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Rapid Tooling for Functional Prototyping of Metal Mold Processes Literature Review on Cast Tooling

Michael D. Baldwin, Patrick W. Hochanadel

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Rapid Tooling for Functional Prototyping of Metal Mold Processes
Literature Review on Cast Tooling

Michael D. Baldwin
Liquid Metal Processing Department
Sandia National Laboratories
Albuquerque, NM 87185-1134

Patrick W. Hochanadel
Department of Metallurgical and Materials Engineering
Colorado School of Mines
Golden, CO 80401

Abstract

This report is a literature review on cast tooling with the general focus on AISI H13 tool steel. The review includes processing of both wrought and cast H13 steel along with the accompanying microstructures. Also included is the incorporation of new rapid prototyping technologies, such as Stereolithography and Selective Laser Sintering, into the investment casting of tool steel. The limiting property of using wrought or cast tool steel for die casting is heat checking. Heat checking is addressed in terms of testing procedures, theories regarding the mechanism, and microstructural aspects related to the cracking.

TABLE OF CONTENTS

Figures

INTRODUCTION

The use of cast dies and tools for die casting and forging applications has long been recognized as a viable alternative to the orthodox method of machining similar material from wrought stock. Nevertheless, cost and time restrictions associated with developing a first-run precision casting, coupled with the uncertainty regarding the performance of castings, placed widespread use of cast tools beyond the current state of technology. With recent advances the in ability to produce castings with much shorter lead times, there is increasing need to characterize the performance of precision cast materials.

The limiting property of these materials for die casting is "heat checking", in which network cracks are formed from stresses brought about by the constraint of the subsurface material on heating and cooling. There are two main theories regarding the nature of heat checking. One is that the material fatigues due to the heating and cooling cycles, and the other is that creep damage accumulates during each heating cycle, and after a certain number of cycles, decohesion at the grain boundaries occurs. Regardless of mechanism, heat checking is caused by the accumulation of damage (creep or fatigue) from the thermal cycling encountered on the surface of a die.

There are many variables which affect the severity of heat checking. These include the maximum and minimum temperatures experienced by the die in a casting or forging cycle, the temperature difference between the maximum and minimum temperatures, the frequency of heating and cooling, the heating and cooling rates during the cycling, the external loads, the die material and its processing, etc. Many of these factors are not true variables in many die casting or drop forging operations. For example, the die caster has a set pouring temperature and a set mold preheat temperature, thus establishing the temperature conditions, as well as the heating and cooling rates per cycle. In general, the external loads also are not a variable. However, the material and its processing may be altered to yield the best resistance to heat checking.

One of the main groups of tool steels are hot work tool steels, which, as the group name implies, are used in various hot-work applications such as die-casting, forging, extruding, punching and other applications which form materials at high temperatures. In addition, hot-work tool steels are used in ultra-high strength structural parts and are currently finding new, lower temperature applications, in which high strength and reasonable toughness are required. Probably the most popular hot-work tool steel is AISI H13 tool steel, which has a typical composition of Fe - 0.4 C - 5.0 Cr - 1.5 Mo - 1 V, and will be used as the general focus for this review.

AISI H13 tool steel belongs to the general class of chromium-molybdenum hot-work tool steels, and to a further subclass of these materials, all of which contain 5 percent chromium[1]. The 5-pct-Cr hot-work tool steels were developed primarily for die casting aluminum alloys. Another major use of this material is for drop-forging dies and inserts. These applications require that the materials air-harden from relatively low temperatures, resist heat checking, withstand impact stresses at elevated temperatures, retain hot hardness at operating temperatures (resist softening), show little size change in hardening, resist scaling (oxidation resistance), resist erosion by molten metals (namely, molten aluminum and brass), and be cost-effective.

AISI H13 steel was designed, in the wrought form, to have good toughness in conjunction with good red (hot) hardness. The thermal shock resistance of H13 steel is such that products made from wrought H13 steel may be cooled, in service, with water or other flushing media.

RELEVANCE OF TOOL STEELS AS A CAST MATERIAL

The notion of using cast tool steels may be attractive to the die manufacturer, as well as the user of the die material because of the overall cost effectiveness. The overall cost effectiveness may be improved with cast die steels in two ways: (1) shaping operations may be minimized, and (2) processing prior to or following shaping of the material may be simplified. If the cast die or insert provides a lifetime comparable to a wrought die or insert, which some suggest, then the overall cost is reduced even more.

A plethora of information pertaining to the processing, microstructure and mechanical properties of various wrought hot-work tool steels is available[1-14]. In addition, numerous publications exist which address the attractiveness of cast dies in terms of cost and die life[15-37]. However, technical information regarding the cast hot-work tools steels is lacking, and very little information exists correlating the microstructure (as-cast or heat treated-cast) with resulting properties.

The notion of replacing wrought tooling with cast product is not new. The drive for cast dies has existed in the United States[38] and Czechoslovakia[39] since the mid 1950's, and in the early 1960's, Japan reported work on cast die steels[16,17]. However, in the late 1960's and early 1970's, the interest to utilize cast tooling tapered off in the United States, probably due to the lead-times (and therefore costs) involved in developing the prototypes. The time involved to create a casting that meets specifications also contributed to the dwindling interest in cast tool steel components. Throughout the 1970's and 80's, Russian and European metallurgists performed extensive research on the casting of 5% chromium die steels and various other tool steels[35,40-43]. Much of their work, however, discussed the usefulness of cast dies and molds in terms of actual production tests, and little pertinent correlation between the microstructure and properties was completed. Recently, Chinese investigators have shown interest in cast H13 steel[36].

The recent advent of rapid prototyping coupled with precision investment casting has recently caused reconsideration of cast dies[44]. The computer-controlled pattern fabrication is performed quickly by using Stereolithography or Selective Laser Sintering. From this type of a program, both cost and time may be effectively minimized so that previous limitations are now not as critical.

Besides cost in developing the prototypical dies, another problem may be casting microsegregation, which is inherent in all castings, and to some extent, in many wrought products. This phenomena, coupled with porosity, produces deleterious properties in castings, and although cast tools may appear to be cost-effective, their life may be shorter (depending on post-casting heat treatments), thereby decreasing the cost effectiveness. A number of papers reported comparable and possibly longer die life for cast dies or components in which the castings were sound[18-21,26,45]. Some investigators claimed that the cast structure aided in the delay of heat checking, and the cracks did not seem to penetrate as deeply as those in the wrought dies; thus, the re-tooling of dies or inserts which are cast tend to be less costly than the wrought counterparts[26]. This, however, was neither correlated with the soundness of the casting nor the actual microstructure. If the heat checks in cast molds or inserts are shallower than the heat checks in the wrought molds, more re-tooling operations may be performed on a cast insert prior to the mold losing its dimensional tolerances. Industrial tests have shown that wear in the cast materials is less than those in the wrought materials. According to Torrey[19,21], the wrought material, because of the smaller grain size, has a greater amount of grain boundary energy, and at higher temperatures, there is a higher tendency for intergranular rather than transgranular decomposition; thus it seems likely that cast dies will not damage as badly as wrought materials due to galling at higher temperatures. Rostocker[45] thermally cycled tapered disc samples made from both cast and wrought H11 tool steel. He found that the cast material resisted heat checking better than its wrought counterpart.

PROCESSING OF WROUGHT H13

H13 tool steel is typically cast into ingots and subsequently hot worked. Special melting processes such as vacuum arc remelting (VAR) or electroslag remelting (ESR) might be used to improve the cleanliness of the microstructure. Various thermal processing steps follow hot working to obtain desired properties. After hot forging, A three-step heat treatment of these materials will produce desired properties: (1) annealing, which provides a uniform distribution of spheroidized carbides in an equiaxed ferrite matrix (performed to ease machining), (2) hardening (austenitizing and quenching), which creates a martensitic structure void (or nearly void) of carbides and (3) tempering, which improves the toughness and thermal stability of the martensitic matrix. After this basic heat treatment is done, optional surface hardening steps such as carburizing, nitriding, chromizing, or a combination of these may be performed to improve the wear-resistance of the surface layer.

The steel is typically annealed in a controlled atmosphere furnace in the austenite or austenite plus carbide phase field. H13 steel, is annealed at about 840-870°C (~1550-1600°F) for at least one hour at temperature[1,2]. Because of the air-hardening nature of this material, a slow furnace cool of 30°C/hr (~50°F/hr) is recommended to prevent non-uniform or non-equilibrium structures (such as martensite or bainite) and to allow for ease in machining.

Hardening (austenitizing and quenching) follows as the next step in the thermal processing of H13 steel. Austenitizing is usually performed in a controlled atmosphere to prevent carburizing or decarburizing, which, if not controlled, may induce early die failure via heat checking or erosion. The austenitizing temperatures range from 1010°C to 1040°C (1850°F to 1900°C), and the soaking times are generally about 20 minutes at temperature, followed by air cooling (oil-quenching is an option)[1,2]. More rapid quenching, however, may result in quench cracks. In addition to allowing for the fully martensitic structure after cooling, austenitizing also allows for the dissolution of the complex alloy carbides present in H13 steel.

Next, tempering is used to improve the toughness and the hot stability of H13 steel. Since the operating temperatures are on the order of 500-650°C in many industrial applications, it is only logical to temper the material at or around these temperatures. In fact, the tool material should not be tempered at temperatures much lower than the desired operating temperature to optimize the thermal stability of the microstructure and properties. In addition, multiple tempering (at least double) is suggested to further increase the thermal stability of the structure. Each tempering step should be followed by air cooling. As expected, an increase in tempering temperature results in greater toughness and impact properties at the expense of the hardness. The first temper is performed to precipitate carbides, which aid in retaining strength at elevated temperatures. This step also allows the retained austenite to transform into martensite. The second step tempers any martensite which formed on cooling after the first temper. A further discussion of the microstructural aspects on heat treatments will be presented in a later section.

PROCESSING OF CAST H13 STEEL

An examination of the property specifications of a cast tool steel must be made prior to selecting a reasonable heat treatment. According to Lubalin[15], desirable and necessary properties include the following characteristics:

1. The castings must be deep air hardening, temper resistant at high temperatures, and stable during and after heat treatment.
2. They must retain a high hot hardness (or elevated tensile properties) to resist wear resistance, erosion and heat check initiation, while remaining tough enough to avoid propagation of heat checks after heat checks have nucleated.
3. The surface composition must be controlled (i.e. decarburization or carburization must be prevented) to prevent further surface stress distributions which may result in a higher susceptibility to heat checking.
4. Machining stress should be minimized.
5. Grain size must be critically controlled.
6. Porosity, whether shrinkage or gas, must be avoided so that a sounder casting will be produced.
7. Casting tolerances must be accurate enough to minimize the difficult machining.

In addition to these properties, Rostocker[45] has shown that cleanliness of the microstructure is very important as well.

Although this list of characteristics is by no means complete, it does address the major concerns of the die casting and forging die industries. It should be evident that no single material will be perfect.

Two types of processes are used for casting H13 tool steel: (1) investment casting and (2) the ceramic mould or Shaw process.

Investment casting, often called lost wax casting, is regarded as a precision casting process to fabricate metal parts from almost any alloy. Although its history lies to a large extent in the production of art, the most common use of investment casting in more recent history has been the production of components requiring complex, highly toleranced, often thin-section castings of high quality. While a complete description is beyond the scope of the discussion here, the salient features of the process will be briefly described, along with a description of how Rapid Prototyping and Manufacturing (RP&M) has been incorporated.

Unlike sand casting where a single pattern can serve to produce a large number of molds, a new pattern is required for every investment casting. These patterns, typically produced in injection molding machines, are made from wax specifically formulated for this use. Once a wax pattern is produced, it is assembled with other wax components to form a metal delivery system, called the gate and runner system. The entire wax assembly is then subsequently dipped in a ceramic slurry, and covered with a sand stucco coat and allowed to dry. This dipping and stuccoing process is repeated until a shell of approximate 6 to 8 mm (1/4 to 3/8") is applied. Once the ceramic has dried, the entire assembly is placed in a steam autoclave to remove most of the wax. After autoclaving, the remaining amount of wax that soaked into the ceramic shell is burned out in an air furnace. At this point the shell is empty. It is then usually pre-heated to a specific temperature, and filled with molten metal. The hot mold assists with filling of intricate shapes and thin sections. Once the casting has cooled sufficiently, the shell is chipped from the mold, and the desired casting cut from the gates and runners. As can be seen, the process requires that a pattern be destroyed for each metal casting produced. A slightly more detailed step by step description of the process is presented below.

Steps to Produce an Investment Casting

1) Design Casting
2) Design Gating System
3) Design & Procure Die
4) Produce Wax Part
5) Produce Gate/Runner Assembly
6) Fabricate Shell
7) Dry Shell
8) Dewax

9) Fire Shell
10) Pour Casting
11) Remove Shell
12) Remove Gates
13) Hot/Cold Straighten
14) Heat Treat
15) Inspect
16) Ship

In a direct sense, the onset of RP&M processes has had an impact on the process by eliminating only a few of the steps listed above, namely the design and procurement of the wax injection die. However, it has been the cost and the time associated with manufacture of this tool that has contributed substantially to the perception of investment casting as an expensive and time consuming process, not at all suited for rapid production of prototypes. In the past, a complex injection tool for a structural casting in the 5 lb. range could cost upwards of $50,000 and require 30 to 50 weeks of lead time. Compounding this problem was that more often than not, the "final" design was in fact not so final. Changes to the die before it ever left the toolmaker were not uncommon. With the advent of RP&M processes, the time required to make the first patterns has been shortened from several months to often only a day. As a result, the investment casting industry is experiencing some remarkable changes that are slowly changing commonly held opinions regarding its agility.

Implementation of rapid prototyping processes in investment casting is acheived by substituting the Stereolithography and Selective Laser Sintering patterns for the wax patterns in the above process. In addition to the pattern, the major change in the process is dewaxing, performed in a flash-fire dewax and not the steam autoclave. Substituting a material other than wax for the pattern is not unique, but the changes to the process that this entails presents some unique challenges to the foundry.

Most published information about the properties of cast H13 and other die steels are based on experiments conducted using Shaw cast steels[15-23,25,26,31,32,37]. Similar to investment casting, the Shaw process utilizes ceramic molds and produces high quality, close tolerance castings.

However, the Shaw process is slightly different in that it utilizes a cope and drag ceramic mold. Foundry sands (and flours ground from them) are mixed into a ceramic slurry with a liquid binder and a gel agent. The slurry is then poured over a pattern, and the slurry dries to form a shell within minutes. The shell is then withdrawn from the pattern and torched with an open flame to introduce microcracks. These microcracks aid in giving the mold thermal shock resistance. After the introduction of the microcracks, the mold may now be baked to remove the volatiles present in the binder and gel agent, thus eliminating gas production due to the mold. The microcracks are usually large enough so that the mold may accommodate the thermal expansion of the ceramic particles, thus preventing dilation of the mold, which in turn decreases the porosity. In addition, this process allows for the metal to be poured in as cold a mold as possible. Methods similar to the rapid prototyping described above may also be utilized to develop patterns for Shaw cast tool steels.

Although little information exists regarding the heat treatment of cast H13 tool steel, logic dictates that a homogenizing treatment should be carried out prior to hardening to minimize (or, if possible, eliminate) the segregation effects which tend to degrade the properties. As Grabin, et al.[41] has shown, a homogenizing temperature of 1150°C and higher, coupled with a sufficient holding time, allows the primary heterogeneity of a cast tool steel to be decreased or effectively eliminated. Following the homogenizing treatment, hardening and tempering should be done in a manner similar to heat treatment of the wrought product. This type of thermal processing has been found to yield properties comparable to, but still somewhat less than, the properties of the wrought material.

MICROSTRUCTURE OF H13 STEEL

The microstructure of H13 steel is variable depending upon the times and temperatures of the various thermal processing steps previously mentioned (hot work or homogenizing, annealing, austenitizing, quenching and tempering). These steps are all important in producing a material with desired microstructural features and corresponding properties.

Typically, the first step after casting is a homogenization treatment, which is done to decrease or, if possible, eliminate the microsegregated areas inherent in castings or ingots. In the case of wrought materials, the hot work may be both a forming and homogenization process, which occurs in the austenite or austenite-carbide phase regions. The microstructure which generally follows hot work is that of coarse prior austenite grains, in which the austenite transformation products remain (such as martensite or bainite). Segregation, however, is never completely eliminated, and this is apparent even in wrought materials (banding). After hot-working H13 steel , annealing is done to provide a more uniform structure. Non-uniformities are intrinsically present following hot working operations. These include variable grain sizes and carbide distributions, as well as a banded microstructure. If some of these non-uniformities, such as variable carbide distributions, are allowed to remain (i.e. insufficient or no annealing), these inhomogeneities may persist throughout processing and influence the service life of the material. Thus, the desired microstructure after annealing is spheroidized carbides in a matrix of equiaxed ferrite grains, a microstructure which yields a uniform hardened structure (i.e. less susceptible to quench cracking from large carbides) and ease in machining. This in turn will increase the service life of the material and ultimately decrease cost.

In castings, the homogenization step is a high-temperature soak for periods of three to four hours. During this soak, the areas of segregation decrease in size. The driving force for this action is the composition gradient, or more fundamentally, the chemical potential gradients which exist due to segregation. This heat treatment typically occurs in a single phase region (austenite) to prevent separation of phases. Besides reducing the microsegregation, a homogenizing heat treatment may allow for a more refined grain structure. However, at temperatures greater than 1150°C, excessive grain growth will probably occur. As the cast material is heated into the fully austenite phase region, new austenite grains nucleate and grow. When the homogenizing temperature is too high, excessive austenite grain growth may result, leaving a non-uniform structure with carbide precipitates, distributed at prior austenite grain boundaries. Besides excessive grain growth, interdendritic regions may liquate, as Yikuan and Renyu have observed[36]. If this occurs, the segregated regions may be enhanced and stabilized rather than decreased. In addition to compositional gradients, castings may have porosity. Therefore, this high temperature homogenizing treatment may be enhanced with an application of a positive pressure. This type of treatment is called hot isostatic pressing (HIPing). It has been shown by Strode that HIPing enhances the strength and ductility of a Shaw cast H12 tool steel[43]. Although the cast material may have somewhat lower properties in comparison with the wrought material (at least in the longitudinal direction), cast materials do not usually exhibit directional properties such as exhibited by the wrought materials.

Following the homogenizing treatment(s), whether they be hot-working or thermal homogenizing, hardening is done by austenitizing the material and quenching from the austenitizing temperature. Schmidt[8] has shown that both the austenitizing temperature and the quench rate following the austenitizing have an influence on the microstructure and the corresponding properties of wrought H13 steel. He revealed that by utilizing an austenitizing temperature greater than 1025°C (1875°F), a higher degree of vanadium carbide dissolution and a higher degree of austenite grain coarsening occurs. After quenching and tempering, the effect of increasing the austenitizing temperature on the properties was manifested as a general increase in hardness capability and a decrease in the Charpy V-notch impact energy values (at both ambient and elevated temperatures). Furthermore, the austenite composition is dependent upon the times and temperatures of austenitizing, since a higher degree of carbide dissolution results in a richer austenite composition. The composition defines the martensite start and martensite finish temperatures (M_S and M_F, respectively), the hardenability of the material and the precipitation of carbides which may occur during the tempering. If the austenite is too rich (i.e. fewer carbides), the M_S is lowered, and the amount of retained austenite at room temperature increases; thus, the amount of hardening may be insufficient. However, this may not be much of a problem, since the desire during austenitization is to completely dissolve the carbides, thus preventing carbide coarsening and/or lengthening along grain boundaries. Also, this dissolution of carbides is probably not as harmful as first implied since during the second (and/or subsequent) tempers, some retained austenite will transform to martensite. If the composition of the austenite is too lean, not only are some carbides retained, but also the M_S is raised. The leaner austenite will have decreased martensite hardenability, and the probability of lower amounts of martensite and higher amounts of bainite is increased. Increasing the amount of bainite typically decreases impact energy values. If the alloy carbides are not completely dissolved, they may remain throughout the thermal processing of the material, and these carbides, if bulky and continuous along prior austenite grain boundaries, are pernicious to the properties of the H13 steel. In addition, if more of the vanadium carbide is dissolved during austenitizing, a more intense secondary hardening will occur on tempering. If the austenite composition is heterogeneous, the ensuing properties will be affected adversely in a manner similar to those observed in cast materials with insufficient homogenization. At any rate, the resulting microstructure is normally martensite (or martensite with bainite), retained austenite, and retained alloy carbides.

The quench rate also influences the properties of the H13 steel. If the quench rate is too slow, bainitic transformations from austenite may occur. The tempered bainitic structure, as discussed by Schmidt[8], results in properties less attractive than those of tempered martensite. This is due to both the presence of interlath MC carbide platelets which may form during the slow quench and the presence of M_2C and $M_{23}C_6$ carbides which form during tempering. Therefore, a rapid quench may be considered to produce a fully martensitic structure and avoid the formation of bainite. Schmidt has also found that these bainitic structures are prone to a 595°C temper embrittlement, regardless of the austenitizing temperature employed. The cause of this embrittlement, however, was not adequately addressed. It was noted, nonetheless, that a similar phenomena occurred in martensitic structures resulting from an overly high austenitizing temperature (1100°C and above). However, air cooling is typically sufficient to avoid the formation of excessive bainite.

The tempering temperatures and times are usually varied to acquire desired combinations of hardness and toughness. The tempering stages of H13 tool steel, as given by Branco and Krauss are[7]:

STAGE I-	Precipitation of transition iron carbides within the martensite plates or laths at temperatures between 100°C and 200°C.
STAGE II-	Transformation of retained austenite to cementite (Fe_3C) and ferrite at temperatures between 200°C and 600°C.
STAGE III-	Precipitation and coarsening of Fe_3C at temperatures greater than 200°C.
STAGE IV-	Precipitation of alloy carbides (such as MC, M_2C, etc.) at temperatures greater than 400°C.

Branco and Krauss[7] further explain that other stages, such as clustering and aging phenomena, as well as segregation of impurity atoms to various interfaces may transpire. In hot work tool steels, the diffusion-controlled coarsening of the alloy carbides may be delayed by alloying with multiple alloying elements, which tends to stabilize the fine carbide dispersions found in H13 steel, hence yielding good elevated temperature properties (e.g. hardness and creep). In fact, the resistance to over-tempering (i.e. a decrease in the strength from tempering too long) has been shown to be detrimental to the heat checking resistance of wrought H13 steel. In H13 tool steel, the most common precipitate for secondary hardening is vanadium carbide, which may dissolve appreciable amounts of molybdenum. The molybdenum that dissolves in vanadium carbide stabilizes the vanadium carbide and yields higher hardness. This precipitates at a tempering temperature of 500-550°C. The V:C ratio in H13 steel is hypo-stoichiometric, but other carbides may form due to the excess of carbon. The second most common precipitate in H13 steel is molybdenum carbide (Mo_2C), which may contain dissolved chromium and vanadium. In addition to these carbides, chromium-rich $M_{23}C_6$ precipitates, which may have molybdenum dissolved in them are typically found in H13 steel. These carbides are less resistant to coarsening and tend to decrease the creep resistance of H13 tool steel.

H13 steel is not typically tempered at the maximum secondary hardness since a distinct embrittlement occurs. However, the strength may be optimized without excessive deterioration of impact toughness by a number of processing or alloying methods. The first method is to utilize as high an austenitizing temperature as possible to dissolve as much of the carbide precipitates without causing excessive grain growth. The second method is to use a V:C ratio close to stoichiometry, which also optimizes the creep strength, and alloy such that the mismatch between the matrix and the precipitate is maximized. To achieve this, an increase in the molybdenum (or Mo:V ratio) is desired. This alloying step would also aid in stabilizing the VC, since the replacement of VC by $M_{23}C_6$ and coarsening of the vanadium carbide would be slowed. Increased silicon would also increase the lattice-precipitate mismatch. Other alloying steps to aid in prevention of recovery, recrystallization and excessive grain growth include titanium or niobium additions, since these form fine carbide/nitride precipitates which pin grain boundaries and do not allow for excessive grain growth.

Much of the available information involving the cast microstructure of H13 tool steel is that given by Arai and Komatsu[46], and Yikuan and Renyu[36]. Arai and Komatsu[46], using various industrial processes, quenched the metal from the liquid state yielding variable cooling conditions. These processes included splat cooling, plasma welding, E-beam melting, sand casting, and, in certain cases, atomized powders and TIG welding. The results presented for H13 included light micrographs showing an as-splat cooled sample, an as-sand cast sample, an as-forged bar sample, and cast and splat samples after soaking each at 1000°C for 25 minutes and oil quenching. The resulting microstructure of the sand cast H13 steel samples appeared to have equiaxed cells, with small amounts of carbide particles decorating grain boundaries. X-ray diffraction lines, spacings, and intensities of as solidified and as annealed H13 steel splat samples were shown, and carbides identified in as-quenched and post-heat treated material (2 hour temper at various temperatures).

A useful pseudobinary diagram for the Fe-C system with 5% Cr has been given by Bungardt, et al[47]. At the carbon level of the H13 alloy (0.37%), one may expect to see a peritectic solidification. However, Asai and Komatsu judged that H13 steel solidified first as primary austenite, and no δ-ferrite was observed by any of the techniques utilized in their investigation[46]. In a paper by Hammar and Grünbaum[48], a projected liquidus curve in the Fe-Cr-C system showed that at the carbon and chromium levels of H13 steel, primary δ-ferrite solidification should occur. This should be the case, since the projection does not take into account the presence of ferrite stabilizing elements found in H13 steel such as vanadium and molybdenum. However, since no evidence of δ-ferrite was found, either none had formed, or if δ-ferrite had formed, all of it transformed to austenite according to the peritectic reaction.

Carbides which were found to precipitate in the H13 steel were determined (via x-ray diffraction) to be M_2C + MC type, rather than M_6C + MC type. More than likely, although not quantified, the M_2C was molybdenum-rich and the MC was vanadium-rich, as Arai and Komatsu stated[46]. After quenching and upon heating, the M_2C was found to precipitate at 650°C and dissolve at about 750°C. Thus, these carbides do not typically form on tempering, since the tempering temperatures are usually around 600°C. Rather, they probably form on cooling during an earlier thermal treatment.

Asai and Komatsu[46] used the presence of vanadium and molybdenum to explain the lack of $M_{23}C_6$ in the as-crystallized structure (for splat cooled samples). In the splat samples (RS samples), the vanadium and molybdenum was found to remain in solution (no apparent segregation of vanadium and molybdenum was observed in RS samples). However, it was unclear whether or not this behavior was observed for the sand casted samples. In alloy systems with greater than 1% vanadium, the precipitation of M_2C was found to occur by Asia and Komatsu[46], but very little M_2C was found in their samples.

Yikuan and Renyu[36] demonstrated the effect of homogenization on the microstructure and unnotched impact properties of cast H13 tool steel. The material investigated was H13 steel cast into a Chinese standard flower ingot configuration, and the "petals" were used to obtain material for unnotched impact specimens. Heat treatments included optional homogenization at temperatures of 1150°C and 1250°C for various times (3 to 5 hours), annealing at 860°C for 3 hours, austenitizing at 1060°C for 20 minutes, and double tempering at 620°C for 2 hours (per temper). These investigators showed that the homogenization treatment decreased the segregation compared to the as-cast microstructure, and homogenized material displayed better impact properties than the material which did not undergo a homogenization treatment. When samples were not homogenized, the segregation effects could be observed microstructurally, even after the hardening and tempering heat treatments. From microscopy and impact data, they determined that the optimal homogenization treatment was 1250°C for 4 hours. Four hours was sufficient to decrease the segregation while preventing a boundary melting phenomenon. The melting was observed in material heat treated at the same temperature for 5 hours or longer. A definite correlation between the unnotched impact energy values and the homogenization heat treatments was observed. In samples that were not subjected to homogenization, the impact energy absorbed was about half that of the samples which were homogenized. Microvoid coalescence was the major characteristic of the samples which were homogenized prior to the normal heat treatments. In samples which were not homogenized prior to normal processing, the fracture surface also exhibited microvoid coalescence, but these samples also exhibited "strips" of brittle fracture, which was attributed to the interdendritic regions. This observation was believed to be the reason for the low (about one-half) impact energy values compared to values for samples which were homogenized. It is evident that few, if any attempts have been made to characterize the as-solidified microstructure of H13 steel with variable cooling conditions. In addition, the resulting properties of cast H13 steel solidified from different cooling rates have been equally neglected.

SERVICE CONDITIONS OF HOT WORK DIE STEELS

In both die casting and drop forging, similar conditions are experienced by the die steel. Both applications introduce either high impact loads or moderate pressures, and both have similar heating and cooling cycles. A typical thermal cycle experienced in the surface layer of a die casting die is seen in Figure 1[49]. This thermal cycling is generally thought of as being most detrimental to the die life, but the wear resistance is also important. The wear properties of H13 steel are important because of new, lower-temperature applications (such as plastic extrusion dies) which have recently surfaced.

Schindler has stated, however, that the die casting pressure is really not important (i.e. has an order of magnitude lower in influence) compared to thermal cycling for heat checking damage[50].

HEAT CHECKING

Heat checking is the term used by many to define the cracking found on the surface of hot work tool materials after being in service. Both manufacturers and users of die casting dies, forging dies, hot rolls, ingot molds, etc. are all concerned with this phenomenon. Other common names for the heat checking phenomenon include craze cracking, network cracking or hot checking, etc. Many believe this problem is one of fatigue produced by cyclic thermal stresses experienced from heating and cooling a material many times (i.e. damage is accumulated) when the material is constrained in some way. Others believe that heat checking is a manifestation of short time creep, in which creep damage accumulates with each impending cycle. Thermal fatigue is the term used to define the damage, regardless of mechanism. Danzer and Sturm[49] have explored the idea that creep plays an important role in heat checking. In their investigation, they analytically modeled a typical stress cycle experienced in a pressure die casting die, and determined that short-time creep may actually be the dominant mechanism for heat checking.

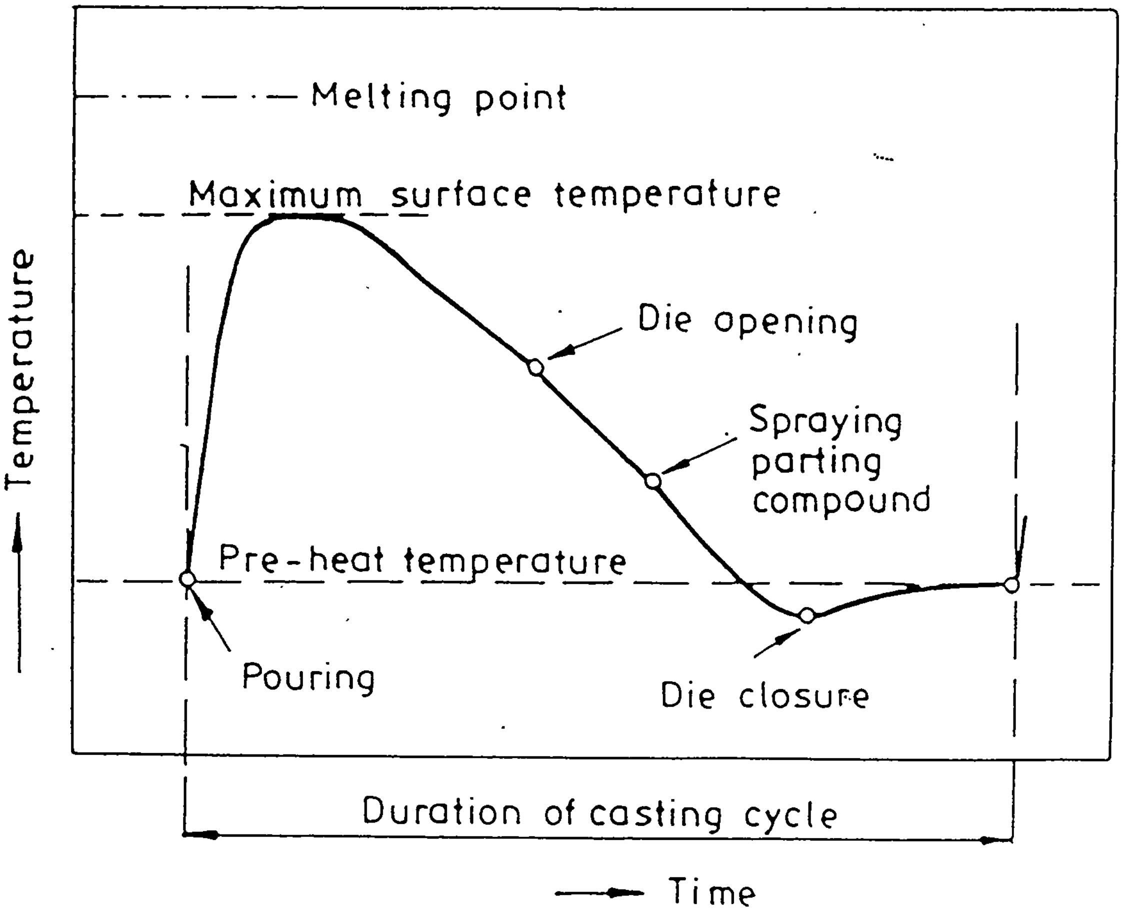


Figure 1. Typical thermal cycle experienced by the surface of a die casting die [49] .

Fundamentally, the stresses in dies and inserts are brought about because the expansion or contraction which would normally result from heating or cooling is constrained by the material underneath the surface layers of the component. If the local stress is sufficient, deformation may occur, and with sufficient deformation, microcracks may initiate. From an industrial viewpoint, failure occurs after a period of time usually much less than the expected life of the component. The dies or inserts are normally removed from service after the heat checks have grown to a critical length. Figure 2 shows schematic diagrams of the temperature distribution, stress distribution, and mechanical strain distribution experienced by die casting dies both on heating (during casting) and on cooling (after the casting has been extracted from the die)[51]. It should be noted, however, that not all of the components for mechanical strain are shown, since the net strain (surface movement) was assumed to be zero. This was taken into account in the calculations of Noesen and Williams[51], who attempted to utilize Coffin's low cycle fatigue model and determine the number of cycles to failure (i.e. number of thermal cycles to initiate cracks) based on the cyclic plastic stain experienced in each cycle. It can be seen from these diagrams how both the stress and mechanical strain exhibit alternating amplitudes in compression and tension. These reversals in stress and strain amplitudes are believed to cause crack nucleation. After these cracks have initiated, propagation of the cracks may occur by a number of methods. In die casting and forging, continued thermal cycling, oxidation wedging and impact may all be mechanisms for propagating heat checks.

The life of a die may be lowered even more if there are microstructural changes which occur during the thermal cycling. These changes may include bulk phase transformations such as re-austenitization on heating, transformations to martensite or bainite on cooling, and over-tempering of the surface of the material, in which temper carbides are coarsened and the hardness of the surface layers decreases dramatically. The bulk phase transformations which may occur introduce transformation stresses which accentuate the thermal stresses and decrease the die life. The over-tempering reactions would simply decrease the resistance to flow of the surface material, thus increasing the surface plasticity and decreasing the life of the die.

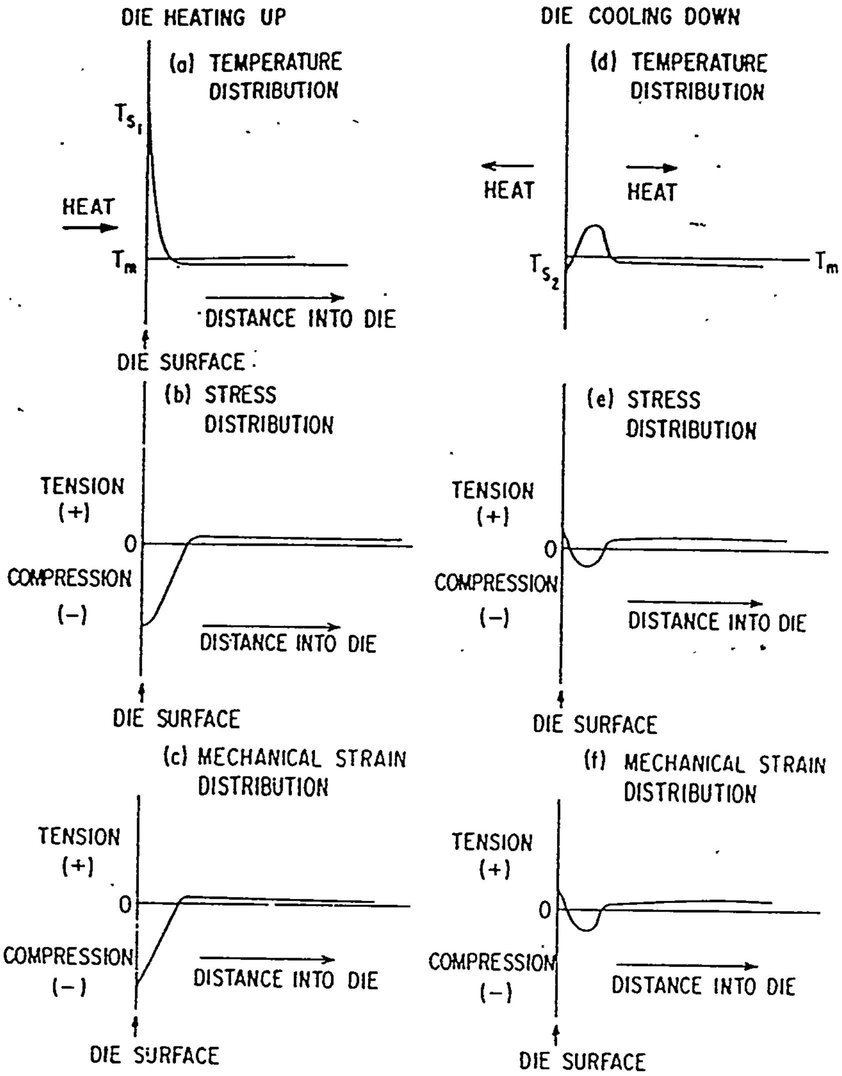


Figure 2. Schematic temperature distribution, stress distribution and mechanical strain distribution experienced by die casting dies on heating and on cooling (adapted from Noesen and Williams [51].

PREVIOUS EXPERIMENTAL HEAT CHECKING WORK

The first thorough examination of thermal cycling was performed in the mid-1950's by Northcott and Baron[52]. They developed a wedge-shaped thermal fatigue specimen, which was heated by a high frequency induction coil and quenched by a water-cooled copper block. After a certain number of cycles, the sample was removed, sectioned, and microscopically examined. The advantages of their experimental design included ease of modeling stresses and temperature control. The major disadvantage was the complex experimental configuration and the large sample size.

Glenny and Taylor[53] first used a fluidized bed system for heating tapered disc samples in 1959. These 1.625-in.-diam. by 0.275-in.-thick samples were tapered to a point. In 1969, Rostocker[45] expanded on Glenny and Taylor's idea by utilizing fluidized beds for both heating and cooling 2.25-in.-diam. by 0.125-in.-thick disc samples. The major advantages of this experimental design included the small sample size and low-cost fluidized beds. One disadvantage was the difficulty in machining tapered discs, especially since many of these materials in the as-cast condition exhibit quite high hardnesses (higher than 50 HRC). In addition, the temperature distribution in the discs did not duplicate the conditions experienced in actual dies or inserts.

In the mid-1960's, Wollering and Oertle[54] studied heat checking with a rotating specimen holder and a oxy-acetylene flame as the heat source. Immediately following the heat cycle (4-5 seconds), samples were rotated into a water bath for quenching. This automated test left concentric rings due to the circular nature of the flame, thus indicating the temperature distribution on the surface. The sample dimensions for this investigation were 1 inch x 1 inch x 1.5 inches. Results from this test were similar to those from the wedge test. After a certain number of cycles, the samples was microscopically examined to determine the extent of heat checking. Advantages of their setup included low-cost, as well as the ability to test several samples at once. The most considerable disadvantage, however, was that good temperature control (and thus stress distribution control) was lacking.

In the early 1970's, a "dip test" was developed at Case Western Reserve University by Benedyk, et al[55]. A similar test was used by Danzer, et al. in the 1980's[56]. The Case Western test utilized a rectangular sample (2 in. x 2 in. x 7 in. long) with a 1.5-in.-diam., 6.5-in.-deep hole drilled into it so that the sample could be internally water-cooled. The sample was then dipped into a bath of molten aluminum and systematically extracted and cooled for a number of cycles so that the sample encountered a thermal cycling experienced by actual die casting dies. After a certain number of cycles, the sample was examined under light to count and measure the cracks. Although this test best simulated the environment encountered, it was only useful to test larger sections; however, in thinner sections, such as those which may be encountered in precision investment cast dies, examination of the heat check susceptibility of thinner sections (<2 inch thick sections) was not possible. In addition, separation of heat checking mechanisms (i.e. purely thermal fatigue/creep or liquid metal-die interactions) was not possible for this testing configuration, and samples were difficult to machine.

Although resistance-heating was used previously in thermal fatigue experiments, resistance-heating was first utilized by Korostolev and Kunilovsii[57] to investigate heat checking and die life in the mid-1970's. Although not much information was available concerning the details of their test, advantages of this type of test include small sample size and the ability to control heating and thermal gradients. However, an external constraint was utilized rather than an internal constraint, as is the case when resistance heating is used. In addition, localized overheating of the crack tip may be realized from a buildup of current.

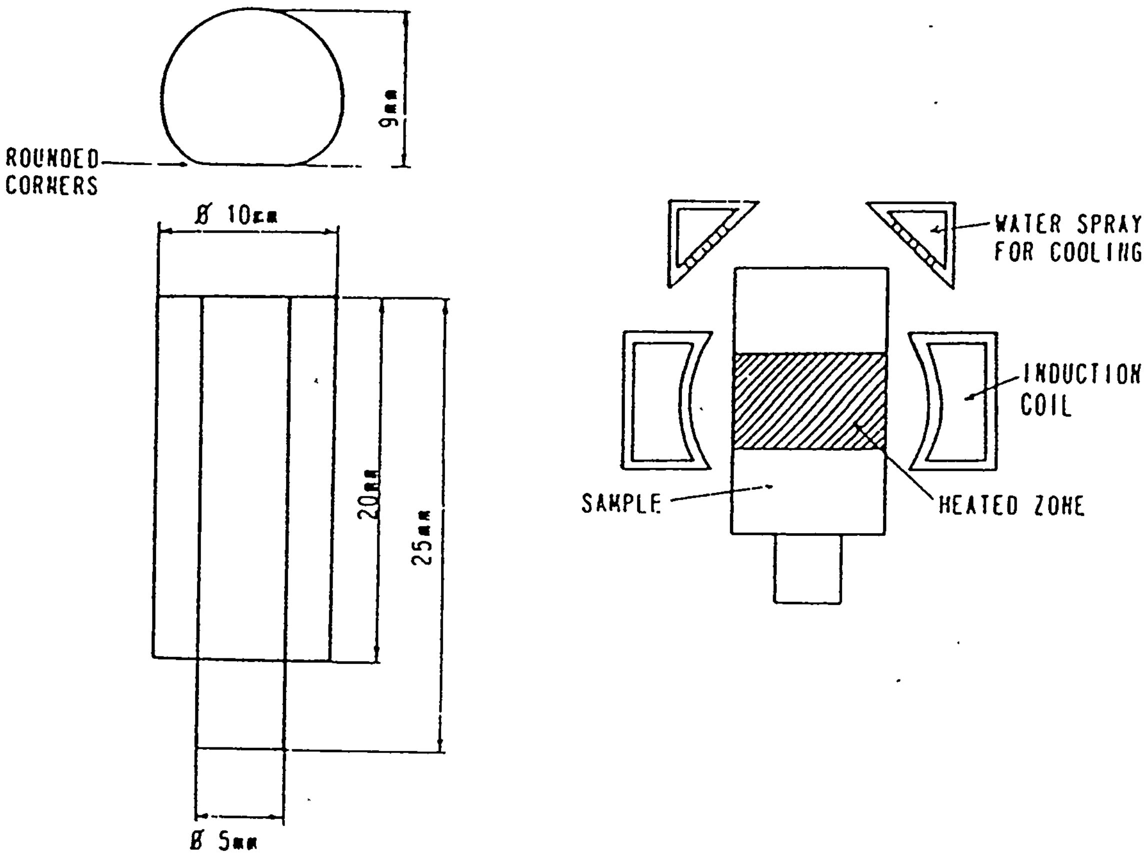
More recently, high-frequency (HF) induction heating of small scale samples has been performed by both Holmes and McClintock[58] using disc-shaped samples and by Malm, et al.[59] using cylindrical shaped samples. After the disc-shaped samples were heated, quenching was performed by forced air. Holmes et al.[60] performed thermo-elastic finite element analysis on the discs to predict the stress-strain behavior of the material during the heating and cooling. Since the disc-shaped samples were so thin in this test, the constraint was not the same as that experienced by dies (i.e. the temperature gradient through the material did not occur in the discs).

Malm, et al.[59], from Uddeholms AB, used HF induction heating of 10-mm.-diam. by 20-mm. long cylinders for an industrial test to determine heat checking susceptibility of various hot work die materials. A flat 5-mm. wide was ground along the length of the sample, so that heat checks would initiate and grow in this area and microscopical examination could be performed after a certain number of cycles. The sample geometry is shown in Figure 3a. After heating, the sample was then water sprayed for quenching. A schematic diagram of their setup is shown in Figure 3b. A temperature gradient (and thus constraint) was created such that it was similar to that experienced by dies and molds. Geller, et al.[61] used HF induction heating to characterize heat checking in various heat-resistant materials. They demonstrated how a material similar to H13 steel in composition resisted heat checking and interaction with liquid metals better than many other die casting mold materials.

MICROSTRUCTURAL ASPECTS OF HEAT CHECKING

Provided no significant microstructural changes (such as bulk phase transformations) occur on one thermal cycle, heat checking is caused by the simple cycling of temperatures. Berns[5], who studied various H13 steels in terms of microstructure and properties, observed that some thermal fatigue cracks initiated at or near nonmetallic inclusions and quasi-eutectic carbides in as-cast induction-heated and water-quenched samples. However, his treatment of this phenomenon was rather light, and the study concentrated more on other mechanical properties. Zleppnig, et al.[62] have also observed that carbide particles or inclusions were associated with many heavy cracks. These second phase particles, however, were always larger than 15 μm in diameter, regardless of chemical composition or physical characteristics.

Sample homogeneity and second-phase distributions were also shown to have an influence on thermal fatigue on cast superalloys, as Woodford and Mowbray have demonstrated[63]. They used an experimental apparatus similar to Glenny and Taylor[53] (thin discs in fluidized beds) to demonstrate the effect of microstructure on thermal fatigue cracking. It was apparent how cracking



(a)

(b)

Figure 3. Uddeholm's heat checking test [59] : (a) sample geometry and (b) schematic of their setup.

proceeded along networks of different types of carbides. For instance, in various Co-based alloys, cracks propagated along networks of $M_{23}C_6$ or MC carbides. In various cast Ni-based superalloys, a similar behavior was observed. It was not apparent, however, whether the carbides were splitting ahead of the crack tip. In addition, they found that cracks tended to grow in interdendritic regions, where the eutectic carbides were observed. In a coarse-grained Ni-based superalloy, some cracks were observed to initiate at grain boundaries. These investigators also revealed that fine-grained material inhibited the thermal fatigue crack initiation, but that the coarse-grained material slowed the propagation.

Udoguchi and Asada[64] have discussed, in terms of dislocation theory, what they believe occurs during these repeated thermal cycles. According to these investigators, the reversal of stress or stress amplitude, if large enough, may exceed the plastic limit, which, in turn, yields repeated plastic strain values. This repetition increases the vacancy density on the surface layer to the point where vacancies are supersaturated. Finally, the vacancies coalesce to create voids, which grow into microcracks through continual absorption of generated vacancies. This argument may also be used to describe grain boundary decohesion which occurs during creep.

Another mechanism which could definitely play an important role in the heat checking behavior of materials is the susceptibility of the material to degrade microstructurally (loss of hardness or ductility). This could be caused by the over-tempering reactions, which include the transformation of secondary hardening precipitates (such as VC) to coarse precipitates which are not as effective in strengthening the matrix (such as M_6C or $M_{23}C_6$). A model regarding the structural degradation of the matrix was discussed by Engberg and Larsson[65] in terms of carbide coarsening. A number of investigations have shown that softening occurs during thermal cycling[14,45,55,66-68]. Rostocker[45] discovered that the softening of H13 steel occurs in areas of both high and low thermal strain on thin disc samples. Benedyk, et al.[55] have shown that softening occurred more rapidly on the corners of the "dip-test" sample, and this softening was more rapid than that observed on a conventional isothermal low-cycle fatigue test at a temperature similar to the peak temperature of the thermal cycle. Microstructural evidence of the carbide coarsening and precipitation was also presented by these investigators. Rostocker[45] and Benedyk, et al.[55] have further shown that this softening is more rapid than predicted by over-tempering the microstructure. Although these results are somewhat inconclusive, studies have shown that the rate of softening is accelerated in the 500-900°C temperature range by the application of thermal stress.

SUMMARY

The major research effort on cast tooling occurred during the 1950's through the 1970's. It is believed that the decline of research in the 1970's and 1980's was mainly because the time and cost required to produce a wax injection mold or a wooden pattern in order to cast a part was approximately the same as simply machining the part from wrought material. With the advent of rapid prototyping in the 1990's, the time required to produce the die or pattern has vastly decreased and cast tooling has become a popular topic again. Issues such as performance of cast tooling are now being revisited. The impact of the lower mechanical properties associated with cast tooling can be reduced if the tooling is designed to be a casting from the beginning. Some have claimed that as-cast tooling has a service life as long, if not longer, as conventionally machined tooling. In addition to the time reduction realized with cast tooling, increasing the complexity of a part is inexpensive relative to machining. Also, rapid prototyping enables the rapid generation of a pattern and therefore the flexibility for design changes that does not exist with machining.

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