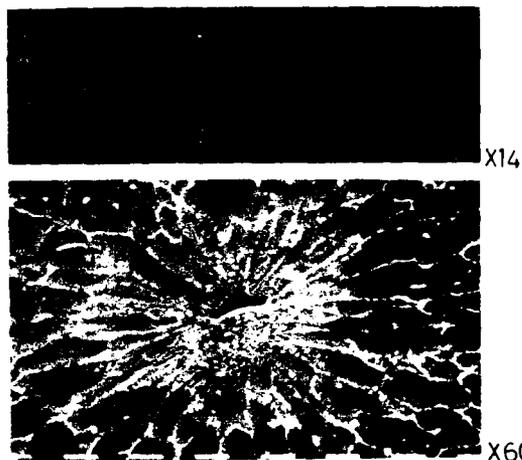


Fig. 10 The remained hole of HIPed specimens. Grains near the hole were stretched and some cracks initiated in the stressed zone.

- treatment and castings are upgraded simultaneously.
- Heat treatment can somewhat dilate the diminished holes, including GAS HOLE and HYDROGEN POROSITY. However, MICROSHRINKAGE are not recovered for the sake of the vacuum state in the as-cast condition.
- Mechanical properties are not greatly enhanced through HIP in our experiment. It needs some corrections about heat treatment conditions in the post-HIPPING of castings for higher mechanical property requirements.

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Investment Castings

THE 1988 INVESTMENT CASTING INDUSTRY: TODAY AND BEYOND

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THE 1988 INVESTMENT CASTING INDUSTRY: TODAY AND BEYOND

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THE INDUSTRY...

Investment casting is a relatively recently developed manufacturing process. The basic procedure is a foundry type. And, it is a metal making process. The mere association with these two classifications, casting and steel making, leads people to categorize investment casting in one of the two areas. To place the investment casting process in either category is like comparing a truck food stop with a four star restaurant and summarizing both as "eating places."

Basic metal making in America has gone through a retching shakedown in the last 15 years. This was once the premier industry in America. Today, less than 50% of the number of people who were employed in steel making in 1968 are in the industry. The biggest company names in steel making have disappeared, along with the vast labor forces.

While the general casting industry is experiencing a resurgence this year, 2,000 "casting" locations have gone out of business in the last 20 years. This is a far cry from the case of the unique and small investment casting industry.

There are just over 6,000 casting locations left in the U.S. These include 2,750 sand mold casting locations, 1,000 no-bake mold casting houses, 650 permanent mold, 600 die casting manufacturing sites, and numerous others such as shell molding

and plaster, centrifugal and evaporative pattern casting types.

Investment casting plants form less than 6% of the total "casting" industry in the USA. There are only 350 investment casters in the U.S. and another 350 investment casters in the rest of the world. The bulk of our industry and markets are in the USA, and U.S. firms lead the world in investment casting technology. This industry is a growth industry in dollars, not in new companies.

The Fredonia Group, a research organization, estimates the entire 1986 "casting" market at approximately \$15 billion. The investment casting industry accounted for \$2.5 billion, or about 17%.

Growth rates for the investment casting industry have averaged 6% to 9% since virtually becoming an industry in 1946. The percentages will continue at that rate for the foreseeable future.

Approximately 65% of the market is aerospace. The top three firms account for over 60% of all U.S. sales dollars. The government is the largest purchaser. The Department of the Air Force accounts for 61% of all government purchases; the Navy's share is 19%. The Army purchases 17%.

The trend favors larger firms as the technology, and particularly the cost of that technology, continues pushing toward larger sizes, thinner, stronger, lighter, or new materials castings, improvements in the production process, and greater plant integration. These trends will require such firms to have a larger sales base over which to spread overhead costs.

MILITARY/AEROSPACE DEFENSE PURCHASES:

In April of 1987 the U.S. government formalized what had been known, that the investment casting industry is a strategically necessary industry for the United States. The military items purchased by the government are typically categorized in 10 product

areas.. The bulk, 57%, of the dollars spent are on blades and vanes. Four of the top five firms account for about 85% of the blade and vane market, and one company clearly dominates this market in sales volume. The other product areas are 9% other turbine parts, 8% missiles, 5% aircraft frames, 4% tanks, 4% electronics, 1% weapons, and the balance on sundry items.

IMPORTS:

The investment casting industry is not impacted by foreign imports as other industries. Imports impact our markets only on higher volume castings vulnerable to lower cost foreign producers. Examples include golf clubs, hand guns, and valves and pump components. Overall, this accounts for a small percentage of our types of markets.

The investment casting industry is "job shop" oriented, compared to the more "production" casting processes. And we require a mature infrastructure of outside suppliers for successful operation. The materials going into our processes are technically exacting and needed in a timely fashion. Therefore investment casters cannot readily succeed in less developed countries, despite the ever alluring low labor costs.

Our procedures, design parameters, and part features tend to require extremely close and constant communication with customers. This promotes or necessitates the need for a close proximity to customers, and therefore tends to preclude casting sales across country lines.

Data on imports and exports of investment castings are not available from official U.S. statistics, but we estimate that the U.S. is a net exporter.

Imported castings are recently from Taiwan, the clear leader, South Korea, Japan, Israel, and Spain. Taiwan, Republic Of China, makes 12 million golf heads a year and ships 93% of those to the U.S.A.

RAW MATERIAL IMPORT RELIANCE:

The investment casting sector is dependent on foreign suppliers for numerous metals, most of which are stockpiled, as well as other items like wax and wax blending materials. Selected ceramic shell making additives, such as zirconium are also import dependent.

Important alloying elements, and key country sources, in the investment casting process include: chromium (from South Africa, Zimbabwe, Yugoslavia), lithium, molybdenum, columbium, tantalum and silicon (from Canada, Brazil, Norway and Venezuela), tungsten (from Bolivia, Canada and Republic of Korea), and manganese (from South Africa, France, Gabon and Brazil).

Some of the most important base metals for aerospace applications, such as cobalt and most of the range of more exotic alloying metals, are imported from politically unstable countries or countries that, from a logistical standpoint, we may find it difficult to import base metals in a national emergency.

Of all strategic metals used in investment casting of critical military parts and components, only molybdenum (of which the U.S. has large reserves) is not subject to government stockpiling.

There is no U.S. mine production for manganese, cobalt and chromium. Were it not for scrap reclamation, reducing our dependence, the net reliance on foreign sources for these three materials would be 100 %.

Manganese is essential to all iron and steel production as a de-sulfurizing, deoxidizing, and/or alloying element, and has no satisfactory substitute. Nickel is a substitute for cobalt and chromium in some applications, but with a loss of effectiveness.

LABOR AND OTHER FACTORS AFFECTING THE I/C INDUSTRY:

The investment casting process remains labor intensive, especially if high production volumes are involved. Labor represents over 60% of value added for the industry as a whole, compared to about 4% for all manufacturing.

Investment casting foundries specialize by size and weight of castings. The capital and equipment needs of the company increases directly with the size and weight of the casting. For this reason, estimating capacity is very much a moving target that responds to a number of production variables and physical limitations.

CAPACITY:

We estimate our industry is averaging 78.5% capacity utilization at the present time. Surges can be handled with relative ease.

Most manufacturing processes are disciplined with good definitions of the physics and mechanics of the processes involved; investment casting is not. An investment casting plant compares to an athletic team that performs well, or not.

There continues to be a great deal of art as opposed to science in our industry. It takes a good deal of finesse to pull all the variables together and routinely produce profitable castings.

The majority of investment casting houses, nearly 87%, handle less than 100-pound castings. Ten foundries can handle 300-pound castings. Two of these, pushing at the very limits of the technology, are capable of processing 2,000-pound weights or more.

IDENTITY CRISIS:

As an industry, our single biggest problem is that most people do not know what an "investment casting" is.

Our second biggest problem is that design engineers do not know the design benefits, applications, and features of "investment castings" compared to other castings or manufacturing processes.

THE DRIVING TECHNOLOGY:

Our industry's products are "net or near net shape," especially compared to the other casting methods. We are high-tech in sophistication by comparison, but these terms are relative.

The investment casting industry has had one indigenous driving force that has distinguished it from any of the other "casting" methods. This is the constant benefit and technology spinoffs of the aerospace sector. Aerospace investment casting has, and will continually contribute to the other areas.

THE PROCESS . . .

Despite its modern, high-tech applications in today's market, investment casting is not a recent development. Evidence of the investment casting technique can be found as early as 3,000 years ago.

The history of the "lost-wax" process stretches backward in time beyond Egypt's pyramids to the Shang Dynasty in China. Chinese art works of bronze were done with detail to match much of the high precision investment casting of today. There is also evidence of the process being used by the ancient Aztecs of Mexico, as well as Colombian Indians. By the mid 1500's, the lost wax technique

was considered "ordinary procedure" for making bronze statues and other works of art, but its practical side had not yet been realized.

At that time, the artist made a wax model which was the original work of art. The wax was pliable and easy to work in detail. This wax image was covered with many coats of a milky slurry or plaster until a strong shell developed over the wax. Later, this plaster was heated, the melted wax was allowed to escape, and a molten metal was poured in to replace the wax. When the metal cooled, the metal sculpture was created.

INVESTING:

The history of the name evolved from the dipping process whereby the ceramic slurry is "invested" onto the pattern or wax. The ceramic slurry is the key distinguishing element to this process compared with any other. Precision casting, precision dip casting, ceramic casting, and similar names exist in an effort to identify this unique process. There are several additional proprietary named processes such as Monoshell, Replicast CS, Shellvest and other systems.

Industrial America paid little attention to the skill and art of the investment casting process until World War II. The war brought on a sudden, urgent need to provide finished precision parts for arms and aircraft. Traditional machining, welding and assembly were too time consuming and costly. Investment casting was used to economically produce many complex parts. Know-how from the dental industry was combined with techniques used by jewelers to produce patterns. Thus, low-cost investment casting produced staggering numbers of parts desperately needed during the war.

The creative use of investment casting during wartime spurred a new generation of creative thinkers. Today, investment casters combine centuries of experience with high technology to

provide applications limited only by man's ability to dream up new problems needing solutions.

THE BASIC TECHNIQUE:

The basic technique for investment casting is the same for just about every application whether the end product is a piece of jewelry, a gear on a piece of heavy equipment, or an intricate part on a spacecraft.

First, a wax pattern is created. The wax replicates the part you wish to duplicate. The color of the wax varies with manufacturers or with intentional color coding for in-house identification. The wax patterns are mounted on more wax forms. The Xmas tree appearance of this combination brought about the resultant name, tree. The tree is dipped into a slurry mixture. The repeated dipping builds up the coating or shell. The wax is removed by heat and a void, representing the part is left.

For art forms, the pattern of the part to be made was done by sculpting the wax. In industrial applications, this pattern is generally made by injecting wax or plastic into a metal form or die.

The die is an aluminum tool that is more or less sophisticated dependent upon the application or requirements. The tooling is needed for constant replication with exacting consistence.

Patterns can be designed for coring internal passageway requirements, and are run on the typical modern wax injection machines.

The design may be as simple as the head of a golf club or as complex as a minute valve system for a space capsule. A wax pattern is made for each casting in the exact geometry of the part to be completed.

THE GATE:

At the mechanical heart of the process, one or more channels or gates are included as a part of the pattern. This channel or gate has several functions:

First, it is used to attach the pattern to a larger channel called a sprue or runner. Several castings can be attached in this way, forming a cluster of the same pattern. Thus, many parts can be produced at the same time.

Secondly, the channel or gate on each part also provides a passage for draining out the wax when it is later melted. Finally, the gate or channel provides a passage for the molten metal entering the mold cavity during the casting operation. Several gates or channels may be added to a part to insure its strength as the molten metal solidifies.

A pouring cup, which will receive the molten metal, is attached to the runner, which, in turn, is attached by gates to the pattern.

Like the age-old process used in sculptures, this tree or cluster is dipped in a ceramic slurry, drained and coated with a fine ceramic sand. This process can be more or less automated.

After drying, the process is repeated again and again using progressively coarser grades of material until a self-supporting shell has been formed. Once the shell is dry, it is heated to allow the wax to run out the gates, runners and pouring cup. The molten metal is then poured into heated shells.

After the poured molds have cooled, the mold material is removed from the casting by mechanical vibration and chemical cleaning. Parts are cut from the cluster and smoothed.

While many parts are ready to use at this point, others are ready for machining or other secondary

operations before shipping. Hipping or hot isostatic pressing, is becoming increasingly important secondary operation for the densification of castings to eliminate porosity and improve fatigue, ductility and other properties.

CASTING METHODS:

Both air and vacuum castings are important in investment casting, and most investment castings--air and vacuum--are gravity poured.

Air casting is used for such common investment cast alloys as aluminum, magnesium, copper, gold, silver, platinum, all types of steel, ductile iron, most cobalt alloys, and nickel base alloys that don't contain reactive elements. Zinc alloys, gray iron, and malleable iron are not usually investment cast for economic reasons, but if they were, they would be air cast.

Most vacuum casting is done in investment casting. Vacuum casting provides cleaner metal and often superior properties. This sometimes provides an incentive to cast some of the normally air melted alloys in vacuum. However, its major use is for alloys which cannot be cast satisfactorily in air, such as the gamma prime nickel alloys, some cobalt alloys, titanium, and the refractory metals.

Centrifugal casting uses the centrifugal forces generated by rotating the mold to propel the metal and facilitate filling. The mere mechanics of this investment casting provide some limitation to the applicable applications.

The cost of any part increases in direct proportion to the preciseness of its dimensional tolerance requirements. This is also true with investment casting, but many times less costly than competing processes. Today's technology has made precision practically limitless, with advancements in investment casting technologies being developed almost daily.

COMPETING PROCESSES . . .

Investment casting is a net or a near-net shape manufacturing process. Besides a near-cousin relationship to the "casting" industry, investment casting competes on the periphery of several processes that deal with much closer tolerances than "casting" per se: machining, forging, die casting, permanent mold casting, stamping, powder metallurgy, roll forming and weldments.

Each process offers its own indigenous benefits, and it should be pointed out if it is not already apparent, that an informed design engineer can make a big difference in the cost and ultimate success of any product. The unique thing about investment casting is that it typically combines many of those benefits of the other processes into one. For example, by design and alloy choices, forging strengths and machined tolerances can be combined; roll formed and/or stamped part assemblies can be replaced; and other combinations are possible as well. Even without the combination benefits, investment castings offer many more advantages.

BENEFITS OF INVESTMENT CASTINGS:

Many of the benefits can be summed up in these categories:

- COST SAVINGS
- RELIABILITY
- DESIGN IMPROVEMENTS
- ALLOY SELECTION
- PROPERTIES ENHANCEMENTS
- CLOSER DIMENSIONAL TOLERANCES
- FINISH BETTERMENT
- MACHINING/PROCESSING SAVINGS

- REPLICATION
- COMPLEXITY
- TIME
- PRODUCT RE-INTRODUCTION

Virtually all investment castings are sold on a cost-justified basis (versus other manufacturing processes). Additional cost benefits can be realized in thin wall, light weight, cored, stronger, more quickly produced, or more quickly re-introduced products.

Design efficiencies and creativity are enhanced by investment casting the parts. Alloy selections are virtually unlimited giving new possibilities and properties to old applications. The other listed benefits such as self evident.

In choosing from alternative metal forming techniques, engineers balance a number of qualitative and cost considerations. In the case of investment casting, a balance must be struck between what are usually high up-front unit costs, and the indirect savings in machining, material, and other unnecessary steps to finish an investment cast part. The investment casting also has important qualitative advantages such as dimensional accuracy and shape complexity that are typically included in weighing costs.

These advantages of investment casting vary from one casting application to another according to dimensional tolerances, other specification requirements, and production volume. In some cases, alternative casting processes may be qualitatively acceptable substitutes and more cost effective.

Precision requirements or the difficulty of working a material can sometimes be beyond the capabilities of competing processes. Gas turbine blades and vanes are a good example.

The thinness of metal sections, dimensional precision and accurate curvature, hollow internal structure, toughness of the metals used and the low degree of variance that can be tolerated from one casting to another, are simply beyond the cost effective means of other casting and forging processes.

MORE ADVANTAGES OF I/C OVER COMPETING METAL FORMING PROCESSES

QUALITATIVE:

- WIDER RANGE OF METALS
- HIGHEST DIMENSIONAL TOLERANCES
- HIGHEST LEVEL OF SURFACE DETAIL
- HIGH SHAPE COMPLEXITY
- HIGH LEVEL OF CONSISTENCY BETWEEN CASTINGS
- NEAR NET SHAPE

COST SAVINGS:

- LOWER METAL SCRAP RATES
- REDUCED FINISHING REQUIREMENTS
- REDUCED ASSEMBLY REQUIREMENTS
- LOWER LABOR SKILL REQUIREMENTS

By reducing scrap rates associated with metal removing procedures of alternative machining processes, investment casting provides considerable materials cost savings.

Scrap rates for investment casting are among the lowest of any metal casting process, at 3% to 5% or less. This is significantly below those associated with forging and machining, which can run as high as 70%.

Part of the cost savings is due to the lower weight of most casting that stems from design advantages,

and the rest derived from reductions in the machining steps.

Investment casting provides the highest degree of precision and lowest amount of post-casting finishing.

THE INSTITUTE . . .

The investment casting industry story would not be complete without mention of its 38-year old Institute. The INVESTMENT CASTING INSTITUTE is the educational, statistical, promotional, and informational forum for this unique, high-tech industry.

Beginning with an idea in 1950, the organization conducted its first annual meeting in New York City in 1953. Formally chartered as a nonprofit corporation in Texas in 1980, the INVESTMENT CASTING INSTITUTE today boasts world-wide membership, with its highest concentration in the United States.

INSTITUTE membership covers the widest population of commercial and aerospace investment casters, ferrous and non-ferrous applications, and a wide variety of product types, production methods, plant sizes, pouring capacities and personalities. This is an exciting industry on the threshold of many evolving sciences. Judged strategically important to the defense and economy of several nations, the investment casting industry is a small and personable industry that is unique in the sense that one can literally meet, and get to know all the participants world wide.

Because of the rapidly expanding market, increased technology, and specialized interest within an evolving industry, the institute divisionalized in 1987. The new sectors within the institute include three major Divisions:

Divisions...

The Commercial Division accounts for the largest number of investment casting foundries in the

Institute, in the United States, and in the world. This is the typical starting point for someone wishing to get into investment casting. It has the most competitors and the lowest dollar market volume.

The Aerospace Division accounts for the greatest dollar volume for our industry. In the Non-Ferrous Division, the largest alloy poured is clearly aluminum. We group titanium in this category as well, but not the nickel or cobalt alloys.

Titanium is one of the fast growing areas in our industry, especially large integrated structural applications. A second area of growth and current attention is the special aluminum alloys, which reduce weight, increase speed, and lower fuel consumption in the gas turbine engines.

We also have a Suppliers Division and an International Division, since our membership extends around the world.

The INSTITUTE seeks to serve the industry by serving as its voice in terms of promotion, statistics, education and communication.

Informing...

AS THE STATISTICAL VOICE of the industry, the INSTITUTE conducts ongoing and constantly enhanced statistical programs to reflect industry changes. A monthly statistical program that tracks a representative population of the overall industry feeds back numbers on monthly sales, bookings, tooling sales, backlogs and employment trends. In addition, a yearly confidential analysis of compensation practices is conducted. A cost of sales -- sales and profit analysis is a valuable tool to owners and managers alike. Finally, the annual spring management meeting is the setting for the early "How's Business Round Table" forum providing insight into the industry's sales analyses by casting types, wage and salary changes, collection practices, capacity utilizations, safety and health issues and anticipated forecasts. Industry-wide surveys are a regular part of the overall statistical program and will be conducted on an

international basis starting with the 1988 fiscal year survey. Many of the statistics generated by the institute serve as a basis for lobbying, industry standards, and market perspectives.

The INSTITUTE has routinely acted as an industry clearinghouse of specifications and suggested standards. For decades it has been the sole source of evolving technology dissemination that has advanced the industry and the investment casting sciences. The INSTITUTE works with aligned industries and associations to develop suggested industry standards for such items as quality controls for raw materials, weld repair, metal specifications, and others.

Promoting...

AS THE PROMOTIONAL VOICE of the investment casting industry, the INSTITUTE'S activities are geared to inform potential end users, designers of investment castings, and the public at large of the inherent benefits of investment castings. Advertising campaigns, news articles, presentations to other closely related associations, appearances in major trade shows and exhibitions, and the publication of numerous books, manuals, papers, brochures and pamphlets are a few of these undertakings.

Communicating...

AS THE COMMUNICATION VOICE, the INSTITUTE seeks to be constantly aware of news in the industry and to pass the most up-to-date information to its members, users, design engineers, schools, universities, and involved governmental agencies. A consultants directory is available, an international directory is scheduled for publication, a speakers bureau is being added, and statistical programs are constantly being expanded to meet industry needs..

Educating...

AS THE EDUCATIONAL VOICE of the industry, the INSTITUTE'S activities take many forms. Training courses, volumes of proceedings, numerous manuals, videos, films, guides, suggested standards, and other available material provide up-

to-date information for both the veteran and the novice investment caster. The *Handbook* has been an industry standard for years.

Service...

To better serve the industry, the INVESTMENT CASTING INSTITUTE is organizing and hosting *INcast Country*, the industry's first international show and exhibition in conjunction with its annual technical meeting in Nashville, Tenn. Oct. 9-12, 1988. Providing service is what the INSTITUTE is all about -- service to the industry, to the company, and to the individual involved in this dynamic industry.

Looking to the future...

As was pointed out earlier, the investment casting industry has had one indigenous driving force that has distinguished it from any of the other "casting" methods, the aerospace sector. There are constant technology spinoffs from the aerospace area. This has and will continually contribute to the other areas of investment casting.

We do not anticipate technological substitutions for our products. The recent advancements in aligned industries such as ceramics, will spur additional product demand for us. For example, if ceramic or composite parts can be manufactured to replace superalloy investment castings in the hot section of the aerospace turbine engines as in the NASP, X-30 hyperspace plane, the high-speed civil transport, or advanced fighter projects, the engines will be able to run faster and hotter. This will mean

that superalloy investment castings will be needed before and after these parts. A whole family of additional engines and aircraft will be created. This means more business, not less.

Virtually every industry and market has been served by investment castings but the penetration has only begun. Additional applications will be uncovered in the various markets as design engineers learn more of the advantages and benefits of investment castings. And there is a pronounced trend for all manufacturing to move to higher quality, net shapes. This means more demand for investment castings.

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AEROSPACE INVESTMENT CASTING IN THE U.S.A. 1988

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AEROSPACE INVESTMENT CASTING IN THE U.S.A. 1988

Thierry Thys
Precision Founders

HISTORY...

Investment castings were first manufactured in the United States at the turn of the century. These castings were used for dental implants and jewelry. Small-scale industrial production started in the late 1920's with the manufacture of dental and printing equipment.

Large scale industrial production did not begin until the early 1940's. During World War II the military need for large quantities of turbo-charged aircraft engines prompted the US government to subsidize a small experienced dental and medical laboratory (Austenal). It promptly expanded its investment casting operations and developed and licensed mass production methods. Several million turbocharger blades in a cobalt-based alloy were produced as a result.

The industry also expanded into the production of parts for small arms and related war products. At the end of World War II industrial plants such as General Electric, Westinghouse, International Nickel, and Bendix, as well as specialty producers such as Austenal, Haynes, Arwood and Crucible Steel, had investment casting operations. US production exceeded 35 million dollars annually.

POST-WAR DEVELOPMENTS...

The industry collapsed dramatically after the war, with annual production falling to less than 5 million dollars. The foundries that remained produced castings primarily for the textile and food processing industries.

The association of investment casting with aircraft engines was re-established in 1948 when the military aviation services expanded their use of turbojet engines that required high temperature airfoils and related hardware. Still, rotating parts such as turbine blades were almost exclusively machined from bar. This preference reflected not only the state of the investment casting process, but also a deep-seated prejudice on the part of the engineering community against the use of castings for critical applications.

Production of investment castings for jet engines re-established the industry. By 1953 it has surpassed its peak World War II volume. The end of hostilities in Korea brought a 2-year decline followed by 3 years of growth. By 1958, annual production had reached a volume of 70 million dollars.

The launching of the Soviet satellite "SPUTNIK" in 1957 somehow convinced US military planners that manned aircraft was doomed and that all future effort should go into the building of missiles. Production and development projects involving jet engines were quickly terminated and, by 1959, the investment casting industry had shrunk to less than half its 1958 peak. Fortunately, by this time, development of jet engines for civilian aircraft and the successful introduction of the Boeing 707 had begun to free our industry from its dependence on military planning. And the limitations of forging technology had begun to free American designers of some of their preconceptions about investment castings.

In the early 1960's, turbine designers recognized that existing and projected forging alloys would not meet the higher temperature requirements for future jet engines. With reluctance and a determination born of necessity, the turbine manufacturers decided that the future of jet engines would depend on the use of investment cast turbine blades. At the same time, vacuum

technology was introduced to take advantage of new high-temperature nickel alloy systems that were impossible to melt in air.

BEGINNING OF AEROSPACE QUALITY CASTINGS . . .

In order to satisfy the demands of these designers, a new kind of casting came into being: the "premium quality" or "documented" casting. This segment of the investment casting world transformed overnight, establishing a supervised and controlled environment that could not be imagined by its commercial counterparts.

For the first time, the customer monitored all stages of casting manufacture. Ingots used in the process required extensive mechanical and chemical testing before they could be poured into castings. 100% X-ray and fluorescent dye-penetrant inspection became standard. Methods were developed for efficient 100% gauging. Once a process was established, no deviations were permitted. The number of documents that tracked the parts and process grew exponentially. Throughout, the emphasis was on consistency and accountability.

Another development at the end of the 1950's had a far-reaching impact on the future of investment castings. This was the introduction of the ceramic shell. Prior to 1958, all castings were restricted in size to what could be contained in a block of solid ceramic, usually no larger than 10" x 10" x 18". There was no theoretical size limit, but the cost of using an expensive ceramic imposed an economic limit on size.

The development of the monolithic ceramic shell freed investment casting of size restrictions. By the early 1960's, parts weighing several hundred pounds were being cast. The 1960's also saw the first attempt to automate shell production for greater levels of consistency.

Melting equipment also kept pace with the changing requirements. The standard 10-pound indirect arc furnace, which had been used since the early 1940's, gave way to induction melting furnaces. This improved temperature control and eliminated the risk of carbon contamination. The induction melting furnaces were also available with any desired capacity.

New markets were opening up. Extensive development and production of microwave tubing and electronic devices provided a growing demand for investment cast parts in aluminum and copper-based alloys.

The first half of the 1970's saw the introduction of titanium alloys to the investment casting industry. The technique of Hot Isostatic Pressing, (HIP), which eliminated internal shrinkage, made titanium castings economically practical.

Economically, our industry tends to lag 12 to 18 months behind the normal business cycle, but the trend has always been up. The industry broke through the 200 million dollar sales barrier in the mid sixties. In 1969, the sales volume had dropped to 200 million, 2/3 of the previous year, only to rebound to 320 million 4 years later. The 1975 recession was followed by rapid growth, responding to the needs of a new generation of military and commercial aircraft, and a booming economy.

The era of 1976 to 1979 saw the 400, 600, 800 and 1 billion dollar barriers fall. 1980 saw another peak but the industry slipped just below the 1 billion level in 1981. 1981 was virtually the last recess and was the prelude for unparalleled expansion that doubled the previous high, reaching 2.1 billion in 1987. Our industry will exceed 3 billion this year, 1988.

You can gain a better understanding of the USA investment casting industry by becoming familiar

with its segments. The following discussion looks at our industry in two ways: by alloy type and by market:

Vacuum alloys comprise 40% of the market, (1987 figures); followed by stainless alloys at 32%. Aluminum hovers at the 12% area, titanium has grown to 6%, copper base alloys are 2% leaving all others at 2%.

Defining the same market by dollar value, airfoils comprise 36%, turbine structural follows at 18%, ferrous commercial (non-documented) are 14%, ferrous non-turbine at 12%, Aluminum documented at 9%, titanium at 6%, aluminum non-documented at 3%, and the copper (also non-documented) alloys are in the remaining, 2% area.

For this discussion, we will briefly explore three of the major market segments: Aluminum; vacuum-cast super-alloys for gas turbines; and structural titanium. These segments are of vital interest because they are the testing ground for new developments in our technology.

ALUMINUM INVESTMENT CASTINGS . . .

The size of aluminum investment castings, the complexity of their design, and the alloys in which most are cast have not changed significantly in the past 12 years. A casting measuring 48" x 24" x 24" is close to the maximum size today, as it was 18 years ago. Some improvement in thin-wall sections have been made but typically wall thicknesses still have a lower limit of .035 for small areas, and the bulk of the castings are still cast in the silicon-magnesium aluminum alloys known as A355, A356, A357. High strength alloys, such as the silver-bearing KO1, failed to live up to their original promise.

What is new, is how these proven alloys are chemically balanced, and the techniques by which they are handled. The focus is now on reliability and improved properties.

A critical measure of quality is the dendritic-arm spacing. A typical premium-grade aluminum investment casting will have a dendritic arm spacing of less than 50 microns, compared with commercial castings, which typically have spacings of 100 microns. The mechanical properties throughout the casting, resulting from this smaller dendritic-arm spacing, will show a minimum of 50 KSI UTS, 40 KSI YTS, and 5% elongation. This is significantly better than the commercially available 35 KSI UTS, 28 KSI YTS, and 3% elongation. This smaller dendritic-arm spacing is the result of carefully balancing the alloying elements and promoting rapid directional solidification. Occasionally, solidification is accompanied by high-frequency vibration or mechanical oscillation.

At present, emphasis has shifted from static testing to dynamic loading and to the establishing of criteria for damage tolerances. These castings have become so reliable that engineers are designing structures to operate where stresses are sufficiently high to cause predictable damage. Notch toughness has become the critical mechanical test. Damage tolerance levels have advanced from 11 KS/in to between 16 KS/in and 23 KS/in.

For higher strengths, producers of premium-quality investment castings are exploring matrices of aluminum and non-metallic particles. Current work indicates mechanical properties of 60 KSI UTS with a 20% increase in modulus. One shortcoming is elongation which falls below 2%.

Surprisingly, there seems to be very little effort directed at increasing high-temperature properties, and aluminum investment castings rarely operate above 315 degrees F.

Premium-quality aluminum investment castings are used extensively in radar antennas, electronic chassis, and flight control structures.

VACUUM-CAST SUPER-ALLOYS . .

The requirements of aircraft turbines have spurred the advancement of investment casting technology. The need for lightweight structures, reliability, and high temperature capability has been directly responsible for the development of vacuum casting, preformed ceramic cores, and the introduction of titanium, as well as directionally solidified and single-crystal turbine blades. Newer engines with more advanced technology have fewer parts, produce more power from a pound of fuel, and last longer than their predecessors.

Higher operating temperatures in jet engines translate to greater efficiency, making this the central focus of turbine engine designers. Modern turbines now operate with gas temperatures that exceed the melting point of the alloy used in turbine blades and vanes, and we are closing in on the theoretical temperature limits of the cobalt and nickel-alloy systems presently in use. Directionally solidified and single-crystal components appear to be the last outpost on the high-temperature frontier.

Future improvements will come primarily from improved systems of cooling. This need for improved cooling is being answered by the use of increasingly complex cooling passages formed by ceramic cores. In many ways, improvement in engine capability is directly linked to the advanced use of ceramic cores.

TITANIUM STRUCTURAL CASTINGS . . .

A goal that is almost as important as withstanding increased operating temperatures is the development of lightweight reliable structural components. Cast titanium was developed to fill this need. Work on investment cast titanium started in the late 1960's, and several designs were scheduled for use on the American supersonic transport. Results were encouraging, but problems with internal voids, and a crack-sensitive outer layer (alpha case) delayed the use of titanium castings until 1974, when the first production castings appeared on the Pratt & Whitney F100 fighter engine.

The development of less reactive ceramic molds and cham-milling solved the surface problems. Hot Isostatic Pressing (HIP) sealed the inner voids. Concurrent with these developments was the introduction of melting and handling equipment that made even larger castings possible. Production castings measuring over 50" in diameter are in general use as primary structures. Castings just under 60" are in development. The size is limited not by the process, but by the available HIP units, the largest of which are 60" in diameter. For comparison, nickel-based vacuum alloys have been cast in this size range and used in production jet engines for over a decade.

The latest efforts of titanium castings are directed at producing rotating compressor parts. A joint industry - Air Force effort has resulted in engine qualifications and production of a small (8" dia.) compressor fan. Hollow compressor blades 12" to 18" in length are currently under development.

Titanium castings are limited to operating temperatures of less than 900 degrees F, and there is active development of alloys which is expected to raise this limit to 1100 degrees F. For structures that must operate at higher temperatures, castings in nickel-based vacuum alloys with wall thickness of

.035 in. hold considerable promise and will appear shortly on production engines.

CONCLUSION . . .

We expect turbine engines to continue to increase their use of investment castings, even though the projected value of flight turbine engines produced in the USA will level off at about 10 billion dollars a year. The reason is simple. Early jet engines made little use of investment castings. Less than 3% of their weight was investment cast. By the early 1970's, that percentage had increased to 7%, and the investment cast weight of current designs are estimated to be between 25% and 30% of the engine weight.

Our concentration on just three market segments does not imply that development and progress is not being made in all areas. In fact, the lower value of the dollar has made American products more competitive on world markets, and concurrently, the commercial segments of our industry are enjoying the highest rates of growth.

New equipment and automation, which brings the latest in consistency and control, are constantly upgrading the industry. The trail-blazing efforts

and new technologies developed for the turbine engine producers are quickly adapted and applied by commercial producers.

Throughout the investment casting industry, improved quality systems employing statistical quality control, computerized cost accounting, and shop floor control are providing engineers a quick and accurate picture of the status of the process. They also provide tools for quickly evaluating and correcting potential problems before they affect the product.

In the near future, we expect that the current work on computer modeling of solidification will be available to the foundry engineers, and will allow them to deal with the fundamental mechanism that controls the basic soundness of cast parts.

We can't expect the explosive growth of the past 10 years to continue forever, but the prospects for the immediate future are very encouraging.

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COMMERCIAL UTILIZATION AND APPLICATION CONVERSIONS OF INVESTMENT CASTINGS

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ABSTRACT

Commercial Investment Castings have evolved into a viable and cost effective means of producing components with repeatable dimensions and close tolerances. Additionally, smoother surface finishes and increased detail far exceeds the same features in forged, welded, or machined and joined components. The selection of a manufacturing method for commercial application components has been based on the above factors, mechanical design requirements (operation stress, temperature, and integrity), cost and weight.

This paper identifies and quantifies commercial investment castings design conformance capabilities. It details recent advances and variations in the process which improves the castings ability to meet or exceed component design criteria. It further cites specific castings which have been successfully converted and field-proven without sacrificing existing design function, weight or performance characteristics.

Magnesium Castings

MAGNESIUM ALLOY SAND CASTINGS IN AEROSPACE APPLICATIONS

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ABSTRACT

The paper attempts to answer the question "Why magnesium anyway— and why sand castings in particular?"

"What features are of particular interest to the aerospace designer?" — low weight very attractive strength to weight relationships, ability to produce parts by all the normal processing methods in extremely complex forms, excellent properties at room and elevated temperature and good corrosion resistance. These requirements are currently being met very satisfactorily as never before by a range of alloys suited to particular requirements.

Improved methods of melting and pouring contribute also to a much improved cast product.

Some interesting properties, perhaps not fully appreciated at the present time, such as the very good damping capacity of the metal have also led to some preference for magnesium as well as its extremely good machinability.

Regarding applications — a large number of slides showing typical applications in the aerospace field illustrate this best.

CRUCIBLE INDUCTION MELTING OF MAGNESIUM ALLOYS

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ABSTRACT

Melting magnesium alloys in a fossil fired furnace has always been a tough and demanding job. We live in an age when employers must be more responsible with regards to safety and environment in the workplace.

Induction melting can improve safety and environmental conditions in magnesium sand foundries. At the same time, the job of melting can be made easier, metal quality can improve and cost savings can be realized.

INTRODUCTION

When magnesium alloys are melted in an oil fired furnace, we see many drawbacks. The most obvious include,

1. Heat and smoke pollution.
2. Oxidation of surface metal by water produced from combustion.
3. Turbulent atmosphere above the melt.
4. Fear of crucible wall failure.
5. Pollution from melting flux and grain refining chemicals.
6. Worker fatigue and a general unattractiveness of the job.

Aluminum-like magnesium-is a reactive metal. When our foundry converted from oil to high frequency induction melting of aluminum alloys, eliminating the water associated with combustion, yielded less aluminum oxide (metal loss) and less hydrogen porosity.

That same water (from combustion) caused the magnesium to oxidize.



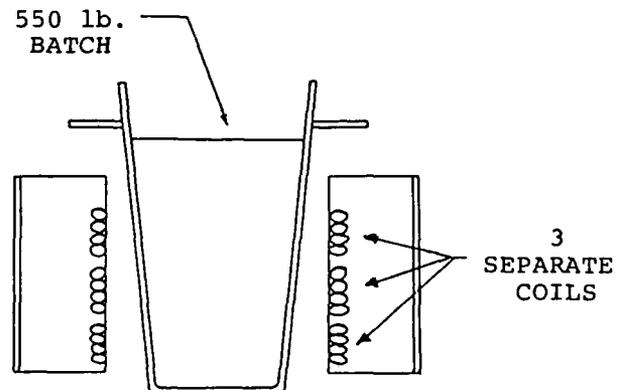
It also reacts with the magnesium melting flux, to produce corrosive hydrochloric acid.



Therefore, removing water vapor near molten magnesium should reduce production of magnesium oxide, hydrogen, and hydrochloric acid.

If we were to utilize high frequency induction melting for our magnesium requirements, we needed a unit that would melt batch sizes of 130 to 150 pounds.

The existing technology available, utilized line frequency (60 cycles), the smallest of which melted 550 pounds of magnesium. The unit employed the concept of sectional crucible heating, which was used to prevent crucible melt-down, with the molten magnesium acting as a coolant.



EXISTING TECHNOLOGY

FIGURE 1

The three separate coils (figure 1) would be energized in sequence, as metal was added to the crucible. With three separate coils, this unit could prove very costly. Therefore, we decided to use existing single phase high frequency (3,000 cycles) technology currently being used by our aluminum casting operation.

THE EXPERIMENTAL MELTS

Our initial experimental melt of magnesium was done on a Pillar - high frequency - 135 kilowatt lift swing induction furnace. We first placed a #80 clay graphite crucible into the lift swing. Forty pounds of alloy AZ91C was melted. SF₆ was used for protection of the molten magnesium (see figure 2).

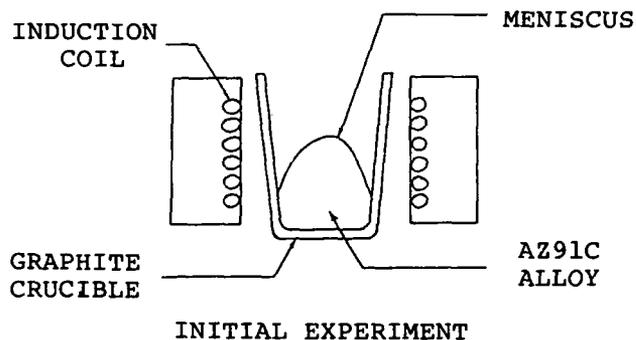


FIGURE 2

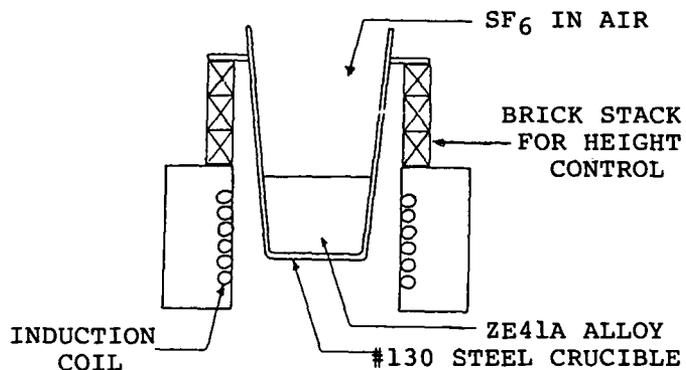
It was our observation that the magnesium melted much more rapidly as compared to an oil fired furnace. The stirring action induced by the secondary current swirled the melt around quite rapidly. Because magnesium is lightweight, this stirring action was greater than when we melted aluminum. Therefore, we restricted our power input to about 30% to prevent throwing the magnesium up over the top of the crucible.

We knew from past experience, that clay graphite crucibles were subject to failure from "shock cracks" and "hot spots", and we needed a crucible of great reliability. Magnesium reacts with many refractories, but has little affect on ferrous materials.

Our next step was to place a #130 (low carbon) steel crucible into the lift swing induction coil (figure #3). This time we would induce the secondary

current into the wall of the steel crucible, in order to melt the magnesium through heat transfer. At the same time, we would eliminate the rapid stirring action experienced with our initial melt in the clay graphite crucible.

Alloy ZE41A was used for this second experimental melt. SF₆ was again used for protection.



MOVING-CRUCIBLE EXPERIMENT

FIGURE 3

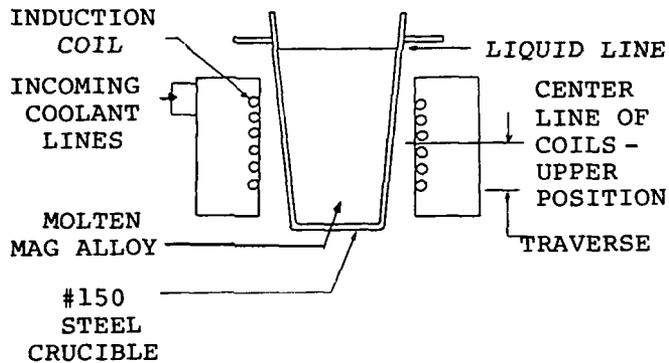
The steel crucible was positioned in the coil so as to heat the base of the crucible only; therefore, establishing a melt bath as rapidly as possible. The quicker the melt bath was established against the crucible wall, the less time we exposed the steel to a "red hot" state. This reduces wear and scaling of the steel crucible. Once the melt bath was established, the temperature of the steel was reduced by the magnesium "coolant".

As more metal was added to the crucible (and the liquid line began to rise), we lowered the crucible further into the induction field. This was done by removing bricks placed under the crucible arms. This "lowering" of the crucible kept the induction field in the middle of the melt. It also distributed wear and tear on the crucible.

During the melting operation, we noticed a small amount of stirring just inside the crucible walls. Upon completion of filling the crucible with magnesium, we made sure no stirring took place at the top of the melt. This was accomplished by keeping the top of the coil below the melt surface.

THE SCANNING INDUCTION MELTER

We had now acquired enough data to design a new melting furnace to suit our needs. The idea of moving the crucible up and down through the induction field proved cumbersome. We felt it would be easier to move (scan) the induction coil over the required distance. A hand crank mechanism was used.



SCANNING FURNACE

FIGURE 4

The power supply for the unit would be 100 kilowatts. The unit has a melting rate of approximately 200 pounds per hour (approximately 750 KWH per ton).

The power supply would be housed in a separate room with a remote control panel near the furnace. Placing the power supply into a separate room reduces exposure to SO_2 gas. Many magnesium sand foundries utilize sulfur in molding sands. In casting operations, these sands give off SO_2 which is very corrosive to copper wiring and plumbing. To insure pipes and wires in the power supply would not be affected by SO_2 , we used a special coating on them. Foam seals were used on the power supply and remote panel doors. Positive air pressure in the power supply room could also be employed.

The furnace pit employs a "bottomless" concept. If a crucible were to fail, enough depth is available so that metal can drop well below incoming coolant lines.

To avoid "thermal runaway", a temperature control interface is used in conjunction with the monitoring thermocouple.

SUMMARY OF ADVANTAGES AND DISADVANTAGES

The benefits of melting magnesium with induction seemed to be endless. The first was elimination of smoke pollution associated with the old fossil fuel method. Process heat was lowered dramatically, proving very beneficial on a hot summer day. The job of melting was now easier and the workplace safer and cleaner.

The noisy roar of the oil furnace was gone. With the induction furnace, some ringing of the crucible can be heard but lowers when metal is added to the steel crucible. Overall, noise levels are greatly reduced.

With the induction process, no water is produced, as when fuel is burned; therefore, less oxidation of the surface metal takes place. When melting flux is used, the resultant drier conditions require the use of far less flux. The lower humidity also generates less hydrogen absorption into the melt.

The calmer atmosphere over the melt makes it easier to utilize SF_6 mixtures for protection, and eliminates the need to fabricate hinged covers on the crucibles.

Crucible life is increased 2 to 3 times (through reduced scaling) as opposed to life in oil fires. The reduced scaling also keeps pot distoration down. Excessive heating of the crucible bottom can occur, however, if power is applied too rapidly. This problem can be remedied by leaving a 10 percent heel of metal in the crucible. This also increases melt rates of subsequent melts.

Anyone that has seen a crucible fail in an oil fire knows it can be a hair raising experience. The violent burning that can occur is partially caused by air from the blower rushing around the crucible. With induction, no rapid movement of air takes place. Crucibles can actually be used to their limit in the induction furnace.

Environmental hazards from flux and grain refining agents are still present. With the calm atmosphere and low process heat, we were now able to install a ventilation hood for exhausting gases from the work area.

Worker absenteeism is reduced. The same workers that ran the oil fires work steadier with the induction system - a direct result of the improved working and environmental conditions.

For zirconium containing alloys, quality was greatly enhanced. With the

reduction of process heat, the melter would not mind doing a longer stirring job, dissolving more zirconium into the melt. Typically, we used to get .4 to .6 percent Zirconium in our K1A alloy in the oil fire. We now average .6 to .8 percent Zirconium and have actually gotten as high as 1.0 percent in some melts. Similar results are obtained in alloy ZE41A, thus increasing our physical properties.

Tools, pouring basins, skimmers, and thermocouples used in the melting process are readily heated by placing them on top of the coil while power is being applied.

Temperature control with the induction melter is more precise than the oil fired furnace. It is easier to apply a percent input on the remote panel as opposed to adjusting a flame on the oil furnace. With induction we can manually hold a melt bath temperature to $\pm 3^{\circ}\text{F}$.

Maintenance of the unit involves checking the cooling system and applying a refractory wash to the coil. Downtime with induction has been no more than for oil.

Cost may be the biggest disadvantage of induction melting. Our first unit cost approximately \$100,000. Our second unit cost less because we were able to utilize a used power supply (previously used for melting aluminum).

THE FUTURE

To further improve the process of induction melting, we plan to experiment with crucibles of different alloys. Stainless alloys will scale less and conduct the electromagnetism differently.

In the future it may be possible to utilize induction in order to constantly stir alloys like K1A. This would keep zirconium in solution so that K1A may be die cast. (K1A is a high damping alloy, and is a unique application for valve covers in autos.)

Still higher frequencies for induction could yield less stirring action through the non conducting crucibles. Smaller more efficient power supplies are now available. The stirring action of induction could also be used for metal matrix compounds.

ACKNOWLEDGMENT

The officers of Yankee Casting Co., Inc. and Yankee Magcast Div. gratefully acknowledge Pillar Industries, Inc. in the joint effort put forth in developing

the scanning induction furnace.

AEROSPACE MAGNESIUM CASTINGS— CURRENT STATE OF THE ART

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Abstract

Magnesium aerospace castings were extensively used during World War II and continued thru the Vietnam War. After this many designs were switched to Aluminum and new design use in magnesium decreased. The magnesium industry has answered this challenge with alloy improvements in both strength and corrosion resistance. These developments, as well as improvements in foundry technology, will be covered in this paper. In addition better techniques for surface protection methods have been published and disseminated.

Magnesium And Magnesium Alloys are the lightest weight metal widely used as a structural engineering material. In addition to its light weight (0.065 lb/in³) it offers many other good engineering characteristics. These include easy machinability, very castable, good weldability and excellent dampening characteristics. It has moderate strength and is plentiful through out the world.

In view of these facts, why is it not more widely used? The major reason most cited is poor corrosion characteristics. Rated as electronegative to most metals, mating where moisture can become an electrolyte requires special techniques. Also, if bare metal is exposed to salt or other corrosive atmospheres high corrosion rates can result. Another impediment to extensive use is a Youngs modulus of 6.5×10^6 lb/in² where rarely have yield strengths over 30 KPSI been developed. This is in spite of the fact that on a strength to weight ratio it compares favorable with many aluminum and steel alloys. Recent foundry advancements and alloy developments in many areas demand a second look at this material for many engineering applications.

CASTING METHODS DEVELOPMENT

Many changes have been made in the way castings are made, processed and evaluated. In the mold making area the trend has been away from green sand molding and into core or dry sand molding. There are several reasons for this. The main one is better dimensional control and the ability to make complex shapes with greater dimensional accuracy. Green sand molds are soft and friable, where as dry sand molds are hard and resist abrasion. Sand from dry sand molds can now be reclaimed which keeps costs down.

Many new core making techniques have also enhanced the quality and capability of core making. One of these areas is the blowing of cores instead of hand ramming. This results in lower costs, better dimensional control and more dense cores. The new blowing equipment can produce both large and small cores.

New sand binders systems have led to many new no bake cores. The sand is mixed with a binder, a catalyst is added or induced causing the core to harden immediately or in a short time. Many of these binders evolve low gas, eliminating the need for core vents. They also have good strength and break down well after casting. Complex pipe core down to 3/16" can now be produced with supports spaced at about 8 to 12 inches.

New gaging techniques for locating cores accurately are being employed resulting in greater dimensional control and nearer net shape castings. This accuracy also allows the casting of thinner walls down to .100". On some parts depending on size, tolerances of ± 0.020 " and be obtained rather than the normal ± 0.030 ". The use of special sands such as zircon with a lower coefficient of expansion also aids in this dimensional control. These finer sands also give better surface finishes, down to 125 RMS in air passages and fluid flow areas.

Low pressure pouring techniques are also being employed for sand castings. The molten metal is put under pressure and forced up into a tube immersed in the melt. The mold sets on top of the furnace. Pressures are controlled to give smooth quiet mold filling. An over-pressure is applied after the mold is filled to aid feeding. The advantages of the process are; ability to run thinner walls, lower gate to casting ratios, lower pour temperature and provides oxide free metal in the casting. Another development in the melt and pour areas is the use of SF₆ (sulfur hexafluoride) as a cover gas in the melting and processing of magnesium alloys. This significantly reduces the use of flux and minimizes the possibility of flux inclusions in the casting. It also improves melt quality and overall cleanliness in the melt room.

Newer and better core removal methods have also been devised. Special vibratory techniques do the job with less manpower and reduced potential damage to the casting. Magnesium presented some special problems in this area due to the high dampening properties. Most of the sonic and resonant frequency units were ineffective. A random frequency impact unit was the only one found to work well.

With regard to gate and riser removal, many methods have been explored but the good old bandsaw still works the best. In removal of riser and gating stubs, high speed routers now remove metal rapidly without generating heat. Robotics have also come into the foundry. A unit can now be used to clean the casting, removing chill marks and core fins. A fixture is required to position the casting, but once done the robot takes over. It can change tools automatically for the various cleaning operations. Another place where robots have replaced man is in the area of sand blasting. The problems with silica dust and non-uniformity of blast are eliminated. The unit does the work quicker and doesn't need a coffee break. Different blast media can also be used for varied applications.

As mentioned before reclamation of the sands used to make the molds led to a significant increase in dry sand molding. Most of these sands were reclaimed by attrition methods. This left a less than prime quality sand and recovery rates were about 80%. The new thermal systems result in 95% recovery with sand quality as good as new. Maintenance on this new equipment is also greatly reduced. A cleaner environment is also maintained.

In metallurgical quality, separately cast test bars no longer fill the bill. Periodic destructive testing of castings to guarantee high strength levels is a must.

Statistical Process Control is also used in many areas of today's foundries. It can be used from the core room to the shipping dock

to assure product quality. Core preability and hardness as well as dimensions can be checked by statistical means. Casting dimensions and metallurgical quality can also be plotted and controlled.

Real time radiography is another area being looked at to control quality and reduce costs. At the present time investigation indicates that savings and quality would not be enhanced on complex parts produced in smaller quantities.

ALLOY DEVELOPMENT

One of the major causes for the decline in magnesium usage is its corrosion potential. In answer to this problem a new alloy was developed with corrosion characteristic in salt atmospheres similar to aluminum. This alloy is a high purity version of the old alloy AZ91C containing 9% Al and 0.75% zinc. The removal of heavy metal elements such as copper, iron and nickel to extremely low levels changes the corrosion rate in salt atmospheres by 100 fold. The processing of the castings after pour must also be controlled so as not to contaminate the surface. Standards and specifications have been issued for this material and they control the corrosion rate to a maximum of 50 mils per year.

Care must still be used in attaching this material to other metals but there is methodology for this which I will cover later. The point is the self corrosion of the material has been virtually eliminated but the galvanic potential with other material still exists.

EQ21A is another new alloy development. It is a lower silver version of QE22A which was the highest yield strength magnesium alloy with tensile strengths of 38 KPSI, yield strengths of 28 KPSI and 3% elongation. The alloy was developed when silver prices were about \$40.00 per oz. Room temperature properties are slightly lower but elevated temperature properties are slightly higher.

The newest alloy in the magnesium arsenal is WE54 a yttrium containing alloy developed for elevated temperature use. This alloy has room temperature properties similar to QE22 but retains high strength up to 580°F. Another virtue of the alloy is that its corrosion rate is similar to that of AZ91E. The main detractor for the alloy is that with the high cost of yttrium the alloy is expensive for normal use. It has also been found that at moderate temperature (400°F) for extended lengths of time (example 2000 hrs) there is a pronounced loss of ductility.

Several other new magnesium alloys are in the development phase. Magnesium Electron Ltd. is working on low cost alloys with elevated temperature capabilities for the automotive industry.

SURFACE PROTECTION

Major advancements have been made in surface protection schemes for magnesium alloys. There were good methods in the past but many of them were not rigorously followed.

Some of these were flawed and not as thorough as the newer systems. Design of castings should be such as to eliminate water traps. This can be done by core holes in areas where water would accumulate in normal operation. When connecting magnesium to dissimilar metals a wet assembly procedure should be used. This puts a buffer between the dissimilar metals. This applies to inserts, studs, bolts and washers. A surface sealing procedure should be followed due to the porosity of most anodic films. Sharp edges should be avoided as they are hard to protect.

Different protection schemes are advisable dependent upon the environments the parts are expected to see. There are three basic methods, one for minimum protection, one for aggressive environments and one for very aggressive environments. The methodology of each can be obtained from M.E.I., RD Box 251, Flemington, N.J. 08822. The title is "Surface Treatments For Magnesium Alloys".

METAL MATRIX COMPOSITES

Magnesium offers some excellent advantages in usage with most fibers. It is readily wettable to the fiber. This is essential to the modulus increase of the material.

There are two main approaches to producing magnesium/metal matrix composites. The first is with discontinuous fibers immersed in the melt. Here good wetting occurs and with proper pouring systems, uniform volume % fiber distribution occurs. The other technique involves using continuous fiber preforms. The preform is then infiltrated with molten magnesium. Greater preferred orientation strengths can be obtained this way but wetting problems are more difficult.

These materials are very hard, have good wear resistance and increase strengths by a factor of 2.

SUMMARY

The magnesium industry has not stood still in recent years. A great deal of work has been and is being done to solve user problems with magnesium alloys. New developments offer solutions to these problems as well opportunities for new applications. Foundry technology improvements now enable foundries to produce more complex castings with higher mechanical properties in the casting. New alloys and metal matrix composites offer

additional opportunities in usage.

As such magnesium should be considered for many application where weight is a prime concern. The industry has and will continue to address the problems of the many potential users of this material.

Ferrous Metals

IMPLEMENTING ADVANCED MATERIALS THROUGH INTEGRATED CASTING DESIGN AND MANUFACTURING (ICDM)

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ABSTRACT

Recent advances in Computer Aided Design, Computer Aided Analysis, and Advanced Material Technology have ushered the casting industry into a new era of component manufacturing. The days of multiple paper transfers of ideas and efforts, coupled with trial and error engineering are ending. The following paper reviews a system for complete integration of electronic transfer of efforts from initial product design, tooling construction, manufacture and testing, and validation of process capability. This is currently being used at CMI for developing such new component programs as Austempered Ductile Iron (ADI) connecting rods and High Silicon-Moly Ductile Iron exhaust manifolds.

The key of the successful implementation of ICDM rests in several areas:

1. Communication:

A cross disciplined Product Team from the customer and vendor representing such areas as Sales, Design, Quality, Accounting, and Manufacturing communicates throughout the phases of product development.

2. Computer Assistance:

Component ideas are developed on 2-D and 3-D CAD stations. Through the assistance of various analytical programs such as SWIFT, designs that reflect better casting practice are developed. This shortens the prototype and preproduction phases. Both prototype and production tooling is cut on the CAD systems. The pattern makers 3-D interpretation of a 2-D blueprint is a thing of the past.

Computer Numeric Control Coordinate Layout Machines verify both tooling and product process capability.

Finally, In Process Electronic Communication of data from such devices as spectrometers, microprocessors, temperature sensors, bar code readers, etc. help maintain the process capability designed into the part and manufacturing process.

3. Advance Material Development:

The demands are greater for stronger materials, thinner components, higher operating temperatures, and greater reliability. To shorten the time from new material development to implementation, more cost effective approaches to experimentation and parameter design are required. Such techniques as Taguchi Fractional Factorial Design of Experiments are being used at CMI for such things as chemistry and heat treatment optimization for Austempered Ductile Iron. Although traditional SPC techniques help maintain current levels of reliability, in order to obtain "World Class Standards" such as

under 200 parts per million, defective optimizing parameter design through changes is required. Such things as optimized chemistry, controlled nodule counts, clean microstructure, heat treatment cycles designed for selected chemistry, etc. will result in dramatic improvements in mechanical properties of ADI and part performance.

Integrated Casting Design Manufacture is total system representing the least cost approach to successful new product development. Couple this with optimized Advanced Materials, such as ADI, and you have a plan for continued growth in the Casting Industry through the next century.

FLOW CONTROL TECHNOLOGY FOR BOTTOM POURING FERROUS ALLOYS

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ABSTRACT

A recently completed study has identified reoxidation as the major contributor to the problem of oxide-macroinclusions in iron and steel castings. While reoxidation of the steel can take place at several points during the processing cycle, a particularly troublesome area is the flow of the molten steel from the time that it exits the bottom pour nozzle until it flows into the mold cavity. Two technologies are under development which offer promise in minimizing the formation of oxide macroinclusions and these will be described in detail.

The first area of concern is the point at which the metal exits the bottom pour nozzle. Due to tangential and torsional forces, the stream tends to be unstable and break up causing excessive exposure. Additionally, most foundries throttle the nozzle in trying to use the nozzle/stopper-rod assembly as a flow control valve. This causes flaring of the stream and further exposure to the atmosphere. A nozzle has been developed featuring a castellated entrance configuration which gradually changes to a cruciform configuration at the exit. A conventional stopper seating area is maintained. In addition, a refractory composition is used which minimizes build-up of alumina in the nozzle. Plant experiences show significant improvement in pouring stream integrity with these nozzles.

A second major location of inclusion formation is the gating system. Most large steel castings are bottom-gated using fireclay tile assemblies. Although these assemblies are satisfactory for many applications, they do require a significant amount of space in the molding flask and the assembly process itself is also labor intensive. Further, conventional gating systems rely only on flotation for inclusion removal. A pre-formed, drop-in, ceramic gating system has been developed to

simplify bottom gating of large castings and ingots combining both flotation and centrifugal processes for inclusion removal. The upper surfaces of the ceramic insert may be coated with tailored refractory materials in order to maximize inclusion capture and retention.

A RECENTLY COMPLETED RESEARCH PROJECT conducted at Southern Research Institute under the sponsorship of the Steel Founders' Society of America, a number of industrial sponsors, and the United States Department of Commerce has firmly established that the major source of oxide macroinclusions in steel castings is reoxidation of the metal during pouring and flow through the gating systems into the mold cavity. Because the same physical-chemical conditions that cause reoxidation defects in steel castings are present in ductile iron production, the results of this research can be extended to help eliminate dross formation.

One step in the processing sequence that is particularly prone to generate reoxidation products is improper bottom-pouring. Due to tangential and torsional forces on the exiting metal stream, flaring may result. A unique bottom-pour nozzle has been developed to greatly minimize this problem (1,2) and plant experiences will be described in the paper.

OXIDE MACROINCLUSIONS

For as long as steel castings have been machined, inclusions visible to the naked eye have been seen on the machined surface; predominantly on the cope surface. Sometimes the inclusions became pronounced enough to be identified in the as-cast surface.

In order to form, oxide macroinclusions must have a source of both oxygen and the metal that forms the oxide. Since steel castings contain elements that react strongly with oxygen, control of the oxygen sources is

the obvious key in oxide macroinclusion control. Oxygen available to react with the molten steel can come from air itself, moisture in refractories or molding sand, or directly from the reduction of oxides in refractories, slags or sands. A further source of macroinclusions is the mechanical entrapment of existing oxides such as refractories, slags and sands.

REOXIDATION - There is an overwhelming amount of evidence that reoxidation is a major cause of oxide macroinclusions. Reoxidation is the reaction of elements of the deoxidized steel with oxygen. This oxygen can come from the air, reactions with the slag or refractories or the reduction of the moisture in refractories or the mold. Reoxidation is progressive as shown schematically in Figure 1. The original deoxidation results in small oxide particles, typically 10 microns in size, composed exclusively of alumina (Al_2O_3) in aluminum deoxidized steels. Reoxidation progresses from the formation of large alumina particles to a multiphase globule to a single phase globule as the excess oxygen combines with the dissolved aluminum then silicon and then manganese and iron.(3).

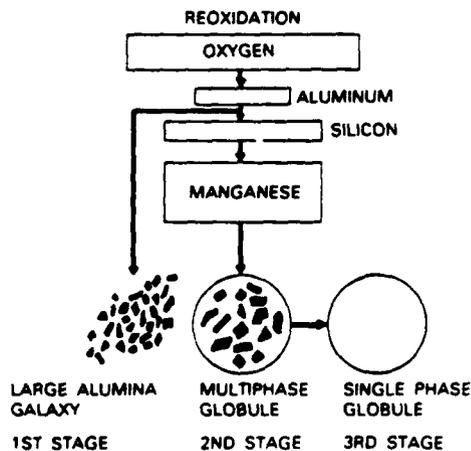


Fig.1 Stylized formation of inclusions in as-cast steel deoxidized with aluminum (>0.01% Al).

Examination of the inclusions yields important evidence of the importance of reoxidation. Typical oxide macroinclusions are whitish or greenish white powders. The composition of the inclusions are pure alumina crystals in a matrix of alumina (Al_2O_3) + silica (SiO_2) + manganese oxide (MnO). The pure alumina areas indicate that the inclusion started out as a product of a reaction of oxygen with the dissolved aluminum since pure alumina does not occur in most refractory products which are normally a mullite composition.

DROSS DEFECTS IN DUCTILE IRON - Flinn (4) states that if a ductile iron casting is cleaned carefully, a white powdery substance is often found on the cast surfaces. The defect is encountered on cope and sidewall surfaces, just as are the macroinclusions in steel. When ductile iron was first produced there was a good deal of astonishment at this defect, which was not present in gray iron and therefore traceable to the 0.04% Mg in ductile iron.

It can readily be seen that the physical chemistry involved in ductile iron dross formation is very similar to that of the formation of reoxidation defects in cast steels, and this has been well-documented in the literature.

FLOW CONTROL

One location of potentially severe reoxidation is that between the nozzle and pouring cup during conventional foundry bottom pouring. A conventional bottom pour nozzle/stopper rod system is an excellent "on-off" valve, but does not function well as a flow-control valve. Since the normal jobbing steel foundry operation involves multiple openings of the nozzle on each heat, and the pressure head in the ladle changes with each casting poured causing a variable flow rate, the normal condition observed is that of severe stream flaring between the nozzle and pouring cup. This flaring results in pouring stream breakup with the consequence of severe reoxidation.

A typical steel foundry bottom pour ladle is illustrated in Figure 2 and the flow patterns at the nozzle entrance are illustrated in Figure 3. The typical iron foundry automatic bottom pouring system is illustrated in Figure 4, and it is easily seen that the flow patterns are identical to those described above.(5)

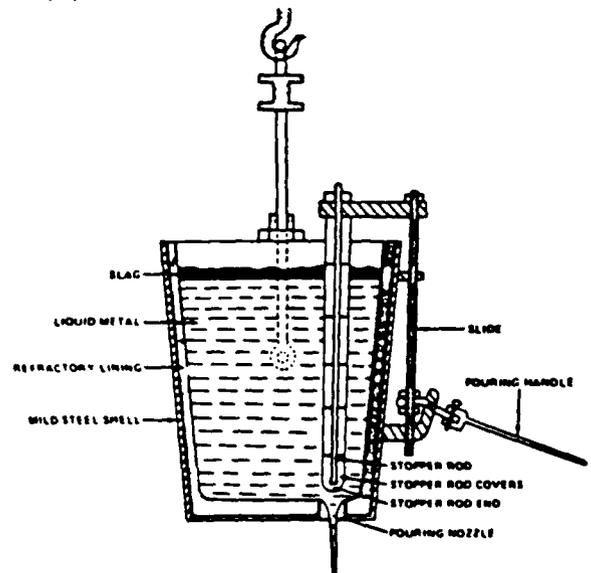


Fig. 2 - Bottom-pour ladle as used in steel foundries.

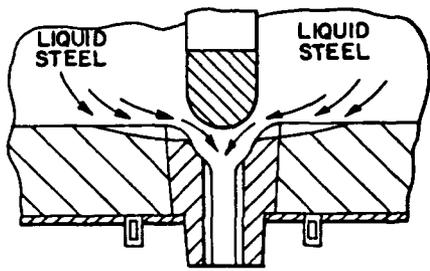


Fig. 3 - Metal flow patterns with bottom-pour nozzle and stopper rod assembly.

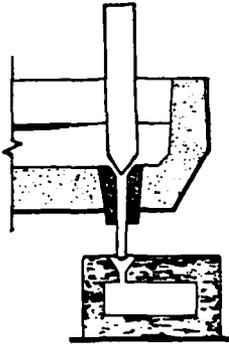


Fig. 4 - Typical automatic pouring system.

As mentioned previously, the bottom pour nozzle stopper system does not make a good flow control valve. The nozzle is typically smooth walled with limited extension past the seat area. In addition, except at full open position, the throttling of the nozzle causes significant stream breakup as shown in Figures 5 and 6. Because of the need to produce a number of castings out of one ladle of steel, and because of the variation of flow rate out of the nozzle determined by the amount of metal in the ladle (pressure head effect), throttling is common practice in steel casting production. Nozzle design can minimize stream breakup.

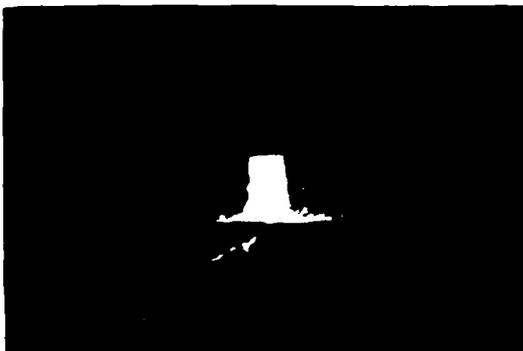


Fig. 5 - Conventional round bottom-pour nozzle pouring stream. Wide open @ 1/1000 sec.

One promising idea is the use of a non-circular cross section to avoid vortexing or stream breakup. A cruciform nozzle has been used with some reported success in Czechoslovakia. These nozzles are shown in Figure 7. This nozzle design was used by the author with some success in a United States foundry but no reliable source is presently available. This, in part, led to this development project.



Fig. 6 - Conventional round bottom-pour nozzle pouring stream. Throttled @ 1/1000 sec

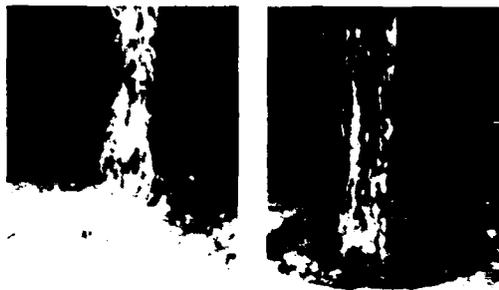


Fig. 7 - Nozzles with round and cruciform openings having same surface area.

Another modification that has been used to avoid stream breakup and air entrainment is the use of a castellated nozzle as shown in Figure 8. (6) The normal circular nozzle seat area is maintained in either geometrical modification, but the design change is made for the purpose of avoiding stream breakup. A nozzle design that combines both of these elements offers an advantage over the standard nozzle that is now commonly used.

CURRENT NOZZLE DEVELOPMENT - Based on the above consideration, an advanced bottom-pour nozzle was developed for steel foundry applications. The nozzles are made of fused silica using a proprietary high-energy compaction process and a high-temperature ceramic bond. The use of fused silica eliminates the build-up of alumina particles in the nozzle throat when pouring aluminum killed steels (2). Photographs of the exit and entrance of the new nozzle design are shown in Figures 9 and 10. They include both the castellations before the seat area and the cruciform cross section on

a. Pouring Stream without and with Castellated Nozzle.



b. Castellated Nozzle before and after Use.



Fig. 8 - Reduction of air entrainment of a pouring stream by use of castellated nozzle.



Fig. 9 - Exit of improved design nozzle.

the discharge from the seat area. This design is intended to achieve a pouring stream superior to conventional nozzle designs, and this has been proven in plant trials. An example of a stream from a throttled stopper rod-nozzle assembly using the new design is shown in Figure 11. This photograph compares directly with that of Figure 6; in fact, it was from the same foundry.

This nozzle design is now available in several sizes and is being used in a number of North American steel foundries. Development continues to refine the metal flow characteristics and evaluate refractory properties (7).

NOZZLE FOR DUCTILE IRON APPLICATIONS - Due to the similarity of the oxide formation problem in ductile iron operations, consider-

able interest has been shown by this segment of the industry. This is particularly true with automatic pouring systems which use a conventional stopper rod/nozzle assembly to control flow rate based on electronic and/or laser detection and measurement of liquid metal level in the mold (5). Observations by the author have indicated considerable flaring of the stream when the stopper rod is adjusted to control flow rate, resulting in splashing and severe stream breakup.

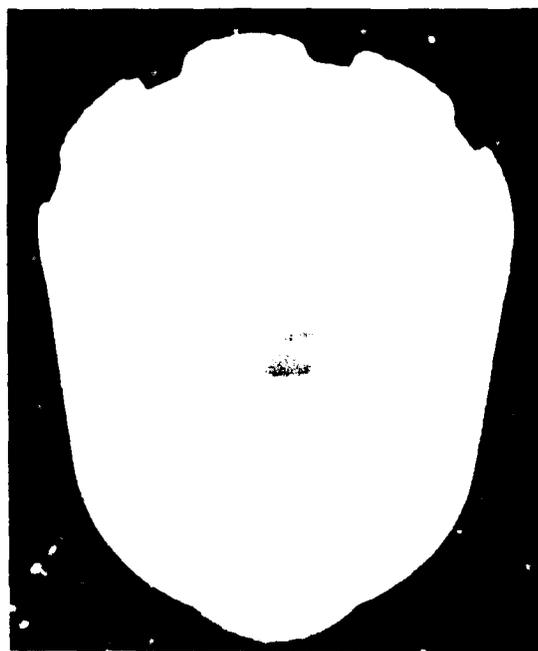


Fig. 10 - Entrance of improved design nozzle.

Based on the above observations, a smaller size nozzle (1 in. dia. equivalent) was developed to fit commonly used automatic pouring systems. Due to the smaller size restrictions, the castellated entrance could not be accommodated, however, the cruciform exit is retained. The initial nozzles were manufactured using the same fused silica refractory as was used with the nozzles for steel applications. It was realized that the fused silica composition would be unlikely to provide the life required in this application however the intent was to verify the nozzle design parameters for ductile iron use.

Initial trials validated the above reasoning. A pour of 2500 lbs of ductile iron was made to verify that the same flow characteristics that were observed with steel would be realized with ductile iron. The metal (3.65 - 3.75 C, 2.45 Si) was poured at a temperature of 2650°F (1454°C). The pour stream remained compact throughout the duration of the pour and no flaring was observed. In addition, virtually no erosion of the cruciform exit was observed.

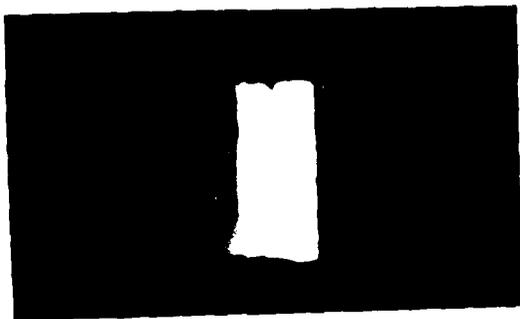


Fig. 11 - Stream from improved design nozzle. Throttled @ 1/1000 sec. (Direct comparison with Fig.6).

Based on the above results, nozzles were made of high-purity tabular alumina. These were successfully used in a high-speed, automated molding line averaging 3500 openings over a 9.5 hour shift without noticeable erosion or stream deterioration.

GATING SYSTEM DESIGNS - At least in larger sizes, the majority of steel castings are poured using a conventional bottom gating system (Figure 12). As described in the literature (8) this approach depends on flotation for inclusion removal. This further then requires a relatively long linear flow distance and a low linear velocity in the horizontal runner. Although this is a perfectly acceptable approach, it does involve two difficulties:

1. The required horizontal flow distance required considerable flask space.
2. The gating system must be assembled from many individual refractory holloware shapes requiring considerable expenditure of manpower, being subject to misalignment, and also being subject to human error in assembly.

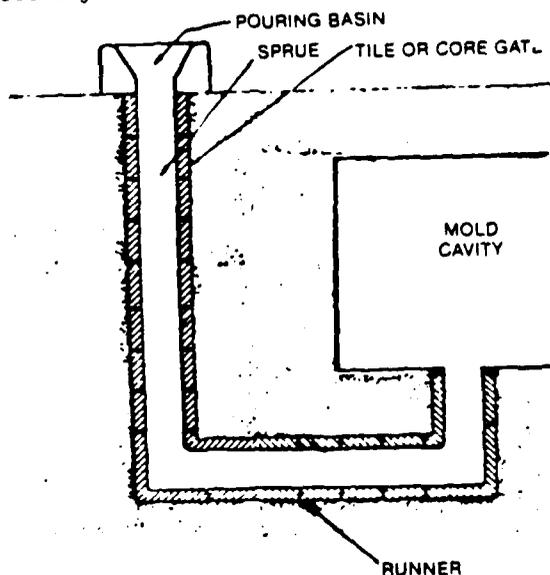


Fig. 12 - Schematic drawing of typical bottom gating system used for casting steel.

In an effort to circumvent these problems, a patented pre-formed refractory gating system insert has been developed (9). As shown schematically in Figure 13, the insert consists of a central, vertical sprue channel surrounded by a spiral runner. The nature of the runner cross section and spiral are such as to optimize inclusion capture, incorporating both centrifugal and flotation mechanisms of inclusion removal.

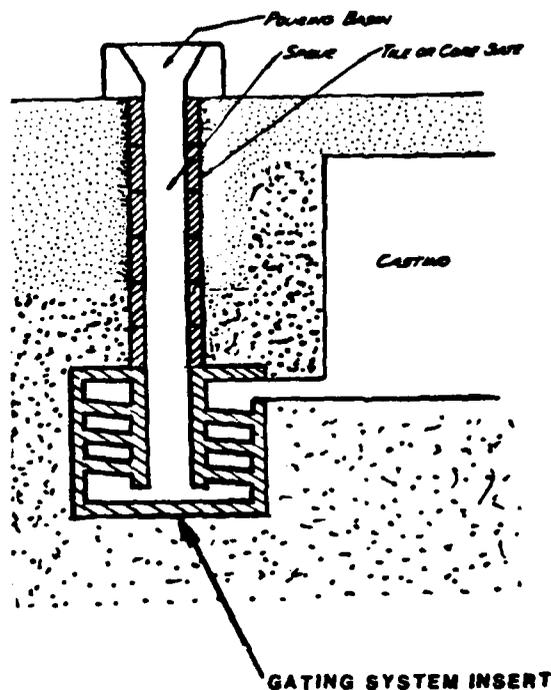


Fig. 13 - Gating system insert.

The gating system insert is assembled from individual flights incorporating the above characteristics as shown in Figure 14. Plant trials in a number of foundries have shown that macroinclusions can effectively be trapped against the upper surface of the flow channel. Development work continues on arriving at optimum refractory compositions and using refractory coatings to maximize interfacial attraction with respect to the anticipated inclusion chemistry.

One objection to the use of the gating system insert is that it may result in a reduction in casting yield (shipped weight/poured weight). It is desirable to rig several castings off of the centrally located gating insert to improve the yield situation. Also as casting size (pouring weight) increases the factor becomes of less importance. Along these lines, an insert which connects directly to downsprue tile of 4 in to 5 in. is being developed. In this size range the improvement in cleanliness far outweighs any loss in yield.

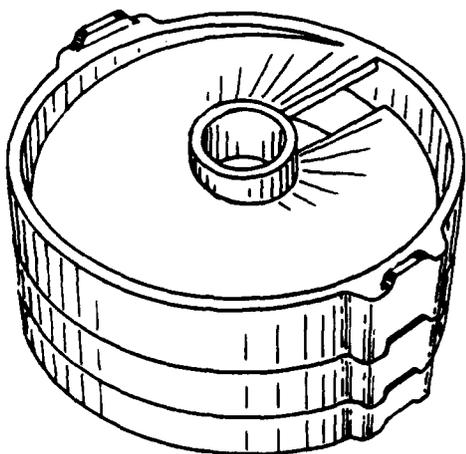


Fig. 14 - Schematic representation of gating system insert assembly.

CONCLUSIONS

Recent research in flow control as applied to the pouring of ferrous castings has resulted in the following:

Nozzles - Initial development of a laminar flow, bottom-pour nozzle has resulted in a nozzle which reduces flaring to a minimum when the nozzle is throttled as a consequence of the variables normally encountered in a jobbing iron or steel foundry. This nozzle also is resistant to nozzle blockage due to the precipitation of alumina in the throat/seal area. The refractory/binder system has proven very resistant to erosion and also thermally stable at the temperatures encountered. These characteristics all lead to a reduction of oxide macroinclusions in steel castings and gross defects in iron castings.

Gating System - A preformed refractory gating system insert has been developed which incorporates both the centrifugal and flotation mechanisms of inclusion removal. Various refractory compositions are available and the surface chemistry of the interface may be modified by appropriate choice of mold coating.

Work continues on both of these flow control devices to further refine design and materials.

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NUMERICAL SIMULATION OF FLUID FLOW AND HEAT TRANSFER FOR THE FILLING OF CASTING

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Abstract

A computer model has been developed based on a computational fluid dynamics technique, called SOLA-VOF, and the principle of the conservation of thermal energy for the 2-D fluid flow analysis and 3-D heat transfer analysis for the filling of casting. It can simulate the flow pattern of the molten metal and the corresponding temperature variation of the molten metal and the mold during the filling of thin section castings. Four simple square castings, all with different set up/mold material, were tested with the model. The results are presented. Attempts were also made to simulate the 3-D fluid flow phenomena in a simple block casting with the SOLA-MAC technique. The preliminary results are also shown.

Introduction

The development of computer aided engineering analysis (CAE) systems for castings is occurring at a rapid rate. The overall objectives are to reduce the reliance on the art and experience of the individual designer and replace it with the systematic analysis of scientific principles and the accumulated experience of many foundry engineers.

It is known that in the casting process many physical phenomena occur simultaneously or in a rapid succession. The casting is initially empty and then the molten metal is poured into the mold cavity through the gating system. At the same time during filling, the heat transfer occurs through conduction and convection. As the casting is filled, heat transfer then dominates. The molten metal loses its heat through the mold and solidifies and subsequently cools down. The relationships between these phenomena and how they affect the quality and cost of the castings are thus major concerns to the foundry engineers. Fluid flow, heat transfer and stress generation are recognized as the most important effects among these phenomena. A complete and practical CAE system should therefore, consider each of the factors involved.

Reviewing the literature, it is noted that extensive effort has been made on heat transfer analysis for solidification. Fluid flow analysis, especially for the filling of casting, has also started to receive attention in recent years (1-3). A family of

computational fluid dynamics techniques, called the Marker-And-Cell (MAC)(4), Simplified Marker-And-Cell (SMAC)(5) and Solution Algorithm - Volume of Fluid (SOLA-VOF) (6), have been adapted for this purpose. These techniques with capability of treating free surface flow were chosen because the fluid flow phenomena during the filling of casting is highly transient with the domain of the molten metal expanding and its boundary to be determined. The applications have shown considerable success. It was possible to obtain important information such as the flow pattern, velocity profile, extent of impact on the mold, extent of flow violence, and last filled area for the gas to be squeezed to.

Despite the encouraging results, very little effort has been made on the heat transfer analysis for the filling of castings. It will be very desirable for it to be conducted because it can provide very important information such as the proper superheat of molten metal to pour and the temperature distribution of the molten metal and the mold right after the casting is filled for the subsequent solidification analysis.

When examining the characteristics of the heat transfer phenomena during the filling of casting, it is known to be a combined mode of (1) advective heat transfer in the melt through the bulk fluid motion, (2) convective heat transfer between the moving melt surface and the atmosphere and between the melt and the mold, (3) conductive heat transfer in the melt and in the mold. Moreover, to evaluate the amount of heat transfer through bulk fluid motion, the amount of fluid in and out of every cell needs to be known. In MAC and SMAC, this will be rather difficult to do for certain cells such as surface cells.

The first part of this study was to adapt SOLA-VOF for the 2-D fluid flow analysis. It then employs the principle of the conservation of thermal energy and makes use of the information obtained from the SOLA-VOF analysis, most importantly the amount of fluid in and out of every cell, for the 3-D heat transfer analysis to obtain the flow pattern as well as the temperature variation of the molten metal and the mold during the filling of casting.

Due to the constraint of the 2-D fluid flow analysis, the applications of the above model were limited to castings of uniformly thin cross sections. To improve the capability of the model, attempts were also made in this study to extend the fluid flow analysis to three-dimension to simulate the filling phenomena of more actual castings. The methodology is similar to the 2-D cases. However, the complexity should be appreciated.

Mathematical Models

Fluid Flow Calculations

A set of conservation equations of mass and momentum were solved numerically. The equations are as follows. Continuity equation (conservation of mass)

$$D = \frac{\partial u}{\partial x} + \frac{\partial v}{\partial y} + \frac{\partial w}{\partial z} = 0 \quad (1)$$

Navier-Stokes equations (conservation of momentum)

$$\frac{\partial u}{\partial t} + u \frac{\partial u}{\partial x} + v \frac{\partial u}{\partial y} + w \frac{\partial u}{\partial z} = -\frac{\partial P}{\partial x} + \nu \left(\frac{\partial^2 u}{\partial x^2} + \frac{\partial^2 u}{\partial y^2} + \frac{\partial^2 u}{\partial z^2} \right) \quad (2)$$

$$\frac{\partial v}{\partial t} + u \frac{\partial v}{\partial x} + v \frac{\partial v}{\partial y} + w \frac{\partial v}{\partial z} = -\frac{\partial P}{\partial y} + \nu \left(\frac{\partial^2 v}{\partial x^2} + \frac{\partial^2 v}{\partial y^2} + \frac{\partial^2 v}{\partial z^2} \right) \quad (3)$$

$$\frac{\partial w}{\partial t} + u \frac{\partial w}{\partial x} + v \frac{\partial w}{\partial y} + w \frac{\partial w}{\partial z} = -\frac{\partial P}{\partial z} + \nu \left(\frac{\partial^2 w}{\partial x^2} + \frac{\partial^2 w}{\partial y^2} + \frac{\partial^2 w}{\partial z^2} \right) + g \quad (4)$$

where D : divergence

u : x component of the velocity vector

v : y component of the velocity vector

w : z component of the velocity vector

P : pressure / density

ν : kinematic viscosity

g : the gravitational constants

In 2-D fluid flow calculations, w is assumed to be zero and u, v are not functions of z. When solving these equations for u, v, w, and P a SOLA scheme was used. It derives a pressure adjustment term as follows.

$$\delta P = -D / \left(\frac{\partial D}{\partial P} \right) \quad (5)$$

Initially, a tentative velocity and pressure field is introduced into the system. While not satisfying the zero

divergence requirement, a pressure adjustment term, δP , is calculated based on equation (5) and the amount of divergence. The pressure adjustment terms are then used to adjust the tentative pressure field and subsequently the tentative velocity field. The process is iterated until zero divergence is satisfied for all full cells. The advantage of this algorithm is that pressure and velocity are both included in the iteration and thus converges faster.

An important problem in computing transient flows having free surface is to monitor the evolution of the fluid domain. As mentioned earlier, SOLA-VOF was employed for the 2-D fluid flow calculations. In SOLA-VOF, a fluid function value $F(x,y,t)$, is designated to each cell, unity for full cells, zero for empty cells, and numbers between zero and unity for surface cells. After the velocity field is obtained, the amount of fluid transported in and out of a cell is estimated. The F value originally designated to that cell is then varied based on the net change of fluid transported. The new F distribution then forms a new fluid configuration. Due to the fact that SOLA-VOF recognizes the evolution of fluid domain based on the variation in F value for each cell, it is then important to estimate as closely as possible the amount of F transported in and out of a cell based on the calculated velocity field. In SOLA-VOF, a Donor-Acceptor scheme is employed precisely for this need. With the amount of fluid in and out of each face of a cell estimated, it is also then possible to calculate the amount of thermal energy advected through each face of a cell. Details of the scheme can be found in reference (6).

In 3-D fluid flow calculations, a simpler marker scheme was employed to monitor the evolution of the fluid domain. To alleviate the heavy load on computer memory and computation due to the large number of markers used in the scheme, a marker rearrangement algorithm was employed. Details of the algorithm can be found in reference (7).

Heat Transfer Calculation

To account for the heat transfer during filling, a conservation of thermal energy principle is employed. The thermal energy in and out of each face of a cell is transported through

conduction / convection. The conductive heat flux, whether it is within fluid or within mold, can be evaluated with the Fourier's First Law. The convective heat transfer between melt and its surrounding (either atmosphere or mold) can be calculated with the Newtonian cooling law. The advective heat transfer through bulk fluid motion, however, can be calculated only when the amount of fluid in and out of a cell can be evaluated, which is precisely one of the major parameters included in the SOLA-VOF calculation. In mathematical form, it can be expressed as follows.

$$Q = C_p \cdot \rho \cdot \Delta V \cdot \delta F \cdot T \quad (6)$$

where Q : thermal energy advected through a cell side in that time step
 ρ : density
 C_p : heat capacity
 T : temperature of the donor fluid
 δF : amount of F transported
 ΔV : volume of the donor cell

As the thermal energy in and out of every cell side / melt surface is calculated, the net thermal energy change is then used to compute temperature variation of the molten metal or the mold in that cell.

Results and Discussion

To illustrate how the techniques can be applied to predict the flow pattern and temperature variation of the molten metal and the mold during the filling of castings, four casting cases were tested.

The first two cases are horizontal castings with thin cross section. The geometry and dimensions are shown in Figure 1. The molten metal enters through the ingate on the left with inflow velocity, 500 cm/sec. These two cases have the same casting configuration, and conditions. However, the first case uses metallic mold, while the second case uses sand mold. The cross section of the castings is assumed to be equal to one cell dimension, so that the flow phenomena can be treated in two-dimensions, while heat transfer is treated in three dimensions.

First, the flow pattern and velocity profile for these two horizontal casting calculated with the SOLA-VOF technique are the same as

shown in Figure 2. In the beginning, the molten metal is at the ingate, ready to enter the casting. As the simulation shows, the melt enters in a straight fashion, hits the wall on the right, and then expands towards the two side walls. At time = 0.08 second, the molten metal is seen to move backward towards the ingate, and cling close to the two side walls, then to reach the wall on the left. At time = 0.12 second, the front moving melt is seen to meet the inflow stream, and two pockets are formed in the center on both sides of the incoming metal stream. The pockets are gradually filled from the left and at time = 0.198 second, the casting is completely filled.

The corresponding temperature distribution of the molten metal and the mold in the same layer for the case of metallic mold is shown in Figure 3. The melt enters the castings at 700 degree C. Color graphics is used to display the result of the temperature calculation. Different colors are used to represent different temperature ranges. At time = 0.04 second, it is seen that the molten metal has lowered its temperature. At time = 0.08 second, the front running melt along the two side wall and at the corners has cooled down quite significantly and the mold opposite to the ingate has been heated up. At time = 0.10 second, the melt front has further cooled down while the melt temperature at the centerline and near the opposite wall has remained relatively unchanged. The mold near the two side walls also raises its temperature. At time = 0.12 second, part of the melt temperature in front of the ingate has cooled down. At time = 0.13 second, the melt temperature has shown a rather wide variation. At time = 0.16 second, two spots near the ingate and in the central region are the coldest. At the end of filling, the central spots are seen to raise its temperature while the melt near the ingate and at the corners are the coldest. The surrounding mold has all been heated up.

The same temperature calculation was made for the horizontal casting with sand mold. The results are shown in Figure 4. The color graphs demonstrate that the temperatures of the molten metal and the mold do not show noticeable change during filling.

The third and fourth cases are vertical castings with thin cross section. The geometry and dimensions are shown in Figure 5. They are similar to the horizontal castings except gravity now plays an important role in the flow behavior of the molten metal and subsequently in the heat transfer process.

The flow pattern and velocity profile calculated with the SOLA-VOF technique for the vertical casting is presented in Figure 6. At time = 0 second, the molten metal is seen at the ingate, ready to enter the casting with inflow velocity 110 cm/sec. At time = 0.1 second, the molten metal enters the casting like a bore, then hits the wall on the back, and jumps up high. At time = 0.45 second, the molten metal move towards the ingate. At time = 0.55 second, the molten metal reaches the wall on the front. At time = 0.6 second, the molten metal climbs upward along the front wall, and a vortex develops near the ingate. At time = 0.65 second, the molten metal has almost filled the casting.

The corresponding temperature distribution of the molten metal and the mold in the same layer for the metallic mold is shown in Figure 7. The melt enters the casting at 700 degree C. At time = 0.1 second, the melt tip has cooled down. At time = 0.25 second, the front running melt has cooled down quite significantly while the mold at the bottom and on top of the ingate has been heated up. At time = 0.452 second, the molten metal has reduced its temperature from the ingate to the opposite wall in a shell like manner. At time = 0.55 second, the metal near the top wall is the coldest. At time = 0.60 second, the distribution of the melt temperature still exhibits in a shell like manner. Near the end of filling, it can be seen that the hot metal stream just exits from the ingate is surrounded by several shells of melt with temperature reduced and the outmost melt is the coldest.

In Figure 8, it shows the temperature variation of the molten metal and the mold for the vertical casting with sand mold. It again shows no significant temperature changes has occurred to the molten metal and the mold during filling.

For the 3-D fluid flow calculations, a simple block casting was tested. The geometry and

dimensions are shown in Figure 9(a). The molten metal enters the mold through an ingate on the bottom with inflow velocity 100 cm/sec. Due to axial symmetry, calculation was performed only in one quarter of the casting as shown in Figure 9(b). The computed 3-D flow patterns and velocity profiles are presented in 2-D sections. Figure 10 shows the 2-D velocity profile on the Y-Z plane near the central plane of the casting for eight different times. Figure 11 shows the 2-D velocity profile on the X-Z plane near the central plane of the casting for the same times. Figure 12 shows the 2-D velocity profile on the X-Y plane on the bottom of the casting for the same times. It should be realized that the presentation of 3-D flow pattern is a rather difficult task. It takes some imagination to visualize what is happening from the 2-D sectioned plots.

Conclusion

By the SOLA-VOF technique and energy balance method, the fluid flow and heat transfer analysis can be conducted for the filling of castings. The result of the combined analysis can provide the information, such as the flow pattern, the corresponding velocity profile, and the temperature variation of the molten metal and the mold. The information are important to the design of casting and the subsequent solidification analysis.

By the SOLA-MAC technique, calculations for 3-D filling pattern can be attempted. The results were preliminary. The complexity should, however, be appreciated.

Acknowledgement

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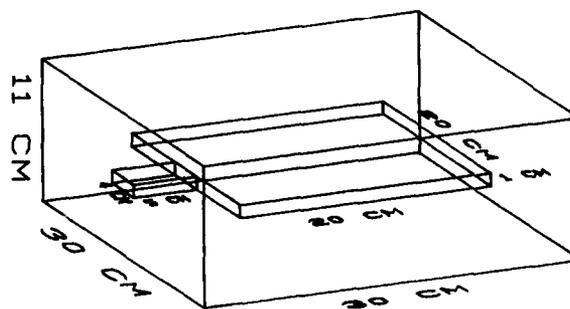
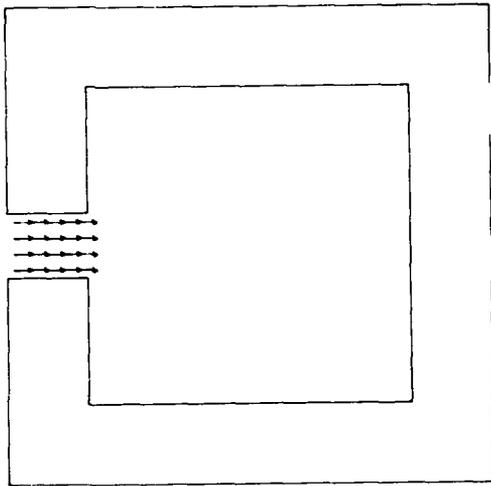
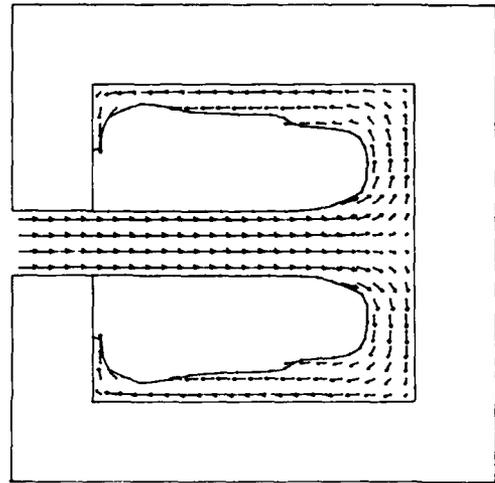


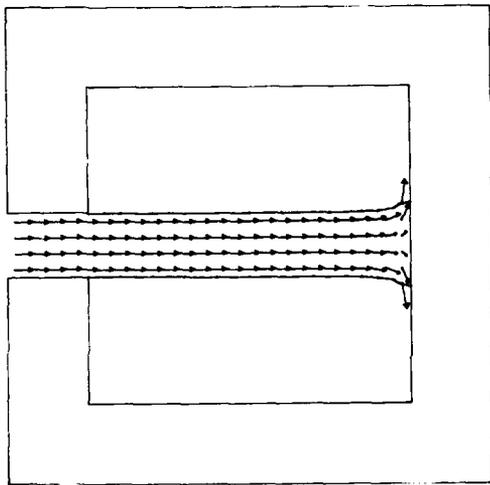
Figure 1. The geometry and dimensions of the horizontal castings



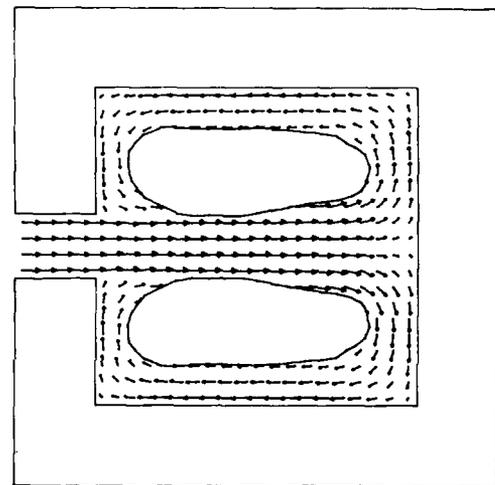
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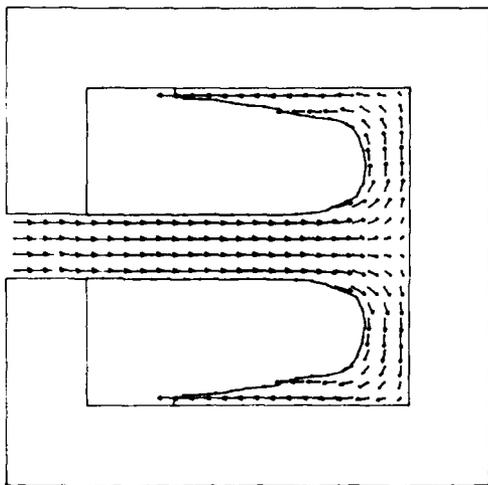
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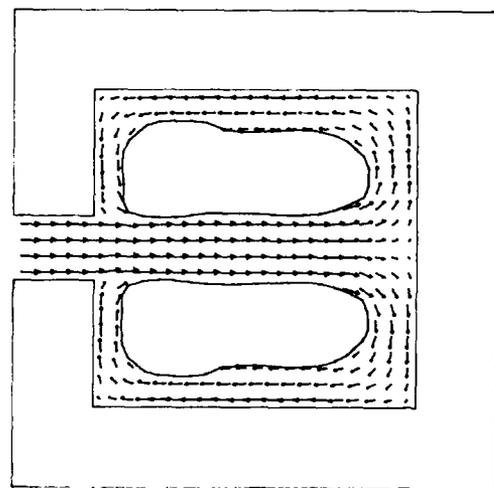
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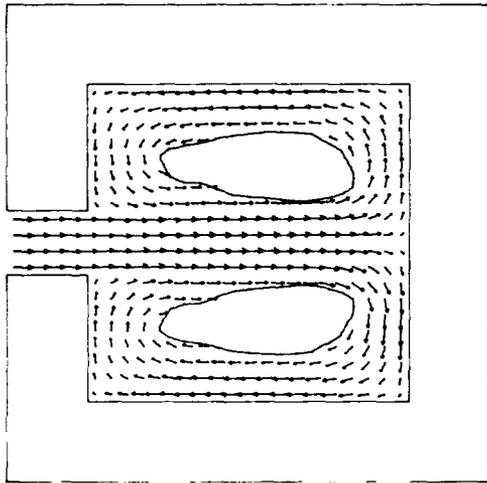
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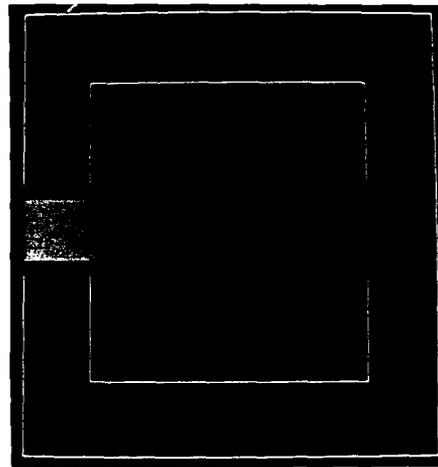
(c) Time = 0.08 second



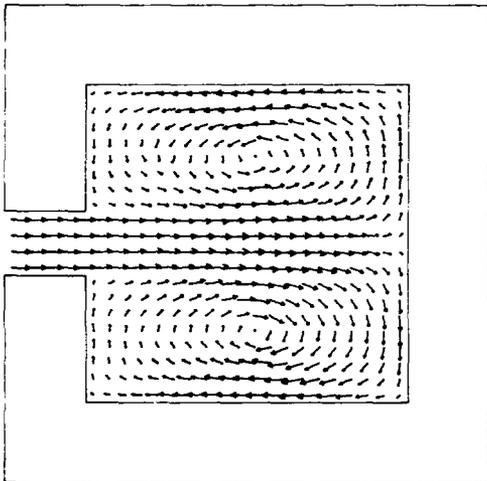
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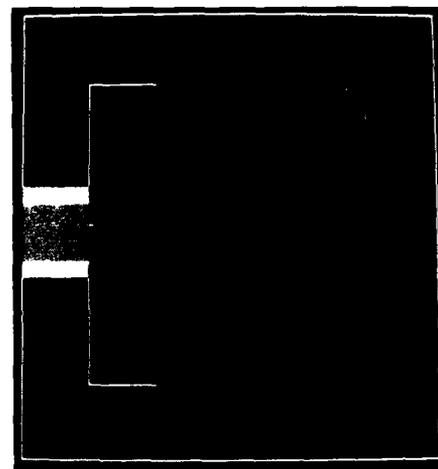
(g) Time = 0.16 second



(b) Time = 0.04 second

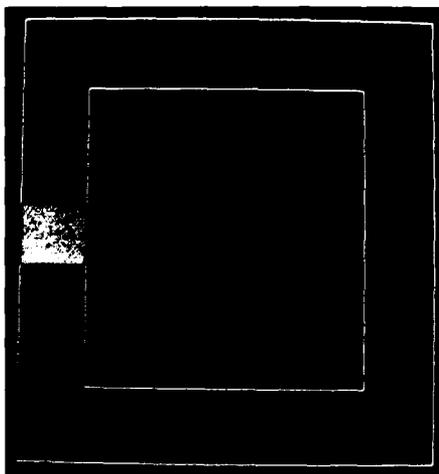


(h) Time = 0.198 second

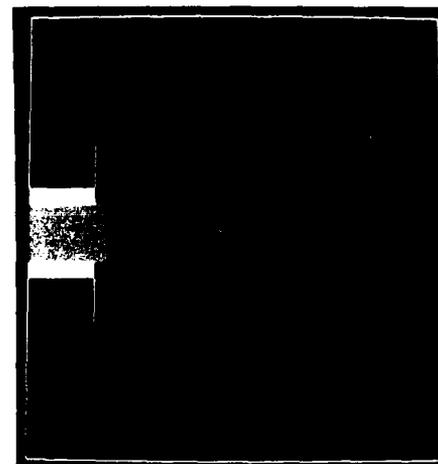


(c) Time = 0.08 second

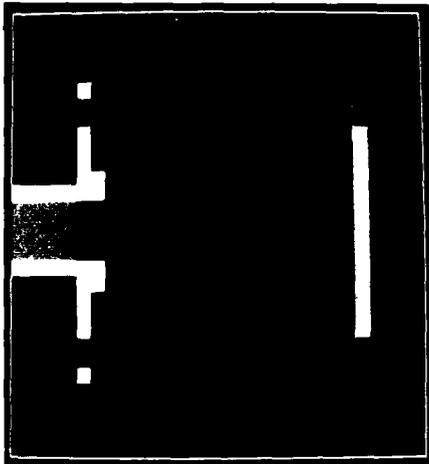
Figure 2. The flow pattern of filling the horizontal castings



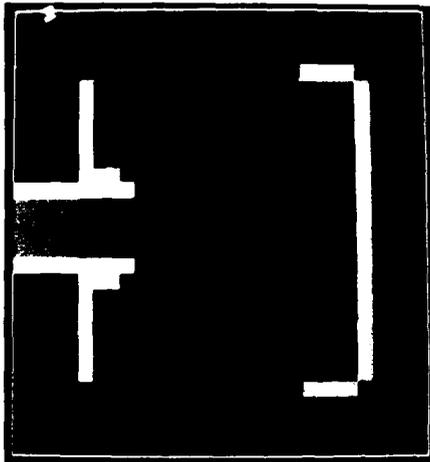
(a) Time = 0.00 second



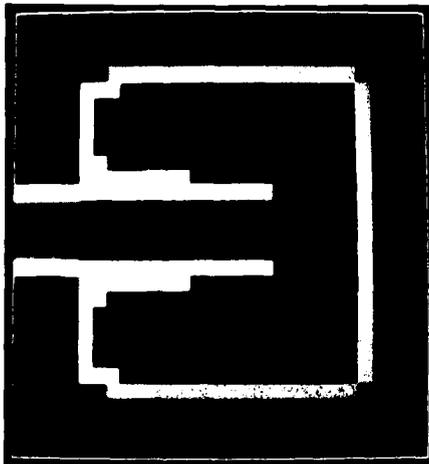
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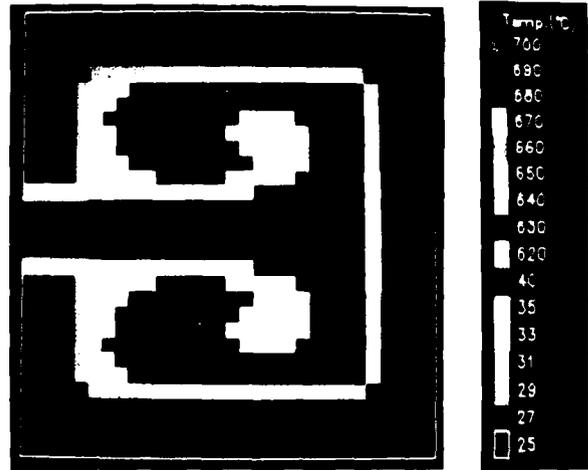
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(f) Time = 0.13 second

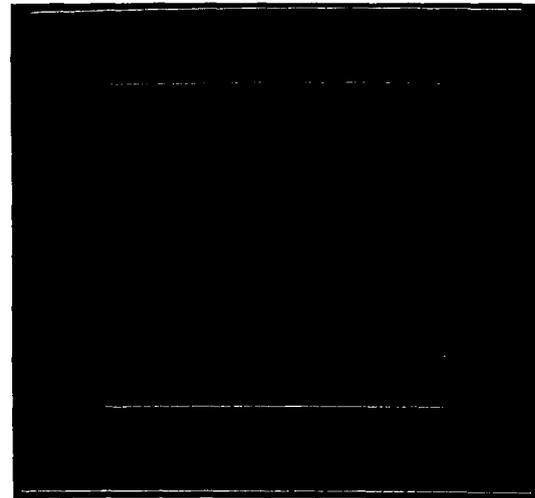


(g) Time = 0.16 second

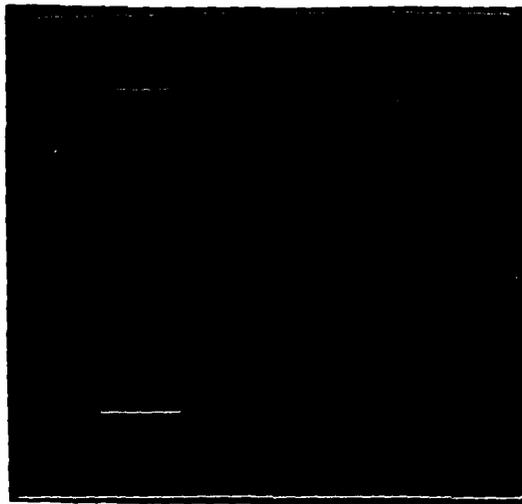


(h) Time = 0.198 second

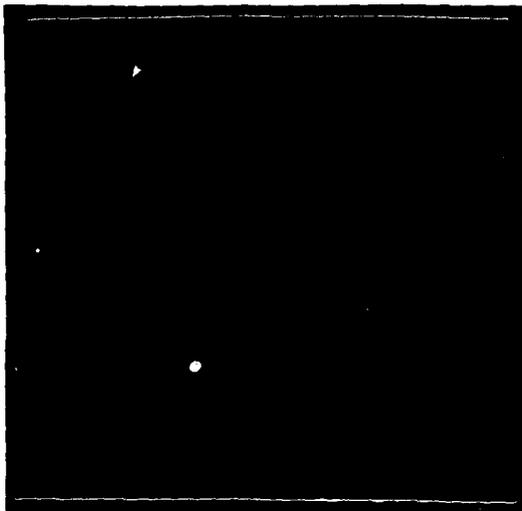
Figure 3. The temperature distribution of the molten metal and the mold for the filling of a horizontal casting with sand mold.



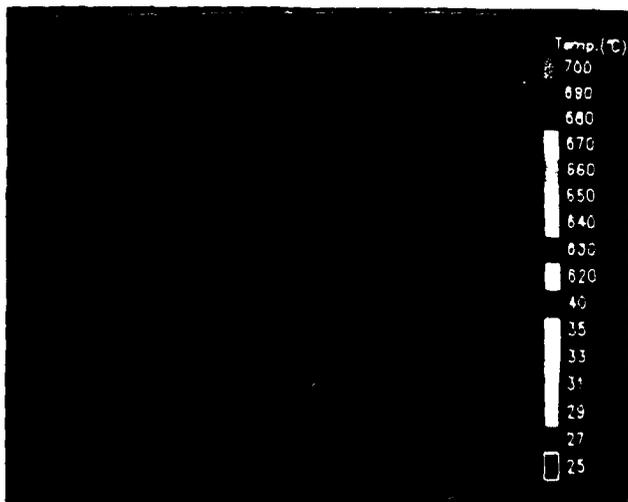
(a) Time = 0.04 second



(b) Time = 0.08 second



(c) Time = 0.13 second



(d) Time = 0.198 second

Figure 4. The temperature distribution of the molten metal and the mold for the filling of a horizontal casting with sand mold.

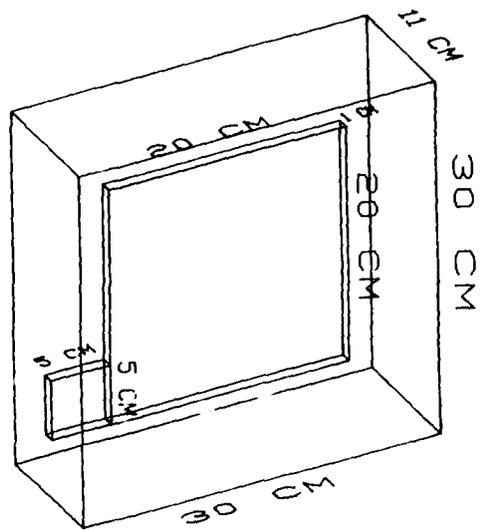
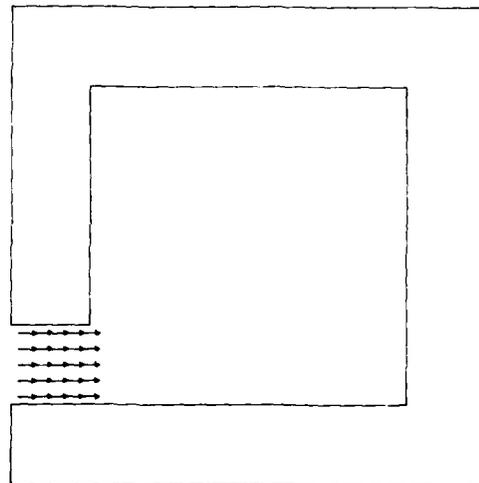
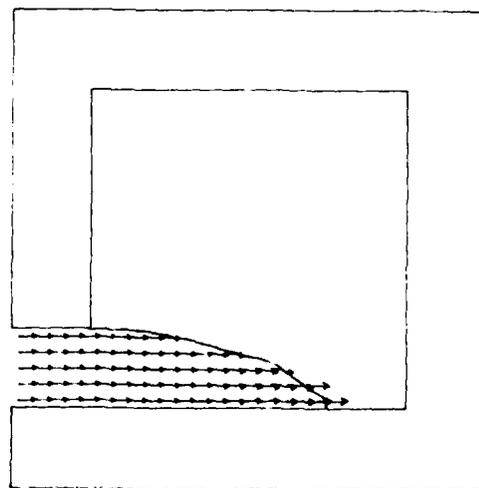


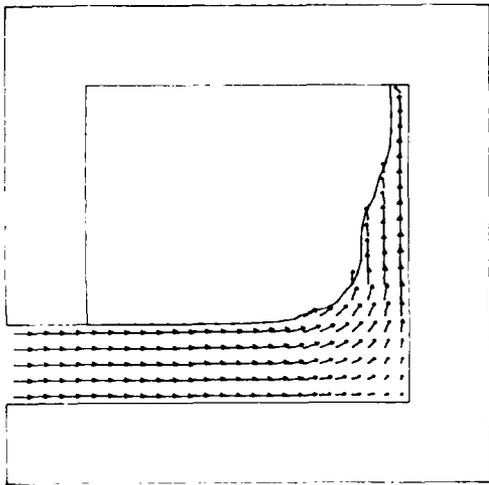
Figure 5. The geometry and dimensions of the vertical castings



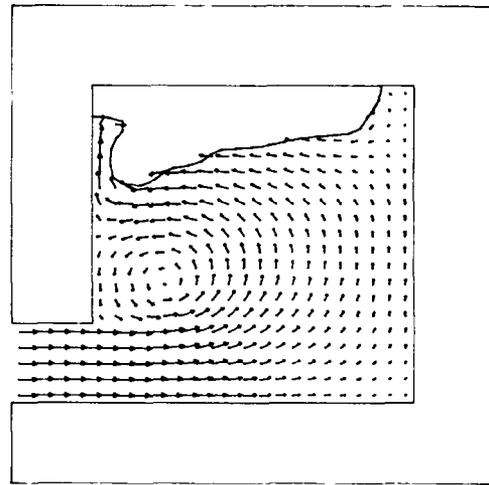
(a) Time = 0.00 second



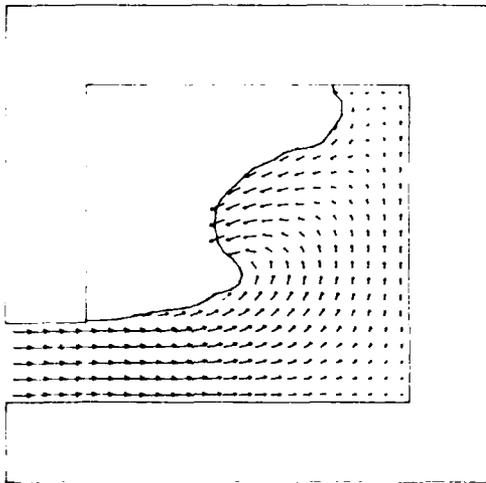
(b) Time = 0.10 second



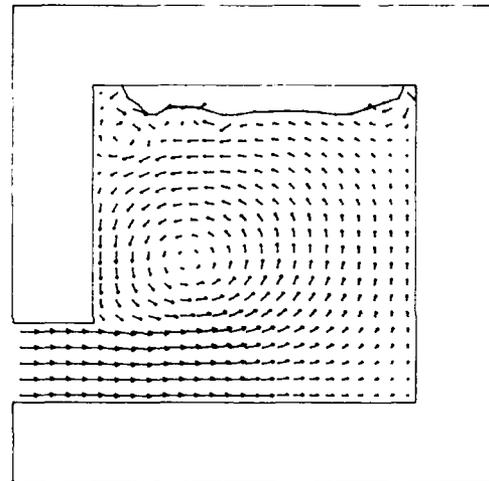
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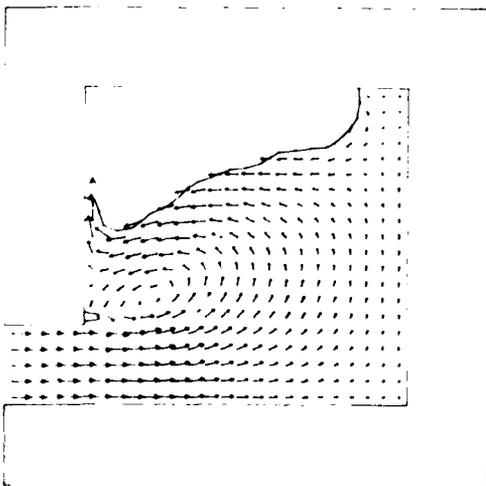
(f) Time = 0.6 second



(d) Time = 0.45 second

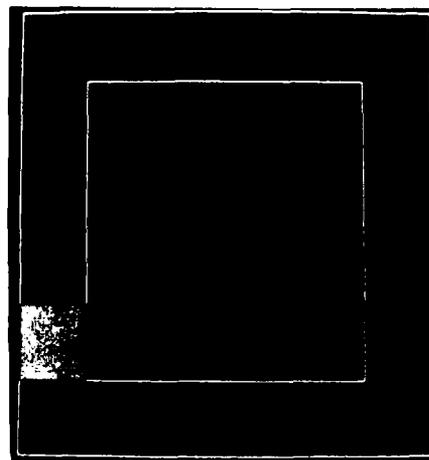


(g) Time = 0.70 second

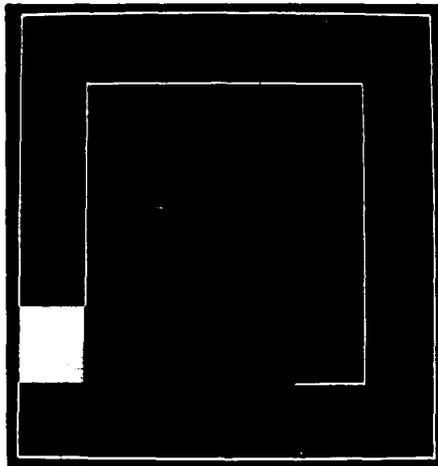


(e) Time = 0.55 second

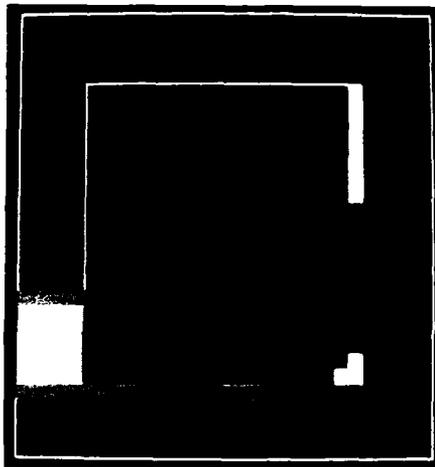
Figure 6. The flow pattern of filling the vertical castings



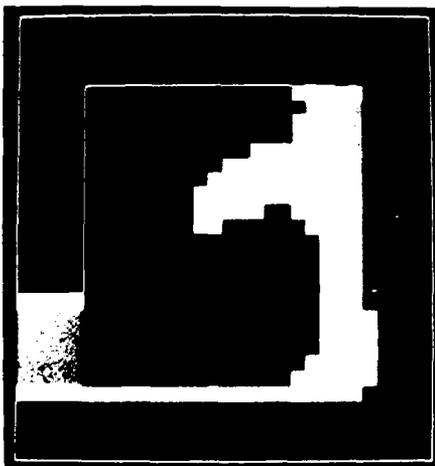
(a) Time = 0.00 second



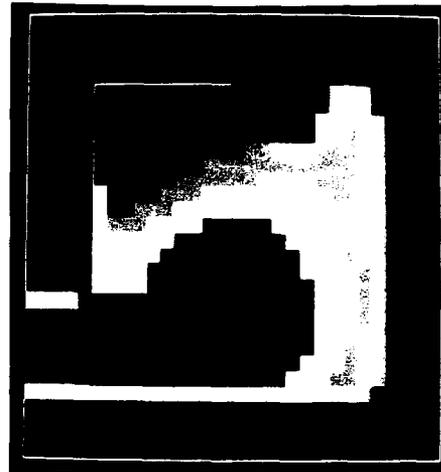
(b) Time = 0.10 second



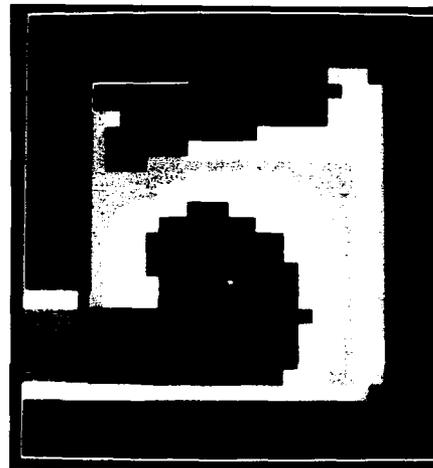
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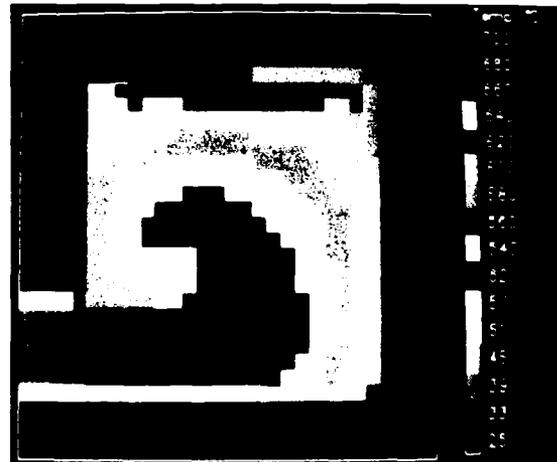
(d) Time = 0.45 second



(e) Time = 0.55 second

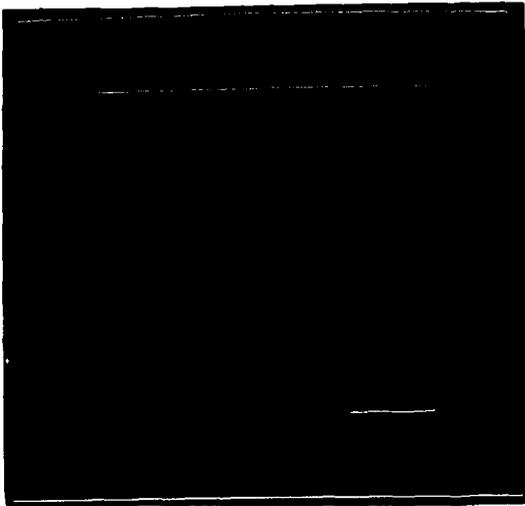


(f) Time = 0.60 second

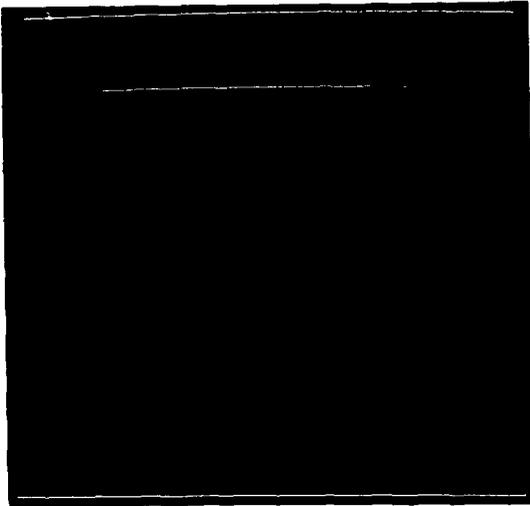


(g) Time = 0.70 second

Figure 7. The temperature distribution of the molten metal and the mold for the filling of a vertical casting with metallic mold.



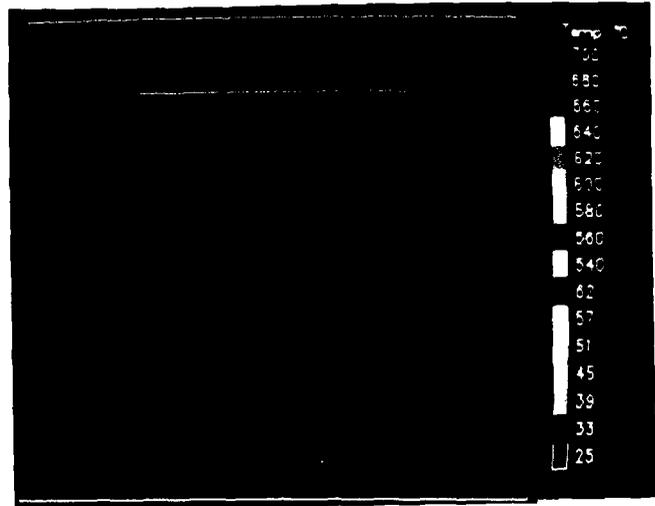
(a) Time = 0.10 second



(b) Time = 0.25 second

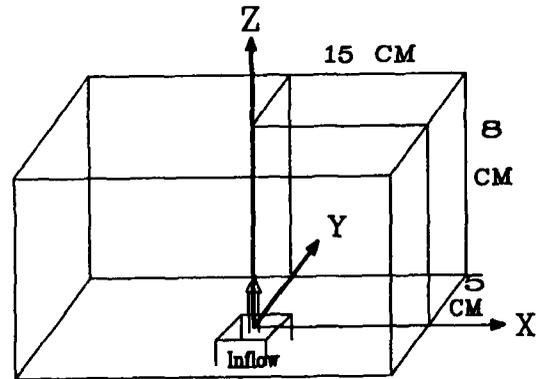


(c) Time = 0.45 second



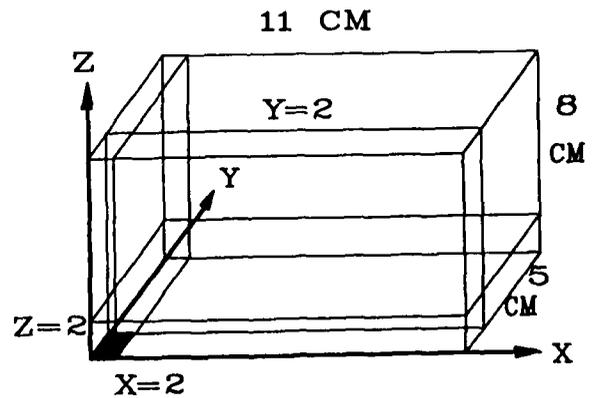
(d) Time = 0.70 second

Figure 8. The temperature distribution of the molten metal and the mold for the filling of a vertical casting with sand mold



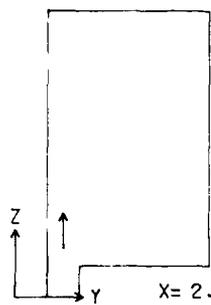
Inflow velocity = 100 cm/sec

(a)

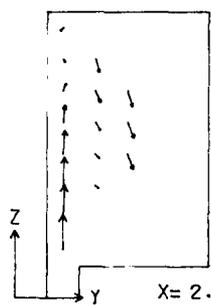


(b)

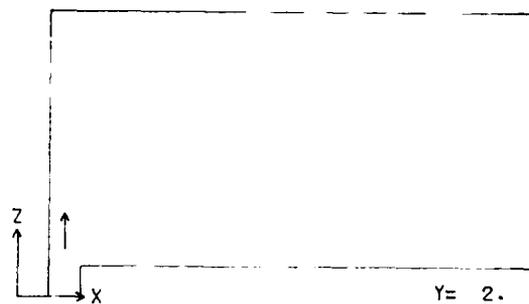
Figure 9. The geometry and dimensions of the block casting.



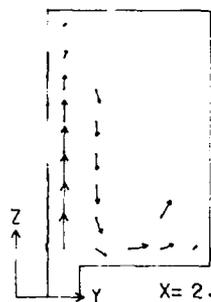
(a) Time = 0.00 second



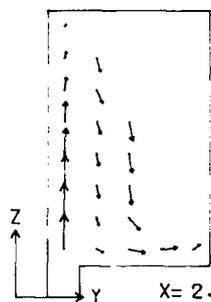
(b) Time = 0.20 second



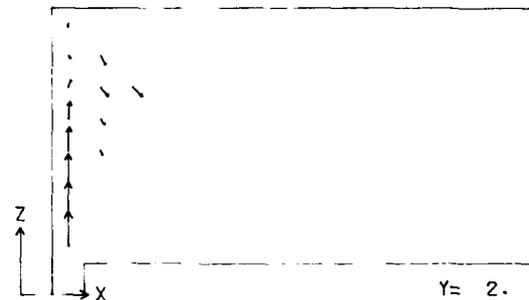
(a) Time = 0.00 second



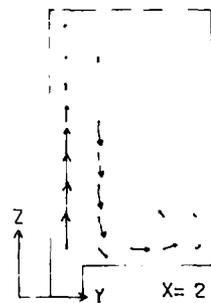
(c) Time = 0.40 second



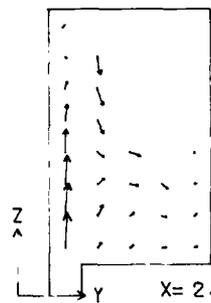
(d) Time = 0.60 second



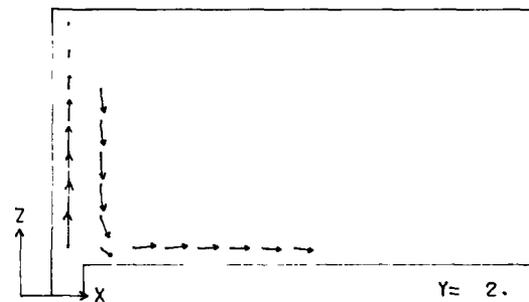
(b) Time = 0.20 second



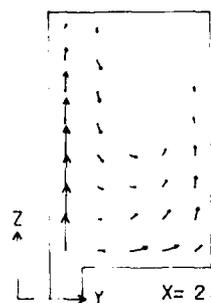
(e) Time = 0.80 second



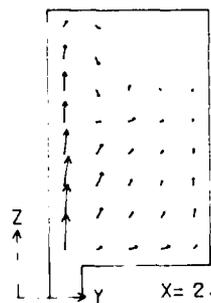
(f) Time = 1.40 second



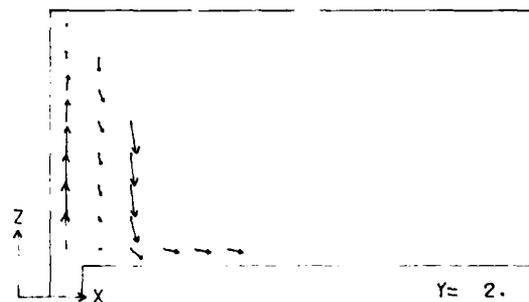
(c) Time = 0.40 second



(g) Time = 1.60 second

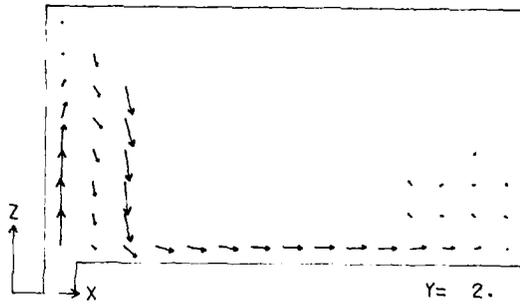


(h) Time = 2.00 second

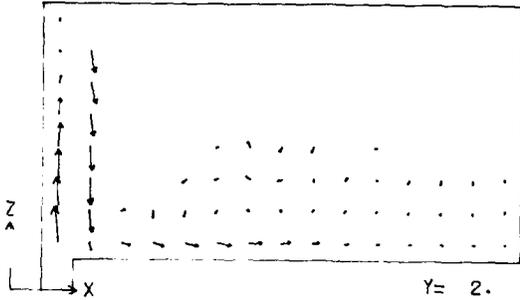


(d) Time = 0.60 second

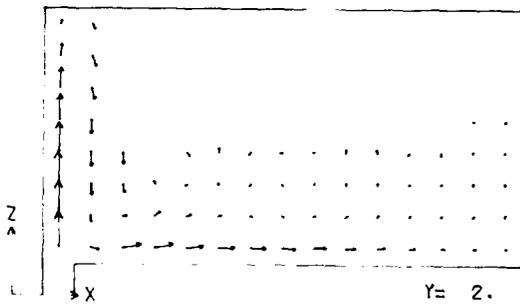
Figure 10. The 2-D velocity profiles on the Y-Z plane near the central plane of the castings.



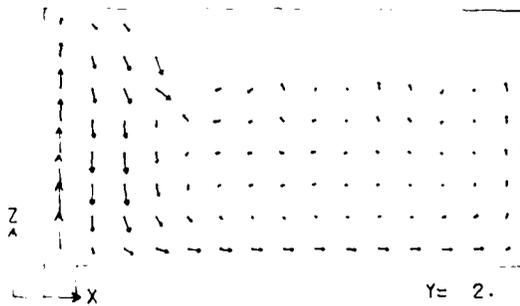
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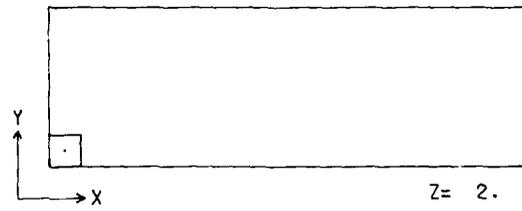
(f) Time = 1.40 second



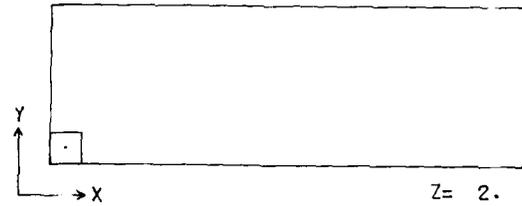
(g) Time = 1.60 second



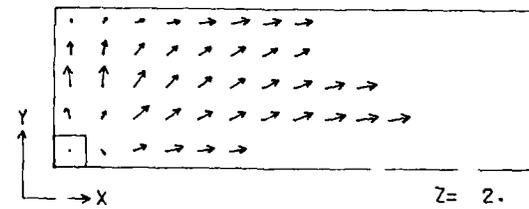
(h) Time = 2.00 second



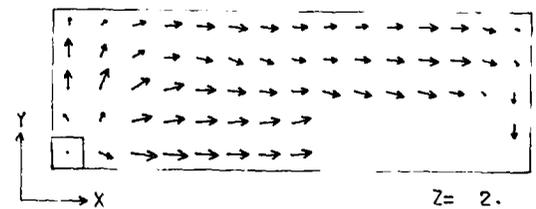
(a) Time = 0.00 second



(b) Time = 0.20 second

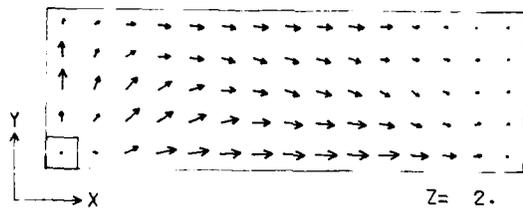


(c) Time = 0.40 second

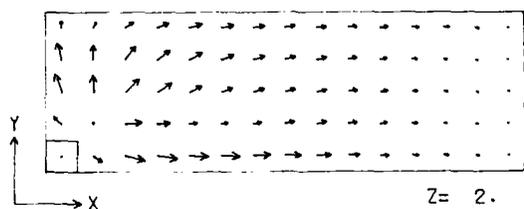


(d) Time = 0.60 second

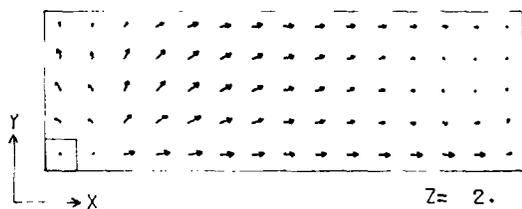
Figure 11. The 2-D velocity profiles on the X-Z plane near the central plane of the castings.



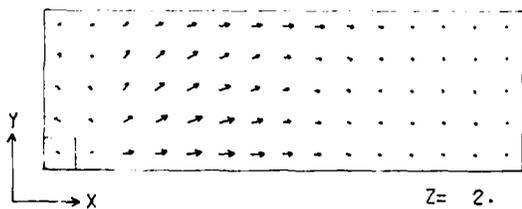
(e) Time = 0.80 second



(f) Time = 1.40 second



(g) Time = 1.60 second



(h) Time = 2.00 second

Figure 12. The 2-D velocity profiles on the X-Y plane near the central plane of the castings.

ADI, AN ENGINEERING MATERIAL

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ABSTRACT

The unique properties of isothermally transformed cast irons have been observed since the 1930's. Practical applications of those materials, however, have remained minimal due to a fundamental lack of process knowledge and inadequate processing facilities. In recent years much research has been done on the properties of Austempered Ductile Iron (ADI) but in each instance the research has centered around only a few process variables.

Much confusion still exists concerning the process, microstructure and terminology of Austempered Ductile Iron (ADI). The intent of this paper is to address and clarify these issues.

There are two required processes in the production of ADI: metal casting and heat treating. The desired properties are achieved only when the two processes are coordinated and when they compliment each other.

The mechanical properties of ADI are shown in this paper as they are influenced by these two processes, especially by the chemical composition and the heat treating cycle.

ADI's remarkable properties were documented by numerous researchers and process engineers. However, very little explanation has been offered in the literature for the mechanisms by which these properties are achieved. Some of these mechanisms are discussed in this paper. In addition, the issue of equipment and facilities development will be

addressed.

BACKGROUND

Atmosphere Group Inc., and its affiliated companies Atmosphere Furnace Co., Controlled Atmosphere Processing Co. and Applied Process Inc. have over 25 years of austempering experience. In 1986 AFC acquired a multi-year research contract with the Gas Research Institute to perform a comprehensive study cataloging the interaction of the many process variables and their relationships to the final properties of ADI and to develop foundry based, gas fired technologies for producing ADI. This paper summarizes some of the results of that ongoing research and commercial developments that have affected the implementation of ADI. Effects of both the casting process and the heat treating process will be discussed.

Austempered Ductile Iron (ADI) has emerged as a new engineering material in the past few years. It has a unique matrix structure and remarkable physical properties. It is strong, wear resistant, light, inexpensive and has high design and manufacturing flexibility.

An ADI part is produced by a casting process. New casting technologies allow the production of complex and near net shape parts.

The properties in ADI are highly versatile. Widely varied mechanical properties are achieved in a component without changing the casting practice or the chemical composition. The same part can be heat treated to display high ductility, high fatigue strength and high impact strength or high hardness, high

tensile and yield strengths and high wear resistance.

A properly executed heat treatment produces a uniform matrix structure throughout the casting, (i.e., the physical properties are similar regardless of the location in the part).

The cost of strength in ADI is lower than in any other engineering material. A comparison of strength properties and their cost in as-cast ductile iron, forged steel and ADI will be discussed.

WHAT IS ADI?

ADI is an alloyed and heat treated ductile iron. Its manufacturing process and base chemistry are similar to those of conventional ductile irons, (nodular irons). It is usually alloyed, however, with either nickel, copper and molybdenum or any combination of these elements. The only purpose of alloying is to increase the hardenability and not to increase strength or hardness as it is erroneously believed by many. Nickel, copper and molybdenum delay pearlite formation in ductile iron. The addition of these elements makes it possible to cool a casting from the austenitizing temperature to the austempering transformation temperature without the formation of pearlite or other high temperature transformation products during quenching.

The amount of nickel, copper and/or molybdenum depends on the quench severity of the quenching medium, and the casting thickness. When the quench severity of the heat treat equipment is high, less alloying is needed. A typical heat treat cycle for ADI is shown in Figure 1.

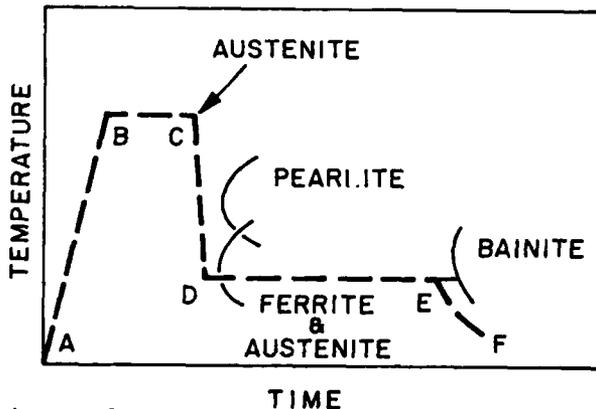


Figure 1. The part is heated to the austenitizing temperature (AB) and is held at temperature (BC) until the matrix is transformed to FCC austenite and until this austenite is saturated with carbon. The casting is then quenched (CD) at a rate

sufficient to avoid the formation of pearlite into a quenching medium, (usually molten salt), at a temperature above the martensite start temperature, (450-750 degrees F). The part is then held at this temperature (DE) for a time sufficient for the austenite to transform to a matrix structure of acicular ferrite and high carbon austenite, (Figure 2). The part is then cooled (EF) to room temperature. No subsequent tempering is required.

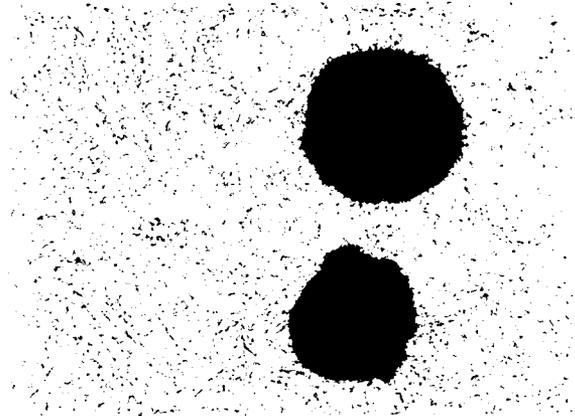


Figure 2.

The bainitic reaction temperature, commonly called the austempering temperature, determines the final mechanical properties in ADI. Higher austempering temperatures produce high ductility, high fatigue strength, high impact strength and low hardness. At lower austempering temperature ADI displays high yield and tensile strength, high hardness and wear resistance.

Unlike in steel, the austempering reaction in ductile iron is a two step process. First, the austenite decomposes to acicular ferrite and high carbon austenite. In the second step, when the sample is held for longer times at the austempering temperature, the high carbon austenite decomposes to iron carbide and the optimum ADI properties are lost.

In steel, austenite decomposes directly to bainite and the intermediate structure of ferrite and high carbon austenite does not occur, (or occurs so rapidly that it cannot be observed). Even today many ferrous metallurgists treat cast iron as a special steel with graphite particles dispersed in it. This is a misconception that has prevented wider application of ADI.

The following is a list of the most significant differences between steel and cast iron:

The carbon content in cast iron is always higher than its solu-

bility in any phase and at any temperature. This is not true for steel.

Steel has a fixed carbon content. Regardless of the thermal history the carbon content in steel remains constant. (The exception is when steel is intentionally carburized.) In cast irons the graphite particles serve as carbon sinks or carbon sources. Because of these graphite particles the matrix carbon content may vary between 0.01 and 1.75 weight percent depending on the thermal history of the casting. In cast irons carbon is free to move between the matrix and the graphite particles. The carbon content of the matrix is variable.

Unlike steel, cast iron solidifies through an eutectic process. The eutectic solidification causes significant solute segregation which, in turn, causes non-uniform carbon solubility in the matrix. Cast irons have higher silicon contents (2.5%) than a plain carbon steel.

These differences have significant effects on the carbon kinetics, and on the austempering process. It takes only seconds to complete the austenite/bainite transformation in plain carbon steel, while it may take hours or even days in ADI.

Once these differences are recognized and their effects on the carbon kinetics are understood, austempered ductile iron production becomes as simple and reliable as that of other conventional steel or iron processes. In fact, as part of the GRI research, mathematical equations are being derived to allow users to specify the chemistry and heat treatment cycle necessary to produce the desired properties in a given part configuration.

MECHANICAL PROPERTIES

No engineering material can be successful without written specifications describing its properties. Recently the Ductile Iron Society has submitted a draft proposal for six grades of ADI to the ASTM for inclusion in the ASTM standards for ductile iron. (The six grades are shown in Table I.)

TABLE I
THE PROPOSED SIX ASTM STANDARD ADI GRADES

GRADE	TENSILE STRENGTH		YIELD STRENGTH		ELONGATION (%)	IMPACT ENERGY*		TYPICAL HARDNESS BHN (B.I.D)
	(MPa)	(Ksi)	(MPa)	(Ksi)		(J)	(ft-lbs)	
1	860	125	550	80	10	105	75	269-321 (3.4-3.7)
2	1035	150	690	100	7	90	65	302-363 (3.2-3.5)
3	1200	175	830	120	4	60	45	363-444 (2.9-3.2)
4	1310	190	965	140	1	40	30	388-477 (2.8-3.1)
5	1450	210	1105	160	-	-	-	418-512 (2.7-3.0)
6	1585	230	1240	180	-	-	-	444-555 (2.6-2.9)

* TESTED AT 22°C ± 7°C

The six ADI grades represent a wide variety of properties. The lower grades are designed to have high ductility, elongation, and impact strength with relatively low hardness and tensile strength. With increasing grade number the hardness, yield and tensile strengths increase and the impact strength and elongation decrease. The lower grades are applicable to moving components requiring higher fatigue strength, parts requiring machining after heat treatment, and parts subjected to severe impact loads. The higher grades are applicable to parts requiring high strength and wear resistance where impact strength is less critical.

Strength properties of forged steel, conventional ductile iron and ADI are compared in Table II.

TABLE II
TYPICAL MECHANICAL PROPERTIES

	FORGED STEEL	DUCTILE IRON	ADI*
YIELD STRENGTH, KSI	75	70	120
TENSILE STRENGTH, KSI	115	100	160
ELONGATION, PERCENT	10	3	10
HARDNESS, BHN	262	262	286
IMPACT STRENGTH, FT-LB**	130	40	120

* GRADE NO.2

** UN-NOTCHED CHARPY AT RT

In this table a micro-alloyed, medium carbon forging steel and a high strength ductile iron are compared with a Grade 2 ADI. In this comparison elongation, hardness, and impact strength were similar. The yield and tensile strengths for ADI were significantly higher. The typical manufacturing cost of the

aforementioned materials is:

- Forged steel \$1.00 per pound
- ADI \$.65 per pound
- Ductile Iron \$.45 per pound

It is seen from the above comparison that the cost of strength is the lowest for ADI. It should be mentioned that the density of ADI is also ten percent lower than that of steel. Therefore, an additional ten percent cost reduction is realized by using ADI in place of steel.

The values for the various grades of ADI listed in Table I are not typical, rather, guaranteed minimum properties. Typical yield and tensile strengths for Grade 1 ADI would be 100 ksi and 140 ksi respectively. The values for Grade 6 would typically be 200 ksi yield strength and 240 ksi tensile strength. The properties listed in Table I allow for variations in production or processing.

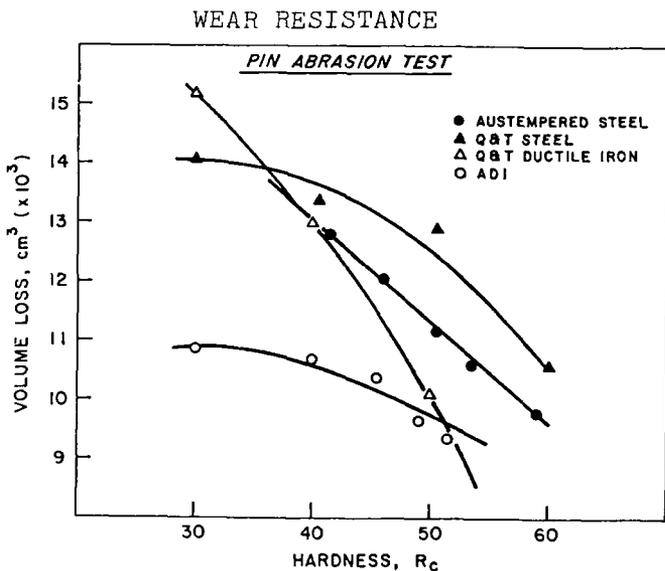


Figure 3.

Figure 3 shows a comparison of wear resistance between ADI and quenched and tempered steel, austempered steel and quenched and tempered ductile iron. The volume loss was determined with a standard pin abrasion test. (1)

It can be concluded from the test that ADI has higher wear resistance under these wear conditions than steel at any hardness level regardless of the processing technique. It also had better wear characteristics than quenched and tempered ductile iron at most hardness levels. The results of this test confirmed earlier studies by other researchers.

The wear resistance of ADI at low hardness is important. The combination of high ductility, fatigue strength,

impact strength and wear resistance is unique to ADI. Applications such as high performance crankshafts, connecting rods, and output shafts are examples of applications that can benefit from this combination of properties.

There is another unique feature of ADI that increases wear resistance; strain induced martensitic transformation. During heat treating it is possible to produce metastable austenite in the matrix which transforms to martensite under high contact stress. For moving components that do not transmit significant power but are exposed to severe wear conditions, the above combination of properties may be the answer.

Camshafts which are engaged with roller followers experience 220-250 ksi Hertzian stress. These high stresses may transform metastable austenite to martensite in the contact surface areas creating a highly wear resistant skin on the part.

This phenomenon is often erroneously called "work hardening" when, in fact, it is not. In work hardening the crystal lattice is distorted and interlocking dislocations are created. The crystal structure does not change. In ADI the FCC austenite transforms to BCT martensite. The hardening in ADI occurs due to a change in the crystal structure. There is a volume expansion during the phase transformation which may cause some plastic flow locally. The surface hardening, however, is caused principally by phase transformation.

THE EFFECT OF THE CHEMICAL COMPOSITION ON THE HEAT TREAT CYCLE

The major alloying elements in ductile iron influence the heat treat cycle and the physical properties of ADI. These elements are silicon, manganese, nickel, copper and molybdenum. It is important to know these influences to obtain optimum properties. The interaction of these various elements determines the consistency, or lack of consistency, of the mechanical properties. The most important parameters influenced by these elements are; the austenitizing temperature, carbon solubility in the matrix, segregation, the effect on pearlite formation, and the tendency to stabilize austenite or form carbide, (thus the effect on transformation time). Table III shows the relative effects of the various elements on these parameters.

TABLE III

CHARACTERISTIC AFFECTED	Si	Mn	Ni	Cu	Mo
Austenitizing temperature	↑	↓	↓	→	↑
Carbon solubility	↓	↑	→	→	↓
Tendency to segre- gate (none, moder- ate, strong)	M	S	N	M	S
Pearlite formation (D=delays it)	-	D	D	D	D
Austenite stabilizer	NO	YES	NO	NO	YES
Carbide former	NO	YES	NO	NO	YES

The effects of each of the major alloying elements is as follows:
SILICON - Silicon is a graphitizer, and as such, promotes graphite formation during solidification. It is readily accepted by the solidifying austenite and segregates around the graphite nodules and in the dendrites. It accelerates the diffusion of carbon and lowers the carbon solubility in austenite. It accelerates both the nucleation and the growth of acicular ferrite and the rejection of carbon into austenite during the austempering reaction. It, only marginally, delays pearlite formation but it strongly inhibits the precipitation of bainitic carbide during the austempering process. For this reason it is the most important element in the formation of the acicular ferrite/carbon rich austenite ADI structure.

Silicon greatly increases the eutectoid temperature and replaces the eutectoid line with a loop in which austenite, ferrite and graphite are in equilibrium. This is an important effect to be considered during the development of the heat treatment cycle. Not knowing the effect of silicon on the eutectoid temperature may result in incomplete austenitization and significantly lower strength in the austempered part. Conversely, austenitizing at an elevated temperature may reduce the toughness of an ADI part.

MANGANESE - Manganese is a carbide forming element. It freely substitutes for iron in cementite and acts as a carbide stabilizer. It lowers the eutectoid temperature and is a strong austenite stabilizer. It slows carbon diffusion

and increases carbon solubility in austenite. It delays the formation of pearlite mainly by increasing the carbon content of austenite. (Next to molybdenum, it increases hardenability more than any other common alloying element.) It also delays the second stage reaction, (i.e., the formation of bainitic carbide), during the austempering reaction.

Manganese is rejected by austenite during solidification and segregates at the cell boundary regions. The manganese content increases exponentially at the cell boundary with increasing bulk content. In these regions the carbon solubility is also raised.

Both carbon and manganese stabilize austenite. In the cell boundary region the austempering reaction is slow and may not be completed even after long austempering cycles in castings with significant cell boundary segregation. Manganese marginally lowers the yield and tensile strength of ADI parts.

If castings can be produced with sufficiently high nodule counts, manganese can be very beneficial. In fact, it is a low cost means of significantly increasing hardenability and, in the higher grades, (i.e., grades 4, 5, and 6), it actually increases the toughness of the austempered part while only marginally reducing the tensile and yield strength.

NICKEL - Nickel delays the formation of pearlite and is often used to increase hardenability. It slightly lowers the hardness, tensile and yield strength but has many beneficial effects. It does not segregate significantly, it increases ductility and impact strength and lowers the eutectoid temperature. It can compensate for high silicon in a casting.

Nickel has disadvantages. Its price is high and unpredictable. It has a moderate effect on hardenability and thus large amounts must be added to through harden heavy sections.

COPPER - Copper is often used to increase hardenability. Its effectiveness, with respect to hardenability, is similar to that of nickel up to about .8%. Above .8% its effectiveness is diminished because of a significant reduction in toughness and elongation. Along with delaying pearlite formation, it also delays the onset of bainitic carbide formation during the austempering reaction. Copper does not stabilize austenite.

Copper has peculiar segregation characteristics. It segregates at the graphite nodules forming a shell around the nodule. This copper rich

shell acts as a barrier to carbon diffusion into, and out of, the graphite nodule. Ductile irons containing copper have higher pearlite contents and higher hardnesses in the as-cast condition than do unalloyed irons. In copper alloyed irons pearlite can have a carbon content that is higher than the eutectoid composition. The macrosegregation of copper is insignificant.

For ductile irons alloyed with copper, longer austenitizing times may be required to fully saturate the austenite with carbon. This is especially important when a lower austenitizing temperature is used.

MOLYBDENUM - Molybdenum is the most potent of the common alloying elements for delaying the formation of pearlite during quenching, (i.e., increasing hardenability). It has, however, a few disadvantages. It raises the eutectoid temperature. It segregates substantially at any level, but especially when the concentration exceeds around 0.5%. It stabilizes austenite, especially at high austenitizing temperature.

Molybdenum is a strong carbide former. It has the tendency to form fine, stable carbides in the cell boundary regions.

Molybdenum increases the tensile and yield strength. It is most beneficial when used in combination with either copper or nickel. The best results have been observed when the molybdenum content does not exceed 0.3% by weight.

DISCUSSION OF PROPERTIES

There is still confusion about what ADI is. Because of its acicular structure and the temperature at which it is formed it is often called bainite. This is technically erroneous. By definition bainite consists of ferrite and iron carbide. The ADI structure consists of ferrite and carbon rich austenite and should therefore NEVER be called bainite.

Most of the confusion stems from steel metallurgy. In steel, austenite decomposes directly into bainite. There is no intermediate structure of ferrite and austenite. In cast irons the bainitic reaction is a two step process.

In the first step, austenite decomposes to acicular ferrite and high carbon, stable austenite. The austenite is usually saturated with carbon. (Earlier work by Kovacs⁽²⁾ identified carbon contents in excess of 2.4% in

fully reacted, stable austenite. Rundman and Rouns⁽³⁾ recorded carbon contents in excess of 2.2%.) In the second step, the reacted austenite decomposes to ferrite and carbide, (bainite), similar to that in steel.

What is "stable" austenite? Some elements such as carbon, manganese and molybdenum have a tendency to stabilize austenite by various mechanisms. At the beginning of the austempering reaction ferrite nucleates at the grain boundaries. As the ferrite grains grow, carbon is rejected into the remaining austenite. As the ferrite volume continues to increase, the austenite volume decreases and the carbon content of the austenite increases further. The higher the carbon content is, the more stable the austenite becomes.

Stability is defined in two ways: thermal and mechanical. Unreacted austenite may transform to martensite when a sample is cooled to room temperature after austempering. Austenite which is the result of incomplete austempering, will transform to martensite when the sample is cooled to a temperature below the martensite start temperature. However, in fully reacted high carbon austenite the thermal transformation may be completely suppressed. Samples cooled in liquid helium at 7.9 degrees Rankin (-452 degrees F) exhibited no martensitic transformation.

Martensitic transformation may occur when a sample is strained. This is called stress induced martensitic transformation or stress transformation. ADI may be thermally and mechanically stable or thermally stable and mechanically unstable. If not properly processed it can be made both thermally and mechanically unstable.

Solute segregation causes instability in ADI. As was mentioned earlier, silicon segregates at the graphite nodules, molybdenum and manganese segregate at the cell boundaries, remote from the nodules. Silicon decreases the carbon solubility while manganese increases it. Silicon accelerates carbon diffusion while manganese slows it. In the cell boundaries, where both manganese and carbon contents are high, the austempering reaction is slow. Near the nodules, where the carbon level is low, due to high silicon, the reaction is fast. Some samples can have certain areas where the reaction is fast and bainite is present while at other areas in the same casting the reaction has not even started. For this reason it

is important to minimize solute segregation, specifically that of molybdenum and manganese.

It was mentioned earlier that hardness and strength increase when the austempering temperature is lowered. The phases are the same in ADI regardless of the grade number. All ADI consists of acicular ferrite and carbon rich austenite. The difference between the various grades is the morphology. In lower strength Grade 1 the microstructure is coarse, while in high strength Grade 6 it is fine.

The strengthening mechanism in ADI is inherently contradictory with that of conventional ferrous alloys. In the higher strength grades of ADI both the ferrite content AND the hardness are higher than in the lower strength grades. Higher ferrite contents traditionally have meant lower strengths and hardness.

The explanation for this phenomenon is as follows. In the higher grades of ADI the grains are much finer and feathery. The grain boundary surface area in the high strength grades is orders of magnitude higher than that of the lower grades. The two phases, ferrite and austenite have different crystallographic structures. Ferrite is BCC. Austenite is FCC. The grain boundary between the two is highly incoherent. Dislocations are pinned down and interlocked and their movement is greatly restricted in a higher grade ADI.

Matrix structures in cast irons are easy to describe and refer to. These structures are often named after prominent metallurgists such as Bain, Pearl, or Marten. ADI has not yet had a name coined for its distinctive structure. As soon as a name for the ADI structure is determined the confusion with bainite will be eliminated. It has been suggested that the structure of acicular ferrite and carbon rich stable austenite be referred to as "Ausferrite", suggesting a composite structure of austenite and ferrite.

EQUIPMENT CONSIDERATIONS

Until recently four basic types of equipment have been used for austempering:

1. Belt furnaces with dry salt quenches that are highly productive but cannot handle large, heavy parts.
2. Electrically heated salt-to-salt lines that can physically handle

large parts but still lack the quench severity to through harden thicker parts even with heavy alloying of the iron (contamination of the quench salt by the austenitizing salt also lowers the quench severity).

3. Batch and semi-continuous furnaces with hot oil quenches are also used but are limited to a maximum quench temperature of 475 degrees F. These units also exhibit low quench severity and an ever present risk of fire.

4. Semi-continuous pusher and roller hearth furnaces with salt quenches are used but they lack the flexibility to adjust to short run jobs and varied cycles

None of these types of equipment was well adapted to a "jobbing" or commercial environment requiring high productivity and flexibility.

In 1984 the first batch austemper furnaces with integral salt quenches were introduced on the market. These furnaces combine high productivity, low operating cost and the flexibility to allow the operator to run all types of castings large and small, from Grade 1 to Grade 6. Controlled water additions to the quench increase the quench severity allowing the use of less, or no, alloying to produce ADI properties in ductile iron.

As the demand for ADI grows and high volume production jobs emerge, dedicated, continuous furnace lines will be installed for processing ADI. In the meantime, smaller, more flexible, batch-type units will be employed in commercial heat treat shops to handle the developing ADI business.

The Gas Research Institute has targeted the development of foundry based ADI processing. It is reasoned that if an in-line unit could be developed to utilize some of the waste heat from the casting process to create ADI castings, significant cost savings could be achieved for the casting producer and the end user. Research on several foundry based concepts is ongoing.

New equipment developments have made commercial processing of ADI an economic reality. As newer technologies emerge the competitive position of ADI will be further enhanced.

CONCLUSION

Austempered ductile iron is a low cost, high performance engineering

material that can be successfully produced if the metallurgical mechanics are understood. Recent research contracted by the Gas Research Institute, and others has fully quantified the relationship between the foundry and heat treat processes, and the mechanical properties. New equipment developments also continue to expand the use of ADI. ADI, with its unusual "ausferrite" matrix, provides the designer with a new degree of engineering and manufacturing flexibility that, to date, has not been available in conventionally processed cast irons.

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HEAT TRANSFER—SOLIDIFICATION KINETICS MODELING OF CASTING SOLIDIFICATION

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ABSTRACT

A complete model for solidification of castings should include both macro- and micro-models. At the macro-level, it is required to describe the heat transfer from the metal to the mold, the fluid flow of liquid metal during mold filling, and stress analysis. At the micro-level, the model should include solidification kinetics and fluid flow during solidification. At the present time, the task of including all these models into a comprehensive solidification package is far from being complete, although significant progress has been achieved in each field.

The paper will present the state of the art on coupling two of the required models, i.e. macro heat transfer (HT) with micro solidification kinetics (SK). Various techniques of coupling these two models will be discussed, emphasizing the progress made in SK modeling with respect to calculation of fraction of solid based on nucleation and growth laws. A number of experimental and calculated results will be shown for some typical casting alloys, to include cast iron and aluminum-silicon alloys.